

PROYECTO MULTINACIONAL DE METALURGIA y
PROYECTO ESPECIAL DE ALUMINIO Y COBRE

Sexto Seminario Latinoamericano en el Nivel
de Post Doctorado

SOLDADURA

R. L. APPS

AGOSTO 1980

PMH/A-282

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DISTORTION AND ITS CONTROL

Distortion is the change in shape of a welded component which results from the welding operation. This change in shape may only be temporary but is more often permanent.

The principal cause of distortion is the physical law of expansion and contraction. Generally the metal being welded is differentially heated i.e. heating is not uniform. The metal is subjected to steep temperature gradients and as it is heated it becomes weaker and more easily deformed in these parts which are the hottest. Table 1 shows how the mechanical properties of steel change with temperature.

Table 1

Mechanical Properties of Steel at Elevated Temperatures

Temperature °C	U.T.S. Kg/mm ²	Yield Point Kg/mm ²	Elastic LIMIT Kg/mm ²	Elongation % on 2in.	R.A. %
40	37.8	27.5	24.0	48	62
100	31.5	28.4	23.0	43	66
150	41.0	32.5	21.0	39	65
200	42.0	33.1	20.0	37	64
250	44.2	32.0	12.0	37	64
300	46.2	30.0	9.5	40	66
375	42.0	28.0	8.0	44	68
425	36.5	26.0	7.0	49	76
500	30.00	23.0	6.0	55	84
540	24.00	20.0	5.0	60	90
600	16.20	15.0	3.5	64	94
650	13.50	12.0	2.5	68	96
700	9.50	9.0	1.5	72	98

The tendency to distortion is aggravated by the restraint at the weld joint. That is to say whilst the weld metal is being heated, it is prevented from expanding as much as it normally would by

the colder adjacent metal. On the cooling cycle, the weld metal contracts but is prevented from contracting fully by the surrounding cooler metal. However, the weld metal is put into tension and usually plastic flow occurs in the weld metal as the assembly adjusts in dimensions to the lowering temperature. Obviously as the temperature gets lower, so larger stresses are required to continue the plastic flow in the weld metal.

If one takes a simple example of a square block of steel. If this is heated uniformly and thoroughly, it will expand in all directions; if it is allowed to cool evenly and without restraint of any kind, it will contract to its original shape and size. If however, it is held in a vice and heated so that it cannot expand in one direction, it will only be able to expand in the two lateral directions (see Figure 1). Then upon cooling, it is able to contract in all directions and therefore does not return to its original shape and size, but ends up shorter and thicker (Figure 1c). Hence, a return to the original shape and size is only possible when a part is allowed to expand and contract freely, without restraint.

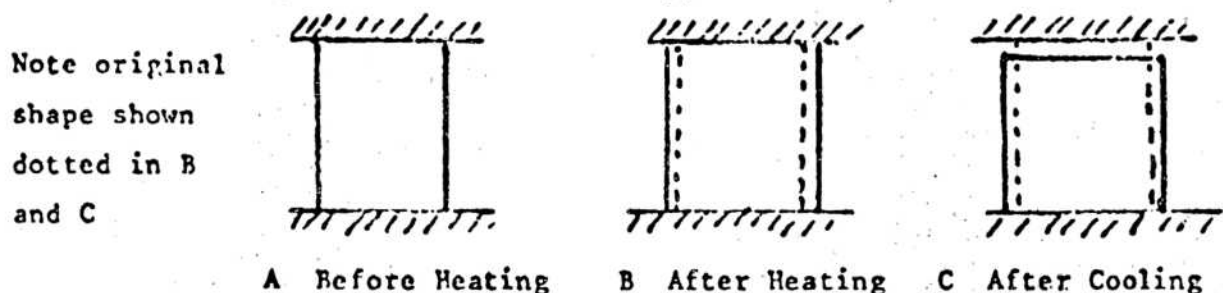


Figure 1. Effect of Heating and Cooling on a Restrained Block of Metal.

When a bar is heated over a small area, the expansion is local and uneven. The mass of cool metal adjacent to the heated area will not expand, and will tend to prevent expansion of the heated metal in all directions except at the surface. Consequently, when the yield point is reached, permanent deformation takes place and upon cooling the bar does not return to its original shape, it has become distorted.

As far as welding is concerned, we are usually heating and cooling a localised area, and distortion will inevitably result unless the parts are restrained from moving in some way. Generally the greater the distortion, the smaller the plastic flow or strain in the weld metal.

The factors which govern distortion are as follows:-

- a) Resistance to the free expansion of the weld metal.
- b) The temperature gradient
- c) Coefficient of expansion of the material and

d) Yield strength of the weld metal and the base material.

Factor (a) will depend on the shape and size of the whole structure and the amount of restraint applied by jigs and fixtures. Factor (b), temperature gradient, will depend on the rate at which heat is applied and how quickly it is conducted away from the weld area. The rate of heating will be governed by the process and the particular process parameters employed. The way the heat is conducted away will depend on the joint geometry and general configuration of the component near the weld area. It will also be affected if preheat is applied.

Control of Distortion

There are three main ways of controlling distortion:-

- a) Forcible restraint of the parts being welded
- b) Weld so as to balance the forces which act to produce distortion
- or c) Stretch the metal, by mechanical means or by heating to counterbalance distortion.

The effect of all three methods is to increase the plastic flow of the weld metal as it cools. In consequence, over application of any of these methods could result in weld metal or h.a.z. cracking. Hence, they should be applied with care, particularly when the metal is susceptible to hot or cold cracking.

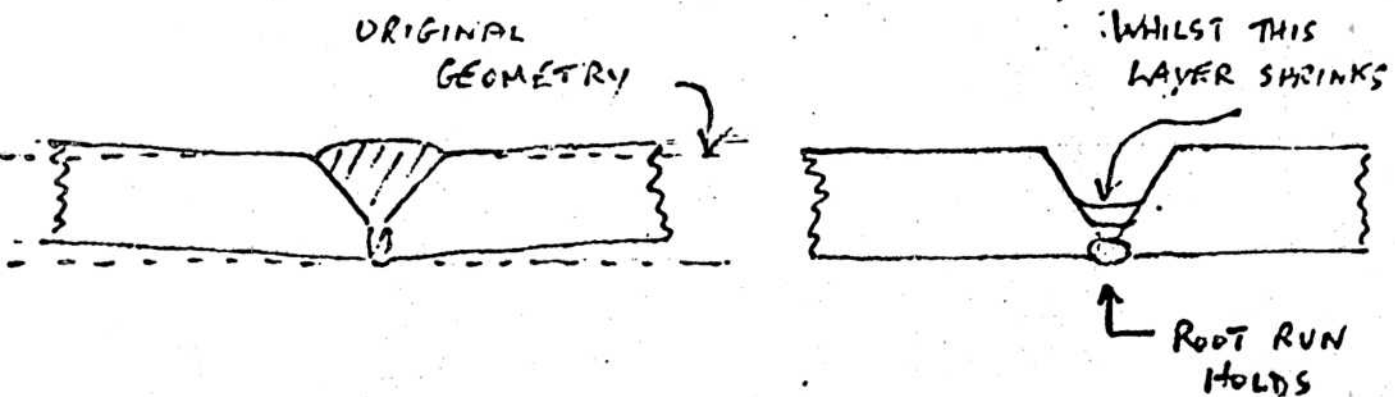
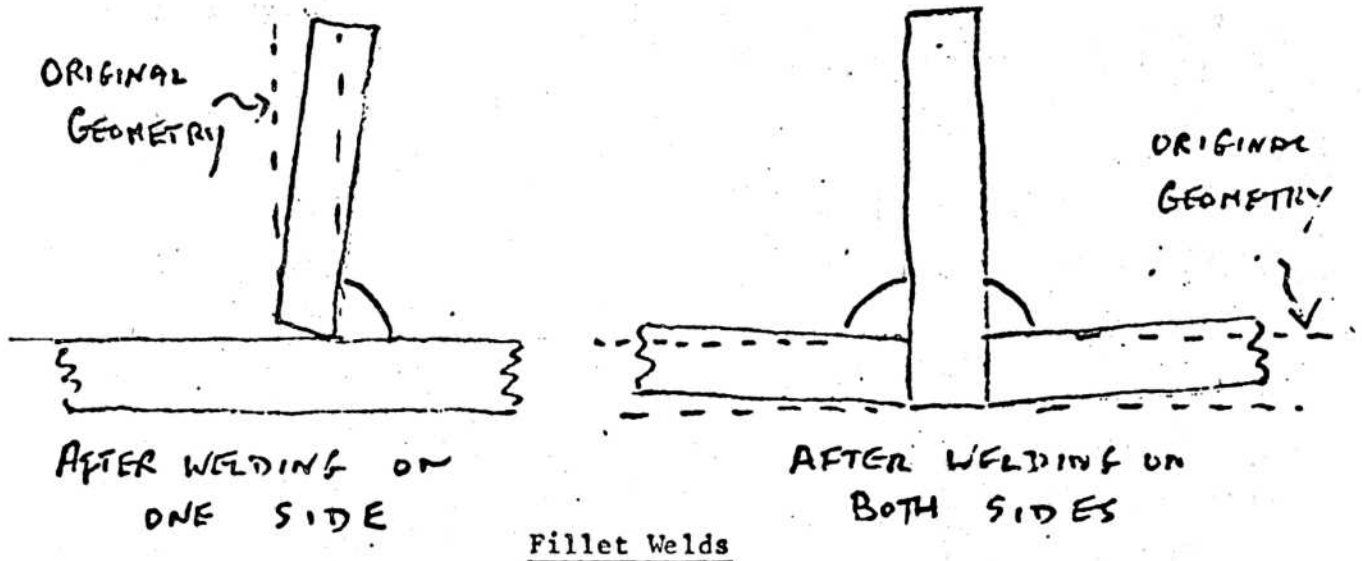
Method (c), stretching, is often done by peening, i.e. stretching the metal by a series of hammer blows, preferably whilst it is still hot. Peening exerts a compressive force on the weld metal normal to the plane of maximum tensile stress and this aids plastic flow.

The greater the amount of weld metal deposited, the greater will be the shrinkage forces, and hence it is undesirable to deposit too much metal. The strength of a joint is determined by its design and nothing useful is added by filler metal in excess of that required to meet the design.

Shrinkage can also be minimised by control of heat input. This can be done by welding at lower amperages or by torch manipulation in gas welding.

One can also minimise distortion by correct welding procedure and by careful control of the sequence of welding. Intermittent welding or the use of a wandering sequence helps to prevent local build up of heat and hence reduces shrinkage.

The effective shrinkage force can be reduced by placing welds as near as possible to the neutral axis. The closer to the neutral axis, the less is the leverage exerted to pull the sections out of alignment.



Butt Welds

Figure 2 Angular Distortion and Shrinkage

An effective way to control and reduce distortion during welding is to restrain the parts forcibly. This method will even eliminate distortion completely, in certain operations, if the restraint is sufficiently great. Greater restraint, however, increases the magnitude of stresses during welding. This applies particularly to the transverse stress, hence the greater the tendency to cracking in this direction.

Planned shrinkage may solve the problem of distortion when design permits the desired arrangement of parts. If the vertical plate of a T weld is positioned out of alignment before welding begins, the shrinkage during welding will pull it into its 90-deg position. Parts can be positioned

with sufficient space between them to compensate for shrinkage. When welded, the contraction of the metal will pull the parts into their correct position.

Shrinkage forces may be balanced one with another either by predistortion, such as prebending the parts to be welded, or by the use of the proper welding sequence.

When the weld metal is deposited at different points of a structure, it is possible to counteract the shrinkage forces of one weld by the normal contraction of another weld. A practical application of this technique is the alternate welding of both sides of the neutral axis of a butt weld.

Another approach is the intermittent welding of structures where the shrinkage forces of Deposit One are balanced by Deposit Two. Deposit Three is applied in such a position that as Deposit Four shrinks, it will counteract the shrinkage of Deposit Five..

In all welded operations where distortion is a factor to be avoided, the direction of welding should be away from the point of restraint, and toward the point of maximum freedom.

Practical Considerations

In general, fillet welding produces longitudinal and transverse shrinkage plus angular rotation of the adjacent plate surfaces. The longitudinal shrinkage is generally about 1/32 inches in 10 feet of weld, and transverse contraction about the same order: i.e. for fillet sizes appropriate to the material thickness (for mild steel).

Butt welds produce longitudinal and transverse shrinkage and may produce an angular rotation of the plate surfaces, depending on the preparation and weld sequence employed. Transverse contractions for mild steel are of the order of 1/16 inch, and longitudinal shrinkage about 1/32 inch per 10 feet of weld. Obviously factors such as fit-up, size of electrode, welding process used will all affect the total weld contraction.

For unrestrained butt welds between mild steel plates, an empirical formula has been suggested as follows⁽¹⁾:-

$$S = 0.18 A^3 t + 0.12 W$$

where S = the transverse shrinkage in inches (or mm)

A = cross section of weld (sq. in.) (mm²)

t = plate thickness (in.) (or mm)

and W = average width of weld (in.) (or mm)

Other metals will give greater or less distortion according to their coefficients of thermal expansion. Aluminium and its alloys show about twice the amount of shrinkage as does mild steel. Stainless steel expands just over 1½ times as much as mild steel.

Points to consider in relation to avoiding or minimising distortion are:-

- (1) Specify and use the minimum amount of weld metal consistent with strength requirements.
- (2) If possible, balance welding about the centre of gravity of the job.
- (3) Allow extra length to cater for possible shrinkage.
- (4) Pre-bend unbalanced sections so that angular distortion will pull back into line.
- (5) Use largest practical size of electrode to deposit the maximum amount of metal in the minimum time.
- (6) Ensure good fit-up.
- (7) If using jigs, ensure that they are strong enough and do not bend as the job moves.
- (8) Cut sections to length after completion of all welding. Equally, drill any important holes, etc. after welding.

The following figures (3-5) illustrate the type of distortion experienced in butt and fillet welds and Tables II and III quantify the amount of angular distortion which may be encountered in butt and fillet welds respectively.

Figure 6 shows the effect of the number of runs on angular distortion of a ½ inch fillet weld (H.V. Position).

Accuracy of fit-up is important because, if the gap is too wide in places more weld metal is required, hence more heat and more distortion. The thinner the material, the more important the accuracy of fit-up becomes.

For the thinner materials a considerable amount of tack welding must be done - tack welds light and close together, or the work must be tightly and accurately jigged so that it will not move.

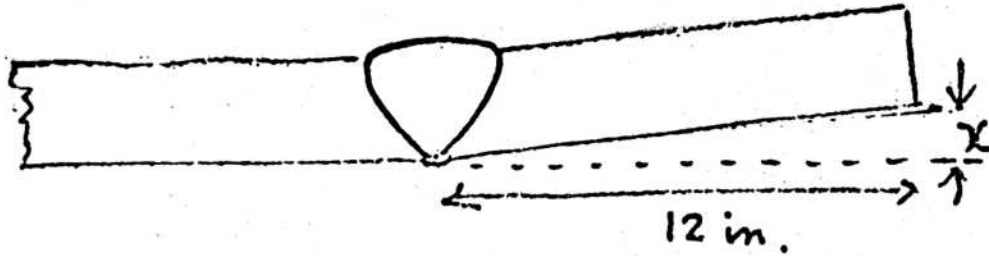
As far as rectifying the effects of distortion is concerned. It is possible to pull back distorted components by the judicious application of heat in the correct places. This usually done by heating with gas blow-pipes.

Figure 7 illustrates the method of straightening the convexed edge of a plate by selective heating of wedge shapes areas using gas blowpipes. By this means, using a great deal of skill and experience on the part of the operator, many distortion effects can be overcome. However, it is generally better and less costly to do everything possible to avoid distortion in the first place.

Referring to Figure 7, it is important that the heat is applied uniformly through the plate section. For this reason, two blowpipes are applied, one from each side. With full heating of two thirds of the width of the plate, heated zone 2 is wide, the straightening effect is 3/16 to 1/2 inch for every 12 feet length of plate. Heating should be to a uniform red heat ($\sim 900^{\circ}\text{C}$) throughout the segment.

Table II

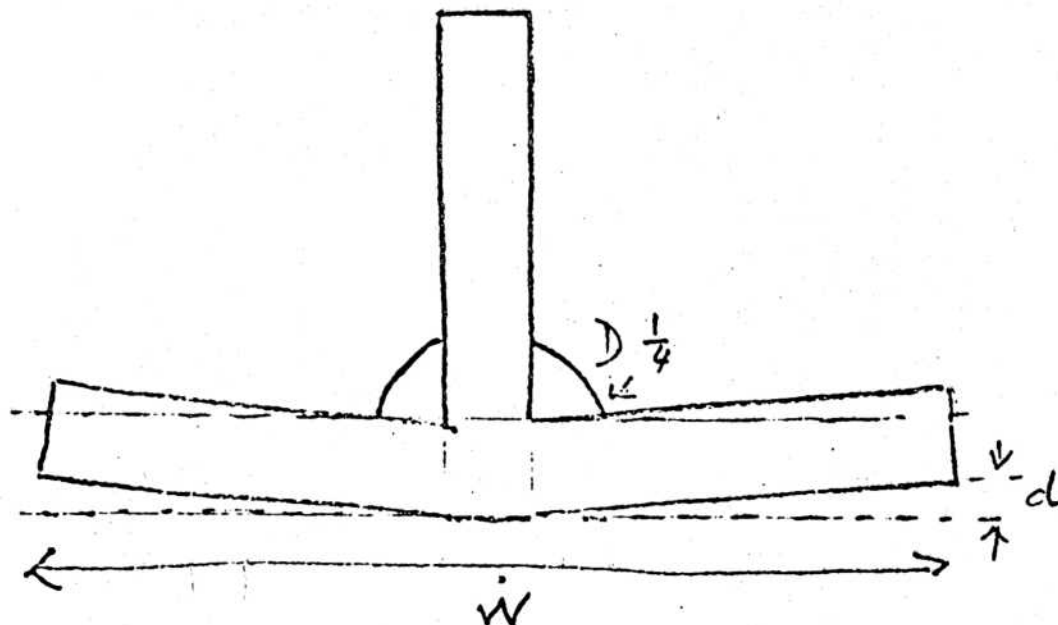
Angular Distortion in Butt Welds



No of Runs	X (in)	No of Runs	X (in)	No of Runs	X (in)
1	3/16	11	15/16	21	1 9/16
2	3/8	12	1	22	1 5/8
3	7/16	13	1 1/8	23	1 11/16
4	1/2	14	1 1/4	24	1 3/4
5	9/16	15	1 3/16	25	1 11/16
6	5/8	16	1 1/4	26	1 13/16
7	11/16	17	1 5/16	27	1 15/16
8	3/4	18	1 3/8	28	2
9	13/16	19	1 7/16	29	2 1/16
10	7/8	20	1 1/2	30	2 1/8

Table III

Distortion in Fillet Welds



Distortion
"d" in
1/64 in
with 1/4
fillet

Width W (in)	THICKNESS t (in)							
	1/4	1/2	3/4	1	1 1/4	1 1/2	2	2 1/2
6	3	2	1 1/4	1	1	-	-	-
9	6	5	3	2	2	2	1	1
12	9	7	5	3	2	2	2	1
15	12	9	6	5	4	3	2	2
18	15	11	7	5	4	4	3	2
21	18	14	9	7	5	5	3	3
24	21	16	10	8	6	6	4	3

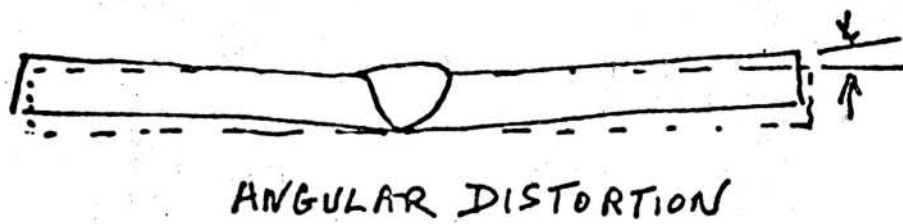
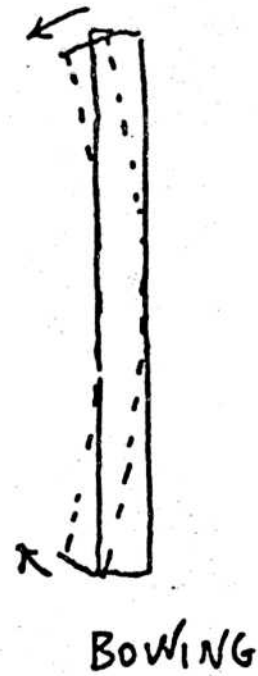
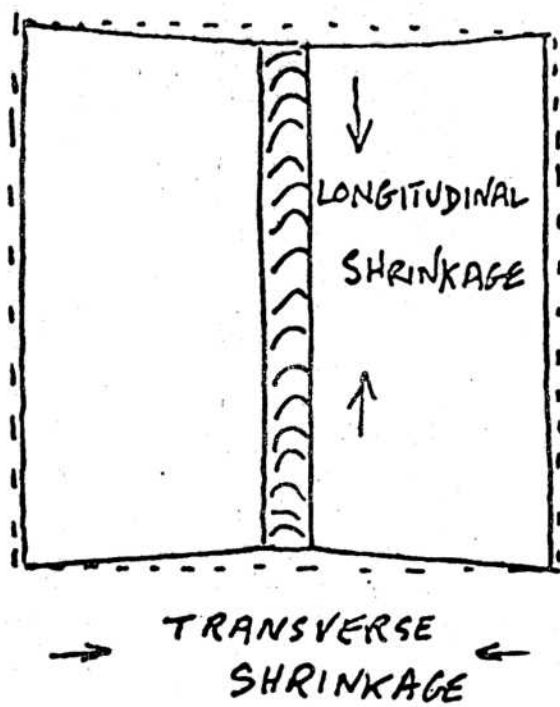


FIGURE 3. TYPES OF DISTORTION
IN BUTT WELDS.

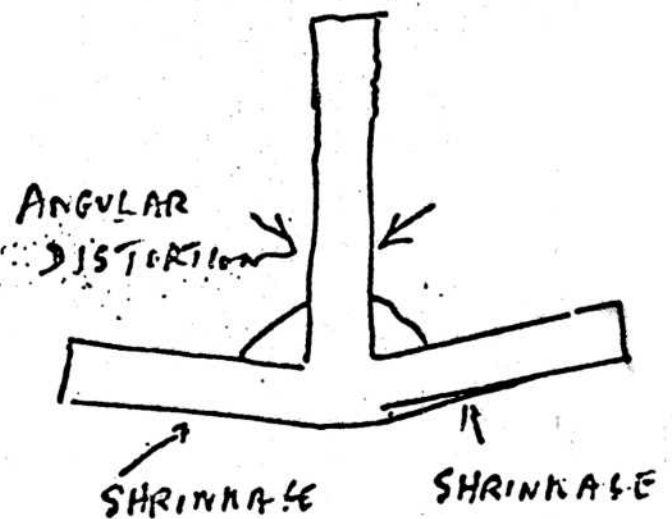
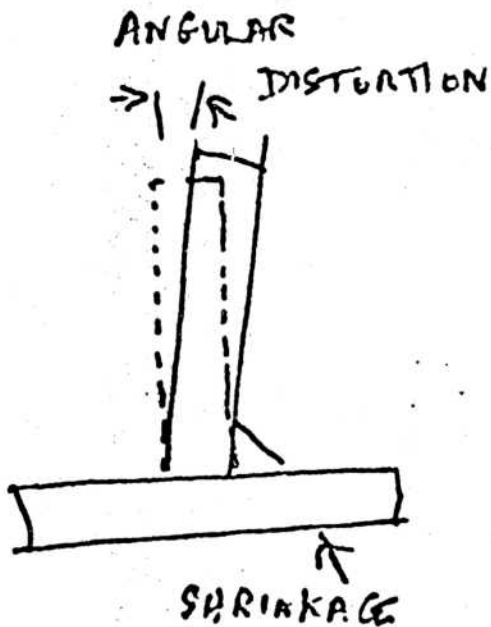
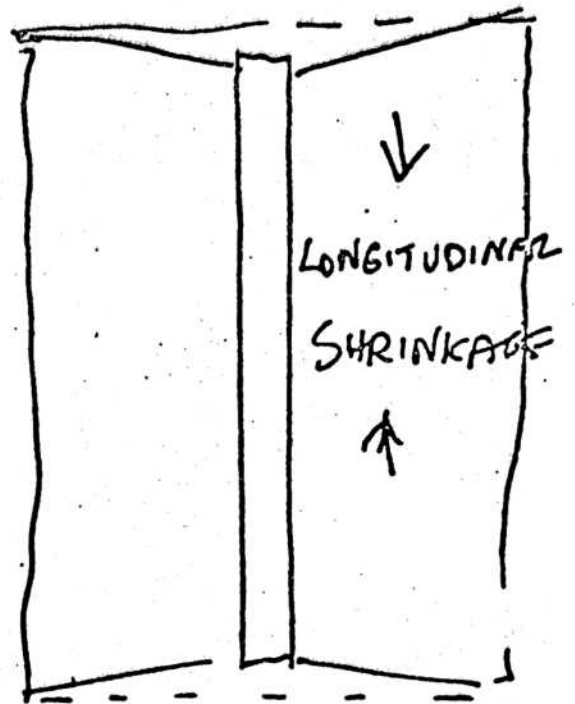
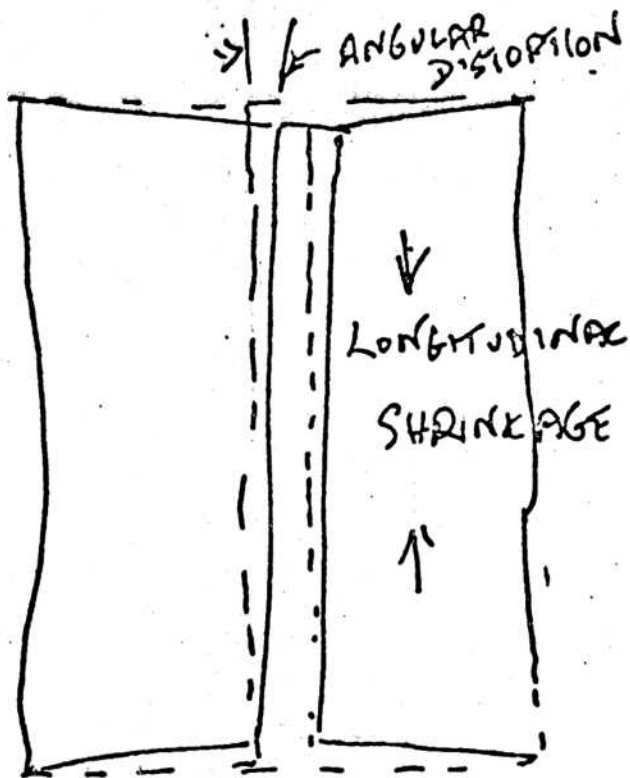


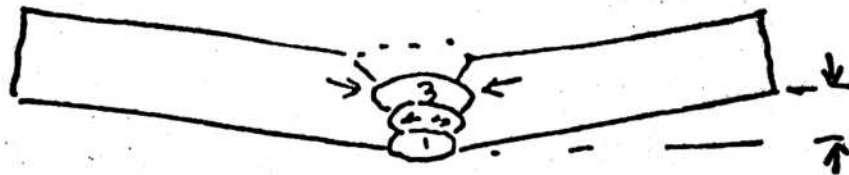
FIGURE 5. DISTORTION EFFECTS IN FILLET WELDS.



FIRST RUN CONTRACTS AND
DRAWS PLATES TOGETHER



CONTRACTION OF SECOND RUN IS
OPPOSED BY THE FIRST RUN



CONTRACTION OF THIRD RUN
INCREASES ANGULAR DISTORTION



COMPLETION OF A BACK RUN
LEAVES CENTRE OF WELD IN COMPRESSION
- OUTSIDE OF WELD IN TENSION.



USE OF DOUBLE VEE AND SYMMETRICAL
WELDING OBVIATES ANGULAR DISTORTION
AND PRODUCES ONLY A SMALL
TRANSVERSE SHRINKAGE

FIGURE 4. ANGULAR DISTORTION
IN MULTI-RUN BUTT WELDS.

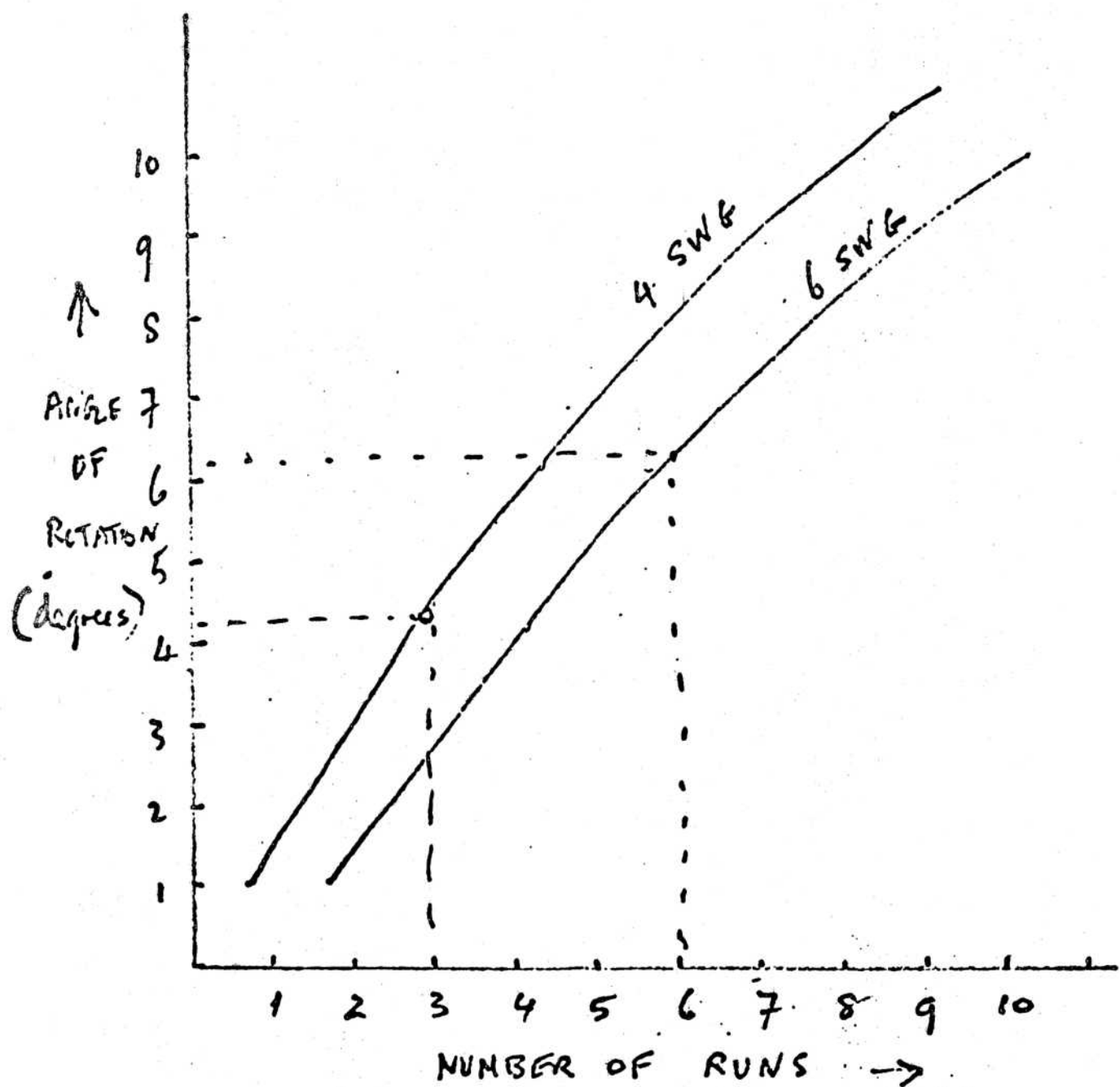


FIGURE 6. EFFECT OF NUMBER OF RUNS ON ANGULAR DISTORTION.
 $\frac{1}{2}$ IN FILLET WELD (HV. POSITION - MANUAL METAL ARC).

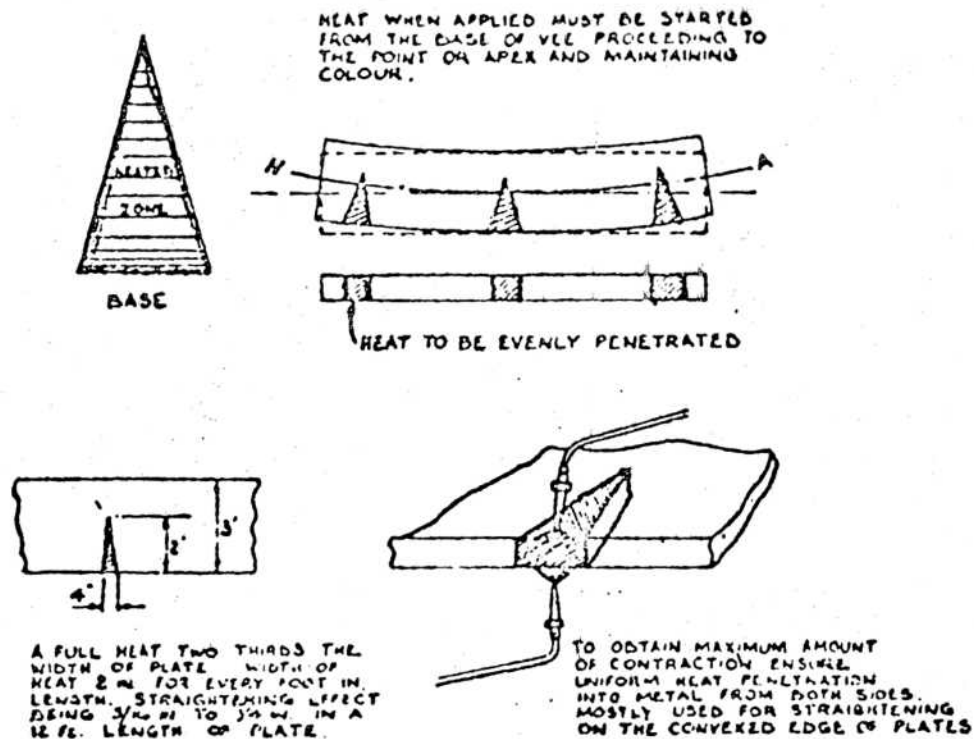


FIGURE 7. METHOD OF APPLYING HEAT TO STRAIGHTEN CONVEX EDGE OF PLATE

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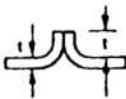
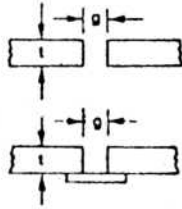
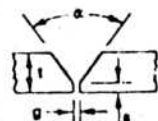
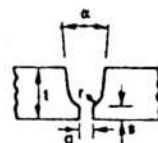
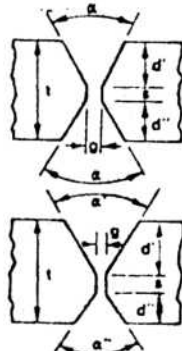
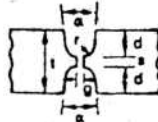
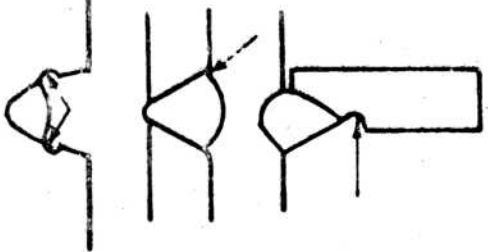
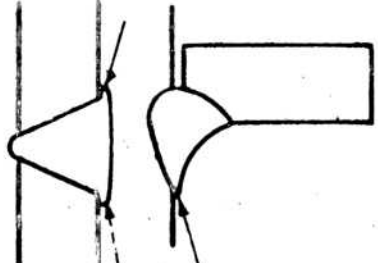
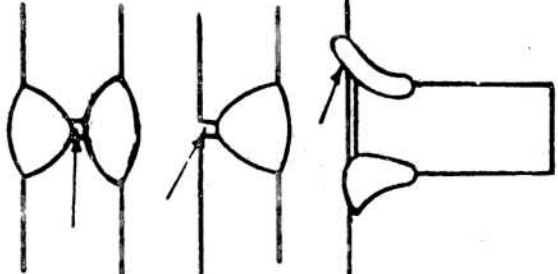
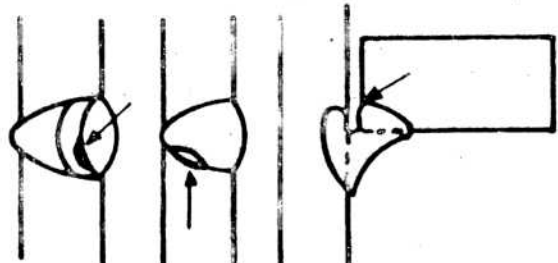
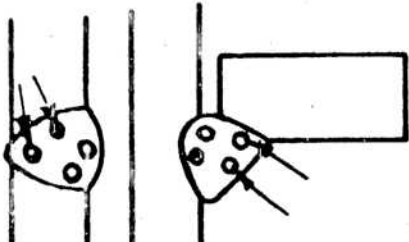
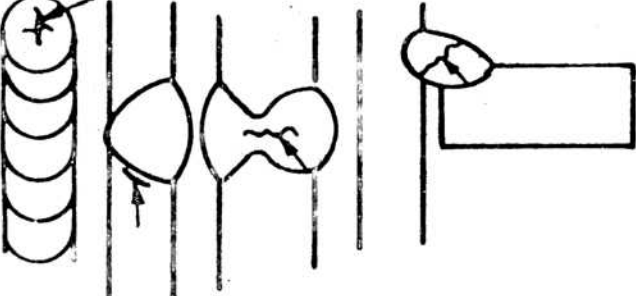
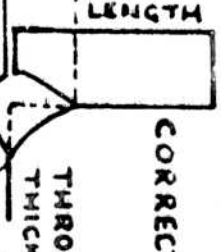
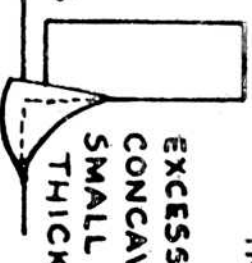
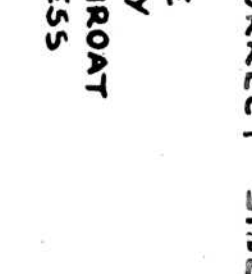
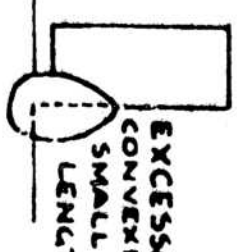
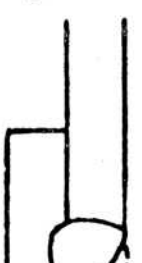
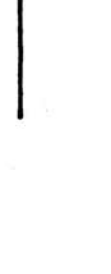
Preparation	Process	Thickness <i>t</i> in.	Gap <i>g</i> in.	Root <i>S</i> in.	Radius <i>r</i> in.	Included Angle α	Depth <i>d</i> in.
(i) Flanged 	Oxyacetylene Atomic Hydrogen Carbon Arc TIG Submerged Arc	1/16 max. 5/16 max.	— —	— —	— —	— —	— —
(ii) Square Butt 	Oxyacetylene, 1 side only both sides Shielded Metal Arc Submerged Arc, both sides Shielded Metal Arc, steel backing Deep penetra- tion, steel backing Submerged Arc, steel backing copper backing	1/4 max. 1/6 to 1/2 1/16 to 3/8 1/8 to 5/8 1/8 to 5/16 3/16 to 5/16 1/4 to 3/8 1/8 to 5/16	$\frac{t}{2}$ to $\frac{t}{2} + \frac{1}{32}$ $\frac{t}{2}$ to $\frac{t}{2} + \frac{1}{32}$ 0 to 3/32 1/32 max. 0 to 5/16 0 to 1/8 5/32 to 3/16 0 to 1/8	— — — — — — — —	— — — — — — — —	— — — — — — — —	— — — — — — — —
(iii) Single-V 	Oxyacetylene Shielded Metal Arc Submerged Arc	1/8 to 1/2 3/16 to 3/4 1/4 to 1 1/2	$\frac{t}{4}$ to $\frac{t}{4} + \frac{1}{32}$ $\frac{1}{16}$ to $\frac{3}{16}$ $\frac{1}{32}$ max.	0 1/8 max. 1/8 to 3/8	— — —	60–70° 60–70° 45° min	— — —
(iv) Single-U 	Shielded Metal Arc Submerged Arc	3/4 to 2 3/4 to 2	$\frac{3}{32}$ to $\frac{5}{32}$ $\frac{1}{32}$ max.	$\frac{1}{16}$ to $\frac{3}{16}$ 1/4 min.	1/4 1/4	20–45° 20° max.	— —
(v) Double-V 	Oxyacetylene Shielded Metal Arc Submerged Arc Shielded Metal Arc Submerged Arc	Over 1/2 5/8 to 1 1/2 3/4 to 2 1 to 3 1/2 1 to 3 1/2	$\frac{1}{32}$ to $\frac{3}{32}$ $\frac{1}{16}$ to $\frac{7}{32}$ $\frac{1}{32}$ max. $\frac{1}{16}$ to $\frac{7}{32}$ $\frac{1}{32}$ max.	0 0 to 1/16 1/4 min. 0 to 1/16 1/4 min.	— — — — —	70–80° 60–70° 45° min. α' 60–70° α'' 60–90° 45° min.	Over $\frac{1}{4}$ Over $\frac{5}{8}$ $d' \frac{2}{3} t$ $d'' \frac{1}{2} t$ $d' \frac{2}{3} t$ $d'' \frac{1}{2} t$
(vi) Double-U 	Shielded Metal Arc Submerged Arc	Over 1 1/2 Over 1 1/2	$\frac{1}{32}$ to $\frac{5}{32}$ $\frac{1}{32}$ max.	$\frac{3}{32}$ to $\frac{1}{8}$ 1/4 min.	1/4 1/4	20°–45° 20°	— —

FIG. 2.3. Joint preparations recommended for the various processes used for welding mild steel. (See "Recommended Joint Preparations for Fusion Welding of Steel", International Institute of Welding, Commission XV, *Brit. Welding J.*, Jan. 1962, 9 (1), p. 13. AWS *Welding Handbook*, Section I, Chapter 6, 1957.)

UNDERCUT	OVERLAP	LACK OF PENETRATION	LACK OF FUSION	POROSITY	CRACKING
					
FILLET WELDS  CORRECT	INCORRECT SIZE AND PROFILE  EXCESSIVE CONCAVITY	INCORRECT SIZE AND PROFILE  SMALL THROAT THICKNESS	BUTT WELDS  CORRECT	 EXCESSIVE REINFORCEMENT	INCLUSIONS 
 EXCESSIVE CONVEXITY	 EDGE OF PLATE MELTED AWAY, SMALL LEG LENGTH	 POOR WELD CONTOUR AND CRATER	 INSUFFICIENT REINFORCEMENT		

WELD DEFECTSDEFECTS IN MILD OR LOW ALLOY STEELS

Should introduce by saying that defects can be introduced as the fault of operator technique, weld settings, faulty equipment, wrong choice of equipment. First two are most usual, third growing with automatic welding.

Aim should be to produce defect free welds, which means careful control at all stages i.e.

- (1) Design and Drawing Office - detailed for welding with full information, and welding where possible.
- (2) Selection of materials - optimum weldability consistent with properties.
- (3) Preparation of materials - cutting to size, edge preparations, cleanliness.
- (4) Equipment - must be adequate for job, well maintained.
- (5) Welding - correct preheat, welding conditions, efficient welder, supervision, electrodes.
- (6) Post-heat treatment, (stress relief or anneal etc.) - correct control - can get cracking if heating too rapid.
- (7) Inspection - use of various methods where justified.

Use of detailed records (allocate responsibility), inspection.

DEFECTS IN ARC WELDS1. Incomplete penetration.

Weld metal fails to reach root of joint or to fuse with root faces of parent plate - voids or slag in root.

Penetration depends on correct electrode size and type in relation to joint (N.B. Penetrating power of electrode must also be considered e.g. - close butt welds need deep penetration electrodes.), correct welding current, suitable joint preparation to allow proper manipulation of electrode. Accurate joint preparation and fit up essential and should be specified.

To obtain complete penetration for double V or V preps. (butt welds) may be necessary to chip out or flame gouge the root of first run - this is normally always carried out on high class work (Class I).

In fillet welds too large an electrode can give lack of penetration but the use of a small electrode does not necessarily give an improvement. Should use largest possible electrode consistent with ease of manipulation and heat capacity of work. Slag inclusions and cracking frequently associated with this defect in fillets - with very low current may merely bridge gap.

Where practicable welds should be made in DH position - use of positioners.

Automatic welding requires care that machine accurately follows joint - or get offset welds, i.e.

Incomplete penetration causes local stress concentrations at root and under pulsating or fluctuating loads will seriously reduce fatigue life. May also reduce strength through shear lack of metal, but not generally serious. May reduce impact strength.

This defect can only be corrected by completely removing defective seam. Surfaces should be re-cleaned and room altered for necessary manipulation of electrode.

2. Lack of Fusion

Discontinuity between weld metal and parent metal, or weld metal and weld metal (i.e. lack of inter-run fusion).

Get boundaries of unfused metal, should not be confused with incomplete penetration.

This defect results mainly from presence of oxides, slag, scale or other non-metallics which prevent underlying metal from reaching melting point. Can normally be avoided by ensuring cleanliness of joint - with multi-runs careful attention should be given to deslagging between runs. If individual runs are badly placed or surfaces hinder deslagging they should be chipped or ground. Care needed on welding current, too low frequently gives lack of fusion but if too high may give excessive deposition rate which may cause operator to increase welding speed so that underlying weld is not heated to melting point.

Defect tends to reduce both fatigue and static strength of joint (impact if in right direction).

Again defect can only be rectified by removing the seam and rewelding.

3. Undercut

Groove or channel in parent metal along a toe of the weld. Gives continuous or intermittent reduction in section of parent plate. May occur in parent plate at toes of welds or at fusion faces of multi-run welds - butt or filler.

Defect commonly associated with excessive welding current but may be caused by incorrect technique such as too high a speed, excessive weaving or wrong angle of electrode. May sometimes be due to inherent welding characteristics of electrode, use of damp electrodes or presence of heavy mill scale.

Fatigue strength reduced even by very slight undercut. Slight and intermittent undercut has no effect on static strength.

Fault may be corrected by adding weld metal where it occurs. Deep undercut, especially if slag present, must be chipped before further deposition, or will give incomplete fusion. Final control of weld should not be greatly altered.

With high tensile steels correct welding procedure should be used at all times.

4. Edge of Plate Melted Off (Lap Joint) Melting of free
edge of lap welds.

Particular case of incorrect weld size and profile - reduction in thickness of plate and reduction in throat thickness.

Caused by incorrect manipulation of electrode, insufficient lap or using electrode that is too large for plate thickness. To avoid melting exposed corner the size of the weld should be $1/16$ " less than the edge thickness.

Effect on static and fatigue strength of lap or strap joint is reduction in so far as reduction in true throat thickness. Corrected by additional run or runs to restore throat thickness to required size.

5. Overlap

Imperfection at toe of weld caused by overflow of metal on to parent plate without fusion. May exist at isolated points or occur continuously. Often associated with fillets where it gives an apparent increase in weld size.

Caused by incorrect electrode manipulation in flat or HV welds allowing weld metal to flow away from fusion zone. Too large an electrode for welding position can cause it. Too low a welding speed can also cause this defect.

Single run fillets over $5/16$ " in leg length, if made in HVpos. tend to sag and cause overlap at toe of horizontal leg. Should use more than 1 run for this size of fillet.

Not a serious fault in static loading but under pulsating or fluctuating loads act as stress raisers and should be corrected.

Overlap can be removed by grinding away excess weld metal but care needed to leave smooth bead and avoid deep grinding marks.

6. Slag Inclusions

Slag or other foreign matter entrapped in weld. Refers to any non-metallics. Usual source is slag formed by electrode coatings but also could form if weld faces, etc. are not clean. i.e. heavy scale, loose rust, dirt and grit.

May be forced in by the arc during welding, in which case it will generally be finely dispersed or it may be trapped in cavities caused by undercut or uneven or convex preceeding runs. In latter cases it will appear as long line of slag.

Light scale oxide from gas cutting, light oil layer generally not important but heavy stuff must be removed. Fusion faces should be smooth and free from irregularities that could cause pockets.

Welding characteristics of electrodes are important, particularly slag viscosity which affects ease of control. Slag should float free and be easily removable after solidification.

Wire brushing and light chipping should remove slag but heavier chipping may be needed with irregular weld surfaces or if the wrong electrode currents and voltages have been used. Often associated with lack of penetration.

For lower runs in V groove the weld bead should have flat or slightly concave surface.

Small globules of slag will not affect static or even fatigue strength but large irregular or elongated lines of inclusions may reduce strength. These may also give rise to radiating hair line cracks. Source of brittle fracture.

Minor inclusions need not be removed but longer inclusions need chipping, cleaning and re-welding.

7. Porosity

Gas pores in a weld may cause :-

- | | | |
|-----|---|---------------|
| (a) | Piping (worm hole) - elongated or tubular | (trapped gas) |
| (b) | Blowholes. | |

Piping and large blow holes (sometimes partially filled with slag) are fairly rare, but small scattered or clustered porosity is quite common.

Caused by entrapment of gas evolved during welding. Composition of plate and wire important especially S content, excessive moisture in covering or scale on weld preparation also casuses.

Use of low welding current, too long or too short an arc and other factors which promote rapid freezing promote porosity - sometimes advisable

to "puddle" weld metal.

Small finely dispersed porosity not generally effective in reducing static or fatigue strength but dense porosity or very large pores can considerably reduce load carrying area and act as stress raisers.

Excessive porosity must be removed and the joint re-welded.

8. Cracks

Discontinuities caused by tearing of metal above solidus, , tearing when in plastic condition or fracture when cold.

May occur in weld metal or haz of plate.

Weld metal cracks may be classified - longitudinal, transverse, crater or hairline (often associated with slag).

~~Long.~~ cracks may be longitudinal (adjacent to edge of weld seam), transverse (which often originate from weld cracks) or "under-bead".

- (a) Weld metal Cracks - caused by highly localised stresses in the joint arising from shrinkage of weld metal and resistance to movement of parts during welding: therefore necessary to give consideration to joint conditions and degree of restraint opposing movement caused by weld shrinkage. Order of assembly should allow shrinkage to take place as freely as possible.

Size of individual runs should be large enough to withstand shrinkage stresses. This is particularly important in welding of heavy plates or sections or under restraint. Rapid cooling increases tendency to cracking.

Badly fitted parts or incorrectly prepared joints frequently result in lack of fusion and penetration or slag inclusions, and these may initiate "hair-line" cracks which can then propagate through the weld.

Longitudinal cracks more likely to occur in first layer of weld, if not cut out may propagate (one of reasons for back chipping). Cracks often result from poor shape or position of deposit in which case it may be necessary to change welding procedure. Long. cracks may also originate from crater cracks and deep craters should be avoided (craters should be examined and chipped where necessary).

Transverse cracks generally associated with joints having high degree of restraint and may extend into parent plate haz, especially if haz has hardened appreciably during welding.

S content of most commercial mild steels does not normally affect weldability. Excess S in plate or core wire may cause weld metal cracking however, due to low m.pt. sulphides. Other impurities, e.g. Ni, Cu, can cause hot cracking.

With automatic welding crater is often large and should therefore use run off plates (manual as well on occasions). These plates should be same thickness as joint. - should be welded for full width to prevent seam opening up during welding and causing cracking.

Crater cracks

- b) Parent Metal Cracking - normally associated with high tensile steels (medium C, low alloy or high alloy steels) but may occur in thick sections of mild steel. Cracks generally within hardened haz and are attributable to composition of parent plate associated cooling rate after welding and degree of restraint, thickness of plate rate of heat input, atmospheric temperature.

Hardness of haz results from rapid cooling after welding and so increases with thickness of plate (rapidity of heat dissipation). Danger of cracking increases, therefore with plate thickness especially in joints of high restraint, therefore rate of cooling should be retarded to avoid formation of excessively brittle zone and reduce rate at which shrinkage stresses are developed. Reduction in cooling rate can be achieved by pre-heating, immediate post-heating or increasing rate of heat input by means of suitable welding procedure, e.g. large runs, heavy weave. Multi-run welds should be deposited so that heat is maintained in surrounding plate and succeeding runs made as soon as possible.

Small tack welding of hardenable steels should be avoided if pre-heating cannot be applied. Avoid arc strikes outside weld groove as may give local hard spot and small surface cracks.

Low temperature pre-heating recommended even on mild steel if temperature is at or below 0°C and chill should be taken off parts above 1½". Use of special electrodes may be advantageous - low H rods. Can also get hot cracking in the haz if very brittle or low melting point grain boundary

Cracks will seriously reduce static and fatigue life and should never be permitted.

Can only be corrected by complete removal of cracked volumes - care needed to avoid leaving cracks. May be necessary to modify welding procedure to prevent recurrence of cracking.

9. Incorrect Weld Size and Profile

Undersize fillet welds, fillet welds of variable size as measured by minimum leg length - may result from excessive concavity or convexity, irregular weld face or unequal leg lengths.

In Butt welds one has lack of reinforcement, excessive reinforcement, or irregularity of deposit.

These defects are generally a result of faulty technique. Convexity or concavity may be a characteristic of the specific electrode but more often is due to incorrect welding current (convexity - low current, concavity - high current.)

Unequal leg lengths in fillet welds generally caused by faulty manipulation or incorrect positioning of the workpiece when welding in the downhand position (i.e. work tilted). Welds of over 5/16" in any other position than dh should be made in more than one run.

If get poor fit up the effective leg length will be reduced.

Excessive, or lack of reinforcement in butt welds is caused by use of wrong welding procedure, and should not arise if joint preparation, rod size and welding current, number of runs, welding speed and, possible, length of run per electrode, are specified and followed. Note, however, that some latitude over current and technique should be allowed to the welder, who will have his own personal idiosyncracies, and on high class work much will be decided by procedure tests and welder approval tests.

Excessive concavity and unequal leg lengths reduce the throat thickness of fillet welds and reduce the ultimate strength of weld. Excessive convexity may give equivalent static strength but often indicates lack of penetration and fusion. May also be associated with overlapping which will seriously reduce the fatigue strength of the fillet.

Both excessive and lack of reinforcement will reduce fatigue strength and the second defect may also cause reduced static strength.

In so far as these faults in fillet welds may indicate other sub-surface defects, it may be necessary to re-weld for high quality work particularly if fatigue strength is important. Otherwise concavity and unequal leg length can be corrected by further (expensive?) runs. Convexity needs the chipping away of excess metal, and care should be taken to leave a smooth surface. (NB. care in cleaning and de-slagging.)

In butt welds reinforcement may be increased or decreased by further runs or chipping and grinding.

Again it should be emphasized that these faults should not occur if the correct procedure is laid down.

10. Poor Surface Appearance

Non-uniform ripples, surface porosity, included slag, faulty junction and craters.

Caused mainly by faulty welding technique although the electrode characteristics can have an effect. Welding position must also be taken into account. Note that accessibility is important and the operator must always be allowed adequate space for manipulating the electrode.

These faults will not normally affect static strength but by acting as stress raisers may seriously reduce fatigue life.

Can be corrected by grinding or chipping and re-welding.

11. Spatter

Main objection is waste of metal and cost of removal. Generally caused by excessive welding current or dampness in the coating. Some electrodes are naturally more prone to spatter than others.

Has no effect on strength and may be removed by chipping and wire brushing the small globules from the surface of the plate.

OXYACETYLENE WELDING

1. Inadequate or Excessive Penetration - caused by insufficient bevelling of edges, too wide a root opening, too great a welding speed, or poor torch and filler wire manipulation.
2. Excessive or Insufficient Reinforcement
3. Undercut
4. Overlap

1 - 4 are visual defects.

5. Incomplete Fusion - Insufficient heating of base metal, too rapid welding speed, gas or dirt inclusions.
6. Porosity - Trapped gases, usually CO or water vapour, may be avoided by correct manipulation, choice of materials, adequate fluxing.
7. Cracking - Hot or cold - depends on choice of materials, welding technique.
8. Hard Spots - Incorrect manipulation, parent material (metallographical characteristics).
9. Excessive Grain Growth in HAZ. - excess temperature.

RESISTANCE WELDING.

1. Lack of Symmetry

Caused by dissimilar thicknesses of same metal, dissimilar metals, different rates of heat extraction, electrodes of uneven sizes or different materials, misshaped electrodes, misaligned electrodes, too much tip radius.

Can lead to indented surfaces, lack of penetration into one sheet with lowering of strength.

2. Surface Indentations

Cannot be eliminated but can be minimised, i.e. by having short welding times keeping heat conduction beyond weld zone to a minimum.

Defects characteristic of welds made with too great a heat generation in weld - expulsion of molten metal often accompanies it. Caused by high welding current, low electrode force, insufficient electrode contacting area, improper alignment of electrodes, dirty metal surfaces, excessive time of current flow, improperly mating surfaces.

3. Surface Melting

Generally a defect

Caused by insufficient pressure or insufficient time with excessive current, or both. It is aided by inadequate electrode cooling, inadequate squeeze and hold times, fouled electrodes or scaly or dirty material surfaces.

4. Excessive Sheet Separation

Pulling apart of mating surfaces of spot weld caused by expansion and contraction of molten metal and forging effect of electrodes. Similar affect caused by poor fit up of sheets. Related to surface indentation.

Caused by steeply tapered electrodes at high pressures acting as a male punch. Misaligned electrodes will also cause it, particularly on one side of weld.

Sheet separation will always occur on gauges of .060" and up.

This defect is undesirable because it tends to increase the bending moment on joint when stressed in shear, and it will also increase distortion of assemblies.

5. Interface Expulsion

Indication of incorrect welding conditions. Usually result of too great a rate of heat generation caused by excessive current, insufficient pressure, undersize electrode contact area, and dirty metal interface, or welding too near an edge.

Results in internal cavitation of weld and a reduction in weld strength.

6. Internal Defects

Cracks, porosity (spongy metal), cavities (caused by expulsion), inclusions (with coated metal).

In general have no significant effect on static or fatigue strength when located in centre of weld nugget even if extend to outside surface. Defects at periphery of weld are very serious however.

Porosity - often present. Caused by overheating, inadequate pressures, premature release of (adequate) pressure, too late an application of forging, pressure, absorbent gases on sheet surface. Porosity can be reduced by using lower rates of heat build up, lower maximum temperatures and higher pressures.

Cracks - generally caused by overheating, improperly applied electrode forces during welding or use of wrong schedules in crack sensitive materials.

In flash and upset welds cracks may be caused by excessive cooling rates or forging at too high or low a temperature.

SOFT SOLDERING

Causes of failure of soldered joints are :-

1. Joint not thoroughly cleaned.
2. Metal was not heated above melting point of solder or was heated too far above this temperature.
3. Wrong flux used, giving inadequate cleaning.

BRAZING.

Voids, lack of wetting of parent material, presence of films, changes in structure of parent material (caused by overheating and, perhaps, diffusion of braze metal, hot cracks (penetration of Cu along grain boundaries) inclusions. Main cause is incorrect fluxing, flux inclusions, overheating and underheating.

CLAD STEELS

These steels demand high quality welds to maintain corrosion resistance and strength. Main causes of trouble are incorrect joint preparation and fit-up, departure from correct procedures.

INERT GAS PROCESSES

Most of defects are common with metal arc defects. An addition is tungsten inclusions.

Common causes of porosity and other troubles are contamination of Argon or CO₂ (generally with water vapour) and contamination of electrode wire, often by adsorption of water vapour on surface.

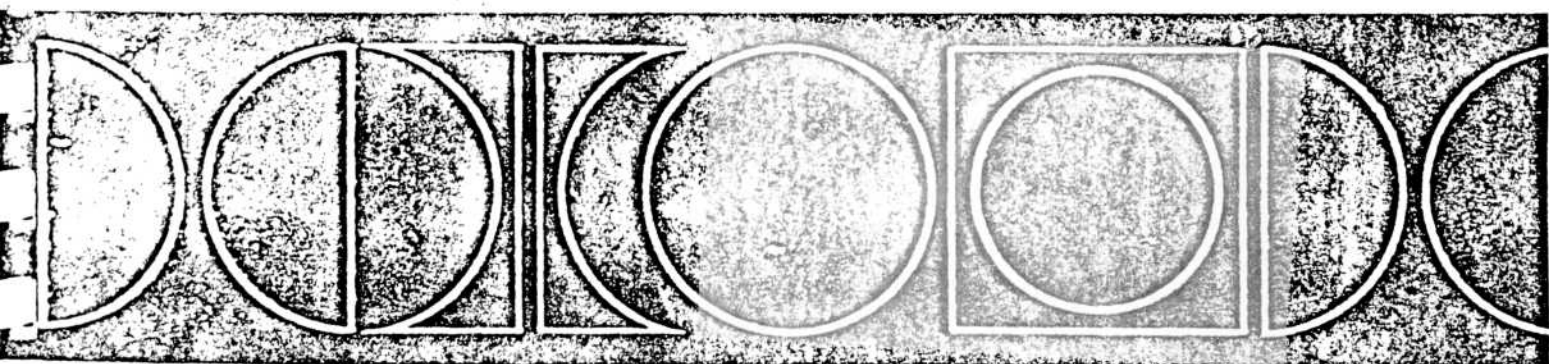
TABLE III

Scope of Non-Destructive Testing Techniques

NDT Method	Position of Defect			Permanent Record	Cost	Safety	Present Usage on Welding
	Surface	Sub-Surface	Internal				
Visual	✓	-	-	No	Low	Good	Wide-spread
Dye Penetrant	✓	-	-	No	High if used widely	Can Cause skin troubles in some cases	Fairly low
Magnetic	✓	✓	-	No	Low	Good	Moderate
Radiographic	✓	✓	✓	Yes	Expensive	Radiation problems	Wide-spread
Ultrasonic	Difficulties with un-dressed welds	✓	✓	Only with expensive recording equipment.	Moderate	Good	Fairly wide, increasing

basic welding metallurgy

School of Welding Technology



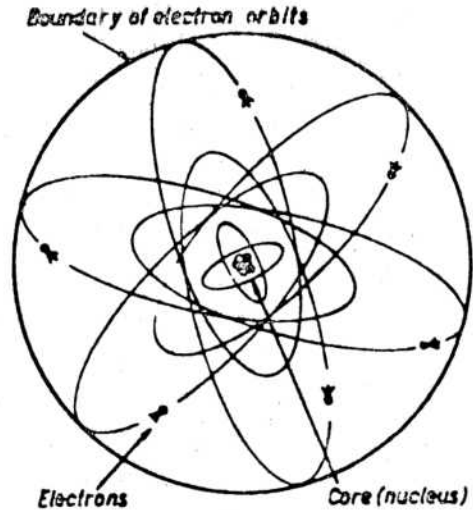
The Welding Institute

Basic welding metallurgy

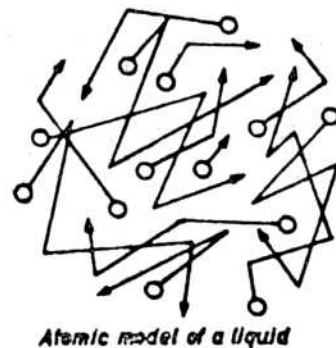
WARNING This section contains some gross over-simplifications but if the non-metallurgist gets some idea of the meaning of some of the esoteric words, that will be useful.

All matter is made up of atoms. Atoms have a heavy core which is positively charged and one or more planetary electrons (negatively charged).

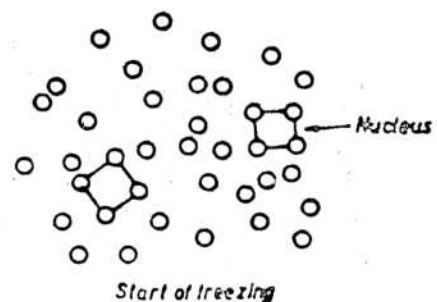
The electrons are a long way from the core. In the sketch, if the core is assigned a diameter of 5 mm then the atom is about 2000 mm in diameter.



In liquid metals the atoms move about in a random manner. The hotter the liquid, the faster the movement.



When pure metals begin to freeze, small geometric groups of atoms (3-dimensional) form, and if they become big enough (NUCLEI), they grow larger to form crystals of solid metal.



Crystals of solid metal have their atoms much closer together than in the liquid and, to repeat what has been said before, arranged in a geometrical pattern.

The number of atoms in a piece of solid metal is enormous and the space surrounding them is vast. It is possible, therefore, for atoms to enter a solid metal, to move through a metal, and for atoms, especially of a gas, to leave a metal. The hotter the metal the faster and easier this action is. The process is known as DIFFUSION.

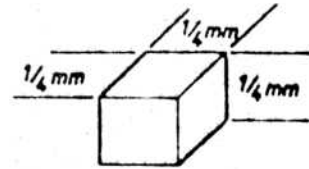
Atoms in a crystal are not in fixed positions, only approximately so. Imagine atoms as ships in a convoy. Each ship tries to keep station relative to each other under normal conditions (at low temperatures in a metal) and makes quite a good job of it. Under conditions of alert (at high temperatures in a metal) there is much more agitation and they move further apart. Melting of a metal can be likened to the ships scattering during an attack.

Subsequent sketches of atomic arrangements in metals show the atoms in fixed positions. Remember that you are looking at atoms as though photographed with a stop-action camera (if that were possible).

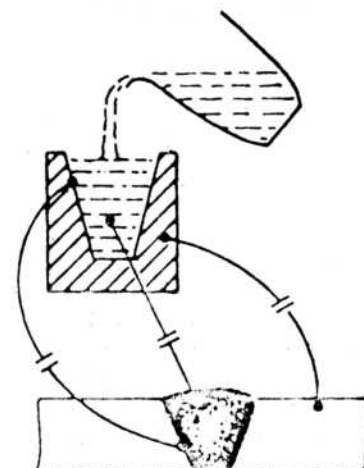
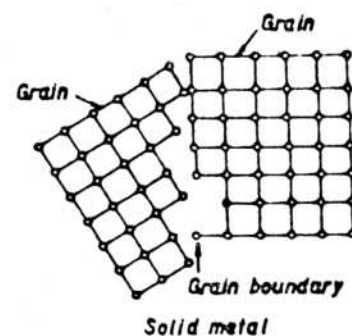
Since the original nuclei were randomly arranged when they formed, it follows that there must be a mismatch when the growing crystals (GRAINS) ultimately meet. The region of mismatch is called a grain boundary. There is a lot of room at grain boundaries in cast metals, and both solid and gaseous impurities tend to collect there. Low melting point impurities are those most likely to be found in grain boundary regions because they remain liquid after the grains have solidified.

Solidification of weld metal

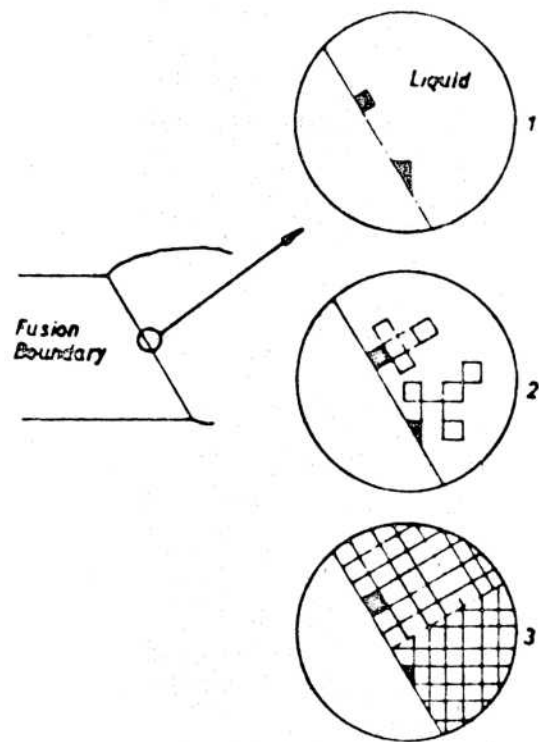
Fusion welding can be likened to casting. The main difference is that in fusion welding a part of the 'mould' dissolves in the 'casting'.



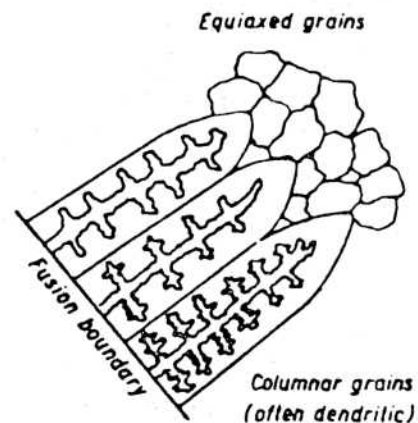
An iron crystal this size contains about 1,000,000,000,000,000,000, atoms (a million atoms along each edge). Even so, this cube of dense metal consists mainly of empty space.



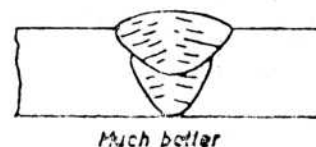
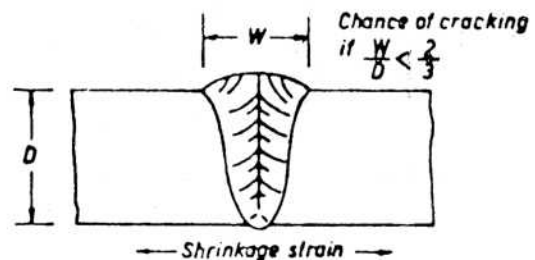
Consider a single run of liquid weld metal situated between two fusion boundaries. Heat is flowing away through, and usually perpendicular to, the fusion boundaries. Nuclei form on the fusion boundaries and solid crystals begin to grow from them. The crystals grow preferentially along the direction in which heat is flowing but at the same time there is growth in other directions leading to eventual infilling and complete solidification.



In alloys, (and impure metals can be considered to be dilute alloys), there is such a wide difference in composition between the grain interiors and grain boundaries that, on etching, a tree-like (DENDRITIC) structure is revealed. If freezing is slow, grains tend to be longer than they are wide (COLUMNAR) and if freezing is fast they tend to measure the same in all directions (EQUIAXED). Both types can occur together in welds.

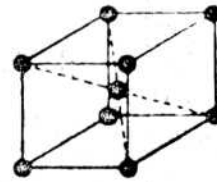


Columnar grains meeting in the centre of deep narrow welds are an undesirable feature, and that is why welds are often made in two or more runs even though the process in use may be able to do the job in one run. (There are other reasons).

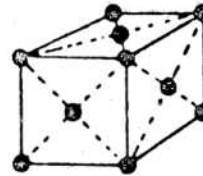


Solid metals

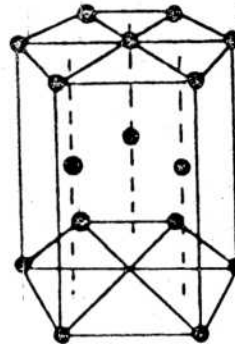
The atoms of a crystalline solid are arranged in a regular three-dimensional pattern called a crystal structure (LATTICE). Only 14 arrangements are possible and only 6 of these are important in metallic structures. The 3 most important are sketched.



Body centre cubic (B.C.C.)
eg Iron at room temp

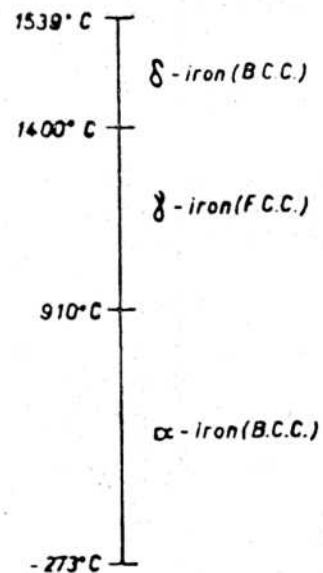


Face centre cubic (F.C.C.)
eg Copper, Aluminium



Close packed hexagonal (C.P.H.)
eg Zinc, Magnesium

Solid iron exists in three forms (ALLOTROPES) when in equilibrium (see later for the significance of this remark). Thus:



A few other metals behave similarly - but none of such commercial importance as iron.

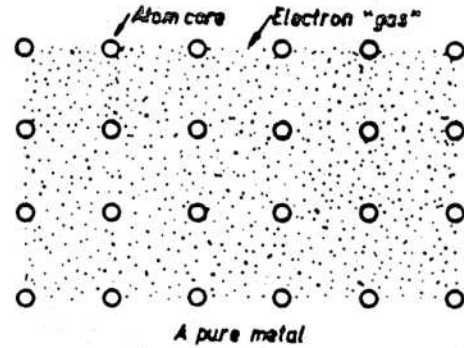
Alloys

Alloys consist of metals, or metals and other elements, in amounts deliberately chosen to produce certain desirable properties. These properties are usually different from, but often related to, those of an alloy's pure constituents.

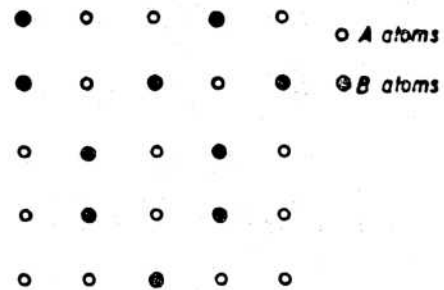
Many alloys are SOLID SOLUTIONS. The term 'solid solution' is difficult to appreciate. Consider

separately a cup of unsweetened tea and a spoonful of sugar. Each is recognisable as a separate entity. Stir the sugar in the tea. Is the sugar now recognisable as sugar? If it were possible to freeze the sweetened tea without the sugar separating out - that would be a SOLID SOLUTION.

Atoms in pure metal might look like this in 2-dimensions.



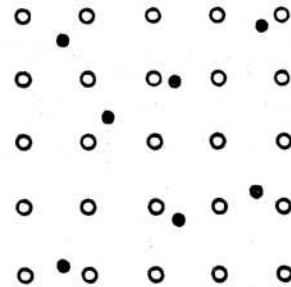
On adding an alloying element, the atoms of the added element may replace some of those of the host metal to form a substitutional solid solution.



A substitutional solid solution
(e.g. Copper-nickel alloys).

or

If the added atoms are very small relative to the host atoms they may occupy the interstices between the host atoms.

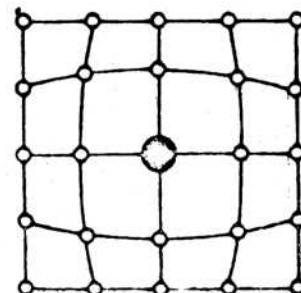


An interstitial solid solution
(e.g. small amounts of carbon in iron)

Elements in solution cause strengthening by distorting the lattice.
Special names are given to some of these solid solutions thus:

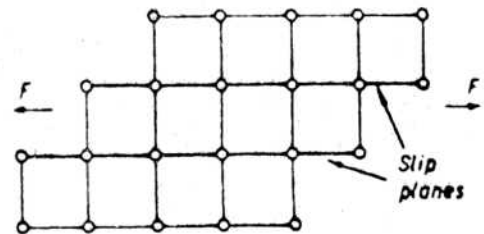
Those based on α -iron are called FERRITE

Those based on γ -iron are called AUSTENITE



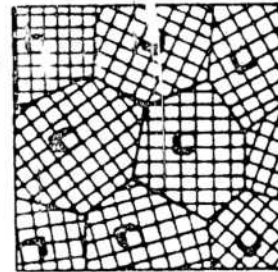
Effect of a large atom

Forces sufficiently high will deform the material permanently. Plastic behaviour
Plastic deformation is introduced by rolling, extrusion, forging etc. and material so treated is strengthened.



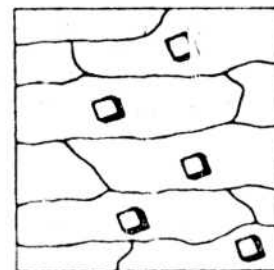
Plastic deformation
(causes permanent change of shape)

Starting with a material like this, and which is soft and weak,



Before rolling (Annealed)

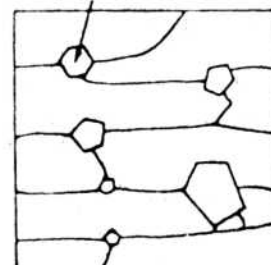
we can roll it and thus harden and strengthen it.



Cold rolled

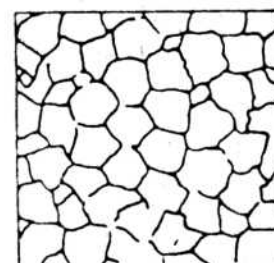
On heating to a suitable temperature so that the atoms can move about appreciably, new undistorted grains begin to form at certain sites in the deformed grains. If conditions are suitable, the distorted grains will be replaced by new grains which may be smaller or larger than the original grains (depending on many complex factors). There will be an accompanying fall in hardness and strength.

Undistorted grains forming at temperature



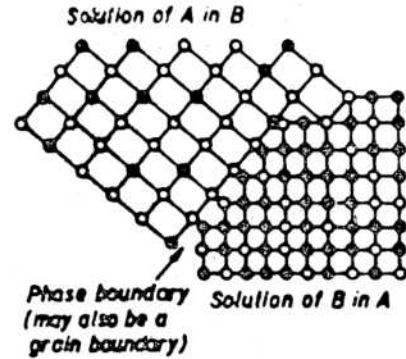
At an early stage of annealing

Thus, work hardened material is softened by annealing. Some metals and alloys (aluminium, copper, cupro-nickel, brass et al.) can only be strengthened by work hardening. If you weld such work hardened material then some part of the heat affected zone (HAZ) will be annealed and there is nothing you can do about it except keep its width to a minimum by controlling the heat input.



Annealed (Recrystallised)

There are circumstances when the added atoms cannot completely dissolve either substitutionally or interstitially and this gives rise to mixed atomic groupings within a single alloy, or to the production of chemical compounds from the elements present. Each distinct grouping of atoms in an alloy is called a PHASE and we speak about SINGLE-PHASE and MULTI-PHASE alloys.



Strengthening of metals and alloys

1. Grain size

Dirty grain boundaries are weak. If the grain size is small, impurities are spread thinly over the large area of grain surface. If the grains are large, the same amount of impurity is concentrated over a smaller area and is therefore more harmful. Small grains are desirable, indeed strength is

proportional to $\sqrt{\frac{1}{\text{Grain size}}}$

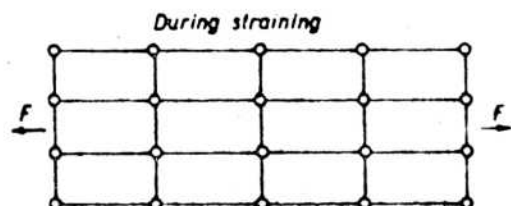
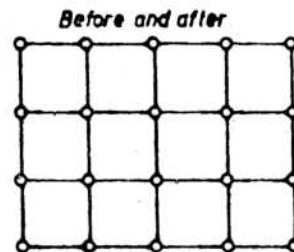
e.g. If the grain size is made 9 times smaller the strength is 3 times greater (don't take this too far.)

2. Solution strengthening

As the amount(s) of alloying element(s) increases in solid solution the strength usually increases. Increase the carbon content of very pure iron from 0.0001% to 0.005% and the strength increases by a factor of 4. (Don't take this too far either.)

3. Work hardening (and annealing)

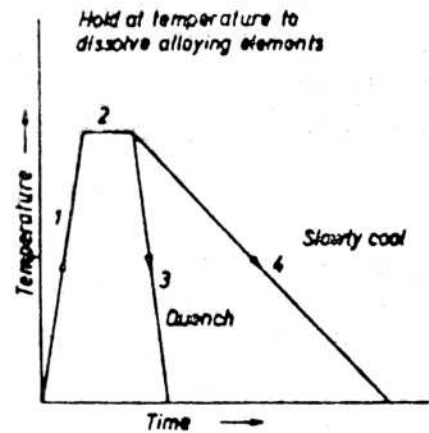
If a force is applied to a crystal it can be deformed. If the force is small, the crystal will regain its original dimensions when the force is removed. Elastic behaviour.



Elastic deformation
(returns to original shape when F removed)

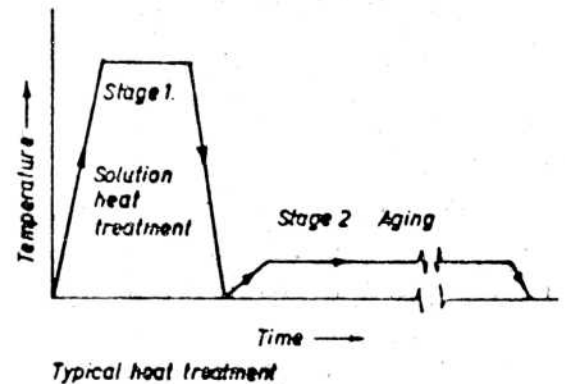
6. Precipitation hardening (age hardening)

Certain alloys will dissolve an appreciable amount of alloying element(s) when hot (but solid) and reject some of the alloying element(s) in the form of large particles of compound(s) when slowly cooled. If the hot material is quenched the alloy becomes super-saturated and will contain more of the alloying element(s) than would normally be soluble at room temperature. 'Duralumin' is a classic example. If the super-saturated alloy is then heated to some intermediate temperature, the compounds are precipitated as very fine particles which harden and strengthen the alloy.

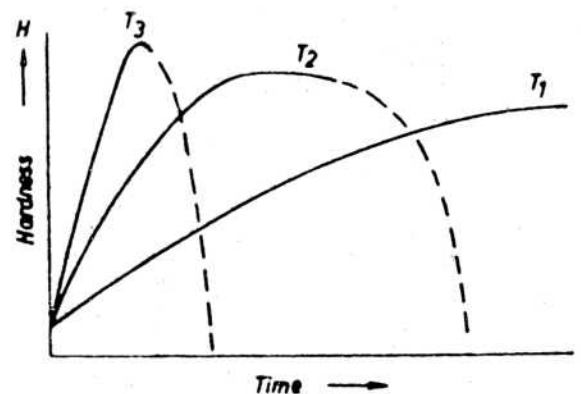


After 1+2+3 - Supersaturated solid solution

After 1+2+4 - Alloying elements rejected in the form of large particles of compound(s)



Typical heat treatment

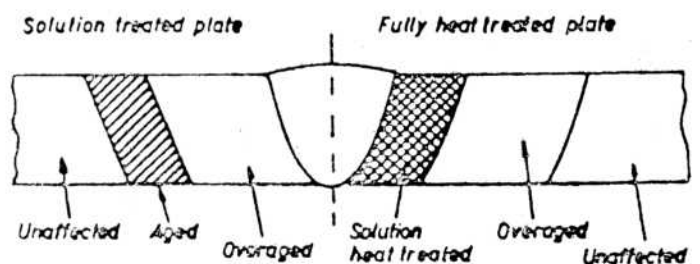


Hardness after aging at temp. T

$$T_1 < T_2 < T_3$$

(If aging is continued after peak hardness is reached, softening - OVERAGING occurs)

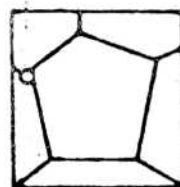
In welding, therefore, the HAZ of age-hardenable alloys responds in a complex way to the thermal cycle.



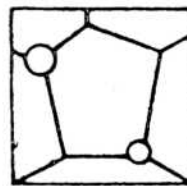
5. Martensitic reactions

Typical of this type of reaction is the behaviour of many steels although some non-ferrous alloys can be strengthened by this mechanism.

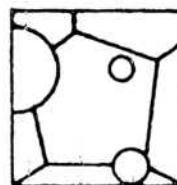
If a plain carbon steel (C 0.06–0.8%) is slowly cooled from a temperature at which the structure is austenitic the final structure is a mixture of ferrite and the compound iron carbide (CEMENTITE). When the ferrite is forming it rejects carbon which diffuses into the remaining austenite. The last austenite to transform deposits alternate plates of ferrite and cementite which have a characteristic microstructure which has been given the name PEARLITE. Pearlite is not a phase, it is a mixture of phases.



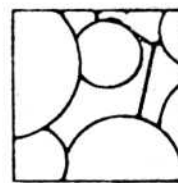
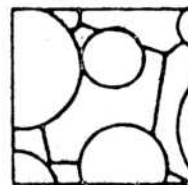
1 Austenite



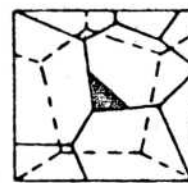
2 Ferrite begin to form.....



3 and continues..... 4



5 to form.....



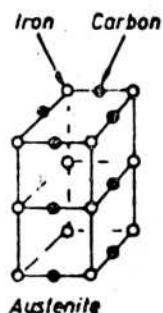
6 the end result being ferrite plus pearlite.

If instead of being slowly cooled, the austenite is rapidly cooled (QUENCHED) there is insufficient time for the above reactions to take place and a new phase, MARTENSITE, is formed. Martensite is hard and brittle, unless its carbon content is very low, and can therefore be an undesirable constituent in many steels.

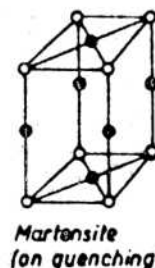
In some welding processes cooling is rapid (e.g. in spot welding) whereas in others it is slow (e.g. in electro-slag welding).

In welding processes where the thermal cycles are intermediate between quenching and slow cooling instead of martensite we may get a constituent which is neither martensite nor ferrite plus iron carbide. This constituent is called BAINITE and its mechanical properties are akin to those of martensite in alloy steels which are fast cooled, and more like those of ferrite plus iron carbide in low carbon steels welded with high heat inputs i.e. fairly slowly cooled.

There is an appreciable volume change when martensite forms from austenite and this may cause cracking especially when hydrogen finds its way into the material (from electrode coverings, grease, scale etc.,).



becomes



Pre-heating will reduce the cooling rate and may help in the production of a softer and more ductile HAZ.

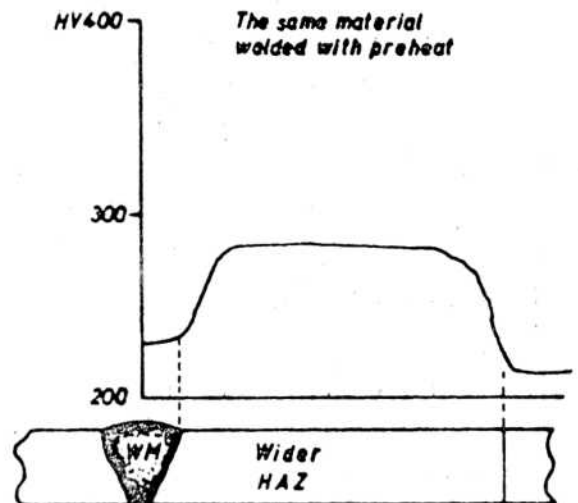
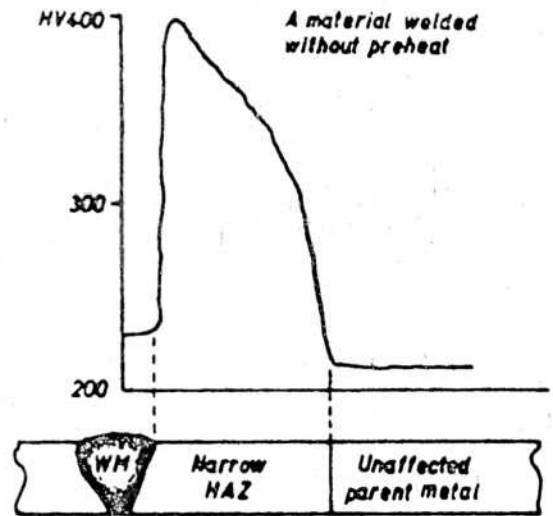
Pre-heating, however, has a more important effect; it delays cooling and enables hydrogen to escape before it can cause damage to cold HAZ material.

A rough and ready view of the situation is that if any part of the HAZ has a hardness of over 350 Vickers units - look out for trouble.

Alloying elements other than carbon have an effect on this type of behaviour and attempts have been made to assign equivalents to these elements. For example, manganese is about $1/3$ as powerful as carbon in its effect.

$$\text{i.e. } \frac{\text{Mn}\%}{6} = \text{C}\%$$

So, in a carbon manganese steel we can use the concept of carbon equivalent.



Thus, carbon equivalent (C.E.) = Actual carbon content + $1/6$ of the Mn content

BS 4360 also gives

$$\text{C.E.} = \text{C} + \frac{\text{Mn}}{6} + \frac{\text{Cr} + \text{Mo} + \text{V}}{5} + \frac{\text{Ni} + \text{Cu}}{15}$$

for a certain range of steels.

Beware CE values greater than about 0.4% especially in thick sections. Remember too that other elements have a carbon equivalent which must be taken into account.

A major problem lies in obtaining a suitable definition of weldability that can be interpreted in quantitative terms. If we turn to the International Institute of Welding we find that Commission IX have defined weldability:

"A metallic material is considered to be weldable, to a certain degree, by a given process, and for a given purpose, when a continuous metallic connection can be obtained by welding using a suitable procedure, so that the joints comply with the requirements specified both in regard to their local properties and their influence on the construction of which they form part". This verbose statement does nothing to advance our search for a definition that we can express quantitatively. The A.W.S. Welding Handbook gives a definition which is shorter but no more helpful:

"the ease with which the effects of welding can be controlled". Thus it would seem necessary to abandon any hopes of expressing weldability in a simple numerical manner unless, as has been done in some cases, we are willing to restrict the term weldability to some specific aspect of welding. This latter approach has been adopted in some cases, for example with underbead cracking, but can hardly be related to the widely accepted meaning of the term weldability.

It is suggested therefore that the looseness of the term "weldability" should be accepted and that the factors that contribute to weldability should be enumerated, so that these separate factors can be assessed individually. Then, by consideration of each factor it should be possible to gain at least a qualitative assessment of the weldability of specific materials, although a true measure of weldability can only be gained by performance in actual service.

FACTORS CONTROLLING WELDABILITY

The weldability of any material will be governed by several factors,

involving:

chemical properties
physical properties
metallurgical properties
design of the actual component
to be fabricated.

It may also be suggested that welding processes should be added to this list since many materials cannot be welded by all processes, but invariably the limitations set upon possible welding processes and positions are controlled by the material properties as affected by the thermal cycle.

Design

Design is rarely mentioned in any discussion on weldability although it may often be the final arbiter in deciding whether a fabrication can be welded. Two aspects are of particular importance: accessibility and restraint. If accessibility is limited it will prove difficult, or even impossible, to produce defect-free welds and the subsequent performance in service will be impaired. Similarly, heavy restraint may lead to high residual stress or cracking.

Chemical Properties

The oxidation resistance, nature of surface films and impurity content of a material are of considerable significance in determining the ease with which it can be welded.

If a material oxidises readily, and is not protected during welding, the weld pool will become contaminated. In some cases a joint cannot be obtained whilst in other cases inclusions and oxide films will drastically reduce weld quality. Oxidation resistance will also affect the choice of welding process; a material such as mild steel can be welded by most processes without excessive precautions whereas the reactive metals such as titanium and zirconium need considerable protection if embrittlement is to be avoided.

Surface films may offer a barrier to good cohesion, as in the fusion welding of aluminium, or may affect the production and distribution of heat, e.g. in the resistance welding of heavily scaled sheet. The ease with

which these films can be disrupted and removed either before or during welding (by flux or arc action) will affect the ease with which a material can be welded and also the choice of available processes. Surface films may also affect emission characteristics so that intermittent films may give variable heat production.

Finally, impurity content can have a considerable effect on weld quality since it may lead to cracking or porosity. The deleterious effect of sulphur in promoting cracking in ferrous and nickel-base alloys is well known, as is the effect of lead in promoting hot shortness in copper-base alloys, but much more work is required on the effect of other impurities and minor alloying additions.

Physical Properties

Several physical properties of the material may affect its weldability by certain processes. Thus the thermal and electrical resistivities will influence resistance spot welding characteristics and in the case of high conductivity metals such as aluminium and copper will make welding difficult or even impossible. A high thermal conductivity combined with a high coefficient of thermal expansion can lead to excessive distortion whilst these properties, together with the elastic modulus, will influence the level of residual stresses. The viscosity of the molten material will influence fluidity in the weld pool and may be of importance in positional welding, although operator technique is generally the overriding factor. Finally, emission characteristics may be a problem in arc welding if they result in arc instability - this rarely occurs, but it can cause difficulties in the automatic arc welding of magnesium and its alloys.

Metallurgical Properties

Any consideration of weldability must dwell largely upon the metallurgical properties and the welding engineer must consider two separate factors:

- (1) the effect of the weld thermal cycle on the parent metal,

and

- (2) the structure and properties of the weld metal.

The welding cycle will result in an heat affected zone in the parent material adjacent to the fusion zone. The severity of this thermal cycle will depend partly upon the welding process and partly on the dimensions of the structure, typical cooling rates being $30 - 110^{\circ}\text{C}/\text{minute}$ for oxyacetylene welding, $110 - 4000^{\circ}\text{C}/\text{minute}$ for arc welding and $3000 - 22000^{\circ}\text{C}/\text{minute}$ for resistance spot welding, whilst the effect of the cycle will depend upon the material. Thus in pure annealed metal and in many annealed solid solutions the welding cycle will have little effect other than causing a little grain growth, whereas if these materials are welded in the work hardened condition recrystallization and softening will occur in the heat affected zone, resulting in a low joint strength. If low melting point constituents are present in the heat affected zone, incipient melting may occur, leading to hot cracking.

With precipitation hardening alloys the strength and hardness depend on a finely balanced structure which is almost certain to be affected by the welding cycle. The overaged region in particular can only be restored by re-solution treating and re-aging, which is impossible with many structures so that a full strength weld cannot be obtained with certain treatable alloys. An exception is provided by the naturally aging aluminium-zinc-magnesium alloys where much of the strength and hardness is recovered by re-aging at room temperature subsequent to welding. If a wide overheated region exists it may result in all the aging elements being concentrated at the grain boundaries where they are not available for strengthening. In some cases no heat treatment will restore properties.

In mild and low alloy steels the heat affected zone temperature will rise above the A_1 temperature and become wholly or partially austenitic. Cooling from this state may result in ferritic and martensitic structures according to composition and cooling rate. This will generally result in an

increase in hardness and fall in ductility, and if martensite is formed there is a very real danger of underbead cracking. A knowledge of the isothermal transformation diagram, or preferably the continuous cooling transformation diagram will give a good indication of the likelihood of martensite formation, although some care in interpretation must be exercised, since these diagrams cannot account for segregation, austenite grain size etc.

Some creep resisting steels, both ferritic and austenitic, exhibit strain induced precipitates on dislocations due to the thermal cycle in the heat affected zone, which can restrict the ability of the heat affected zone to deform on subsequent heat treatment or service at elevated temperatures. This can result in re-heat cracking which has proved particularly troublesome in high temperature steam mains in power stations. Evidence is now accumulating as to which materials are susceptible to this form of cracking, but much more work is needed.

The anisotropy of properties and, especially the presence of aligned non-metallic inclusions can result in low ductility in the short transverse (i.e. through thickness) direction of steel plate. Certain joint configurations e.g. fillet and T-butt welds, can result in lamellar tearing or pull-out cracking.

The weld metal, which may be of parent plate composition or not, has been heated above the melting point. It may, therefore, be considered as cast metal although this is not a good comparison since its properties may vary markedly from the properties of bulk cast material. Our main concern must be to decide the effect that the process of solidification and cooling will have upon the joint properties and it is suggested that this can best be achieved by considering the weld structure and the likelihood of porosity and cracking.

The weld structure will generally be columnar due to the directionality in the cooling rate, although some equiaxed grains may be found

in the centre or at the fusion boundaries. In many materials the weld metal strength will be slightly below that of the parent metal although it may sometimes be improved by alloying additions to the filler rod, or by peening.

It is fortunate with mild and low alloy steels that weld metal strengths are higher than those obtainable with hot wrought materials of the same composition, due to the fine ferrite grain size and high dislocation density. This increased strength is gained with little loss of ductility and allows the weld metal to be less highly alloyed than the parent material. This improved strength is maintained at slightly elevated temperatures.

With the majority of precipitation hardening alloys, whether aluminium or nickel based, it is virtually impossible to match the properties of the fully hardened material and the only solution at present lies with the designer in positioning the welds away from regions of high stress. Some work is known to be under way at present on the hardening characteristics of aluminium-zinc-magnesium weld metal but initial results are far from conclusive.

Gas-metal and/or slag-metal reactions in the weld pool may give rise to gas, and if this is trapped in the solidifying weld pool porosity will result. Minor porosity is insignificant but large pores or regions of porosity can result in a significant reduction in weld properties.

Weld metal cracking may occur at temperatures near the solidus due to an unfavourable composition or it may occur at lower temperatures due to excessive restraint. Hot cracking depends upon the composition (both alloying elements and impurities) and cooling conditions in the weld metal; it is generally due to the persistence of liquid films at the grain boundaries. Considerable work has been done upon hot cracking in welds and the susceptibility of many weld compositions are known. In most alloys hot cracking can be eliminated by attention to filler composition, although this may be at the expense of other properties. Restraint cracking is poorly documented and is

often attributed to other causes. It is only likely to arise where exceptionally heavy restraint is applied, and even in these cases the danger can be minimised by attention to welding procedure and by use of a ductile weld deposit.

ASSESSMENT OF WELDABILITY

In assessing the weldability of a material for a particular fabrication one must account for all the factors listed above. Most of this assessment can be done by consulting available information, which will enable many points to be eliminated as unimportant.

The chemical and physical properties will be readily available, even with regard to new materials, but information on metallurgical behaviour under welding conditions may be more difficult to obtain since alloys are generally designed with regard to service performance rather than ease of fabrication. Even so most material producers will examine the welding characteristics of new products although they may be rather over-optimistic regarding the ease of welding. The design of the fabrication falls within the sphere of the design engineer but consultation with the welding technologist is essential to prevent the weldability problems that arise, or are aggravated by poor accessibility and high restraint.

Thus, much of the information for assessing weldability is likely to be available. The most difficulty will generally arise in assessing the susceptibility to cracking of a welded structure - which is probably why cracking tests have become synonymous with weldability tests. The need to determine the susceptibility of welded alloys to various types of cracking has led to the introduction of a large number of cracking tests, most of which were originally designed to examine one specific type of cracking in a small group of related alloys. Examples are the Reeve and C.T.S. tests for underbead cracking in low alloy steels, the Houldcroft test for weld metal hot cracking in light alloys, and the various patch tests. Most of these

tests are perfectly adequate for the job they were originally designed for, which was to place a series of alloys in order of susceptibility to a particular type of cracking. Difficulties arise, however, when engineers attempt to extend the scope of tests, or assume that avoidance of cracking in a specific test means that cracking will not occur during actual fabrication.

A cracking test on a new alloy is unlikely to be of use to the welding engineer unless he is able to place the new alloy within the context of cracking susceptibility of other similar alloys for which fabrication experience exists. If this can be done, then a good estimate can be made of the likelihood of cracking in a fabrication. For commonly used materials, even information of this nature is likely to be available in the literature. It is less likely that the effect of segregation or variation in minor alloying additions and impurities within the specification has been determined, and cracking tests may prove extremely useful in this context. Even in these cases, however, a framework of results is needed into which the new cracking tests can be fitted.

DEPARTMENT OF MATERIALS

Cracking

Cracking is perhaps the most serious of weld defects, although fortunately it is not as common as porosity. Three types of cracking may be found in welded joints: cold cracking, hot cracking and delayed cracking. Cold cracking occurs in the weld metal or heat affected zone at temperatures well below the solidus, it may occur during cooling after welding or, in special cases, a few hours after welding. Hot cracking occurs during welding at temperatures at or above the solidus and may occur in the weld metal, heat affected zone or at the fusion line. Delayed cracking has been found in creep resisting steels and occurs in the heat affected zone during post weld heat treatment or in subsequent high temperature service. It has been suggested that a similar form of cracking is possible in any material where excessive grain strengthening occurs.

Cold Cracking

This type of cracking can occur in all materials where ductility is low and heavy restraint exists. The shrinkage strains generated during cooling cannot be accommodated because of the low ductility and cracking occurs at the lower temperatures of the cooling cycle. Cold cracks caused by restraint generally occur longitudinally in the weld metal and is often initiated by slag inclusions or lack of fusion in multi-run welds (Fig. 1). A special case of cold cracking known as underbead cracking can occur in the heat affected zones of hardenable steels when these are welded by a process which introduces hydrogen into the weld. Underbead cracking, which will be discussed in detail in Chapter 4, takes several hours or even days to appear and this can be dangerous since cracking may not occur until after inspection.

Hot Cracking

Hot cracking is more prevalent than cold cracking and more difficult to control; it is essentially inter-crystalline with cracks occurring along grain boundaries or between dendrites. The general mechanism for hot cracking is similar for weld metal, heat affected zone or fusion line cracking although most investigations have been confined to weld metal cracking. The basic explanation lies in the inability of the material to withstand contractional and external stresses imposed during cooling. It is generally agreed that this weakness occurs at temperatures at or above the solidus and is caused by the presence of liquid metal at the grain boundary faces.

Two suggestions have been made to explain weld metal cracking, both of which are supported by some experimental evidence. One school of thought considers that cracking results during solidification when the growing dendrites interlock and prevent free movement of the remaining liquid metal. In this temperature range (the 'shrinkage-brittleness' range) the weld metal

has little strength and no ductility, so that contraction strains cannot be accommodated, resulting in cracking. If sufficient liquid is available and can reach the cracks 'healing' can occur, but otherwise the cracks will persist. The alternative suggestion considers that cracking occurs at or near the solidus temperature during the last period of solidification when the grains are virtually fully formed and only a liquid film persists. The ratio of interphase and grain boundary energies will then define cracking tendency. In contrast to the shrinkage-brittleness approach, it is argued that cracking cannot occur at intermediate temperatures between liquidus and solidus since shrinkage strains are uniformly distributed. In this second suggestion the presence of low melting point segregates which are molten below the equilibrium solidus of the metal are particularly dangerous.

Borland¹ has recently attempted to reconcile the above suggestions and to develop a generalized theory of super-solidus cracking. Solidification may be considered to occur in three stages (Fig. 2).

- i) primary dendrite formation,
- ii) dendrite interlocking, during which both the liquid and solid phases are continuous but the liquid remains capable of free circulation around the dendrites.
- iii) grain boundary development, where solidification is considerably advanced with a semi-continuous network of grains restricting free passage of liquid; a stage which continues to the solidus temperature at which the remaining liquid freezes.

The temperature (a_c) at which dendrites interlock is known as the coherent temperature and cracking is possible once the cooling metal has reached this temperature. However, free movement of the liquid will allow healing of cracks during this stage, whereas in the third stage starting at the critical temperature, a_e , and the solidus temperature, a_{e1} . The width of the critical

The width of the critical range is also important and crack susceptibility should be at a maximum for the composition possessing the greatest critical temperature range, as shown in Fig. 3. The greatest range occurs theoretically at the limit of solid solubility, but although experimentally the variation of crack susceptibility is found to increase to a maximum with increasing alloy content before falling away, the maximum crack susceptibility was found to occur at a lower alloy content than the limit of solid solubility. Pumphrey has suggested that this discrepancy is a result of the non-equilibrium cooling conditions which will alter the limit of solubility. Further depression of the solidus can also be caused by lack of diffusion and, in some cases, by the presence of low melting point segregates. Increase of the alloy content will result in larger quantities of eutectic liquid being available for healing and with a certain critical proportion of eutectic liquid (estimated at 20% for aluminium-silicon alloys) cracking will be prevented.

The existence of a critical solidification range has been argued above for a eutectic system with limited solid solubility but it can also explain hot cracking in other types of alloy systems.

From the above discussion, certain other constitutional factors might be expected to effect susceptibility to cracking. Any factor tending to depress the solidus, such as undercooling or low melting point segregates, will increase susceptibility, conversely the presence of a second phase with a high solubility for atoms that tend to produce low melting point films will be beneficial. The crystal shape and grain size must also be considered. A crystal shape giving a high degree of entrapment is detrimental whereas a fine grain size, such as can be produced by a flat temperature gradient which encourages nucleation will reduce the tendency to cracking.

Forland suggests that the relative values of interphase (solid-liquid) and grain boundary (solid-solid) free energies are of importance in deciding the dihedral angle and hence the distribution of liquid at the grain boundaries. The relation is:

$$\tau = \frac{\gamma_{SL}}{\gamma_{SS}} = \frac{1}{2 \cos \frac{\theta}{2}}$$

where γ_{SL} = interphase energy (solid-liquid)

γ_{SS} = grain boundary energy

θ = dihedral angle.

The ratio, τ , varies upwards from 0.5 ($\theta = 0^\circ$). At a dihedral angle of 0° liquid will cover all grain faces, at an angle of 60° ($\tau = 0.57$) liquid is concentrated at grain edges, whilst at angles above 60° ($\tau > 0.57$) the liquid collects at the grain corners. Where liquid completely covers the grain faces strains are well distributed and fissures or cracks would not be expected, provided that sufficient liquid is available, while at dihedral angles of 60° and above sufficient contact should have developed to give a strong bond and thus prevent fissure formation. With intermediate conditions, however, only small areas of the solid grains may be joined and these may fracture readily under contraction strains. Under these conditions the dihedral angle may prevent healing.

The interphase and grain boundary energies can be found experimentally and it may be possible to deduce alloy compositions that are not susceptible to hot cracking, once the effect of minor additions on the energies have been determined. The main problems still to be evaluated by the theory lie in the effect of crystal shape and size (for example, the significance of columnar growth), and the effect of the moving weld pool.

Hot cracking in the heat affected zone occurs by a similar mechanism to that operating in weld metal cracking, and it is suggested that liquid

films at the grain boundaries are necessary for cracking to occur. The same relationship between the dihedral angle and interphase free energies is applicable. In this case, however, a supply of molten eutectic is not available, so that entectic healing cannot occur. This mechanism of cracking presupposes the existence of strain in the heat affected zone and it follows that if the weld metal solidifies at a lower temperature than the liquated films, stresses will not be induced until the danger has passed, and cracking will not occur. Thus, choice of weld metal can reduce heat affected zone cracking problems, and is of particular importance in the welding of aluminium alloys.

This mechanism of cracking assumes a liquid film and the origin of this film is problematical. In a cast structure, coring will give low melting point segregates at grain boundaries, but this can hardly persist in hot worked material. It may, indeed, be necessary to look at each material individually, and it has been suggested that a concentration of low melting point segregates can occur as a result of phase changes. For example, in austenitic steels, Parks³ considers that ferrite stabilisers, such as silicon and phosphorus, diffuse into the δ ferrite since their free energies of solution are less in ferrite than austenite. On subsequent cooling and as the ferrite transforms a concentration of these elements occurs and may lead to liquation of the ferrite/austenite interface during the weld thermal cycle. An experimental correlation between cracking and the presence of liquid films has not yet been satisfactorily achieved and it is suggested that other mechanisms should be considered.

With fusion line cracking, intergranular cracks run into the heat affected zone at right angles to the weld, and in certain circumstances may also extend into the weld metal. The crack mechanism is again believed to be similar to those for weld and heat affected zone cracking, in that a liquid film is necessary at the interface. It has been suggested that preferential absorption of certain elements occurs at the weld metal/base metal interface, and that these elements may penetrate the grain boundaries, lowering the solid/liquid interfacial energy and forming a base metal liquid film. Cracking will then occur in the presence of stresses and the absence of entectic healing. This suggested mechanism is of interest since, it explains the presence of base metal cracks in material that is free from fusible intergranular segregates. If partial healing occurs from the weld pool these fusion line cracks may, in fact, appear to be heat affected zone cracks.

Delayed Cracking

Delayed cracking is found in high temperature austenitic and ferritic steels after postweld heat treatment or during subsequent high temperature service. These cracks, which have been particularly troublesome in austenitic superheater steam lines operating at 550°C and above, are intergranular and occur in the heat affected zone. In many cases they appear to originate from a minor weld defect such as undercut or a hot tear at the toe of a weld (Fig. 4). During the welding cycle the ductility of a heat affected zone is severely

reduced by the strain assisted precipitation of particles (often carbides) on dislocations. Thus localised strain during postweld heat treatment or subsequent service cannot be accommodated and cracking occurs.

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DEPARTMENT OF MATERIALS

Cracking Tests

Any cracking test must be designed to investigate a specific type of cracking and, preferably, a particular range of materials. Ideally, a cracking test should be:

- a) Economical in material, time and labour.
- b) Simple in interpretation of results.
- c) Capable of giving direct correlation with actual structure.
- d) Free from the human element.
- e) Reproducible and consistent.
- f) Of high sensitivity, to allow distinction between the effects of all welding variables.

These ideal requirements can never be completely reached, indeed some of them are self-contradictory and demand a compromise with the result that cracking tests are really only sorting tests, capable of placing a series of materials in order of susceptibility to a particular type of cracking, but not capable of quantitative interpretation for an individual material. The main difficulty lies in defining restraint as it occurs in welded fabrications, and in putting a quantitative figure to any degree of restraint. Restraint is known to be important in most types of cracking but the quantitative variation of restraint in cracking tests has so far proved impossible to achieve.

It is not possible to consider all cracking tests here. Borland¹ describes 30 tests and even this list is not complete.

The majority of cracking tests were developed for carbon and low alloy

steels, and very few for other materials. Tests for underbead cracking are, of course, restricted to these steels but tests for weld metal cracking have found more general application. Cracking test results may be in the form of crack/no crack indications or on the length of cracking before arrest. The former tests will require several test specimens in which the factors affecting cracking (e.g. restraint or cooling rate) can be varied whereas the latter tests provide a continuous variation in the affecting factors (e.g. restraint in the Houldcroft fishbone test) and theoretically a single test should provide a unique indication of crack susceptibility. In practice, however, control of the variables is never exact enough to allow reliance on a single test and several tests are necessary to obtain a positive result. Most tests involve a butt or fillet weld, others require simply a bead on plate test, but few bear any direct relation to actual welded assemblies. The B.W.R.A. test for delayed cracking in austenitic steels (see below), however, relates directly to the steam pipe header joint in which this form of cracking was first recognised, whilst some industrial companies are known to use test specimens that simulate the more difficult joints that they actually encounter. These latter tests are generally nearer to procedure and welder approval tests than to tests of the crack susceptibility of a specific parent or filler material.

Most tests for heat affected zone cracking are concerned with underbead cracking in steels, and these range from a simple bead on plate test incorporating little restraint, to heavily restrained fillet weld tests such as the Reeve test and Cruciform test. The most developed test for

underbead cracking is the Controlled Thermal Severity (or C.T.S.) test which has been adopted by the International Institute of Welding as the standard test for collaborative work. The C.T.S. test incorporates less restraint than some tests and may give negative results for materials in which restraint makes a large contribution to crack susceptibility. This test has been extended to cover electrodes and preheat temperatures but in these cases great care must be exercised in interpreting the results. The B.W.R.A. test for austenitic steels was developed specifically to examine delayed cracking in steam leaders; it has been suitable for this purpose but is unlikely to be used for other materials.

Tests for weld metal cracking generally involve some form of restraint, e.g. patch test, Lehigh restraint test, although the Murex hot cracking machine applies strain to a fillet weld while it is being deposited. Crack length is taken as a measure of crack susceptibility where restraint is varied continuously in a single specimen whereas the crack/no crack criterion is used in specimens with a specific restraint.

Since no cracking test will give a direct correlation with practice the utility of the more elaborate tests may well be questioned. Unless a lot of background experience is available or a direct correlation with a practical fabrication has been found, the authors see an advantage in using the more simple tests. The larger and more complex the test the greater the cost in materials and labour.

Several cracking tests are described below, covering some of the more widely used tests.

Tests for heat affected zone cracking:

In the simple bead on plate test² used by the Battelle Memorial Institute a small weld bead is deposited on the steel under test using a cellulosic coated electrode; this type of electrode ensures that hydrogen will be present in the heat affected zone. The welding variables are all controlled, i.e. electrode size, current, voltage, and welding speed, as is the initial plate temperature. After welding, the steel plate is held for 24 hours at 15°C to enable the cracks to develop, tempered for 1 hour at 595°C and sectioned longitudinally (Fig. 1). The longitudinal section is examined for cracks metallographically or by magnetic crack detection. The susceptibility to cracking is indicated by the ratio of total crack length to bead length; at least ten tests are needed for reliability.

The C.T.S. test^{3,4} was developed by Cottrell at B.W.R.A. and is based on the assumption that underbead cracking is based on the cooling rate at 300°C (in effect the microstructure of the steel), provided that hydrogen is present. The degree of restraint is assumed to have little effect on the amount of cracking. The test assembly is shown in Fig. 2, the contact surfaces being ground to give efficient thermal contact. The bolting of the two plates and the anchor welds provides some restraint although this is not as severe as with the Reeve test,⁵ from which the C.T.S. test is derived.

Two test welds are deposited in the downhand position as indicated, the assembly being cooled to room temperature before each weld is made. For a given type and size of electrode and fixed welding conditions the cooling rate, and hence the degree of cracking, is greater with the final (trithermal) weld.

After welding the test assembly is kept at room temperature for 48 hours before being sectioned and examined microscopically for cracks.

The first weld has two paths along which heat can be conducted and the second weld three paths, which explains the use of the terms bithermal and trithermal. For any given test piece dimensions the trithermal weld will have a greater cooling rate than the bithermal weld. The unit for heat conduction has been arbitrarily taken as one section of $\frac{1}{4}$ inch thick plate and from this is derived the concept of thermal severity in terms of plate thickness. Thus the bithermal weld in two sections of $\frac{1}{4}$ inch plate has a thermal severity number (TSN) of 2 and the trithermal weld one of 3. For other plate thicknesses the thermal severity numbers are given by

$$\text{TSN} = 4(t + b) \quad \text{for bithermal welds}$$

$$\text{and } \text{TSN} = 4(t + 2b) \quad \text{for trithermal welds}$$

(see Table I) and the relation between TSN and cooling rate is shown in Figure 2 for different levels of heat input and weld sizes.

Cracking at any particular TSN is determined on a crack/no crack basis and the concept of a Weldability Index (Table II) has been proposed using fillet welds of $\frac{1}{4}$ inch leg length. The incidence of cracks at a given TSN (i.e. cooling rate) results in a material being rated from A to G in the Index. A rating of A indicates virtual freedom from underbead cracking whereas a rating of G indicates a very high susceptibility.

The type of electrode coating has a large effect on crack susceptibility and hence the Weldability Index rating, by controlling the amount of hydrogen available in the weld metal. Thus, for the Weldability Index rating of a

material to have any significance it must be prefixed by the coating group of the electrode used: a material may have an Index rating of C or D with rutile electrodes but this may be improved to A for basic electrodes. For most general material tests rutile electrodes are used since these give the greatest susceptibility to cracking.

Because the electrode coating can affect the Weldability Index the C.T.S. test has been used as a test for electrodes whilst it has also been used to investigate the influence of preheat, which affects cooling rate and hence the TSN. In other tests the severity of the restraint has been increased by the use of high tensile weld deposits or by altering the notch at the root of the fillet welds. These variations can give useful information but care must be taken in comparing tests made under these different conditions.

The B.W.R.A. test for delayed cracking in Cr-Ni austenitic steels⁶ is shown in Fig. 3, welds are made with coated electrodes in the sequence indicated. Each run is made in two 180° segments followed by cooling to room temperature. The rigidity of the specimen is often sufficient to cause minor hot cracking at the toe of the weld and these propagate on subsequent heat treatment. A series of tests are necessary to define the temperature range over which cracking may occur.

Tests for weld metal cracking:

A large number of these tests exist ranging from simple pad tests to the production of butt and fillet welds under considerable restraint.

The simplest for weld metal cracking consists of depositing a pad of weld metal on a $\frac{1}{2}$ inch thick plate of the material to be welded:⁷ cracking

is then assessed by micro-examination. This test has been used largely for assessing austenitic weld deposits but since it involves little restraint it does not always give positive results. The effect of dilution of the weld metal by the parent plate is obviously avoided in this test.

Several versions of T fillet weld test exist, including one adopted by the British Standards Institution for testing covered electrodes for mild⁸ and medium high tensile steel⁹. Each test requires a T joint to be produced in the plate material ($\frac{1}{2}$ inch thick for BS 2549), welds on either side of the T being produced successively (see Fig. 4a). Details of plate preparation, preheat and welding conditions are standardised to allow reproduction of the restraint and the second weld is produced under a greater degree of restraint than the first. After visual examination for cracking the fillet welds may be broken open (Fig. 4b) and examined for further evidence of hot cracking.

The Lehigh restraint test¹⁰ was developed to study the effect of restraint on cracking in steel weld deposits; it consists of a slotted plate with a machined centre groove in which the weld is deposited (Fig. 5). The restraint is controlled by the depth of the slots and is measured as the width of the plate between the slots. In general only a root run is deposited and cracking is determined by macro-examination. The test may also be used to study cracking in the weld h.a.z.

The Murex hot cracking machine^{11,12} places a stress on a cooling weld deposit and offers a useful method of distinguishing the hot cracking tendencies of different electrodes. For this test a 3 inch fillet weld is

deposited in the V formed between two $3\text{ inch} \times 2\text{ inch} \times \frac{1}{2}\text{ inch}$ plates (see Fig. 6). One of the plates is rotated at a constant speed, 5 seconds after the welder starts depositing the electrode. A total rotation of 30° is effected giving a maximum fillet angle of 120° . The speed of rotation can be varied for different tests; a speed of $1.10^\circ/\text{sec.}$ has been found suitable for low alloy electrodes. Any cracks obtained tend to be longitudinal and the length of crack a good measure of the crack susceptibility of the electrode, but is less sensitive when considering the effect of plate composition, probably due to the small amount of dilution that occurs.

Patch tests^{13,14} have been used extensively to examine hot cracking in a wide range of ferrous and non-ferrous materials. The most usual test for materials up to $\frac{1}{8}\text{ inch}$ in thickness (Fig. 7) consists of a 6 inch square from which a 2 inch diameter circular patch is trepanned. The patch is welded back into the square and the crack susceptibility expressed as a percentage of total weld length or by the angle subtended by the weld start and point of crack initiation. The restraint will vary throughout welding and cracking will not start until a sufficient level of restraint has been reached. Experiments with austenitic steels have shown that variation of the plate to patch ratio had no effect on the results but that an increase in welding speed increases the severity of the test. A suggestion has also been made that a simple circular welding bead in a square of test material would give the same results as the true patch test; a test made in a light alloy is shown in Fig. 8. Patch tests are also used for thicker materials¹⁵ in which case larger plates are used with an appropriate edge preparation.

The Houldcroft fishbone test was developed to examine the crack susceptibility of light alloy sheet during TIG welding, with or without filler additions. The test has since been used for other sheet material. It is based on the concept that if a weld is started on the edge of a sheet, a crack can initiate and will then propagate until the restraint is reduced. In the fishbone test restraint is progressively reduced by a series of slots of gradually increasing depth as shown in Fig. 9. The test weld is made along the centreline of the specimen and full penetration and a constant bead width is essential; manual welding generally gives a more uniform bead although automatic TIG welding has also been used. The test specimen is not restrained externally but rested on a carbon block which provides good electrical contact but poor heat conduction. The crack susceptibility is measured in terms of crack length, a mean of 6 tests being taken. On the basis of work on $1/16$ inch thick Al-Mg-Si alloys Houldcroft has suggested that, for welding speeds around 8 ins/min. and bead widths of $1/4$ inch, a crack length of less than $1/2$ inch indicates a fully weldable material whereas a crack length greater than $1 1/4$ inches indicates a high crack susceptibility. Although the test is generally discriminatory between different alloys it does fail to distinguish between the crack susceptibility of some alloys of similar compositions e.g. Al-3.5Mg and Al-5Mg alloys, and some scatter is found between nominally identical specimens. It was suggested¹⁷ that these problems resulted from a variation in heat flow pattern at the weld starts and an unslotted portion was added to the testpiece to enable the heat flow pattern to become steady before the first slot was reached. This modification improved the discrimination of

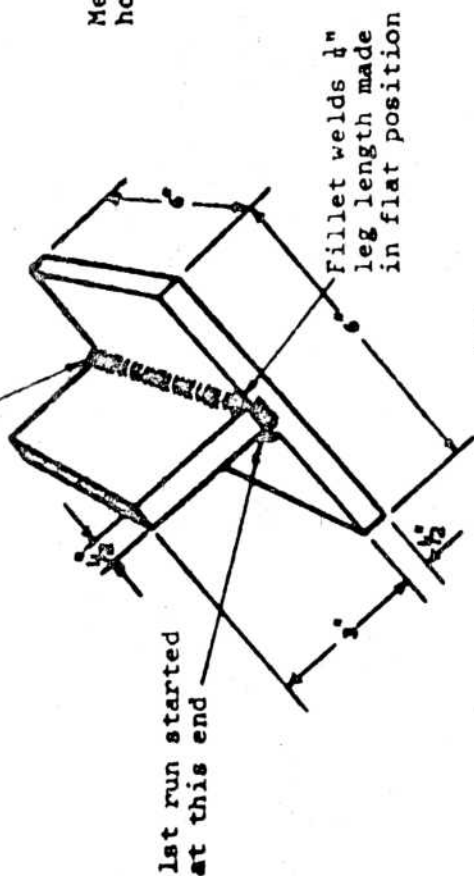
the test at lower welding speeds but did not reduce the scatter.

A recent comparison of the Houldcroft and patch tests^{1a} have shown that the latter test covers a more limited range of crack susceptibilities than the former. Over this more limited range, however, the patch test appears to bear a closer relation to practical welding behaviour.

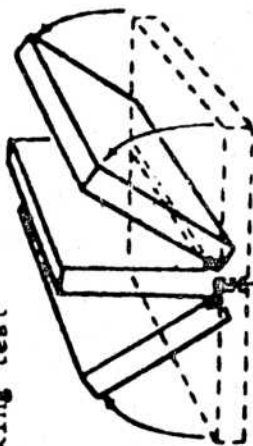
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2nd run started at this end



Method of breaking hot cracking test



(b)

(a)

Fig. 4. British Standard single-T test specimen

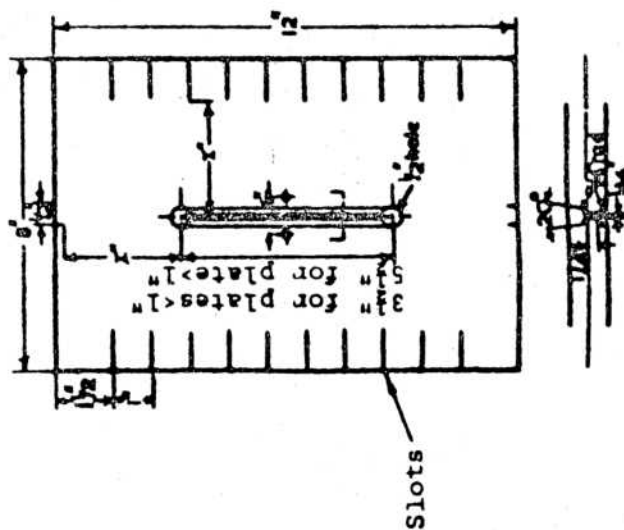


Fig. 5. Lehigh restraint specimen

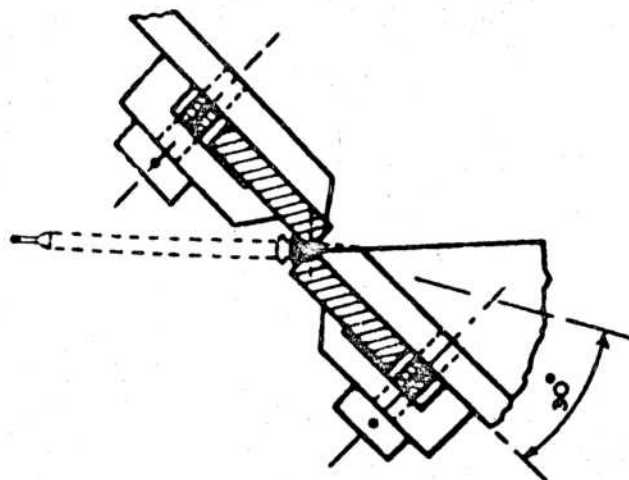


Fig. 6. Principle of Murex hot crack machine

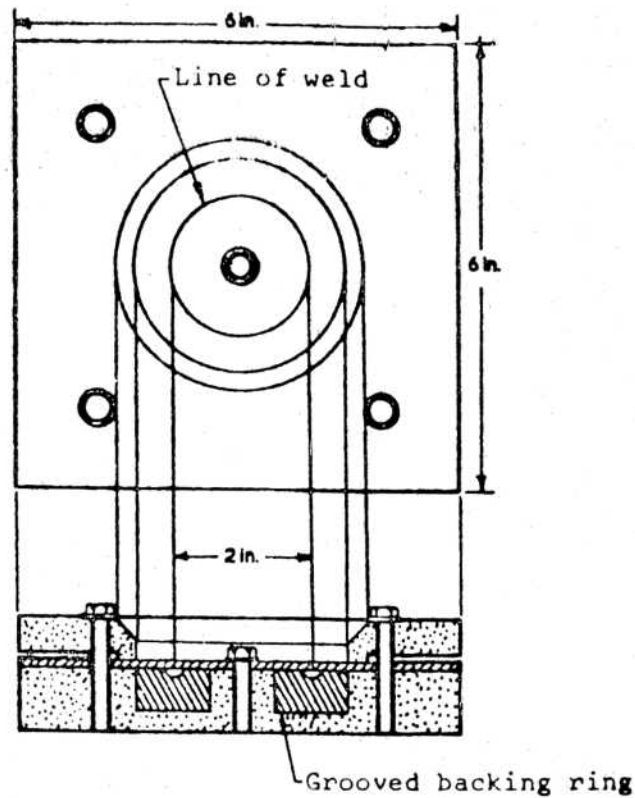


Fig.7. Jig for patch test in sheet material

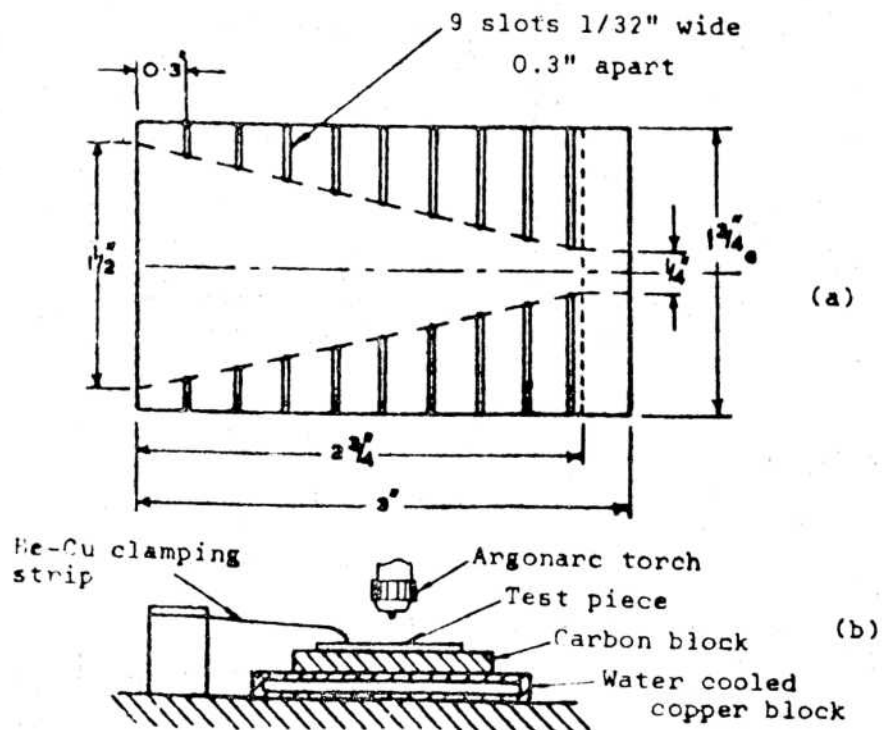


Fig.9. Houldcroft fishbone specimen:
 (a) For $\frac{1}{16}$ in. thick material
 (increased by 50% for $\frac{1}{8}$ in sheet);
 (b) section through test assembly.

1. Preheating - consists of heating the weld joint to a predetermined temperature before welding in order to produce a satisfactory joint. Preheat temperatures vary from 20°C to 600°C depending on the material. Preheating of the whole component is desirable. If impractical localised preheating can be applied preferably over a distance 6-12 times the plate thickness on either side of the joint.
2. Interpass Temperature - maintaining the preheat temperature between successive passes in a multirun weld.
3. Post Heating - maintaining the preheat and interpass temperature after welding is complete.
4. Tempering/Ageing - heating to temperatures below that required for transformation in order to promote precipitation of a second phase.
5. Solution Treatment - heating to temperatures above the transformation temperatures in order to redissolve second phases or precipitates. Rapid cooling then forms a single phase solid solution.
6. Annealing - for steels involves heating to temperatures $20-30^{\circ}\text{C}$ above the upper critical temperature, A_3 , followed by slow furnace cooling. Produces a soft structure.
7. Normalising - for steels involves heating to temperatures $20-30^{\circ}\text{C}$ above A_3 followed by cooling in still air. Useful for grain refinement and consequent improvement in mechanical properties.
8. Quenching - usually involves heating to above the transformation temperature ($20-30^{\circ}\text{C}$ above A_3 for steels) followed by controlled cooling in water, oil, salt bath, or air blast.
9. Stress-relieving - heat treatment applied after welding in order to reduce the level of residual stress. Tempering and ageing effects may also occur. In steels susceptible to secondary hardening or temper embrittlement the treatment may have a deleterious effect on mechanical properties.

This question must be asked, since heat treatments are expensive in both time and money. The advantages and disadvantages of heat treatment, particularly related to welding, require a combination of knowledge and experience to achieve a balance between effective quality and economic viability.

The characteristics of a fusion-welded joint, which comprises weld metal, the heat-affected zone of the parent material, and the unaffected parent material, are:

- (i) change in microstructures, grain size, and properties
- (ii) residual stresses from thermal expansion and contraction leading to distortion.

These factors can be affected by the many variables involved, such as:

- (i) type and quality of material being welded
- (ii) thickness and condition of the material
- (iii) design and geometric complexity of the welded structure
- (iv) type of welding process and welding consumables selected, the heat input, and the thermal conditions imposed
- (v) Service conditions in which the welded structure is operating.

In many cases heat treatment is essential to restore the desired properties and reduce the residual-stress levels.

WHEN TO HEAT TREAT?

In many cases heat treatment is necessary before, during, and after welding.

Preheat treatment

This operation is practised to achieve:

- (i) a reduction in thermal stresses during welding
- (ii) control of cooling rate in the heat-affected zone of a weld
- (iii) control of the rate of diffusion of hydrogen in a welded joint
- (iv) assistance in fusing high thermal conductivity materials.

Preheat is expensive and should be applied only when necessary to achieve a satisfactory joint. Preheat is much more expensive if applied incorrectly, if limited, or even avoided, due to other external factors, such as:

- (i) cost
- (ii) lack of appreciation of the function of preheat
- (iii) inefficiency of equipment for preheating
- (iv) welders' objections to working under 'hot' conditions
- (v) lack of control of temperature
- (vi) tack welding without preheat
- (vii) inspection authorities' stage inspection requirements.

The great importance of composition, microstructure, stress, and hydrogen in the weldability of the thicker carbon steels, low- and high-alloy steels, and castings in these materials, has been recognized in the voluminous published literature available, and to a great extent in practice. In practice, however, the two factors which have given rise to much trouble and expense are inefficient preheat and lack of effective drying of welding consumables.

Whatever post-heat treatment and non-destructive inspection is applied at a later stage, it is nonsensical to start with a cracked component due to lack of attention to preheat and failure to control the major source of hydrogen in the welded joint.

The welding of the range of C-Mn alloy steels have provided many examples of malpractice in welding and heat-treatment procedure. As an example, many inspection authorities insist that the root run of one side pipe and tube welds must be crack-detected or radiographed before proceeding to complete the weld. Surely, wherever possible, there is considerable advantage in applying the correct preheat, then to tack-weld, weld, and preferably post-heat-treat as one cycle, rather than oscillate in the dangerous lower transformation regions. All things being equal, there is a much better chance of achieving

sound welds by this procedure than by satisfying what could in some cases be considered as the 'whim' of the inspecting authority.

Post-heat treatment

Heat treatment after welding may be required

- (i) to develop the optimum mechanical properties in the welded joint
- (ii) to reduce the residual-stress level and thus (a) improve the resistance to brittle fracture or fatigue failure; (b) reduce the risk of stress-corrosion cracking; (c) maintain dimensional stability, particularly after machining.

All the conventional types of heat treatment are applied to welded structures, i.e. normalizing, annealing, quenching, tempering, solution treating, and stress relieving, the latter being by far the most common.

How these treatments are applied depends upon the type and size of welded structures involved. Gas and electricity are the main media, for heating either completely in furnaces, or by specialized induction-heating coils or other flexible devices. Whatever the means, the equipment must meet the following requirements:

- (i) be capable of achieving the required temperature
- (ii) provide uniform heating conditions throughout all section thickness of the component
- (iii) control and preferably record the temperature within the limits of accuracy specified.

Modern equipment can provide these conditions, but adequate attention to maintenance is essential. In the general engineering industry the availability of equipment of various vintages and degree of control leaves much to be desired. Repeat heat treatments, decreed by control tests of hardness and mechanical properties, can create severe problems of late delivery and escalating costs, and sometimes outright rejection.

EFFECTS OF WELDING AND HEAT TREATMENT

Whenever conditions demand a specific heat-treatment cycle, it is essential that procedure tests are carried out. These should be as closely simulated as possible to the actual conditions in practice. This can often be difficult to satisfy from laboratory to shop-floor conditions. Procedure tests will give confirmation of the latitude which can be expected in manufacture, and serve to indicate the likely behaviour of the structure under service conditions.

It is important to ensure that any testpieces required for approval are welded under the same conditions as the structure, and that any heat-treatment procedure is the same, whether an integral or separate testpiece is used. Much time and cost has been expended in failure to give close attention to this logical exercise.

NON-FERROUS HEAT TREATMENT

The heat-treatment of non-ferrous metals is a topic which covers a number of operations but many of the treatments are of the solution treatment and ageing type.

In the heat treatment of aluminium alloy forgings and pressings, for example, furnaces are employed which are capable of operating at any temperature in the range 100 to 700°C. Because many of the components treated in such furnaces may be destined for the aircraft industry, such flexibility must not be gained at the expense of accurate control, correct heat treatment being essential. Throughout the operating range of these furnaces temperature control is automatic to 2°C, so that they are equally suitable for the solution treatment or ageing stage of the cycle. A typical furnace is rated at 200 kW and it can carry out various treatments on up to 3 tons of aluminium alloy at one time with a maximum permissible dimension of 10ft. in the charge. Somewhat similar furnaces (Fig. 3) are capable of operation at higher temperatures - 400°C to 1000°C - and can be used to heat treat forgings fabricated in titanium and "Nimonic" alloys. They approach the low temperature furnaces in controllability, which for the materials involved is naturally all important.

Furnaces covering the above ranges are also quite suitable for other non-ferrous heat treatment operations which are carried out, such as annealing, homogenising and treatments of the "quench-and-temper" type, according to the requirements of the material being processed.

VACUUM BRAZING

The technique of producing brazed joints in vacuum is on the fringe of heat treatment but it is appropriate to mention it here. For example, vacuum brazing has now been developed to the point where it is capable of producing joints in high-temperature, corrosion-resistant alloys which are consistently strong and ductile and do not detract from the properties of the parent metal.

(Fig.6). Under the unique conditions of high vacuum, no flux is required and this makes an important contribution to the quality of the joint. Distortion is also inherently at a minimum. Filler alloys based on nickel, with additions of chromium, manganese, gold, platinum or palladium are used with melting temperatures in the range 900-1300°C. The degree of alloying of the filler with the base metal must be carefully controlled by suitable adjustment of time and temperature and each new application may require some development work and trials to be carried out. Once this has been done the results obtained are consistently reproducible. Useful experience has already been gained in this field (Table IV) and further applications may be anticipated.

A conventional vacuum furnace can braze a large number of assemblies at once, but it is still essentially a batch process and unsuitable for very high production rates when the components are relatively large. A semi-continuous vacuum brazing unit has been described with an entry zone, one or more heating zones and an exit zone. This type of furnace has been used in the fluxless brazing of aluminium automotive heat exchangers at 595 to 620°C. It has been suggested that a similar concept could replace the conventional copper automobile radiator and that such components would be capable of meeting the requirements of present day motor car engines and also future types which would operate at increased coolant temperatures and pressures.

DEVELOPMENTS IN FURNACES FOR STRESS-RELIEVING AND ANNEALING

Welding in its various forms is much better known than vacuum brazing as a joining process and is nowadays somewhat taken for granted. But engineers and metallurgists have long been aware that welding can set up severe internal stresses in a fabricated vessel or component. Also, a large casting may contain internal stress as a result of differential cooling in sections of varying thickness. In either case, thermal stress relief may be employed to avoid subsequent distortion, stress-corrosion cracking or similar problems; for example, carbon-manganese steels may be prone to cracking when in contact with caustic or nitrate solutions and austenitic steels subject to attack at chloride-ion concentrations of only a few p.p.m.

Standard procedures have been laid down for the stress-relief of welded fabrications. For most structures the temperature is raised slowly and

uniformly to between 600 and 650°C although for materials such as the austenitic stainless steels higher temperatures are required. The temperature throughout the structure being treated should vary by less than 25°C although sometimes 20°C is specified and indeed some industrial users of these weldments demand that 10°C be attained. After heating, slow controlled cooling is carried out to avoid introducing further residual stresses.

The distinction between stress relief and annealing in the treatment of welded structures is in practice obscure or one of nomenclature only but where substantial recrystallisation of the metallurgical structure takes place the term annealing is appropriate. Until recently furnaces for annealing and stress relieving were fabricated with relatively heavy insulating materials which absorbed and stored much of the energy supplied during the heating-up period. Because of the basically batch-type nature of the process, this energy was largely wasted and thermal efficiency was effectively only 10-15%. Current furnaces for these applications employ lightweight materials which are very good insulators, such as calcium silicate used up to about 750°C, or ceramic fibres with temperature capability of 1 100°C or higher. These are often referred to as "low thermal mass" (LTM) furnaces. This means that the process not only uses less energy but the cycle time is shorter and better utilisation of the furnace is possible (see Fig.7).

The indirect of resistance furnace is particularly suited to the application of the LTM concept. Obviously, uniform heating and cooling of the structures being treated is of prime importance. In the case of large components, especially where the furnace might become unevenly loaded thermally, the furnace has individually controlled furnace zones so that each part of the fabrication may be heated at the same rate to the desired temperature. To ensure that the intended heating cycle is closely adhered to, an electrical signal is generated which is proportional to the temperature required at any particular point in the cycle. The method frequently adopted is to use a cam with a profile equivalent to the time/temperature curve required. A cam follower driving a potentiometer can then be used to provide the necessary comparative signal. The heating elements are supplied with power as necessary to bring the temperature within each zone to the level required at that particular time. Such a control system is readily applied to an electric LTM furnace and indeed the whole heat treatment cycle may be controlled automatically, very often being carried out at night without supervision and taking advantage of the favourable energy tariffs available.

TABLE IV Some materials and material combinations suitable forVacuum Brazing

Base Materials	Brazing Alloys (Filler Metals)	Brazing Temp. °C
Aluminium alloys 2002, 2219, 3003, 6061	88 Al-12 Si	580-595
Beryllium	88 Al-12 Si, 92.5 Ag-7.5 Cu, 72 Ag-28 Al, 72 Ag-28 Cu, 94 Ti-6 Be, Ti-Cu-In, Ni	780-900
Superalloys Hastelloy X	Ni-Mn-Si-Cu, Au-Pd-Ni, Ni-Cr-Si	900-1260
Inconel 718	Au-Ni-Cu, Mn-Ni-Co-B, Cu-Mn-Co-Ni	
Udimet 700	Ni-Cr-Mn-Si, Pd-Ni-Cr-Si, Pd-Ni-Cr	
Haynes 25	Au-Pd-Ni, Co-Ni-Cr-Si-W-B, Ni-Cr-Si, Pd-Ni-Cr	
TD Nickel	50 Ni-25 Mo-16 Cr-5 W-4 Si, 54 Pd-36 Cr-10 Ni	
TD-Ni-Cr	38 Ni-21 Fe-20 Cr-10 Si-9 Mo-2 Co	
Titanium Alloys Ti-6 Al-4 V Ti-8 Al-1 Mo-1 V Ti-5 Al-2.5 Sn	Ag-Al, Ag-Cu-Ge Ag, Ti-43 Zr-12 Ni-2 Be Ag-Cu, Ti-15 Cu-15 Ni	900-1260
Refractory Metals Columbium	Ti-8.5 Si, Ti-33 Cr, 66 Ti-30 V-4 Be, 48 Ti-48 Zr-4 Be, 75 Zr-19 Cb-6 Be, Ti-13V-11 Cr-1 Al	980-1790
Molybdenum	Ti-8.5 Si, 62 Ti-25 Cr-13 Ni	980-1490
Tantalum	34 Ti-33 Zr-33 V, Ti-30 V, Hf-7 Mo, Hf-40 Ta	1650-2050
Tungsten	90 W-6 Ni-4 Fe, Pt-3.5 B	1093-1920

The main technical benefits of the electric low thermal mass furnace are the precise and uniform control of temperature and, because of the restricted access of air resulting from the virtual sealing of the furnace, the low level of oxidation which the treated structure suffers during the cycle. This means that, while avoiding the complications of a conventional controlled furnace atmosphere, it is not normally necessary to carry out an expensive shot-blasting operation on the fabrication after stress-relieving.

1) Specifications of Welding Requirements

BS639 : 1972 Covered electrodes for the manual metal-arc welding of mild steel and medium-tensile steel

Requirements for all classes of normal penetration electrodes, including hydrogen controlled electrodes, and deep penetration electrodes. Tolerance on diameter of core wire, covering, packing and storage, marking, certification, initial and production control tests, period check tests, retests and test certificates.

BS1719 : Part 1 : 1969 Classification and coding of covered electrodes for metal-arc welding

Classification and coding of welding electrodes based on method of manufacture, flux covering, welding position, welding current and voltage. Appendices on characteristics of flux coverings, determination of diffusible hydrogen, determination of metal recovery.

BS2642 : 1965 General requirements for the arc welding of carbon manganese steels

Parent metals and carbon equivalents; electrodes; control of cracking and excessive hardening of heat affected zone; preheat and inter-pass temperature; welding procedure; workmanship in joint preparation and welding; quality of welds. Appendices on design and types of weld preparation.

BS1821 : 1957 Class I oxy-acetylene welding of steel pipelines and pipe assemblies for carrying fluids

Deals with the shop and site oxy-acetylene welding of steel pipelines of pipe assemblies for carrying fluids suitable for Class I conditions in sizes up to 24 inches diameter.

The specification is subdivided into general requirements and requirements for butt-joints, and branch and sleeve welds. In each case details of the preparation and technique are specified.

BS1856 : 1964 General requirements for the metal-arc welding of mild steel

Specifies the general requirements for the metal-arc welding of mild steel under normal conditions of manufacture and service. The standard includes parts dealing with the following:

BS2493 : 1971 Low alloy steel electrodes for manual metal arc welding

Similar to BS1719:1969 and therefore now due for replacement.

BS4165 : 1971 Electrode wires and fluxes for the submerged arc welding of carbon steel and medium-tensile steel

Materials, details of butt and fillet welds, information required by the welding organization, welding equipment, welding procedure, workmanship, welding operators, supervision and inspection, heat treatment and testing.

The requirements are not all applicable to certain forms of construction, nor does the standard specify the allowable stresses in welds or methods for their testing.

Notes on butt welds and fillet welds are included as appendices.

BS1965 : Butt-welding pipe fittings for pressure purposes

Part 1 : 1963 Carbon steel

Specifies leading dimensions, tolerances and materials for 90 and 45 elbows 180° return bends equal tees, concentric and eccentric reducers and caps, in carbon steel, for butt welding on to pipes.

Details of testing and marking are also given.

Part 2 : 1964 Austenitic stainless steel

Dimensions, tolerances and materials for 90 and 45 elbows, 180 return bends, concentric and eccentric reducers, equal and reducing tees and caps, in wrought and cast austenitic stainless steel, for butt-welding on to pipes. Details of testing and marking.

BS2633 : 1973 Class 1 arc welding of ferritic steel pipe work for carrying fluids

Arc welding using manual, automatic or semi-automatic processes; parent metals, welding consumables, preparation for welding, butt joints, branches, attachments, flanges, testing and inspection, acceptance requirements, procedure and welder approval tests.

BS2971 : 1961 Class II metal-arc welding of steel pipelines and pipe assemblies for carrying fluids

Specifies requirements for shop and site metal-arc welding of steel pipelines and pipe assemblies for carrying fluids suitable for Class II conditions in sizes up to 24 in diameter. The specification is divided into general requirements, requirements for butt joints, branches, sleeve joints and flanges, requirements for the rectification of faulty welds, and testing. In each case, details of the preparation and technique are specified.

BS4515 : 1969 Field welding of carbon steel pipelines

Arc welding by various manual, automatic and semi-automatic processes of butt joints, branch connections, fillet welds and socket welds in carbon steel transmission pipelines. Qualification of welding procedure, qualification of welders, general requirements for workmanship, inspection of production welds, standard of acceptance, rectification of welds.

BS4677 : 1971 Class 1 arc welding of austenitic stainless steel pipework for carrying fluids

Arc welding using manual, automatic or semi-automatic processes; parent metals, welding consumables, preparation for welding; butt joints, branches, attachments, flanges; testing and inspection; acceptance requirements; procedure and welder qualification tests.

2) Specifications for Materials

BS1501 : Steels for fired and unfired pressure vessels. Plates

Part 1 : 1964 Carbon and carbon manganese steels. Imperial units

General and specific requirements of a range of steels for plates to specifications 1501-151, 1501-161, 1501-211, 1501-213, 1501-221 and 1501-224. Chemical compositions, mechanical properties at room temperature and at elevated temperature, low temperature impact properties. Appendices on (i) stress rupture and creep properties and (ii) influence of fabrication and heat treatment on mechanical properties.

Part 2 : 1970 Alloy steels. Imperial units

Requirements for a range of alloy steel plates to specifications 1501-261, 1501-271, 1501-281, 1501-282, 1501-503, 1501-509, 1501-510, 1501-620 and 1501-622. Chemical compositions, mechanical properties at room temperature and at elevated temperatures, low temperature impact properties. Appendices on creep rupture properties and influence of fabrication and heat treatment on mechanical properties.

BS1501 : Part 3 : Corrosion and heat resisting steel: Imperial units

Requirements for a range of corrosion and heat resisting steel plates (including high proof stress steels). Chemical compositions, mechanical properties at room temperature and at elevated temperatures, low temperature impact properties. Appendices on stress rupture properties 0.2 per cent proof stress properties, and influence of fabrication and heat treatment on mechanical properties.

BS1501-6 : 1958 Steels for use in the chemical, petroleum and allied industries

Provides for a series of carbon and alloy steel specifications in the following forms: 1504, castings; 1506, bars for bolting material; 1501, which has been revised and now applies to sections and bars to specifications 1501-240 to 1501-846 inclusive.

For each of these standards there are general clauses covering such features as freedom from defects, margins of manufacture, identification, mechanical tests, inspection, etc, followed by a series of specifications which give the detailed requirements of chemical composition and the mechanical properties. An appendix is included explaining the numbering system which has been adopted whereby the reference number of the steel gives a broad indication of the type of steel to which it refers. Further appendices are included giving comparable British and American specifications and recommendations for repairs by welding to castings in BS 1504.

BS1502 : 1968 Steels for fired and unfired pressure vessels. Sections and bars

General and specific requirements for range of steels for sections and bars to specifications 1502-151, 1502-161, 1502-211, 1502-221 and 1502-224. Chemical compositions, room temperature mechanical properties, low temperature impact properties.

BS1503 : 1969 Steels for fired and unfired pressure vessels. Forgings

General and specific requirements for a range of carbon and low alloy and stainless steels for forgings. Chemical compositions, mechanical properties at room temperature and at elevated temperatures, low temperature impact properties. Appendices on creep and creep rupture properties and stress relieving.

BS4360 : 1968 Weldable Structural Steels

Requirements for a range of structural steels in terms of compositional carbon equivalent and properties, including Charpy V values. Many steels need to be normalised to meet properties. For weldability refer to BS1856 and BS2642.

BS1510 : 1958 Steels for use in the chemical, petroleum and allied industries (Low temperature supplementary requirements to BS 1501-1506)

Specifies additional requirements for steels covered by BS1501 and BS1506 intended for operation under severe conditions at low temperatures or for applications where a high degree of notch ductility is required. Applies to sections and bars having a ruling section not exceeding 4 in.

BS1515 : Fusion welded pressure vessels for use in the chemical, petroleum and allied industries

Part 1 : 1965 Carbon and ferritic alloy steels

Pressure vessels subject to internal pressure. Materials, criteria for establishing design stresses. Calculation of scantlings of cylindrical and spherical shells, domed, conical and flat ends; compensation of openings; manhole and inspection openings; bolted flanges; supports and internal structures. Manufacturing procedure, welder performance tests, post-welding heat treatment. Non-destructive tests (with unacceptable weld imperfections as revealed by radiographs) and pressure tests. Appendices cover: calculation of stresses arising from local loads and thermal gradients; avoidance of fatigue cracking vessels required to operate at low temperatures; pressure relief devices; welded connections; approximate metric equivalents; design stress values for materials for which elevated temperature properties are not specified.

Part 2 : 1968 Austenitic stainless steel

Pressure vessels subject to internal and external pressure. Materials, criteria for establishing design stresses. Calculation of scantlings of cylindrical and spherical shells; domed, conical, flat, spherically domed and bolted ends; flat tube plates; openings and branch connections; bolted flanges; manholes and inspection openings; jacketed construction; supports and internal structures. Manufacturing procedure, weld procedure and welder performance tests, tolerances, surface finish. Non-destructive tests (with unacceptable weld imperfections as revealed by radiography) and pressure tests. Appendices cover calculation of stresses arising from local loads and thermal gradients; avoidance of fatigue cracking; descaling; pressure relief protective devices; welded connections and welded seams; design of openings and penetrations; design stresses; selection of economic flange proportions; stages of inspection.

BS4534 : 1969 Weldable chromium-nickel centrifugally cast steel tubes

Specifies requirements covering 14 grades of steel. Chemical composition, mechanical properties, testing, non-destructive testing either dye-penetrant or acid-etched ring; standards of acceptance, dimensional tolerances and marking. Charts establish acceptable levels of dye-penetrant indications for four classes of tube. An appendix shows the relationship between British and American Standard Specifications.

BS449 : The use of structural steel in building

Part 1 : 1970 Imperial units

Imperial unit version of Part 2. It is not a technical revision of the standard except that it includes the use of steels to BS 4360

Part 2 : 1969 Metric units

Metricated version of Part 1, in accordance with the change to metric system in the construction industry. Also expanded to cover the use of steels complying with BS4360

Addendum No.1 (PD 4064) to BS 449: Part 1 : 1970 The use of cold formed steel sections in building

For use in conjunction with BS449. Those clauses of BS449 which do, and those which do not, apply to the use of cold formed sections are indicated. The methods of design apply to the basic shapes given in BS2994 and to other cold formed shapes that may be required for special conditions. Worked out examples illustrate design methods.

BS4 : Structural steel sections

Part 1 : 1972 Hot-rolled sections

Sizes and properties in metric and imperial units for hot-rolled structural steel sections. Metric designations are included for the first time. Prepared in conjunction and aligning with the BCSA Handbook on Structural Steels 1971.

Part 2 : 1969 Hot-rolled hollow sections

Circular sections (tubes) and rectangular sections (including squares). Dimensional, out-of-square and weight tolerances. Tables of dimensions, weight per metre and properties, e.g. moment of inertia, radius of gyration, elastic modulus.

BS4848 : Hot-rolled structural steel sections

Part 4 : 1972 Equal and unequal angles

Dimensions, rolling tolerances and properties of a co-ordinated metric range of equal and unequal angles. Two tables are included giving those angles in the British Standard range, whilst Appendix A gives those ISO angles not included in the British Standard range. Partially in agreement with ISO/R657 Parts 1 and 2.

BS24 : Railway rolling stock material

Part 1 : 1956 Axles

Locomotive crank axles, solid forged; locomotive straight axles; locomotive straight axles (acid open hearth steel); carriage and wagon axles. Quality of material, heat treatment, selection of test pieces and methods of test. Form of test pieces appended.

Part 2 : 1956 Tyres

Locomotive tyres; carriage and wagon tyres. Quality of material, heat treatment, sampling, methods of test. Form of test piece appended.

Part 3A: 1959 Laminated springs and spring steel

Section 1 : steel for the springs; composition, rolling tolerances, cambering test, identification, protection, bundling, Section 2 : steel for buckles; composition, mechanical tests. Section 3 : quality of material, assembly of springs and buckles, scragging test, load deflection test, tolerances on finished springs. Form of test piece, in an appendix.

Part 3B : 1955 Helical and volute springs and spring steels

Section 1 : steel for the springs; composition, rolling tolerances, identification. Section 2 : springs made from hot rolled steel bars to Sec. 1; quality of material, compression test, workmanship. Section 3 : steel for helical springs; composition, grinding tolerance, identification. Section 4: helical springs made from round ground steel bars to Sec. 3; quality of material, compression test, workmanship, tolerances. Method of calculating number of helical coils, in an appendix.

Part 4 : 1956 Steel billets, blooms, bars and forgings

Section 1 : steel bars, billets, blooms and slabs for forging purpose; quality of material, rolling margins, identification. Section 2 : steel forgings, forged bars and classes C and D rolled bars; material, identification, sampling; tensile and bend tests. Form of test piece, heat treatment, in appendices.

Part 6 : 1957 Steel slabs, plates, sections, bars and rivets for locomotive boilers, locomotives, carriages and wagons

Section 1 : boiler (inc. inner firebox) plates, sections, bars for boilers. Section 2 : plates, section, bars for locomotives, carriages, wagons. Section 3 : frameplates for locomotives, tenders, bogies. Section 4 : slabs for locomotive and tender bar frames. Section 5 : rivet bars and rivets for boilers and rolling stock. Quality of material, sampling, methods of test, in all Sections. Form of test pieces, metric tables, in appendices.

BS970 : Wrought steels in the form of blooms, billets, bars and forgings

Part 1 : 1972 Carbon and carbon manganese steels including free cutting steels

General requirements: information to be supplied by purchaser, product analysis and permitted variations, condition of material on deliver, provision of test material, testing etc. En steels replaced. Specific requirements for carbon and carbon free cutting steels supplied to close limits of chemical composition. Specific requirements for carbon manganese and carbon free cutting steels supplied to specific mechanical properties. Summary tables. Appendices include guidance for the purchase of steels to hardenability requirements by special arrangements.

Part 2 : 1970 Direct hardening alloy steels, including alloy steels capable of surface hardening by nitriding

General requirements. Information to be supplied by purchaser product analysis and permitted variations, condition of material on delivery, provision of test material, testing, etc. En steels replaced. Specific requirements for alloy steels supplied to close limits of chemical composition, specific requirements for alloy steels supplied to harenability requirements and specific requirements for alloy steels supplied to specific mechanical properties. Summary tables.

Part 3 : 1971 Steels for case hardening

This part of BS970 presently does not cover the carbon and carbon manganese steels for case hardening. Until requirements for these steels are added to this part of the standard the provisions of BS970: 1955 continue to apply for them.

General requirements, information to be supplied by purchaser, product analysis and permitted variations, condition of material on delivery, provision of test material, testing, etc. En steels replaced.

Specific requirements for alloy steels for case hardening applied to close limits of chemical composition, chemical composition of individual steels.

Specific requirements for alloy steels for case hardening supplied to hardenability requirements, purchase to hardenability requirements, chemical composition of individual steels, hardenability data.

Specific requirements for alloy steels for case hardening supplied with specified mechanical properties. Summary tables.

Supplement No.1 (1973) to BS970 : Part 3

General requirements, information to be supplied by the purchaser, product analysis and permitted variations, condition of material on delivery provision of test material, testing, etc. Specific requirements for carbon and carbon manganese steels for case hardening supplied to close limits of chemical composition, chemical composition of individual steels; and supplied with specified mechanical properties. En steels replaced. Summary tables. (There are no requirements for these steels supplied to hardenability requirements and therefore no Section 3 is applicable to this Supplement.) Supersedes BS970.

Part 4 : 1970 Stainless, heat resisting and valve steels

Information to be supplied by purchaser, product analysis and permitted variations, condition of material on delivery, provision of test material, testing, etc. En steels replaced. Chemical composition, heat treatment and mechanical properties of individual stainless and heat resisting steels. Chemical composition, condition of supply, approval test and heat treatment of individual valve steels. Summary tables.

Part 5 : 1972 Carbon and alloy spring steels for the manufacture of hot-formed springs

General requirements, information to be supplied by the purchaser, product analysis and permitted variations, frequency of tests etc. En steels replaced. Specific requirements for spring steels supplied to close limits of chemical composition; composition of individual steels. Specific requirements for spring steels supplied to hardenability requirements; chemical compositions of individual steels, hardenability data. Appendix gives standard sizes and tolerances.

BS971 : 1950 Commentary on BS wrought steels (En series)

Properties in relation to design; ruling section; equivalent section; 'mass effect'; weldability. En steels classification (chemical compositions, mechanical properties).

BS2853 : 1957 The design and testing of steel overhead runway beams

Specifies the minimum requirements for the design and testing of fixed overhead runway beams made from rolled sections of steel complying with BS15. It applies to the runway beams and their components only; it does not apply to supporting structures, to travelling trolleys, and lifting appliances operating on the beams, or to crane gantries or rails.

BS3100 : 1967 Steel castings for general engineering purposes

Comprising:

- 592 Carbon steel castings for general purposes
- 1398 Low alloy steel castings for use at elevated temperatures
- 1456 1½ % manganese steel castings
- 1457 Austenitic manganese steel castings
- 1458 Alloy steel castings with higher tensile strengths
- 1461 3% chromium-molybdenum steel castings
- 1462 5% chromium-molybdenum steel castings
- 1463 9% chromium-molybdenum steel castings
- 1617 Carbon steel castings with high magnetic permeability
- 1630 13% Chromium steel castings for resistance to corrosion
- 1631 Austenitic chromium-nickel steel castings for resistance to corrosion
- 1632 Austenitic chromium-nickel-molybdenum steel castings for resistance to corrosion
- 1648 Alloy steel castings for use at high temperatures
- 1760 Carbon steel castings for resistance to wear and suitable for surface hardening
- 1956 1% chromium steel castings for resistance to corrosion
- 4238 Close composition high alloy steels for use at high temperatures
- 4239 Carbon steel castings for case hardening
- 4240 3% nickel-steel castings for case hardening
- 4241 3½% nickel-chromium-molybdenum steel castings for case hardening
- 4242 Ferritic steel castings for use at low temperatures

Process of manufacture, fettling and dressing, heat treatment, sampling and testing, non-destructive testing, rectification and inspection. Chemical composition and mechanical properties of twenty specifications covering fifty-nine grades. Appendices include welding procedures and a tabular summary of compositions and mechanical properties.

BS29 : 1963 Carbon steel forgings for shipbuilding and general engineering purposes

Steelmaking process, chemical composition, minimum forging reduction, heat treatment and mechanical properties, for carbon steel forgings having tensile strengths 22-45 tonf/in² and ruling sections above 6 in.

BS4670 : 1971 Alloy steel forgings

Specifies requirements for alloy steel forgings for shipbuilding and general engineering purposes having tensile strengths between the limits of 600 N/mm² and 1250 N/mm², and ruling sections above 150mm and up to 1000 mm. Includes details of steelmaking, composition, forging and mechanical tests.

3) Codes of Practice

BS153 : Steel girder bridges

Parts 1 & 2 : 1972 1. Materials and workmanship 2. Weighing, shipping and erection

Materials, standards of workmanship, protection against atmospheric corrosion. Notes on rivets, bolts and nuts for erection; measurement for payment by weight, packing marking. Allocation of responsibility during erection; special requirements.

Part 3A : 1972 Loads

States loads to be considered in the design of steel girder bridges.

Part 3B & 4 : 1972 B. Stresses 4. Design and construction

Relaxations in basic stresses, depending on load combinations and methods of stress calculations. Effects of fatigue. Permissible stresses in welds. Formulae and tables for various combined stresses. Allowable loads on roller and spherical bearings. Design of construction details simulating effective length of struts, compression flanges and chords under various conditions. Braced structures; rivetting; bolting; welding.

BS1500 : Fusion welded pressure vessels for general purposes

Part 1 : 1958 Carbon and low alloy steels

Deals with classification of pressure vessels; materials and design stresses; design of cylinder shells, end plates, conical ends, flat covers, tube plates, branches and components, stayed surfaces, manholes and inspection openings, bolted flange connections -full face joint flanges and narrow face joint flanges; jacketed connections and internal structures; manufacture and workmanship; heat treatment; inspection and mechanical tests; welding and radiographic examination; pressure tests and protective devices.

Part 3 : 1965 Aluminium

Two classes of pressure vessels made from aluminium and aluminium alloys; materials and design stresses; design of vessels and components; manufacture and workmanship; inspection and mechanical tests; radiographic examination of welds; pressure tests, marking, and protective devices. Design methods for flanged connections appended.

1500A : 1960 Carbon and low alloy steels

This publication consists of those sections of BS1500 :

Part 1, dealing with the manufacture workmanship, inspection and testing of fusion welded pressure vessels and has been produced in a handy and convenient form for the use of personnel who are primarily concerned with the manufacturing aspects of pressure vessels.

The contents of this publication include the whole of Sections 4 & 5, and Appendix A, of BS1500 : Part 1 : 1958, together with the relevant illustrations and other appropriate information.

BS1515 : Fusion welded pressure vessels for use in the chemical, petroleum and allied industries

Part 1 : 1965 Carbon and ferritic alloy steels

Pressure vessels subject to internal pressure. Materials, criteria for establishing design stresses. Calculation of scantlings of cylindrical and spherical shells, domed, conical and flat ends; compensation of openings; manhole and inspection openings; bolted flanges; supports and internal structures. Manufacturing procedure, welder performance tests, post-welding heat treatment. Non-destructive tests (with unacceptable weld imperfections as revealed by radiographs) and pressure tests. Appendices cover: calculation of stresses arising from local loads and thermal gradients; avoidance of fatigue cracking; vessels required to operate at low temperatures; pressure relief devices; welded connections; approximate metric equivalents; design stress values for materials for which elevated temperature properties are not specified.

Part 2 : 1968 Austenitic stainless steel

Pressure vessels subject to internal and external pressure. Materials, criteria for establishing design stresses. Calculation of scantlings of cylindrical and spherical shells; domed, conical, flat, spherically domed and bolted ends; flat tube plates; openings and branch connections; bolted flanges; manholes and inspection openings; jacketed construction; supports and internal structures. Manufacturing procedure weld procedure and welder performance tests, tolerances, surface finish. Non-destructive tests (with unacceptable weld imperfections as revealed by radiography) and pressure tests. Appendices cover calculation of stresses arising from local loads and thermal gradients; avoidance of fatigue cracking; descaling; pressure relief protective devices; welded connections and welded seams; design of openings and penetrations; design stresses; selection of economic flange proportions; stages of inspection.

BS487 : Fusion-welded steel air receivers

Part 1 : 1960 For pressures not exceeding 500 lb/sq.in.

Class II and III air receivers for design pressure not exceeding 500 lb/sq.in. and the product of the design pressure in lb/sq.in. and internal diameter in inches not exceeding 21,000; does not apply to compressed gas cylinders. Materials and formulae for the calculation of shell and end plate thicknesses, types of welded joints, particulars of welding procedure and weld tests, dishing and flanging of end plates; inlet and outlet connections, access, compensation for openings, hydraulic test. Includes 40 illustrations.

BS487 : Part 2 : 1963 Receivers to Class 1 requirements

Class I air receivers with no limitations on size or design pressure. Types of welded joints with particulars of welding procedure, materials of construction, formulae for calculating scantlings; inlet and outlet connections, access holes, compensation for openings. Inspection procedure, hydraulic test, tests of welded seams and welder competence, marking and certification.

BS1019 : 1963 Photographic lenses. Definitions, methods and accuracy of marking

Defines the parts of a photographic lens, focal length and associated quantities and the aperture and related quantities. Marking requirements, accuracy of markings and sequence of markings are specified. Appendices cover the methods of measurement of focal length, effective aperture and aperture ratio.

BS1113 : 1969 Water-tube steam generating plant (including superheaters and steel tube economizers)

Applies to parts of water-tube steam generating units subject to internal pressure, and similar parts of steam and water heating units used in association with steam generating units. Revised in metric units and also meets requirements of ISO/R 831. Materials and design stresses, design, manufacture, workmanship, inspection and testing.

BS3915 : 1976 Carbon and low alloy steel pressure vessels for primary circuits of nuclear reactors

BS5500 : 1976 Unfired fusion welded pressure vessels

Covers carbon, ferritic alloy and austenitic steels and incorporates 150 proposals.

BS5135 : 1974 Metal arc welding of carbon and carbon-manganese steels

Major standard incorporating new thinking on preheat requirements.

BS4870 : 1974 Approval testing of welding procedures

BS4871 : Part I 1974 Approval testing of welders working to approved welding procedures

BS4872 : Part I 1972 Approval testing of welders when welding procedure approval is not required

Weld Thermal Cycle

The weld thermal cycle will, after the carbon and alloy content of the steel, be the most significant factor in influencing weld and heat affected zone structure and properties. The cycle will depend upon the welding process, the heat input and the workpiece dimensions. Significant aspects of the cycle are:

- (1) maximum temperature reached;
- (2) time at elevated temperatures i.e. at or above A₁ and A₃ temperatures;
- (3) cooling rate.

The relative importance of these factors will depend largely upon the composition and structure. In mild (plain C below 0.25%C) steels, the maximum temperature is the most important factor since it determines the austenite grain size. Time at elevated temperatures may be significant in determining the extent of dissolution of carbides and re-distribution of non-metallic inclusions.

Cooling rate will influence the austenite transformation products and is thus most important with steel compositions or dimensions where transformations to non-equilibrium structures such as bainite or martensite is critically dependent upon the actual cooling rate. Examples include low alloy steels, thicker sections of carbon-manganese steels; and high strength low alloy (micro alloyed) steels.

Structure of Weld Metal

Most deposited weld metal contains only about 0.1%C and no alloying elements other than Mn and perhaps Si. Impurity elements such as S and P are maintained at very low levels.

The structure of the weld metal consists in weld metal with strengths up to 600 N/mm^2 , of proeutectoid ferrite outlining the original columnar grain boundaries with the centre of the "grains" consisting of a more acicular ferrite together with small amounts of distributed carbides.

This ferrite is characterized by a fine grain size (the actual grain size depends upon the heat input/arc energy and cooling rate) and a high dislocation density. Small globular inclusions are often visible in weld metal

deposited by slag shielded welding processes.

Subsequent weld runs give "refined" weld metal due to reheating the original "as deposited" weld metal above the A_{c3} temperature, i.e., re-austenitization. This refined weld metal has a similar grain size to as deposited or unrefined weld metal but is more equiaxed. Refined and unrefined weld metal can be readily detected by macro-etching.

The mechanical properties associated with these weld metal structures are good. Tensile strength is higher than that of hot wrought metal of similar composition with unrefined weld metal being somewhat stronger than refined weld metal. No yield point may be detectable but proof strength is also high and the ratio Proof Stress : Tensile Stress is above 0.75. Ductility remains adequate with tensile elongations generally above 25%, although flux/slag constitution is important in manual metal arc weld deposits. Notch impact behaviour is generally better in refined weld metal but interstitial impurities and alloying elements are more important: basic fluxes are highly beneficial in MMAW and SAW and inert gas shielded processes give even better results. For the welding of steels with proof stress above 550N/mm^2 it becomes more difficult to obtain weld metal of overmatching strength, ductility and notch toughness. It is necessary to use alloying elements, minimise impurity elements and carefully control the welding procedure; further developments in this field are required, however.

Structure in the Weld Heat Affected Zone

The HAZ is defined as the region extending from the fusion boundary to the unaffected parent material. It is usually defined in terms of changes in microstructure although there is evidence that loss of notch toughness can occur in regions outside the visible heat affected zone.

In mild steels four distinct regions can be observed in the HAZ:

- (1) Region of grain coarsening - a coarse Widmanstätten structure heated between 1100°C and the melting point;
- (2) Region of grain refinement - a fine grained intermediate transformation product resulting from heating to just above the A_3 temperature ($900^{\circ} - 1100^{\circ}\text{C}$);
- (3) Region of partial transformation resulting from heating in the temperature range $A_1 - A_3$ ($750^{\circ} - 900^{\circ}\text{C}$);

- (4) Region of spheroidization, a narrow region heated just below A₁ (750°C) where some partial spheroidization of carbide has occurred (very difficult to detect).

Changes in properties accompany the changes in structure, these changes are generally an increase in strength at the expense of ductility. Notch ductility may be improved in the region of grain refinement but may be drastically reduced in the grain coarsened region.

A redistribution of non-metallic (generally by liquidation) may also occur in the grain coarsened region resulting in hot tears or embrittlement.

Regions of grain coarsening, grain refinement and partial transformation may be noted also in the HAZ of low alloy and microalloyed steels. However the initial parent metal structure may not have been the equilibrium structure but a ferrite plus tempered bainite or tempered martensite structure. Equally the steel composition will be more hardenable so that martensite may occur, particularly in the coarse grained region. Twinned martensite is the most dangerous structure being hard and brittle with a high susceptibility to hydrogen induced cold cracking. The hardness and susceptibility of the martensite will depend on the carbon content and in some cases an immediate post weld tempering treatment will be essential.

The Carbon Equivalent Formula

Accurate prediction of the behaviour of the HAZ of C-Mn and low alloy steels requires knowledge of carbon and alloy contents, transformation characteristics under continuous cooling conditions, and the cooling rate. This information is often not available and even where it is known the welding engineer may not have sufficient expertise to use it.

To ease the welding engineer's task many workers have expressed the hardenability of a steel in terms of the effect of each alloying element relative to carbon, which has the most marked effect. The result has been the postulation of "carbon equivalent formulae" which indicate the hardenability of a weld HAZ and, combined with the welding procedure, indicates the probability of hydrogen induced cold cracking. Carbon equivalent formulae are generally formulated after work on a restricted range of steels and only apply to that composition range.

The most used formula (IIW formula) applicable to C-Mn and slightly alloyed steels (approximately 0.13 - 0.25% C, 0-2% alloying elements) is:

$$\text{Carbon Equivalent} = C + \frac{\text{Mn}}{6} + \frac{\text{Cr} + \text{Mo} + \text{V}}{5} + \frac{\text{Ni} + \text{Cu}}{15}$$

The maximum C.E. level to avoid cracking depends upon steel thickness restraint and welding parameters but for thicknesses up to 50mm and ambient temperatures above 5°C the C.E. should not exceed 0.45 HAZ hardness can be approximately related to C.E. by the expression

$$H_v = (1200 \times CE) - 200$$

and to avoid cracking HAZ hardness should not exceed 350 Hv.

It cannot be overemphasized that all carbon equivalent formulae are restricted to specific, often quite narrow, composition ranges. Additionally, while hardenability depends upon all alloying elements hardness and crack susceptibility in the HAZ depends strongly upon carbon content.

HOT TEARING AND BURNING

If a wrought steel is overheated in the austenitic region a re-distribution of non-metallics can occur, aligning them at the grain boundaries. This re-distribution is generally attributed to local melting of the non-metallic inclusions, sometimes with eutectic dissolution of metal. On cooling the liquid non-metallics solidify as grain boundary films which embrittle the steel. This phenomenon is known as burning and cannot be rectified by subsequent heat treatment. If stress is present, the grain boundaries may part during cooling - hot tearing. Re-distribution of non-metallics to give hot tearing and burning can occur during the welding of steels in both the weld metal and heat affected zones; the problem is most prevalent with those welding processes which involve a slow thermal cycle, such as oxy-acetylene welding and electroslag welding. It is fortunately rare in most welding situations.

The weld thermal cycle can give grain boundary films in the HAZ if suitable non-metallics are available; if sufficient thermal stress is present, hot tears will occur instead of plastic deformation within the grains. If hot tearing does not occur the HAZ may still be brittle and incapable of plastic deformation.

HAZ hot tears may be recognised by:

- (1) They are intergranular with respect to the prior austenite grains;
- (2) Their length varies from very short to a few grain diameters;
- (3) They are usually accompanied by re-distribution of non-metallic inclusions and formation of grain boundary arrays similar to burnt steel;
- (4) Hot tears may be observed in structures where there is no chance of HAZ cold cracking.

The significance of hot tears are:

- (1) In thin sections with prolonged heating they may reduce static strength;

- (2) They may initiate fatigue cracks or a brittle fracture;
- (3) Cold cracks may propagate from hot tears after transformation to an acicular microstructure;
- (4) Reheat cracks may propagate from hot tears in ferritic heat resisting steels;
- (5) Burning or tearing may result in reduced ductility or notch toughness.

Hot tearing in the HAZ is aggravated by high sulphur and high carbon contents, so that problems are more likely in medium carbon steels; segregation in mild steels can, however, lead to hot tearing. Experimental evidence points to manganese and iron sulphides being the main source of boundary films rather than silicates or phosphides.

Similar problems can be found in ferritic weld metal: high sulphur (around 0.1%) is needed for hot tearing and although this level is unusual it can occur due to segregation. Manganese silicate can also liquate at austenite grain boundaries but this form of hot tear generally are often referred to as refined zone fissures.

The significance of planar and bulk non-metallic inclusions in weld metal are:

- (1) Grain boundary films can give liquation cracking;
- (2) Non-metallics may serve as nuclei for fissures after transformation to ferrite;
- (3) Non-metallics may cause microcracks on heating by subsequent runs (refined zone fissures);
- (4) Large angular inclusions may cause internal necking and reduced ductility;
- (5) Thin films provide easy fracture paths for fatigue and brittle failure.

The non-metallics available are very important in promoting weld metal hot tearing so that deoxidation practice, filler composition and the type of welding slag (if any) are significant. Rutile slags give good protection from deoxidation so that only a little ferro-manganese is used. These slags cannot cope with high sulphur or silicate contents and the weld

deposit can exhibit porosity, thick cracks filled with iron oxide, thin cracks filled with iron sulphide and refined zone fissures. Oxide-silicate coatings give a highly oxidized weld metal and although weld properties may not be high, the large amount of silicate and oxide inclusions effectively scavenge sulphur and prevent hot tearing. Basic coatings give reasonable but not complete protection from oxidation and also contain ferro-manganese and ferro-silicon. Basic slags were found to scavenge sulphur into complex spheroidal inclusions and to prevent hot tearing at all but the highest levels of sulphur. However, if aluminium, which is sometimes used as a deoxidizer in basic slags, was present in the inclusions, sulphur was not absorbed and hot tearing occurred.

MIG-CO₂ welding using wires deoxidized with manganese and silicon gave no hot tearing and only a little porosity even with high sulphur contents, the sulphur being effectively scavenged by the manganese silicates produced during welding. However the use of wires deoxidized with manganese, silicon and aluminium, which generally give superior welding qualities, again resulted in hot tearing as the sulphur was not effectively removed from the weld metal.

It must be emphasized that hot tearing is unusual in the welding of structural steels with carbon contents below 0.25% and sulphur below 0.05%; problems generally arise through segregation. Equally, any coated electrode can generally be used without problems. If hot tearing or embrittlement problems do arise, however, then attention has to be paid to the steel composition and welding consumables.

LAMELLAR TEARING IN STEEL PLATE

Cracking in steel plate, lying in the plane of the rolling direction was first reported in 1961 and has been observed frequently since that date. It occurs in any joint in which the welding has applied tensile stress in the short transverse direction, i.e. fillet, T-butt and edge joints, and where the ductility in the short transverse direction has been exceeded. The characteristic fracture is step-like and does not necessarily appear at a surface. Originally termed "pull-out cracking" this type of fracture is now designated "lamellar tearing".

Lamellar tearing was first observed in silicon-killed steels of 25 - 50mm thickness containing multi-run fillet or T-butt welds but it has subsequently been found in a wide range of steels from 2.0mm upwards. In fact, there is some indication that thin materials are more susceptible to lamellar tearing if they are restricted from bending and strained in the thickness direction.

Metallographic Features of Lamellar Tearing

Originally, lamellar tears were illustrated which appeared to propagate from the root or toe of a weld, possibly from an initial defect. Although there was some initial confusion with underbead cracking, lamellar tears often traversed the whole of the heat affected zone and could even extend beyond the visible heat affected zone. In other cases the tear terminated at the fusion boundary. Subsequently lamellar tears were observed away from the toe or the root of a weld, and occasionally well away from the heat affected zone. In places, twin tears can occur, or minor discontinuities are visible away from the main tear.

The characteristic stepped formation of lamellar tears may be divided into flat terraces linked by shear walls. Considerable evidence exists to emphasize the importance of inclusions in lamellar tearing. The terraces contain many inclusions identified as silicates, sulphides or mixtures of the two. Silicates tend to form irregular plate-like inclusions at the lower temperatures of rolling, and are often cracked, whereas sulphides are more spheroidal. Silicates may also form sharp extensions from sulphides. Thus silicates appear to be the more dangerous but in the absence of silicates the sulphides can elongate and contribute to tearing. Alumina inclusions are also

significant. The appearance of the terrace surfaces, as revealed by the scanning electron microscope, is largely of ductile fracture, although there is some evidence of small areas of brittle fracture. The shear walls are often undercut and again the appearance is largely of ductile fracture.

High temperature metallography has shown that inclusions tend to separate from the matrix on heating. This separation has been termed "decoherence" and may be due to differential thermal expansion or incipient melting; void formation may also occur. These phenomena partially explain the contribution of the heat affected zone to lamellar tearing.

Parent Steel

Lamellar tearing is associated with low ductility in the short transverse direction. The occurrence of this low ductility has been known to steelmakers for many years but it is only recently that its significance has been realized. Short transverse or through thickness tensile tests often show low reduction of area values of 0-20%, well below those obtained with conventional tensile tests on steel plate. These low tensile ductilities are not solely due to the presence of small non-metallic inclusions.

Steelmaking practice and subsequent treatment is very important in determining susceptibility to lamellar tearing. The significance of treatment prior to welding depends on the type, incidence, distribution and morphology of non-metallic inclusions. In general silicates are worse than Type II sulphides which are worse than Type I sulphides and aluminates because they become brittle at higher temperatures and are therefore likely to crack or shatter during hot rolling. Equally, flat plate-like inclusions, especially if cracked, are worse than linear or spheroidal inclusions. It is possible to list steels in order of approximate decreasing susceptibility to lamellar tearing: fully killed steels deoxidised by silicon or silicon plus aluminium containing silicates and Type II sulphides, fully killed steels containing only Type II sulphide stringers, semi-killed steels containing duplex sulphides with silicate tails, fully killed thick plate with ellipsoidal sulphides, REM treated steels with spherical sulphides, special steels with sulphur content reduced below 0.01%. Rolling practice can also influence inclusion distribution: the lower the finish rolling temperature the greater the danger of cracking the inclusion. Initial rolling of an ingot, to some 75mm thickness, will tend to maintain inclusions in a spheroidal form, so that material above this thickness should be reasonably safe although lamellar tearing has been reported in thicknesses of 200mm.

Thin steel, below 15mm thickness, was considered to be non-susceptible because the material would bend if load was applied in the thickness direction, although as mentioned earlier, this is not always the case. It may be noted that the requirements of high finish rolling temperature and low plastic deformation are contrary to normal good practice.

Any tendency to strain ageing may make a secondary contribution to lamellar tearing in or near to the visible heat affected zone. Some susceptible steels have been shown to exhibit a low ductility trough in the 200^o - 300^oC region whereas a non-susceptible steel plate maintained high ductility at all temperatures. Dangerous temperatures can be incurred and maintained for several minutes during welding. Prior cold work may also contribute to low short transverse ductility. The significance of a ductility dip at temperatures around 1000^oC has not been investigated in detail but there is no evidence that this phenomena is important. Similarly no evidence has been presented to indicate that creep relaxation properties are significant for lamellar tearing.

Hydrogen does not appear to contribute to decoherence at the inclusion/steel interface but its presence does make it easier for cracking to occur between decohered inclusions. Thus good low hydrogen practice will not eliminate lamellar tearing but it can increase the minimum strain level at which it will occur.

Mechanism of Lamellar Tearing

The original view of the propagation of a lamellar tear from some initiating point is now recognised as incorrect. As the steel is strained in the thickness direction, decoherence or, possibly, void formation will occur at steel/inclusion boundaries. Decoherence in regions where clusters of inclusions are present may result in multiple micro-cracks at several levels. In some cases decoherence may occur during the final stages of hot rolling as shown by cracked silicates. Sound metal at the tips of inclusions in the same plane will then suffer heavy plastic straining and eventually fracture in a ductile (or occasionally brittle) manner giving the characteristic terraces. Around a cluster of inclusions on the same plane the sound material will possess good strength and ductility and will resist propagation of a tear. At this stage the weakest link between adjacent terraces will fracture giving shear walls, which are often undercut.

Some sort of tear propagation may be achieved by the tensile strain applied by successive weld runs, which would explain the progressive nature of

Tests for Susceptibility to Lamellar Tearing

The determination of the susceptibility of a particular plate to lamellar tearing is obviously desirable, but considerable disagreement exists with regard to a satisfactory test. Ideally, a non-destructive test is required but even a generally accepted cheap and simple small-scale destructive test would be an improvement over the existing situation.

The only two non-destructive tests that have been suggested involve the use of ultrasonics. The use of conventional longitudinal waves to locate individual inclusions is of very little value, however, since the minimum detectable size is much greater than the smallest inclusions than can contribute to lamellar tearing. Attenuation measurements offer greater hope of success although difficulties exist in separating inclusion attenuation from that caused by grain boundaries. Meyer claims considerable success from ultrasonic attenuation measurements but U.K. workers have been less successful. The discrepancy may be due to the choice of ultrasonic wave frequency, most U.K. work has been below 10 Mc/s whereas 15 - 20 Mc/s has been reported as desirable. It should be noted that ultrasonics is highly successful in detecting lamellar tearing once it has occurred.

Several destructive tests have been devised and used although, they have the general disadvantages of destructive tests for quality control purposes. These tests include the measurement of mechanical properties and actual weld tests.

Metallographic inclusion counts are claimed to be extremely useful but the method is time consuming. Through thickness tensile tests are the most useful method for steel thickness above 25mm but the minimum acceptable reduction of area needs further definition. Current practice requires a 25% RA_z mean for 4-6 specimens of 6.4mm diameter with a 10% or 15% minimum but IIW has suggested a lower mean of 15-20%. Node plates for offshore platforms generally require 30% or even 35% mean RA_z . For larger diameter specimens the mean RA_z can be reduced, for example to 15% for 18mm diameter specimens. With plates of 25mm thickness or greater the complete tensile specimen can be machined from the plate but for thinner plate or where the behaviour of the sub-surface layers is important, extension pieces can be attached by friction, stud or arc welding. In the latter process low hydrogen procedures and 150°C preheat should be used.

A number of tearing tests such as notched bend tests and COD tests have been suggested but remain to be proved. A through thickness bend test on sample slices from suspected plates has given encouraging results but the

test needs to be more closely defined. It is suggested that steels that give crack lengths over 30mm for surface strains below 6% are susceptible and if cracking shows at strains below 2% lamellar tearing is highly probable. Several welding tests are useful for examining crack morphology but are not sufficiently selective. The Cranfield test is very severe and will cause tearing in almost any steel. However, the method of causing tearing by rotation of the fillet stem is not characteristic whilst the use of crack length or number of weld runs to cause detectable lamellar tearing is insensitive. A window frame test is nearer the practical situation but remains insensitive. More recently the U.K. Welding Institute have introduced a "box" welding test which uses edge welds and Lehigh have suggested a "restraint" test which may prove more selective.

Avoidance of Lamellar Tearing

Four methods are in present use to avoid lamellar tearing:

- (1) The use of steel with guaranteed through thickness properties;
- (2) Metallographic examination for inclusions with rejection of unsatisfactory material;
- (3) Use of buttering techniques below the stem of the fillet or T - butt joint;
- (4) Re-design of the joint detail to avoid stressing in the short transverse direction.

The ideal method to avoid lamellar tearing is to use clean vacuum degassed steel with low sulphur content and a guaranteed RA_z value. The cost of such steel is double that of silicon-killed steel but it does have genuinely isotropic properties which ensures more effective utilisation: Z quality steel should always be used when joint design and high restraint makes lamellar tearing probable.

The second method has worked quite well but it is costly and cannot completely avoid lamellar tearing. The third method which aims at removing inclusions and replacing susceptible parent steel by clean weld metal has worked very well and is a good alternative to Z quality steel in all but the worst situations. The fourth method which aims at reducing the direct tensile loading is the most desirable but it does demand that the problem is appreciated at the design stage. Some of the current suggestions such as the use of special forgings are likely to prove more expensive than Z quality plate and may replace lamellar tearing by other problems.

Practical Significance of Lamellar Tearing

It has been difficult to point to service failures that have occurred due to lamellar tears. Since lamellar tears have been generally found in important structures no one has been prepared to risk leaving them and considerable effort has been spent in eradicating the defect. However at least one high pressure boiler drum is known to contain a lamellar tear at a nozzle weld and the tear is being monitored to determine any propagation. More recently, examples of lamellar tears in jacketed reaction vessels have been shown and these have led to failure over a period of years, probably due to thermal fatigue. However, lamellar tearing was not a sudden occurrence in 1961, although the use of thicker plate was increasing dramatically at that time, so that many structures must be giving satisfactory service despite the presence of lamellar tears. Fracture mechanics is attempting to define acceptable limits but until codes and standards are modified it appears to apply the same restrictions of lamellar tearing as are applied to lack of fusion, incomplete penetration and undercut.

THE ROLE OF HYDROGEN

Hydrogen is a major factor in several of the problems that can arise in the welding of steel. These include:

- (1) Fish-eyes;
- (2) HAZ hydrogen induced cold cracking (HICC)
- (3) Fissures;
- (4) Chevron cracks;
- (5) Weld metal HICC.

Hydrogen can be available from several sources. In manual metal arc welding several coatings break down to give a shielding atmosphere that contains hydrogen (often up to 30%), whilst in inert gas shielded processes moisture in the shielding gas may break down to give hydrogen. Moisture on the parent steel and filler wires, often absorbed in rust, or hydrogen present in surface oil grease is also a potential source of hydrogen in the welded joint.

Hydrogen goes into interstitial solid solution in steel, its solubility is greatest in molten steel, and higher in austenitic structures than ferritic structures. On weld metal and heat affected zone cooling the tendency is for hydrogen to be present as a supersaturated solid solution unless cooling has been delayed in order to allow the hydrogen to diffuse to atmosphere.

Much of the hydrogen in solid solution in a ferritic steel structure will diffuse away over a few months storage or by heating at 200° - 300° for 4 - 16 hours. This essentially mobile hydrogen has been termed "diffusible hydrogen". However, not all the hydrogen in solid solution can be removed so easily and a proportion remains in solution - "fixed hydrogen". As the temperature of the welded ferritic steel is raised, more and more of the fixed hydrogen is evolved, but the last remnants can only be removed by vacuum fusion. It must be emphasized that there is no intrinsic difference between fixed and diffusible hydrogen although much of the welding literature does differentiate between the two.

Fish-eyes

These are bright spots of cleavage fracture surrounding a dark spot, occurring on predominantly ductile fracture surfaces that have been

produced by slow straining, for example in a tensile test. Slight heating (to about 100°C) or long storage before straining avoids fish-eye formation. It is notable that high speed fractures such as those obtained in impact tests, do not exhibit fish-eyes, indicating the existence of a diffusion type mechanism. Fish-eyes can often be observed on weld metal fracture surfaces where the weld metal has been deposited from arcs with considerable hydrogen in the arc atmosphere, for example from rutile and cellulosic electrodes.

The pupil of the fish-eye may be an inclusion, defect or fissure and the iris consists of cleavage facets that have undergone some plastic deformation. Since plastic deformation is required for fish-eye formation they are of little practical significance in welded structures except for plastically designed structures.

The first explanation of fish-eyes was that interstitial hydrogen diffused to voids or non-metallic inclusions and combined to form molecular hydrogen. High pressures developed which caused local fracture, triaxiality caused by surrounding steel resulted in the fractures being brittle. This hypothesis is less popular now although the diffusion of interstitial hydrogen towards voids is still considered necessary. The diffusion gives a hydrogen concentration and diffusion gradient around small voids which lowers the cleavage stress locally below the yield stress. Plastic deformation then results in local brittle failure. The hydrogen diffusion gradient only becomes significant when the defects open up so that rapid straining does not allow time for the cleavage stress to be lowered and fish-eyes to form.

Fish-eyes have been found, but only rarely, at plate-like inclusions in weld metal when no hydrogen is present. The cause may be other interstitial elements that lower the cleavage stress or perhaps merely triaxiality at a stress raiser.

HAZ Hydrogen Induced Cold Cracking

HAZ cold cracking occurs in the weld HAZ within the temperature range $+160^{\circ}\text{C}$ to -20°C (some authorities extend this range from $+200^{\circ}\text{C}$ to -100°C) by cleavage; it occurs spontaneously after welding, often after a delay period of several hours. Sometimes the cracking is continuous but at other times it propagates in a series of steps. The cracks do not necessarily appear at a free surface and are often very narrow so that detection may be difficult. Where the cracks lie exclusively in the HAZ below the weld bead they are referred to as underbead

cracks but where they propagate from the toe of the weld they are known as toe cracks.

There are three requirements for hydrogen induced cold cracking:

- (1) a susceptible microstructure;
- (2) hydrogen;
- (3) restraint or residual stress

The most susceptible microstructure is twinned martensite i.e. acicular martensite containing over 0.20% carbon and internally twinned, but bainitic structures consisting of ferrite lathes separated by carbide particles or retained austenite is also susceptible to cracking. Auto-tempered martensite is not susceptible but mixed autotempered martensite and bainite structures may be. Structure and not composition is the important factor and susceptibility is related to the ease in initiating and propagating fracture in the various structures. The significance of alloying elements especially carbon is in increasing hardenability, hardness and the short range stresses.

Hydrogen is introduced to the HAZ from the weld metal: since hydrogen is more soluble in austenite than ferrite and the weld metal re-transforms to ferrite before the HAZ (because of the difference in hardenability), hydrogen in the weld metal migrates to the still austenitic HAZ when the weld metal transforms. On transformation of the HAZ the hydrogen is trapped and contributes to HAZ cracking. Where a susceptible microstructure is present, cracking may be triggered by provision of hydrogen from external sources. Thus chemical plant has suffered from HICC years after construction due to a build up of hydrogen in the HAZ.

At one time stress (or restraint) was thought to play no significant part in HICC but it is now known to be inter-related with hydrogen, presumably by providing a source of energy for crack propagation. The exact significance of stress is difficult to evaluate since stresses will always be present due to transformation and welding.

The amount of hydrogen for HICC varies with the steel and with the restraint. Less hydrogen is needed as the carbon content and/or alloy content is increased, and the hydrogen is very low indeed for cracking in high strength medium carbon steels. Impurity levels are also important since non-metallic inclusions can act as sinks for hydrogen reducing its influence. Cracking at very low hydrogen levels (below 10 ml/100 gm of weld metal) have been found in forged vacuum melted steels with very low sulphur contents and the reason has been attributed to the absence of sulphides and

other non-metallic inclusions acting as hydrogen sinks.

As the restraint is increased, less hydrogen is required to cause HICC. At medium carbon contents (leading to high HAZ hardnesses) or high restraint HICC may occur immediately after welding but as the HAZ hardness, hydrogen content or restraint is reduced the delay between welding and cracking is increased, often up to several days. This feature can cause problems since, if ultrasonic inspection is carried out immediately after welding (which is normally considered desirable, inspection may have been completed before the onset of cracking. However, the delay may have advantages if its extent is known, in that post weld heat treatment may eliminate the susceptibility of the joint to HICC. For highly susceptible steels immediate PWHT of the welded joint after welding may be essential to prevent cracking, by tempering of the HAZ to produce a non-susceptible microstructure or by elimination of hydrogen.

The original explanation of HICC was in terms of interstitial hydrogen diffusion to voids, microcracks, etc. where it recombined to form molecular hydrogen at high pressure. An alternative mechanism still requires hydrogen diffusion gradients around voids but the role of the hydrogen lies in lowering the cleavage stress of ferritic structures and possibly in lowering the M_s temperature to give greater stresses. Transformation with its high tensile strains will form micro-voids at the tips of dislocations, defects or inclusions, hydrogen will diffuse to these voids setting up diffusion gradients with consequent regions of low cleavage stress. The residual long and short range stresses then cause cracking. The delayed and occasional intermittent nature of underbead cracking can be explained in terms of the time needed for hydrogen diffusion.

The whole question of the role of hydrogen in causing cracking in steel weldments is still under investigation and generates considerable controversy. Recent researches suggest that hydrogen encourages whatever mode of cracking is most probable in a given steel, so that hydrogen should not be solely associated with cleavage or intergranular brittle fracture.

The avoidance of HICC in steels is of considerable and growing importance particularly with the increased usage of hardenable steels. Although it may be theoretically possible to eliminate HICC by attention to only HAZ microstructure hydrogen content or restraint it is more usual to try to influence beneficially all three factors.

Thus, where a steel is considered to be susceptible, attempts will be made to reduce restraint by careful attention to weld detail and set-up. Joint cleanliness and welding consumables will be

controlled to reduce weld hydrogen levels to a minimum; in many cases welding electrodes or fluxes will be baked and stored in heated ovens while electrodes issued to a welder may be kept in a heated quiver. Arc energy will be controlled.

The use of preheat is often quoted as a means of controlling the HAZ microstructure by slowing the cooling rate and hence raising the austenite transformation temperature. However this change is rarely achieved in practice and could well be undesirable if it were since it would reduce the strength of the HAZ. Preheat is used more realistically to allow time for hydrogen to diffuse from the welded joint to the atmosphere and so reduce hydrogen content; it may also reduce transformation stresses sufficiently to reduce crack susceptibility. Preheat is often specified for welded joints between certain specific limits: the minimum preheat is sufficient to slow the cooling rate sufficiently to allow hydrogen removal and the maximum preheat is such as to avoid undesirable metallurgical transformations that would lower the fracture toughness of the HAZ. Since preheat levels are chosen to influence cooling rates they will obviously be chosen in relation to arc energy, joint design and steel thickness as well as steel composition.

The susceptibility of any given joint to HICC will depend upon hardenability and potential HAZ microstructure as well as joint design and dimensions. The continuous cooling transformation diagram will indicate hardenability whilst the carbon content particularly, together with other alloying elements will indicate the susceptibility of the resultant microstructure. However, even if all the metallurgical data were available, not all welding engineers are metallurgists so that it is necessary to develop a simple method to assess the susceptibility of any steel composition to HICC. This has resulted in the development of carbon equivalent formulae which by expressing composition in terms of the influence of carbon attempts to indicate hardenability in terms of an equivalent percentage of carbon. Within quite specific limits the CE formula works very well and several formulae have been suggested for various ranges of composition. However, it must be remembered that the CE formula indicates hardenability not susceptibility, which is largely governed by carbon content. Restraint, joint design and joint thickness will also influence hardenability and/or susceptibility and must also be taken into account.

The standard IIW formula:

$$CE = C + \frac{Mn}{6} + \frac{Cr + Mo + V}{5} + \frac{Ni + Cu}{15}$$

applies to conventional structural steels with 0.13 - 0.25%C, 0.0 - 1.6%Mn and up to about 2% of other alloying elements. Standards for weldable steels generally limit the CE to a minimum of 0.43 or 0.45 which means that such steels can be welded without precautions at thicknesses up to about 50mm. At lower carbon contents much higher CE values can be tolerated before HICC becomes a problem. Thus two steels with similar CE values only have similar susceptibility to HICC if the carbon contents coincide. This situation has become more important with the increasing introduction of microalloyed steel which may have CE up to 0.6 but in which carbon content may be well below 0.1%. These steels are far less susceptible to HICC in the HAZ but in some cases the problem may be transferred to the weld metal.

In some recent standards such as B.S.5135 the carbon equivalent of the parent steel has been related to the joint shape and size, the intrinsic hydrogen content of the welding process, the preheat, welding heat input and joint restraint in such a manner that the welding parameters can be chosen to avoid HICC. This approach has been very successful and is likely to be more widely adopted.

Fissures

Fissures are microcracks in mild and low alloy steel weld metals which form in the ferritic temperature range (100° - 200°C) by cleavage without external loading. There has been considerable controversy about their formation, morphology and influencing factors.

In single run weld deposits three types of microcracks have been observed.

(1) Transcrystalline fissures caused by cleavage cracking in alpha ferrite by a combination of high hydrogen content and internal stress. The fissures are very narrow and difficult to detect; they sometimes pass through large spherical inclusions.

(2) Inter-columnar fissures formed by the same mechanism as above but associated with non-metallic particles lying between columnar crystals. These fissures are of irregular appearance and sometimes with parallel multiple paths formed by nucleation on both sides of an inclusion-metal interface. At their tips these fissures may become transcrystalline.

(3) Small hot cracks similar to HAZ hot tears caused by liquid or semi-liquid non-metallic films at grain boundaries have been termed fissures although the mechanism of formation is quite different.

In multi-run deposits micro-cracks have been observed in both refined and non-refined regions but are often more intense in later runs,

even with basic electrodes, and may even be confined to refined regions. Fissures in non-refined zones and some of those in refined zones are presumably of the same variety as for single run deposits. Refined zone fissures, however, have largely been shown to result from the melting of manganese silicates reheated above the A3 temperature, that is a special form of hot cracking.

It is this variety of fissures that has led to much of the confusion about true a⁻ range fissures.

A number of facts are known about fissures although there are some discrepancies.

(1) High hydrogen is generally reported as the major cause of fissures as it lowers the cleavage stress.

(2) Nitrogran has also been reported to have a major effect on fissuring by reducing cleavage stress.

(3) Alloying additions, particularly chromium, molybdenum and vanadium influences fissure formation and are generally considered to be detrimental.

(4) High cooling rates increase the tendency to fissure formation due to a greater retention of hydrogen and higher internal stresses.

(5) Increasing the restraint increases the fissure density; hence weld geometry is important.

(6) Rutile weld deposits exhibit the most fissures; this is attributed to high hydrogen and nitrogen contents, and low ductility.

(7) Non-metallic inclusions often lie in the path of fissures, possibly due to decoherence.

Fissures are generally aligned perpendicular to the direction of welding and hence the greatest internal stress, they are generally small but can develop to macroscopic size in conditions of high restraint and cooling rate. Other evidence indicates that more fissures form close to the weld surface, especially at the root of weld ripples where there are marked stress concentrations. There is a little evidence to show that fissures can link to form macrocracks under favourable circumstances.

The significance of fissures is much disputed. Micro-fissures in small quantities have not been considered to detrimental in weld metal but in large numbers or sizes they may reduce ductility or become an initiation for other forms of cracking or fracture. It may be significant that fissures are most prevalent in rutile weld deposits which exhibit

the lowest ductility.

Precautions against extensive fissuring are those needed for the prevention of other defects. Rutile electrodes should not be used under conditions of high restraint or high cooling rate - if these conditions exist then well-baked basic electrodes should be used. Additionally, electrode condition should be good since spalled coatings may lead to nitrogen contamination of the weld metal.

Mechanical Properties

Even if hydrogen does not result in weld defects its presence in welded joints can lead to a loss of ductility and a reduction in failure stress. These effects could lead to low ductility failures on overloading or to brittle failure.

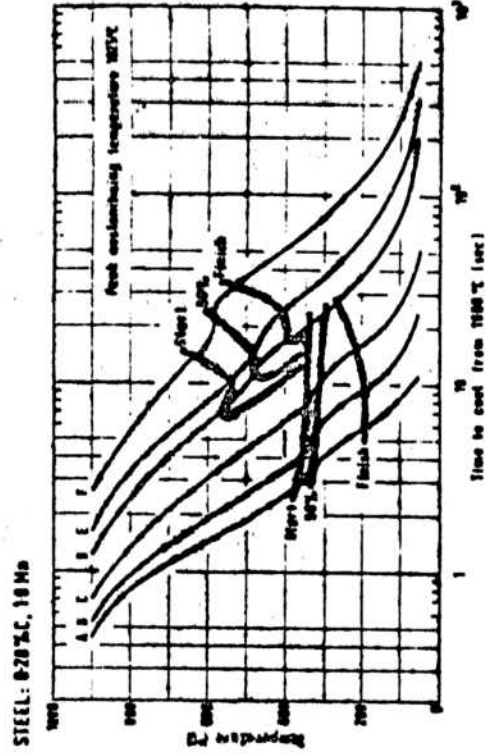


Fig. CCT diagram for 0.20% C, 10% Mn steel

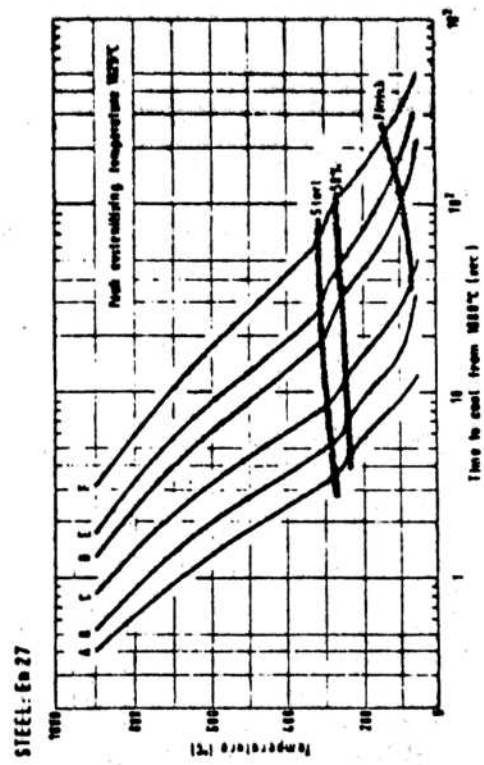


Fig. CCT diagram for Steel En27 (0.3% Ni-Cr-Mn)

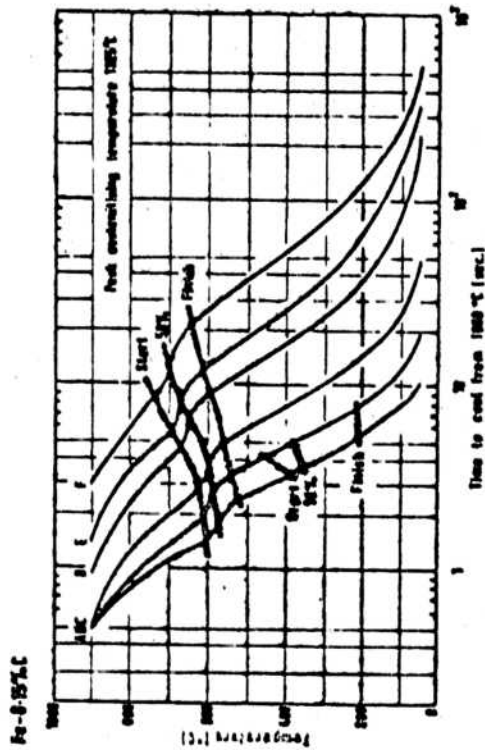


Fig. Continuous cooling transformation (CCT) diagram for a typical mild steel

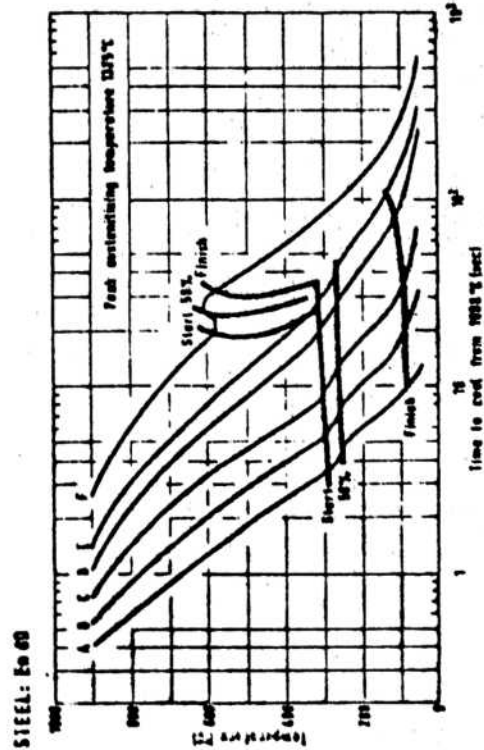


Fig. CCT diagram for Steel En60 (0.45% C)

DEPARTMENT OF MATERIALS

BRITTLE FRACTURE

MAT.110

Steel will generally only fail after considerable plastic deformation but over the last 100 years there have been reports of sudden failures occurring without any prior warning. With mechanically joined steel structures these failures were few and rarely catastrophic but with the advent of welded structures the incidence of failures and the size of cracks increased markedly.

Definition

Brittle failure occurs when the material fractures in a sudden extensive manner with little plastic deformation or absorption of energy, often under relatively low mechanical stress. Failure is generally by cleavage instead of shear.

Recognition

Brittle fracture can be recognised by the suddenness with which it occurs and by the absence of local deformation. Fracture surfaces are characteristic - generally the crystal faces are quite obvious. Brittle fracture often occurs at rather low temperatures although this is not always the case.

The fracture plane is generally approximately perpendicular to the plate surface with less than 2% reduction in plate thickness. There is very little plastic deformation and paint or scale near to the fracture is not cracked or flaked. The fracture surface is bright and granular (crystalline as opposed to fibrous) and often exhibits a chevron pattern with the apices of the chevrons pointing to the fracture source. Many fractures show regions of shear adjacent to the plate surface (shear lips) which accounts for virtually all the reduction in thickness.

Factors Involved in Brittle Fracture

- (1) Liability to brittle fracture is temperature dependent: there is an upper temperature above which a crack cannot propagate without marked plastic deformation, energy absorption and application of considerable stress, whereas below a certain temperature a crack will propagate with little plastic deformation or energy absorption and at stresses below yield. These two temperatures constitute a transition temperature range which can vary widely for steels.

- (2) The presence of a stress concentration, e.g. a notch represents a potential source for crack initiation.
- (3) The geometry of a notch influences crack initiation but not crack propagation. The sharpness of a notch can influence the transition temperature.
- (4) Susceptibility to brittle fracture increases with plate thickness. The effect is partly metallurgical, but greater thicknesses also restrict freedom of plastic flow, while larger defects remain undetected.
- (5) Cracks may be initiated and propagated under a steady stress but only after a high degree of work hardening at the point of initiation. Cracks are more likely to be initiated by an impact load and in this case the work hardening may be less.
- (6) Even in mild steel at low temperatures a certain amount of plastic deformation occurs before the crack develops.
- (7) For propagation to occur the applied stress must be such to ensure the necessary intensity of stress at the crack root and to supply the necessary energy. In general this is supplied internally.
- (8) Brittle fractures are more frequent and extensive in welded than in rivetted structures. Thus welding is a contributory factor:
 - (i) residual stresses may facilitate crack initiation and can provide energy for propagation.
 - (ii) the welded structure provides a continuous fracture path.
 - (iii) local structural changes can facilitate crack initiation and propagation.
 - (iv) weld defects can act as crack initiation sources.
- (9) Metallurgical structure, composition and prior history can affect susceptibility to brittle fracture.
- (10) Irradiation can raise the transition temperature.

Metallurgical Aspects of Brittle Fracture

- (a) Carbon content and Manganese: Carbon ratio.
The transition temperature increases with carbon content although the condition of the carbon is important. Manganese increases pearlite fineness and refines ferrite grain size and is considered to be beneficial.
- (b) Impurities: Interstitial impurity atoms greatly increase the transition temperature.
- (c) A fine ferrite grain size and fine pearlite give the lowest transition temperatures.
- (d) Degree and method of deoxidation: Deoxidised steels have lower transition temperatures than rimming or semi-killed steels.
- (e) Finishing temperatures after rolling: This should be low for the best results.
- (f) Normalising affects the grain size, and hence the transition temperature.
- (g) Cooling rate from hot rolling or normalising temperature: Effects ferrite grain size and pearlite distribution.
- (h) Cold work in manufacture and fabrication. Plastic deformation raises the transition temperature.
- (i) Quench ageing, strain ageing and temper brittleness. All these effects will markedly increase the transition temperature.
- (j) Stress relieving heat treatment: Very beneficial in reducing residual stress but also can remove ageing effects so that it can be metallurgically worthwhile.

Production of Notch Ductile Steels

Steps that can be taken to improve resistance to brittle fracture are :-

- (1) reduction of carbon content
- (2) reduction of phosphorus and nitrogen contents
- (3) increase in manganese content
- (4) use of a killed or semi-killed steel in place of rimming steel.
- (5) ensure a fine ferrite grain size by controlling roll finishing temperature or by normalising.
- (6) reduction of austenite grain size by fine grain practice, e.g. addition of aluminium followed by normalising.

A series of notch ductile steels have been introduced (BS.2762:1956) which give guaranteed impact properties: composition and properties are given in Table II. N.D.I. and II are balanced (or semi-killed) steels, the first "as rolled" and the second normalised. N.D.III is silicon killed and normalised and N.D.IV is aluminium grain refined and normalised.

TABLE NOTCH DUCTILE STEELS (BS.2762:1956)					
Specification	Temp. °C	Energy Absorption (Charpy V-notch) ft.lbs.	Composition, %		
			Carbon (Max)	Silicon	Manganese (Max)
N.D.I	0	20	0.20		1.50
N.D.II	-15	20	0.20		1.50
N.D.III	-30	20	0.17	0.10-0.35	1.50
N.D.IV	-10	45	0.17	0.10-0.35	1.50
	-20	40*			
	-30	35*			
	-40	25*			
	-50	20*			

* According to agreement between manufacturer and supplier.

PREHEAT GUIDE for estimating purposes only and assuming low hydrogen electrodes are used. Joint type, welding process and consumables will affect the choice of temperature for production use.

B.S.1501 Number	Thickness (in.)	Minimum Preheat Temperature (°C)
151 A		
B	< 2	0
C		
161 A	< 2	0
B	2 to 4	50
C		
221	< 1	100
	1 to 2	150
240	< 1	0
	1 to 2	50
	2 to 4	100
503	< 1/2	50
	1/2 to 1	150
	1 to 2	200
620 A	)	
B	) 3	150
625	< 2	250*
ASTM Number	Thickness (in.)	Minimum Preheat Temperature (°C)
A201A	2	0
	2 to 4	50
	4 to 8	100
	8 to 12	150
B	< 2	0
	2 to 4	50
	4 to 8	100

*Immediate stress relief after each weld may be necessary

ASTM Number	Thickness (in.)	Minimum Preheat Temperature (°C)
A203 A	$\frac{1}{2}$	50
	$\frac{1}{2}$ to 1	100
	1 to 2	150
	2 to 4	150
	4 to 6	250
B	$\frac{1}{2}$	50
	$\frac{1}{2}$ to 1	100
	1 to 2	150
	2 to 4	150
	4 to 6	250
A203 C	$\frac{1}{2}$	100
	$\frac{1}{2}$ to 1	150
	1 to 2	200
	2 to 4	200
D	$\frac{1}{2}$	50
	$\frac{1}{2}$ to 1	150
	1 to 2	200
	2 to 4	250
E	$\frac{1}{2}$	50
	$\frac{1}{2}$ to 1	150
	1 to 2	200
	2 to 4	200
A 204 A	1	0
	1 to 2	50
	2 to 4	100
	4 to 6	150
B	1	50
	1 to 2	50
	2 to 4	100
	4 to 6	150
C	1	50
	1 to 2	100
	2 to 4	150

ASTM Number	Thickness (in.)	Minimum Preheat Temperature (°C)
A212 A	≤ 1	0
	1 to 2	0
	2 to 8	100
B	≤ 1	0
	1 to 2	50
	2 to 8	100
A225 A	≤ 1	0
	1 to 2	50
	2 to 4	100
B	≤ 1	0
	1 to 2	50
	2 to 4	100
A285 A	≤ 2	0
	2 to 2	0
B	≤ 2	0
	2 to 2	0
C	≤ 2	0
	2 to 2	0
A299	≤ 1	100
	1 to 2	150
A302 A	≤ 1	0
	1 to 1	100
	1 to 2	150
	2 to 4	150
	1 to 2	150
B	≤ 1	0
	1 to 1	50
	1 to 2	100
	2 to 4	150
	1 to 2	150
A353 A	2	0
D	2	0

4.

ASTM Number	Thickness (in.)	Minimum Preheat Temperature (°C)
A357	≤ 2	250*
A387 A		150
B		150
C		150
D	≤ 3	200
E		300*

*Immediate stress relief after each weld may be necessary.

THE WELDING OF STAINLESS STEELS

When over 11% of Chromium is added to iron it imparts a considerable improvement in corrosion and oxidation resistance - hence the term stainless steels. The corrosion resistance is a result of the development of an impervious surface film of chromium oxide which is self-healing if scratched or damaged. Further improvement in the oxide film can be obtained by a passivating treatment in an oxidising acid, for example nitric acid or a nitric acid-hydrofluoric acid mixture. Increase in the chromium content will improve the oxidation and corrosion resistance. Other alloy additions may be made to increase corrosion (e.g. molybdenum) or oxidation resistance or to give other desirable mechanical properties (e.g. nickel).

The majority of these steels can be readily welded although the oxide film, metallurgical structure and properties may all offer some difficulties, the magnitude of the problem varies with the specific steels.

Types of Stainless Steel

Division of these steels is generally according to metallurgical structure - austenitic, ferritic, martensitic, etc. The most useful steels are those with an austenitic structure containing about 18%Cr and 8%Ni, which accounts for some 90% of all usage of these steels.

<u>Steel Type</u>	<u>Composition, %</u>				<u>Weldability</u>	<u>Examples</u>
	<u>C</u>	<u>Cr</u>	<u>Ni</u>	<u>Mo</u>		
Ferritic	0.05/0.20	13/25	-	-	Quite good-poor	Types 442, 443, 446.
Martensitic	0.15/1.0	13/18	-	-	Poor	Types 403, 405, 410, 414, 416, 420.
Austenitic	0.02/0.10	18/25	8/20	0/5	Good-very good	Types 302, 304, 321, 347, 316, 317, 310.
Ferritic/ Austenitic	0.08	approx.26	approx.5	0/1.5	Quite good	
Martensitic/ Austenitic	0.05	approx.15	approx.5	1.0/2.0	Good	AM350, AM355, 17/7 PH, PH 15/7 Mo, FV520.

The addition of Ni gives a ductile notch tough, corrosion and oxidation resistant austenitic steel with good high and low temperature properties. Austenitic steels have a wide industrial usage based on the above characteristics and some domestic use because of corrosion resistance and a decorative appearance.

The martensitic steels are those which transform to martensite on cooling; they can be brittle and those with higher carbon contents (above 0.2%) are crack prone. Ni additions up to 5% are sometimes added to reduce the crack susceptibility. Except in sheet form, welding can be difficult and special precautions are needed. In sheet form these steels may be used for decoration (e.g. car trim) and in heavier sizes for cutlery, tools, chemical structures and hydro-electric turbines.

At high Cr contents the austenitic phase may be suppressed giving a ferritic structure at all temperatures. High Cr content gives good corrosion and oxidation resistance which may be better than austenitic stainless steels in certain environments e.g. in the presence of Sulphur. With the carbon levels used in the Western world the main problems with these steels are brittleness and low notch toughness below 100/200°C together with heavy carbide precipitation and grain growth in weld heat affected zones. Weldability tends to be from poor to fair depending on carbon content and material thickness, and pre-and post- weld heat treatment may be necessary. Uses are mainly restricted to the oil and petrochemical industries where the specific properties are required; vessel lining is a major use.

A variety of special steels give higher strengths combined with ease of fabrication and corrosion resistance. These steels are designed for processing in the ductile austenitic form and then strengthened by transformation to martensite and precipitation hardening. Most but not all of these steels are readily weldable by the inert gas processes but need one or more post-weld heat treatments to give good joint strengths.

The high price of these steels demand good stock control so that full information should be available for the welding engineer regarding steel type and potential welding problems. It is important that the steel should be identified before welding if subsequent problems are to be avoided. A simple magnetic test will often serve to identify an austenitic stainless steel.

Welding the Austenitic Stainless Steels

After the elimination or minimisation of the standard weld defects, the main need is to maintain corrosion and/or oxidation resistance. In general filler metals with similar alloy additions to the parent steel are used, although often with slightly higher Cr and Ni contents to maintain corrosion or oxidation resistance in the "cast" weld metal and with a correction to eliminate weld metal cracking. The need to match weld metal colour, strength and other properties to those of the parent steel also lead to the use of weld filler compositions similar to the parent steel.

The high coefficient of thermal expansion and low thermal conductivity of these steels can lead to high distortion or, if weldments are rigidly clamped high internal stress. Welding processes with a diffuse

heat source give more problems than those with concentrated heat sources. Distortion leads to problems of fit and residual stress can lead to cracking or stress corrosion in certain media.

To maintain corrosion resistance it is necessary to maintain the appropriate Cr content in solution. Some weld deposits contain up to 10% delta ferrite to reduce the risk of weld metal hot cracking but this has no effect upon corrosion resistance except in the most reactive media.

With 18Cr 8 Ni steels containing 0.05% C or higher the steel can be sensitized by holding in the temperature range 500° - 900°C for several minutes. This treatment results in the precipitation of Cr rich carbides, Cr_{23}C_6 at the grain boundaries but these temperatures will not allow widespread Cr diffusion so that Cr depleted zones with reduced corrosion resistance occur around grain boundaries. Sensitization can occur in part of the heat affected zone during welding and subsequent exposure to mildly corrosive environments results in accelerated corrosive attack in the haz or weld decay. Therefore, plain 18Cr 8Ni steels with over 0.03% C should not be welded unless a post-weld heat treatment at 950°C or 1050°C can be applied to redistribute the Cr or to dissolve the Cr_{23}C_6 . Alternatively low carbon (0.03% or less) stainless steels with insufficient C to deplete the Cr can be used, although these steels are expensive. A cheaper method of avoiding sensitization is to add to the steel an element which has a greater affinity for C than Cr. Ti and Nb (Cb) can be used and lead to the stabilized stainless steels (Types 321, 347). These types are fully weldable but Ti cannot be transferred across a welding arc so that weld filler metals contain Nb. No post-weld heat treatment is used.

Although stabilized steels do not suffer from weld decay they can exhibit knifeline attack in certain media, which is the intergranular corrosion of a narrow region of the haz close to the fusion boundary. This region of the haz is heated to a high enough temperature to dissolve the carbides of Nb or Ti and if the region is reheated to 650°C (perhaps in multi-run welding) precipitation of Cr_{23}C_6 can occur in preference to carbides of Nb or Ti. However if the region is heated to around 870°C the carbides of Nb or Ti are precipitated and corrosion resistance is maintained.

In chloride solutions, welded stainless steels with high internal stress can suffer from stress corrosion cracking. Where service is required in such media a high temperature stress relief above 950°C is necessary.

Despite the generally good weldability hot cracking can be encountered in weld metal, or occasionally the weld haz. This cracking is caused by a low ductility at 1300°C, possibly due to low melting point segregates, and is increased by the presence of S, P and possibly Si. If hot cracking does occur, high impurity levels should be suspected. High

restraint can also increase cracking so that the problem is worse in thick steels (where segregation can lead to high S and P levels). Cracking can be reduced or even eliminated by keeping restraint to a minimum (good joint fit-up!), low weld heat input and rapid cooling. The presence of 4 - 8% of delta ferrite in the austenitic weld metal can reduce the danger of cracking by holding the dangerous impurities in solution. The ferrite does not adversely affect joint properties. Electrodes and filler wires for welding all but the 25 Cr 20Ni steels are often chosen to give this delta ferrite content. The welding engineer can choose an appropriate composition with the aid of the Schaeffler Diagram which relates composition to the ferrite content for weld metal. Dilution must also be taken into account. In general, the aim is to keep a largely austenitic structure with less than 10% delta ferrite.

The weld heat thermal cycle can result in fine precipitation within the heat grains in many austenitic steels which greatly increases the hot strength of the grains. If the welded structure is subsequently heat treated or placed in elevated temperature service, the thermal strains are applied directly to the heat grain boundaries and cracking can occur - re-heat cracking. In many cases an initiation point such as a hot tear at the weld toe or another stress raiser, is required to initiate cracking. Re-heat cracking raises problems of steel selection for high temperature service since steels with the highest creep strength are the most susceptible. Nb containing steels are the most susceptible to re-heat cracking; Mo steels are resistant since they do not contain these fine intra-granular precipitates.

The austenitic stainless steels can form a brittle intermetallic phase, sigma Fe Cr, on prolonged heating at 565° - 925°C. The times for sigma formation are generally several hundred hours even if delta ferrite is present, which accelerates sigma formation. However, several weld metals can form sigma phase during post-weld heat treatment and in the long thermal cycles associated with strip cladding (where delta ferrite contents can reach 14%) sigma phase has been claimed to form during welding. Sigma phase can cause severe embrittlement at room temperatures and loss of creep strength above 700°C.

Most fusion welding processes can be used for the austenitic stainless steels. General welding precautions include the maintenance of good accessibility, wide edge preparations and low restraint. Short arc lengths are necessary to eliminate porosity and heat build-up must be prevented or distortion, high residual stresses or cracking can occur.

Manual metal arc welding generally utilizes basic coated electrodes with stainless steel core wires. Rutile coatings are also used and

give better handling characteristics and bead appearance. Synthetic electrodes with mild steel core wires and coatings loaded with alloying elements are also used. TIG welding is widely used on thinner materials using DC electrode negative or AC. Argon or helium shielding is used and 2 - 5% H_2 and may be added to eliminate porosity. Pulsed TIG is useful with thin material as it reduces distortion and controls penetration. MIG welding with spray transfer is widely used for heavier sections and A/O_2 or A/CO_2 gas mixtures give better wetting and bead appearance than argon alone. CO_2 is not generally recommended because of C pick-up and Si and Mn loss. Short circuiting transfer techniques can be used but better results are obtained with pulsed current techniques. Plasma welding at all current ranges gives good results; A/H_2 mixtures are generally used. Pulsed current techniques at higher currents have been shown to have advantages of easier control, minimal distortion, lack of surface undercut and uniform penetration. Submerged arc welding can be used for the 18/8 steels but the large weld pools give increased cracking tendency and 10% delta ferrite is recommended in the weld metal. Cladding by this process is widely used. Electroslag welding is possible but not widely used. Electron beam welding gives extremely good results and the weld section makes for low distortion. Finally, oxy-acetylene welding is widely used for repair work and minor jobs. Special fluxes and care in cleaning is essential and the spread heat source can lead to high distortion and residual stresses.

No preheat should be used for the austenitic steels and interpass temperatures should not exceed $100^{\circ}C$. Post-weld heat treatment is not generally recommended because of problems of carbide precipitation, sigma formation and re-heat cracking. Where heat treatment is necessary, $950^{\circ}C$ is the minimum for stress relief and diffusion of Cr and $1050^{\circ}C$ is necessary for full dissolution of carbides. Heat treatment times must be restricted to 3 mins. for every 2.5mm (0.1in.) of thickness to avoid grain growth followed by quenching use of water spray or, for thin material, air cooling.

Martensitic/Austenitic Steels

These steels obtain their properties by carefully controlled heat treatments and sometimes by controlled working, so welding problems may be anticipated. Fortunately, most of these steels are readily weldable by the inert gas processes although subsequent post weld heat treatment(s) may be required to retain properties. The rule must always be to follow the manufacturers instructions.

Al containing alloys e.g. 17-7 PH are not weldable by covered electrodes because of problems of transferring Al across the arc while certain other compositions e.g. 17-10 P, cannot be welded because of cracking tendencies.

These alloys are best welded in the austenitic or martensitic condition before precipitation hardening, but many can be welded after hardening. It is generally claimed that only a final precipitation hardening treatment is required for steel welded in the martensitic or fully hardened condition, to restore joint properties, but some strength is generally lost. The best results are obtained by full heat treatment but this is not always possible.

Ferritic Stainless Steels

These steels, with their intrinsic lack of ductility at ambient temperatures require a preheat of 200°C when welded in heavy sections with a ferritic weld deposit. Preheats can be reduced or avoided when austenitic deposits are used. With sheet material no preheat is needed. Manual metal arc welding with matching ferritic deposits or austenitic deposits are used, submerged arc welding with austenitic deposits and TIG welding are also possible. Since no solid phase transformations occur, post-weld heat treatment cannot be used to refine the grain size. However, treatment at $750 - 850^{\circ}\text{C}$ to relieve residual stresses and spheroidise carbides is useful in reducing brittleness and eliminating intercrystalline corrosion.

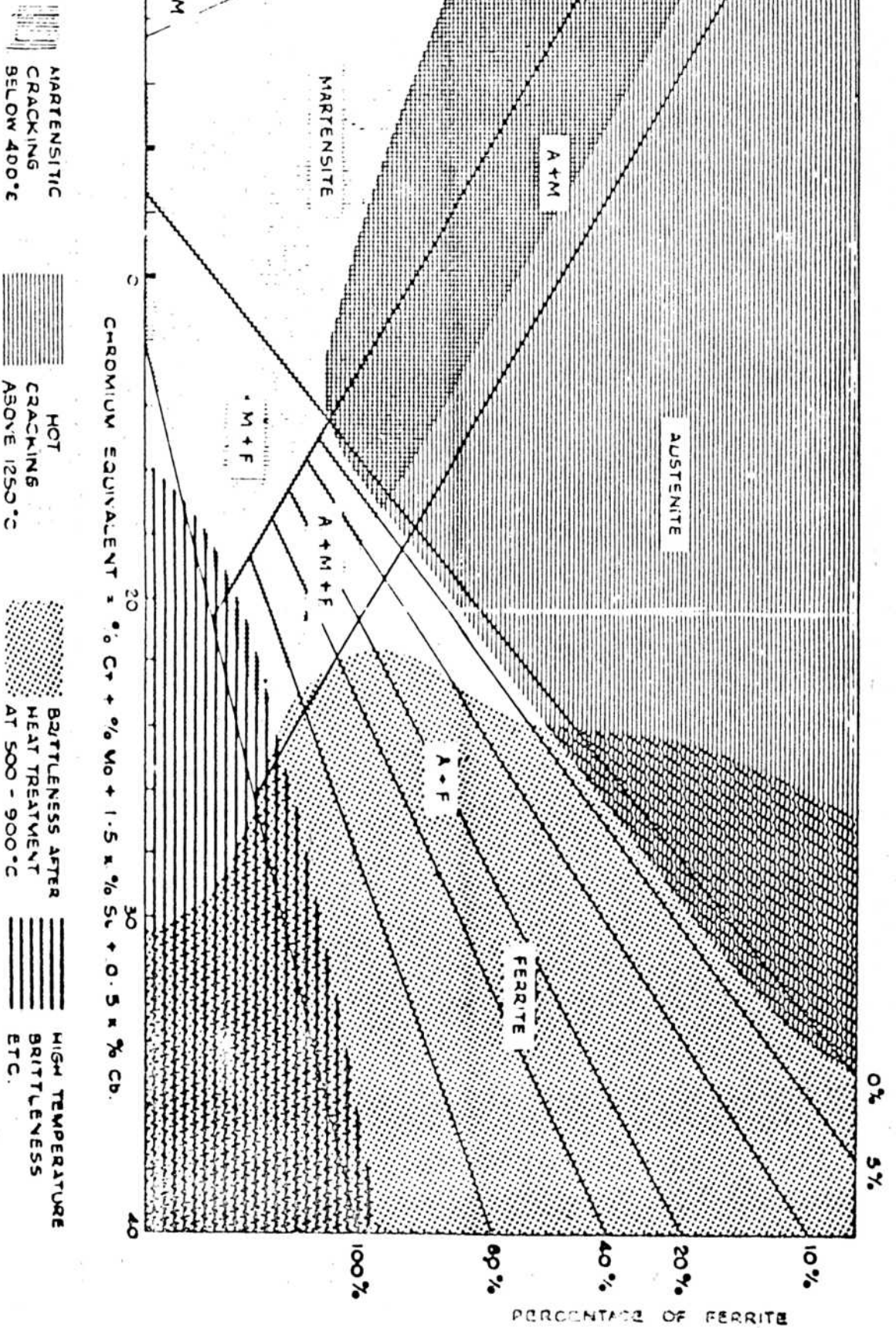
It should be noted that these steels can suffer from weld decay and sigma phase formation.

Martensitic Stainless Steels

These steels can be dealt with directly as martensitic steels and have the usual welding problems of underbead cracking and loss of ductility. Carbon content is of greatest importance.

Although sheet material can be welded autogenously without pre-heat precautions are generally required with thicker sections. In extreme cases, to avoid cracking and maintain joint properties, preheat from $200 - 370^{\circ}\text{C}$ (near or above M_s) is required, followed by slow cooling to 100°C and then immediate post-weld tempering at $750 - 800^{\circ}\text{C}$ is required. Tempering in the range $550 - 700^{\circ}\text{C}$ must be avoided or sensitization can occur.

constitution diagram for stainless steel welds with approximate regions of defects depending on composition and phase balance



METALLURGICAL EFFECTS DURING STRESS RELIEF : AUSTENITIC STAINLESS STEELSAfter C.L.Cole and J.D.James

Material Type	Condition	Stress relief heat treatment			
		200-400°C ~40% peak	350-650°C ~35%	850-900°C ~85%	950-1050°C ~95%
18-8 e.g. 304S15	SHT, SA* weldments*	NME NME	IGP IGP & SIP	IGP IGP & SIP	IGP(C) IGC(C)&SIP(H)
18-8-1 e.g. 321S12	SHT, SA weldments	NME	NME SIP	NME SIP	NME SIP(H)
18-8-0.03C e.g. 304S12	SHT, SA and weldments	NME	NME	NME	NME
18-12-Mo e.g. 316S16	SHT, SA* and weldments*	NME	IGP	IGP	IGP(C)
18-12-Mo-Ti e.g. 320S17	SHT, SA, and weldments	NME	NME	NME	NME
18-12-Mo-0.03C e.g. 316S12	SHT, SA, and weldments	NME	NME	NME	NME

SHT solution heat treatment, rapidly cooled from ~1050°C

SA solution annealed, slowly cooled from ~1050°C

* likely to be sensitized prior to stress relief, unless of small section size and/or low carbon content

IGP intergranular precipitation of chromium carbide (sensitization)

NME no metallurgical effect

SIP strain-induced precipitation of carbides weldments over 0.75 in.

(H) during heating

(C) during cooling (assumed always to be slow)

Description of Heat Treatment

950 - 1050°C giving the maximum relief of about 95%

850 - 900°C providing about 85% relief, but permitting the use of more conventional heat-treatment facilities

550 - 650°C promoting dimensional stability, with about 35% relaxation, but with possible difficulties from sensitization

200 - 400°C giving only about 5% general stress relief, but reducing peak stresses by up to 40%, and therefore suitable for assisting further fabrication, or for reducing the tendency to stress corrosion.

BLE 1 - PREHEAT LEVELS

MATERIAL	CODE	MIN. PREHEAT °C	NOTES
CARBON STEEL	BS.2633	Nil 100 150	Thickness < 19 mm, C ≤ 0.26% Thickness > 19 mm, C ≤ 0.26% All thickness, C > 0.26%
	ASME*	10 80	Thickness < 25 mm and C ≤ 0.3% Thickness > 25 mm and C > 0.3%
C-½ Mo	BS.2633	150	All Thickness
	ASME*	10 80	Thickness < 16 mm or UTS < 48 h bar (70,000 psi) Thickness > 16 mm or UTS > 48 h bar (70,000 psi)
Cr-½ Mo ½ Cr-½ Mo	BS.2633	150	All Thickness
	ASME*	10 120	Thickness < 12.5 mm or UTS < 41.5 h bar (60,000 psi) Thickness > 12.5 mm or UTS > 41.5 h bar (60,000 psi)
Cr-1 Mo Cr-½ Mo 7 Cr-½ Mo Cr-1 Mo	BS.2633	200	All Thickness
	ASME*	150 200	Thickness < 12.5 mm + Cr < 6% or UTS < 41.5 h bar (60,000 psi) Thickness > 12.5 mm + Cr > 6% or UTS > 41.5 h bar (60,000 psi)
½ N1	BS.2633	250	All Thickness
	ASME*	150	All Thickness

ASME Section VIII Division 1 and 2

ASME Section III

Table 3 General PWHT requirements: UK codes: C and C-Mn steels

Code	BS 1500 : Part 1 : 1958	BS 1515 : Part 1 : 1965	BS 1113 : 1969	BS 2633 : 1966	BS 2790 : Part 1 : 1969	BS 3915 : 1965, BS 4208 : 1967
Heating rate	Above 300°C Max. 200°C/h or 200°C/h	Above 400°C Max. 200°C/h or 200°C/h or 50°C/h	Above 300°C Max. 200°C/h or 200°C/h or 50°C/h	Heat to temperature at a rate not exceeding 400°C/h	Furnace temperature raised slowly	Above 300°C Max. 250°C/h or 250°C/h
Cooling rate	$t = \text{max. thickness}$ Down to 300°C Max. 250°C/h or 250°C/h	whichever is the greater $t = \text{max. thickness}$ Down to 400°C Max. 250°C/h or 250°C/h or 50°C/h	whichever is the greater $t = \text{max. thickness}$ As per BS 1515 : 1965	Not specified Joints shall not be exposed to still air until they have cooled to a temperature of 500°C	Cool slowly to 400°C, then cool in still air	$t = \text{max. thickness}$ Down to 300°C Max. 250°C/h or 250°C/h
Temperature and time at temperature	Below 300°C may cool in air $t =$ 1 h/25 mm of max. thickness 600°-650°C Note: 575°C 1-5t h 550°C 2-5t h 525°C 3-75t h 500°C 6t h 475°C 12t h	1 h/25 mm of max. thickness 580°-620°C	1 h/25 mm of max. thickness 580°-620°C Note: 2 h/in of max. thickness, 2 h min. for furnace tempering of manipulated pipe 630°-670°C when $C \geq 0.25\%$ As per BS 1500 : 1958	1 h/25 mm of max. thickness 580°-620°C Note: 630°-670°C when $C \geq 0.26\%$	2½ min/mm of max. thickness 600° ± 20°C	Time $a + b + c \geq 1$ 4 16 580°-620°C Note: $a = h$ in range 600°-620°C $b = h$ in range 575°-600°C $c = h$ in range 550°-575°C No credit for time at temperature less than 550°C Thermocouples to be attached to the metal surface
Temperature measurement	Temperatures specified shall be the actual temperatures of any part of the vessel and shall be determined by thermocouples attached to the vessel	As per BS 1500 : 1958	As per BS 1500 : 1958	Thermocouples shall be in metallic contact with the parts being heated and shall be attached by an approved mechanical method	No mandatory requirements (in 1956 version furnace thermocouples)	Control temperatures during heating and cooling to avoid undesirably high stresses
Thermal gradients	During heating, not more than 150°C within any 4500 mm interval of length During holding not more than 50°C	As per BS 1500 : 1958 During holding entire vessel in range 580°-620°C Heated region $\geq 2.5\sqrt{Rt}$ on each side of centreline of weld Temperature profile as per BS 1113 : 1969 Thermocouples required to check peak and ½ peak temperatures	During heating, not more than 150°C within 4500 mm During holding entire vessel in range 580°-620°C Temperature profile so that length of cylinder on each side of weld at PWHT temp/2 $\geq 2.5\sqrt{Rt}$ Thermocouples required to check peak and ½ peak temperatures	No requirements given	No requirements given	BS 3915: Category A pipework, shell sections, etc. as per BS 1515 BS 4208: Control so that 'temperature gradient not harmful'
Local stress relief	Heated region, $\geq 3 \times$ shell thickness on each side of centreline of weld Harmful temperature gradients to be avoided	As per BS 1500 : 1958	As per BS 1500 : 1958	Temperature to be maintained symmetrical over peripheral band of a minimum width of 3 times width of weld preparation with the weld symmetrically disposed	Not allowed	

Table 4 General PWHT requirements: USA and ISO codes: C and C-Mn steels

Code	ASME Section VIII Division 1	ASME Section VIII Division 2	ASME Section III Category A (Category B as per ASME VIII Div. 1)	ISO 'Pressure vessels' (Draft)
Heating rate	Above 310°C Max. 205°C/h or 205°C/h Need not be less than 38°C/h t = max. thickness			Above 400°C Max. 220°C/h or 220°C/h t or 55°C/h whichever is the greater t = max. thickness
Cooling rate	Down to 310°C 235°C max. or 235°C/h Need not be less than 38°C/h Below 310°C may cool in still air			Down to 400°C Max. 275°C/h or 275°C/h t or 55°C/h whichever is the greater Below 400°C may cool in still air
Temperature and time at temperature	1 h/25 mm (1 h min. for ASME VIII Division 2) 593°C min. Note: When it is impractical to PWHT at above temperature PWHT at lower temperatures for longer times permitted Temp. below that specified 50°F 100°F 150°F 200°F	Time h/25 mm 2 3 5 10		2 min/mm 550°–600°C Note: ½ h minimum 2 h maximum
Temperature measurement	Temperature of vessel; Where more than one pressure vessel are PWHT together in one charge, thermocouples attached to confirm temperature	Only reference related to heating vessel internally	As per ASME VIII Division 2	As per BS 1500 : 1958
Thermal gradients	During heating not more than 120°C within any 4 500 mm interval of length During holding not more than 70°C	As per ASME VIII Division 1 During holding not more than 38°C	As per ASME VIII Division 1 As per ASME VIII Division 2	As per BS 1500 : 1958 During holding entire vessel in range 550°–600°C
Local stress relief	For shell circ. seams width of heated band on each side of weld > 2x shell thickness. For nozzle welds heated band 6x plate thickness. For pipe circ. butt welds heated band 3x width of finished weld	As per ASME VIII Division 1	As per ASME VIII Division 1	For circ. seams, heated region, $\geq 5\sqrt{Rt}$ on each side of centreline of weld. Temp. at edge of heated region $\geq \frac{1}{4}$ peak temp. For nozzle welds, heated band $\geq 2.25\sqrt{Rt}$ on each side of centreline of weld

should not exceed 400°C/h (presumably from ambient), all the codes infer that the heating rate from ambient to 300°–400°C is not critical. Presumably it is argued that stress relaxation only becomes significant in the creep range, and that at temperatures below this, heating rate is relatively unimportant. However, ambient to 300°–400°C is precisely the temperature range in which uncontrolled heating rates can produce excessive temperature gradients and thermal stresses. More than one failure either during stress relief or on subsequent hydraulic testing has been reported from cracks developing during the initial stages of the heat-treatment cycle as a result of thermal stresses being superimposed on the residual welding stresses.⁷

Above 300°–400°C most of the codes specify maxima and minima heating and cooling rates; these range from 200°–250°C/h maximum to the requirement that the rates need not be slower than 38°–55°C/h.

Such general requirements are of little value in guiding the fabricator faced with the problem of avoiding reheat cracking such as that encountered in Cr-Mo-V⁸ and Cr-Mo steels when heating rate above 450°C is critical or with avoiding temper embrittlement when the cooling rate in the range 450°–600°C must be as fast as possible.¹

TEMPERATURE AND TIME AT TEMPERATURE

With the exception of BS 3915 and BS 4208, the general rule in most codes with respect to soaking time is 'one

hour per 25 mm of thickness'. BS 1500 and the ASME codes lay down longer soaking periods when temperatures lower than the recommended minima are employed.

The PWHT temperatures recommended by the several codes are given in Table 5. For alloy steels the UK codes generally indicate that these are given as a guide and that the effect of heat treatments proposed should be considered in regard to any special requirements. However, no instructions are given on how to assess the optimum PWHT temperature.

The UK code recommendations with regard to temperature are reasonably consistent, with the exception of BS 1500, which only applies to C and C-Mn steels and, like the ASME codes, permits the use of temperatures lower than the recommended minimum (see Tables 3 and 4).

ASME codes are unique in that they specify minima values only. The ISO draft recommends a lower temperature range for C and C-Mn steels than other codes, 550°–600°C as opposed to 580°–620°C in the UK codes, and 593°C min. in ASME codes. The maximum temperature recommended in the ISO draft for 3½Ni corresponds to the minimum recommended in the UK codes. For 2½Cr-1Mo and 5Cr-½Mo, the ISO draft permits a much wider temperature range than the UK codes with the exception of BS 2633 as applied to 2½Cr-1Mo.

This latter code is the only UK code attempting to cater for service conditions when creep properties are important, and for others, such as hydrogen service when

Table 5 Detailed PWHT conditions for various materials

Material	Code				ASME Section VIII Div. 1 and 2 ASME Section III		ISO
	BS 1500 : Part 1 : 1953	BS 1515 : Part 1 : 1965	BS 1113 : 1969†	BS 2633 : 1966‡			
Carbon steel	600–650°C 1 h/25 mm	580–620°C 1 h/25 mm	580–620°C* 1 h/25 mm	580–620°C* 1 h/25 mm ($\frac{1}{2}$ h min.)	P1	593°C min.* 1 h/25 mm	550–600°C 2 min/mm ($\frac{1}{2}$ h min., 2 h max.)
C–1Mo	—	620–660°C 1 h/25 mm (1 h min.)	—	630–670°C 1 h/25 mm (1 h min.)	P3	593°C min.* 1 h/25 mm	580–620°C 1 h/25 mm (1 h min.)
$\frac{1}{2}$ Cr– $\frac{1}{2}$ Mo	—	—	—	—			—
1Cr– $\frac{1}{2}$ Mo	—	620–660°C 1 h/25 mm (2 h min.)	630–670°C 1 h/25 mm	630–670°C 1 h/25 mm ($\frac{1}{2}$ h min.)	P4	593°C min.* 1 h/25 mm	620–660°C 1 h/25 mm (1 h min.)
$1\frac{1}{2}$ Cr– $\frac{1}{2}$ Mo	—		—	630–670°C 1 h/25 mm ($\frac{1}{2}$ h min.)			—
$2\frac{1}{2}$ Cr–1Mo	—	660–700°C 2 h/25 mm (2 h min.)	650–700°C 1 h/25 mm	660–690°C or 700–750°C¶ 1 h/25 mm (1 h min.)	P5	676°C min.* 1 h/25 mm	625–750°C 1 h/25 mm (2 h min.)
5Cr– $\frac{1}{2}$ Mo	—	700–740°C 2 h/25 mm (2 h min.)	—	710–760°C 1 h/25 mm (2 h min.)			670–740°C 1 h/25 mm (2 h min.)
7Cr– $\frac{1}{2}$ Mo	—	—	—	—			—
9Cr–1Mo	—	—	—	—			—
$\frac{3}{4}$ Ni	—	580–620°C 1 h/25 mm (1 h min.)	—	590–620°C 1 hr/25 mm (1 h min.)	P9b	620°C min.* 1 h/25 mm	550–590°C 1 h/25 mm (1 h min.)
$\frac{1}{2}$ Cr– $\frac{1}{2}$ Mo– $\frac{1}{4}$ V	—	—	660–700°C 1 h/25 mm (3 h min.)	670–710°C 1 h/25 mm	—	—	—

* See alternative permissible temperature/time combinations (Tables 3 and 4)

† Plate material, for forgings, pipe, tube, temperatures may be different

‡ 630–670°C when C $\geq$ 0.25%

§ Heat treatment of welds only

|| Optimum creep properties

¶ Softening welds where creep properties unimportant

• Max. PWHT temp. 675°C

BS 3915 : 1965 and BS 4208 : 1967 do not have specific requirements relating to low-alloy steels. Generally BS 1515 : Part 1 : 1965 is used to cover such materials

full softening is required. In that BS 2633 clearly delineates the optimum temperature ranges for the development of maximum creep properties and for softening in 2½Cr–1Mo it is more useful than the ISO draft.

The ASME codes are the only ones which comment on PWHT of dissimilar metal joints; they state that in such cases the PWHT shall be that specified for the material requiring the higher PWHT temperature. Clearly it is dangerous to take such recommendations at face value.

The majority of codes do not refer to the use of normalizing as a means of PWHT, although, e.g. in the case of electroslag welds, such a treatment may be necessary to produce properties adequate for design.

The only UK codes referring to normalizing used as a PWHT, as distinct from an operation to remove harmful effects of forming, are BS 2633 and BS 1113. Both these detail the PWHT to be given to pipe taking the condition of the pipe before welding into account. Table 6 reproduced from BS 2633 illustrates when normalizing is considered necessary.

The footnote to the table is most significant. As far as the authors are aware, it represents the only reference in UK codes to the marked drop in weld-metal properties which can occur during normalizing.

TEMPERATURE MEASUREMENT AND THERMAL GRADIENTS

All UK codes, except BS 2796, and the draft ISO recommendations lay down that the temperatures specified shall be the actual temperature of any part of the vessel. ASME codes state that the minimum PWHT

temperatures given shall be the minimum temperature of the plate material of the shell or head of any vessel. The only reference to thermocouples attached to vessels in ASME Section VIII Division 1 relates to the case when more than one vessel is subject to PWHT in one furnace charge and in Section VIII Division 2 and Section III to the case of vessels which are heated internally.

The danger of relying solely on measurement of furnace temperature is well illustrated by the failure of the Immingham pressure vessel for which a PWHT temperature of 620–660°C had been specified.⁷ Although the furnace recording thermocouple indicated that the PWHT had been carried out correctly, the actual temperature reached in one part of the vessel was of the order of 525°C maximum. The vessel failed on hydraulic test, and it was concluded that the incomplete removal of residual stresses and insufficient tempering of weld metal and HAZ were prime factors in causing brittle fracture.

BS 2633 and BS 2790 make no reference to thermal gradients. BS 3915 has the vague requirement that temperature gradients be controlled 'so as to limit thermal stress to an acceptable value and to avoid unacceptable locked-in stresses being present on completion of the operation'. The remainder of the codes listed give a maximum permitted differential during heating and during soaking; BS 1515, for example, specifies a maximum of 150°C within any 4500 mm (15 ft) length on heating. The justification for such requirements is obscure, and they can be interpreted in several ways. For example, it could be taken to imply

MATERIAL CLASSIFICATION	CONDITIONS OF PIPES PRIOR TO WELDING			FINAL HEAT TREATED CONDITION OF PIPES
	As Rolled	Normalised	Normalised and tempered	
Carbon steel, up to 0.26% carbon	A	A	-	-
Carbon steel; from 0.26 to 0.4% carbon	B*	A	-	Normalised
Carbon-molybdenum steel	B*	A	-	Normalised
1% Cr - ½% Mo	B*	A	-	Normalised
1¼% Cr - ½% Mo	C*	D	A	(i) Normalised and tempered or (ii) Annealed
2¼% Cr - 1½% Mo	C*	D	A	(i) Normalised and tempered or (ii) Annealed
5% Cr - ½% Mo	-	-	-	Annealed
7% Cr - ½% Mo	C*	D	A	(i) Normalised and tempered or (ii) Annealed
9% Cr - 1% Mo	C*	D	A	(i) Normalised and tempered or (ii) Annealed
½% Cr - ½% Mo - ¼% V	C*	D	A	Normalised and tempered
3½% Nickel	-	D	A	Normalised and tempered

* Where pipes, on completion of welding, are normalised or normalised and tempered in a furnace the fabricator should ascertain that the weld metals used will be suitable, after the heat treatment, for the services for which they are intended. Normalising temperature and times shall be in accordance with Table 4.

KEY TO LETTERS USED IN TABLE 5.

Post-weld heat treatment:

- A. Welds locally heat-treated (See Subclause 19 h(ii) regarding cooling rate).
- B. Pipes and welds normalised in furnace
- C. Pipes and welds normalised and tempered in furnace.
- D. Pipes and welds tempered in furnace.
- E. Welds locally heat-treated (see Subclause 19 h(iii) regarding cooling rate).

TABLE 6- RESIDUAL STRESS AFTER 1" PLATE (Initial Strain 0.15%)

CONDITIONS					
	670°C/1Hr	550°C/1Hr	600°C/1Hr	575°C/1Hr	550°C/2.5Hr
	h bar: 2 tonf/in ²	h bar: 2 tonf/in ²	h bar: 2 tonf/in ²	h bar: 2 tonf/in ²	h bar: 2 tonf/in ²
RESIDUAL STRESS					
BS 1501 STEEL					525°C/3.75Hr
1511 Gr. 28	-	11.3 : 0.9	2.5 : 1.6	3.5 : 2.3	3.7 : 2.4 (4.5 : 2.9)*
213 Gr. 32	11.2 : 0.3	2.0 : 1.3	3.2 : 2.1	4.3 : 3.1	5.3 : 3.4 (7.0 : 4.6)*
221 Gr. 32	-	-	3.7 : 2.4	6.2 : 4.0	7.4 : 4.8 (7.7 : 5.0)*
224 Gr. 28/ 32	-	-	2.8 : 1.8	3.7 : 2.4	3.6 : 2.3 (4.9 : 3.2)*
271 281 320 Gr. 28-32	- - - 9.6 : 6.2	7.5 : 4.9 4.1 : 2.7 4.3 : 2.8 12.1 : 7.8	12.6 : 8.2 9.0 : 5.8 8.7 : 5.6 15.4 : 10.0	14.6 : 9.5 12.0 : 7.8 10.8 : 7.0 17.0 : 11.0	15.2 : 9.9 12.8 : 8.3 12.1 : 7.8 18.5 : 12.0
BS 1501 STEEL					575°C/10Hr
	700°C/1Hr	675°C/1Hr	650°C/2Hr	625°C/3Hr	600°C/5Hr
RESIDUAL STRESS					
	h bar: 2 tonf/in ²	h bar: 2 tonf/in ²	h bar: 2 tonf/in ²	h bar: 2 tonf/in ²	h bar: 2 tonf/in ²
622 Gr. 31	-	-	6.2 : 4.0	7.7 : 5.0	9.6 : 6.2
622 Gr. 35	2.3 : 1.5	3.9 : 2.5	5.7 : 3.7	7.6 : 4.9	9.1 : 5.9
					10.7 : 6.9 11.3 : 7.3

* Figures in brackets 550°C/1Hr; ISO recommendation

TABLE C - TEMPERING TEMPERATURES AND PWHT RECOMMENDED BY BS MATERIAL SPECIFICATIONS

MATERIAL	BS 1501: PART 1: 1964		BS 1503: 1969		BS 1501: PART 2: 1970	
	Temper	PWHT	Temper	PWHT	Temper	PWHT (1)
Carbon Steel		600 $\pm$ 10°C (1)(2)	580-560°C (1)	580-560°C (1)(2)	640-690°C	630°C min (2)(3)
1 Cr- $\frac{1}{2}$ Mo			620-690°C	620-690°C	640-690°C (4)	630°C min (2)(3)
1 $\frac{1}{2}$ Cr- $\frac{1}{2}$ Mo			620-690°C	620-690°C		
2 $\frac{1}{2}$ Cr-1 Mo			650-700°C	650-700°C	Grade 31	
					650-725°C (4)	630°C min (2)(3)
					Grade 45	
3 Cr- $\frac{1}{2}$ Mo			620-700°C	620-700°C	630-670°C	600°C min (2)(3)
5 Cr- $\frac{1}{2}$ Mo			675-750°C	675-750°C		
3 $\frac{1}{2}$ Ni			580-630°C	580-630°C	580-630°C	(3)
1 Cr- $\frac{1}{2}$ Mo- $\frac{1}{4}$ V			660-700°C	660-700°C		
NOTES	(1) Test coupons cooled in still air (2) Advises that properties adversely affected when PWHT done above 620°C or when time at max. temp. greater than 3 hrs. (1) When design based on creep or stress rupture properties temperature should not exceed 620°C (2) Test coupons heat treated to produce similar effect on properties as that applied to vessel (1) Test coupons taken from fully heat treated plates - no provision for PWHT of test coupons. (2) Advises that properties adversely affected when PWHT done below 630°C/600°C (3) States that if tempering temperature exceeded properties may be adversely affected. (4) When creep or stress rupture not a controlling design criterion higher temperature may be used.					

TABLE 2 - RESIDUAL STRESSES AFTER PENT WELD METAL VERSUS PLATE MATERIAL: CARBON STEEL.

Initial Strain 0.15%

Material	CONDITIONS			
	525° C/0 Hr.	525° C/1 Hr	525° C/10 Hr	600° C/1 Hr.
	RESIDUAL STRESS			
BS 1501 224 Gr 28 Plate Matching Weld Metal AWS/A516 E 7016	h bar : tonf/in ²	h bar : tonf/in ²	h bar : tonf/in ²	h bar : tonf/in ²
	14.1 : 9.1	6.0 : 3.9	3.6 : 2.3	2.6 : 1.7
	20.1 : 13.0	15.9 : 10.3	14.2 : 9.2	11.4 : 7.4

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ALUMINIUM ALLOYS AND THEIR WELDABILITY

1. The Alloy systems

There are many aluminium alloys and grades of 'pure' aluminium (the American Aluminium Association currently lists 50-60 different wrought alloys) but many of these are for very special purposes e.g. foil containers, can ends, lithographic plates which are not relevant to general engineering. It is convenient for descriptive purposes to divide the wrought alloys into 3 main groups - pure aluminium, non-heat treatable alloys and heat treatable alloys. Table 1 lists the more important alloys.

a) Pure aluminium

There are a number of grades of pure aluminium with purities ranging from 99.99% to 'electrical grade' (99.5%) with a guaranteed conductivity of 61.5% I.A.C.S. to 'commercial purity' (99.0%). The principal impurities are iron and silicon. The strength of pure aluminium is low (U.T.S. 70-150 N/mm²) depending upon the amount of (cold work) but the ductility and corrosion resistance are high. It is used mainly for electrical conductors, lightly stressed and decorative assemblies and for some food and chemical plant.

What can be considered as a slightly stronger grade of 'pure' aluminium is the Al-1½% Mn alloy which has a U.T.S. of 90-195 N/mm². This is used widely for sheet metal work, hollow-ware, packaging and vehicle panelling.

b) Non-heat treatable alloys

These are based on the aluminium-magnesium system and obtain their strength principally by solution hardening from Mg and, in some alloys, Mn + Cr although there is some contribution from the work hardening resulting from rolling. Strengths range from a U.T.S. of 180 N/mm² for Al-Mg₂ to 275 N/mm² min. for the strongest alloy Al-Mg_{4.5} Mn. This alloy family is much used in welded construction because of its high strength, weldability and good corrosion resistance. Application include road and rail transport vehicles, pressure vessels, LNG storage tanks and ships superstructures.

The alloys are normally supplied as sheet or plate, less commonly as extruded sections.

c) Heat treatable alloys

These alloys obtain their strength as a result of a precise heat treatment, normally given by the supplier, which promotes precipitation hardening. The specific heat treatment varies with the alloy but will always comprise a high temperature solution treatment (450-500°C for 1-2 hours) followed by a quench to room temperature followed by a precipitation treatment for, say, 16 hours at an elevated temperature (150-180°C) or some days at room temperature.

or a combination of both. Heat treatments are highly specific to a given alloy and need to be closely controlled.

The very high strength aircraft alloys of the Al-Cu-Mg, Al-Cu-Mg-Si and Al-Zn-Mg-Cu alloys with U.T.S. values of up to 500 N/mm² come into this group but they are only welded under very special circumstances and in general engineering terms are usually described as 'unweldable'. A newer class of strong but weldable alloys based on the Al-Zn-Mg system is now being widely used for armour plate and military bridges.

The most important heat treatable alloys in terms of tonnage produced are the Al-Mg-Si alloys which have strengths in the range U.T.S. 150-300 N/mm² and relatively high 0.2% proof stresses (yield strengths) in the range 100-270 N/mm² (the strongest non-heat treatable alloy has a 0.2% proof stress of 130 N/mm²). The alloys are very easy to extrude and are widely used for such applications as window frames, motor car trim, structural members in road and rail transport vehicles, cranes, bridges, roof trusses, etc. These alloys are easily welded but the as welded strengths are naturally much lower (up to 50% lower sometimes) than the heat treated parent material. Consequently their use in welded construction is limited to situations where the joint tensile strength is not the critical factor e.g. as extruded stiffeners fillet welded to plates or in structures where fatigue loading is the deciding factor. Post weld heat treatment to improve the weld strength is rarely practicable and is only ever used on small components.

2. Welding Metallurgy

Compared to ferrous alloys, the welding metallurgy of aluminium and its alloys is very simple because apart from precipitation hardening in the heat treatable alloys, which is a low temperature reaction, there are no phase changes below the solidus. In practice, the main consideration is the selection of consumables (filler alloys) to achieve a weld metal of adequate tensile strength and freedom from solidification cracking (hot cracking). Fracture toughness does not need to be considered because all the weldable aluminium alloys are so tough even at very low temperatures that it is difficult to measure the fracture toughness.

There are 3 types of filler metal normally available: pure aluminium, Al-5% Si and Al-5% Mg, some varieties of the latter containing Mn and/or Cr also. The normal rule is to use pure aluminium or Al-5%Si for welding pure aluminium, Al-5% Mg for welding Al-Mg alloys and either Al-5% Mg or Al-5%Si for welding Al-Mg-Si alloys. This rule can be expressed slightly differently by saying that a filler of 'matching' composition should normally be used but one with, if possible, a higher alloy content than the parent material. The higher alloy content will help to improve the strength of the weld metal and will also reduce the hot cracking tendency (see figures 1 & 2).

It is a good, general rule that the 'as welded' tensile strength of a weldable aluminium alloy (i.e. as measured by a transverse tensile test on a butt weld) is equivalent to that of the parent material in the annealed ('O') condition. A weld is rarely stronger than this although of course it can be weaker if an inappropriate welding technique is used or defects are present.

3. The differences between Aluminium & Steel which are of significance to the Welding Technology

The welding technology of aluminium alloys is different from that of steel in certain important areas - welding processes, metallurgy and quality control standards and it is convenient to discuss this by first describing 5 major differences in the chemical and metallurgical properties between aluminium and steel which are significant to fusion welding:

- 1) Aluminium has a tenacious, rapidly forming and protective oxide film.
- 2) Suitable welding fluxes for aluminium leave corrosive residues and do not have very good electrical properties.
- 3) Aluminium alloys undergo no phase changes in the solid state and have a F.C.C. crystal structure.
- 4) Aluminium weld metal is always weaker and more ductile than wrought material of the same composition.
- 5) Aluminium has a low solubility for hydrogen when in the solid state.

Factors 1 & 2 mean that flux welding processes (manual metal arc, submerged arc and electroslog) are not very satisfactory and cannot be relied on to give good quality weld deposits. Almost all fusion welding of aluminium is done by inert gas shielded arc processes (TIG or MIG) using argon or helium as the shielding gas. Some pre-cleaning to loosen the oxide film is frequently carried out particularly for high quality work.

Factor 3 means that there are no cooling rate sensitive phase transformations (e.g. bainitic or martensitic reactions) as a result of welding and no ductile to brittle transition temperature. Consequently control of weld heat input, preheat and post weld stress relief or tempering are not required for aluminium alloy welding. Preheat is occasionally used on very thick sections (>100 mm) but merely as an aid to obtaining good weld penetration not for any metallurgical reason.

Factor 4 & 5 mean that as the weld metal does not 'overmatch' the plate as is normal in welding ferrous metals there is a fear that small weld defects will be more serious in their effect on weld strength. Also, under given conditions of cleanliness, i.e. a given concentration of H₂ in the arc atmosphere arising from moisture and grease on the plate and consumables, there is more chance of getting porosity in the weld. In practice the effect of weld defects on the strength of welds in aluminium alloys is directly proportional to the loss in thickness they create allowing, of course, for any bending stresses (figure 3 & 4) and since, say, 3% porosity is excessively bad by any standards the significance of porosity in aluminium welds is frequently overemphasised. Table 3 lists some actual test results showing the effect of varying amounts of porosity on the strength of welds in Al-Mg4.5Mn. Because the fracture toughness of weldments in aluminium alloys is always high there is no need to consider weld defects in terms of their tendency to initiate unstable ('brittle') fracture. Therefore the whole subject of the significance of weld defects in welded aluminium structures subjected to static loading is described by figures 3 & 4. For welds subjected to fatigue loading the importance of weld shape is paramount as for welds in steel and B.S. code of practice CP118 'The Structural Use of Aluminium' classifies joint types in terms of fatigue loading and specifies design stresses. Table 4 gives some examples from this standard.

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Table 1

Some Common Wrought Aluminium Alloys

Alloy Type	B.S.	A.A.	I.S.O.	Temper	Plate		% Elongation on 5.65 50	Extrusion		% Elongation on 5.65 50
					0.2% P.S.2 (N/mm ²)	U.T.S. (N/mm ²)		0.2% P.S.2 (N/mm ²)	U.T.S. (N/mm ²)	
99.0 purity	1C	1200	Al 99.0	M		70			60	20
				H4		110-140	6			
Al-2.5%Mg	N4	5052	AlMg2	M		180	14			
Al-4.5%Mg+ Mn+Cr	N8	5083	Al+Mg4.5-Mn	O	125	275-350	14	125	275	14
Al-Mg-Si	H9	6063	AlMgSi	TB				70	130	16
				TF				160	185	8
Al-Mg-Si-Mn	H30	6351	AlSiMgMn	TB	115	200	15	120	190	16
				TF	240	295	8	270	310	8

Temper designations are : M - as manufactured

O - Annealed

H4 - Strain hardened to about half its maximum capability

TB - Solution heat treated and naturally aged

TF - Solution heat treated and artificially aged

Table 2

Physical Properties of Aluminium *

density	2.705 g/ml.
coefficient of expansion	24×10^{-6} per °C
Thermal conductivity	220 W/m°C
meltingpoint +	660°C
elastic modulus (E)	70,000 N/mm ²

* These vary slightly with alloy composition.

+ The solidus of some alloys is below 600°C.

Table 3

EFFECT OF UNIFORM POROSITY ON THE STATIC TENSILE STRENGTH OF BUTT WELDS
IN Al.Mg4.5 Mn

Grade of porosity*	Weld strength (percentage of specified minimum tensile strength of annealed plate)	
	Bead on	Bead removed
Sparse A		100
Scattered A	100	100
Extensive A	100	
Extensive A	99	
Extensive A	100	97
Sparse B		100
Sparse B, scattered A	100	98
Scattered B	92	
Scattered B	100	99
Scattered B, scattered A		95
Scattered B, extensive A	100	
Extensive B, sparse C	88	
Extensive B	97	95
Sparse C, scattered B, extensive A		88
Scattered C	89	
Scattered C, scattered B, extensive A		98
Extensive C	76	74

* As classified in B.S.499, Part 3. where

A : $\frac{1}{64}$ in. dia. pores, 'sparse' : 4t-40t per in² of radiograph

B : $\frac{1}{32}$ in. dia. pores, 'scattered' : 40t-160t per in² of radiograph

C : $\frac{1}{16}$ in. dia. pores, 'extensive' : more than 160t per in² of radiograph

where t = metal thickness (in.)

Table 4

CLASSIFICATION OF WELD TYPES ACCORDING TO THEIR FATIGUE STRENGTHS

(Derived from the B.S. code of practice 'The Structural Use of Aluminium' CP 118)

Class of joint	Weld type	$f_{\max}$. N/mm ² *
II	Continuous, full penetration longitudinal or transverse butt welds with the reinforcement dressed flush with the plate surface and the weld proved free from defects by non-destructive testing.	58
III b	Full penetration transverse butt welds made from both sides with the reinforcement on each side having a maximum height of 3 mm or 1/5th of the parent plate thickness, whichever is smaller and blending smoothly with the parent material.	38
c	Full penetration, continuous, longitudinal, automatic butt welds free from ripple and other transverse surface irregularities and with no interruption in welding either the root pass or the final pass.	
IV a	Continuous, longitudinal fillet welds with no interruption in welding either the root or the final pass.	29
b	Transverse butt welds made from both sides with the height of the reinforcement bead greater than permitted in class III b.	
V a	Transverse butt welds made from one side with an underbead.	24
b	Transverse butt welds made on a permanent backing strip attached with full length fillet welds parallel to the butt welds.	
c	Transverse non load carrying fillet welds.	
VI a	Transverse butt welds made on a permanent backing strip not attached by full length fillet welds.	20
b	Transverse load carrying fillet welds or cruciform welds, either weld being with or without full penetration.	
VII a	Continuous longitudinal fillet welds with unrepaired interruptions.	16
b	T-joints, the welds being with or without full penetration if welded from both sides but with full penetration if welded from one side.	
VIII	Intermittent, longitudinal, non-load carrying fillet or butt welds.	12
IX	Intermittent, longitudinal, load carrying fillet or butt welds.	9

* $f_{\max}$ for an endurance of 10^7 cycles at a stress range of $\frac{f_{\min}}{f_{\max}} = -1$

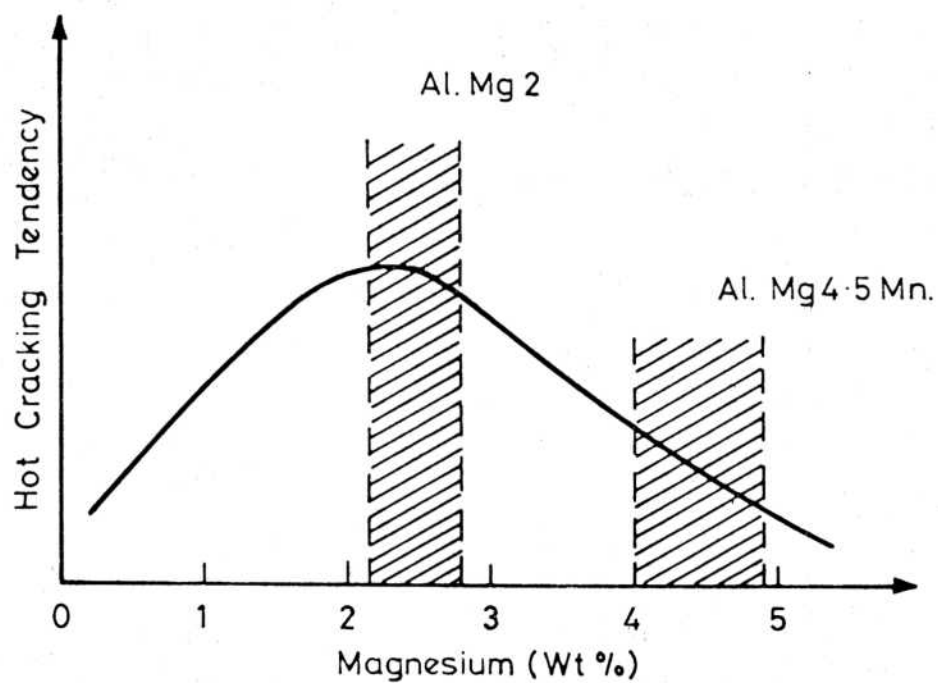


FIG. 1. HOT CRACKING SUSCEPTIBILITY OF Al-Mg Alloys.

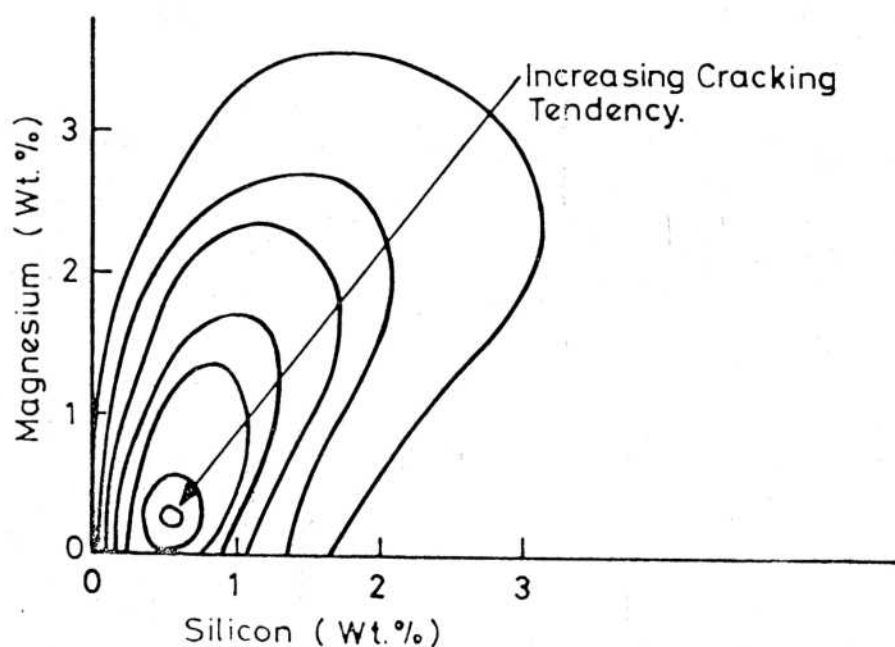


FIG. 2. HOT CRACKING SUSCEPTIBILITY OF Al-Mg-Si Alloys.

Sound Weld	●	Bead On.
	○	Bead Off.
Continuous Defect.	▲	Bead On
	△	Bead Off.
Intermittent Defect.	□	Bead Off.

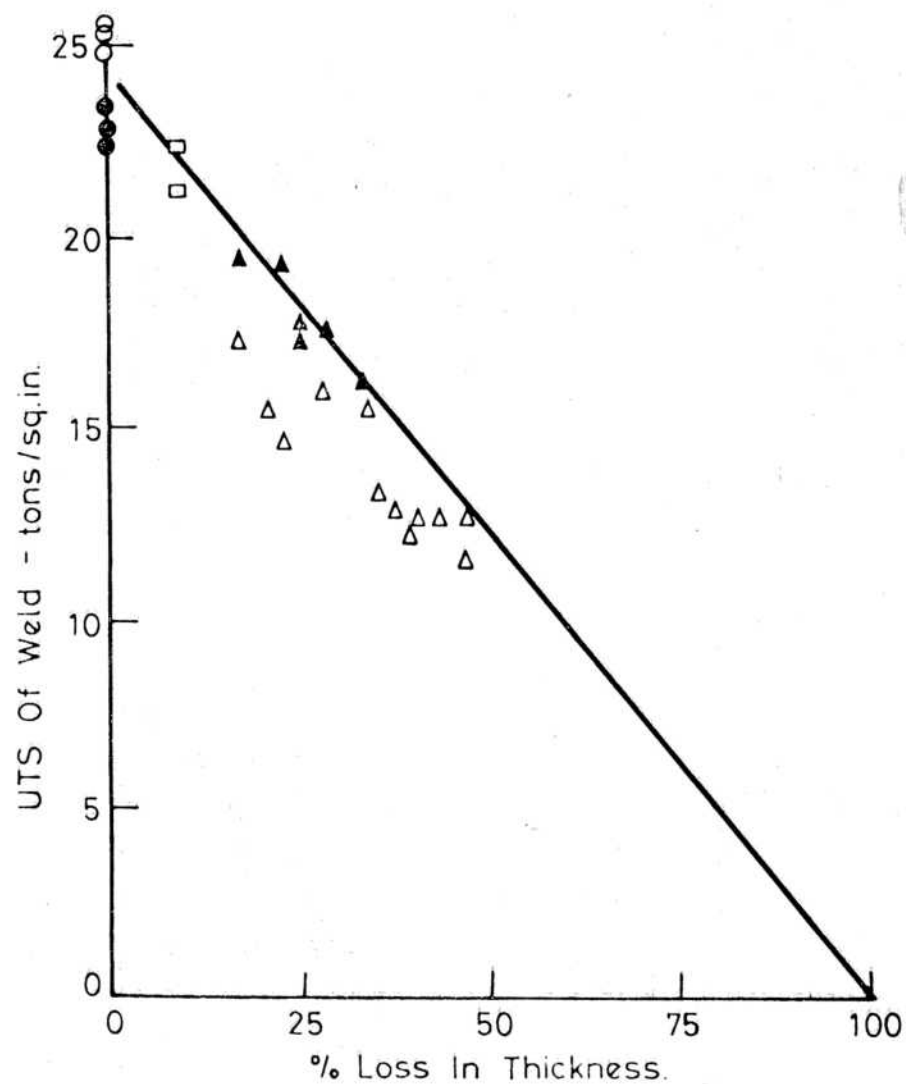


FIG. 3. VARIATION OF UTS WITH DEPTH OF DEFECT.

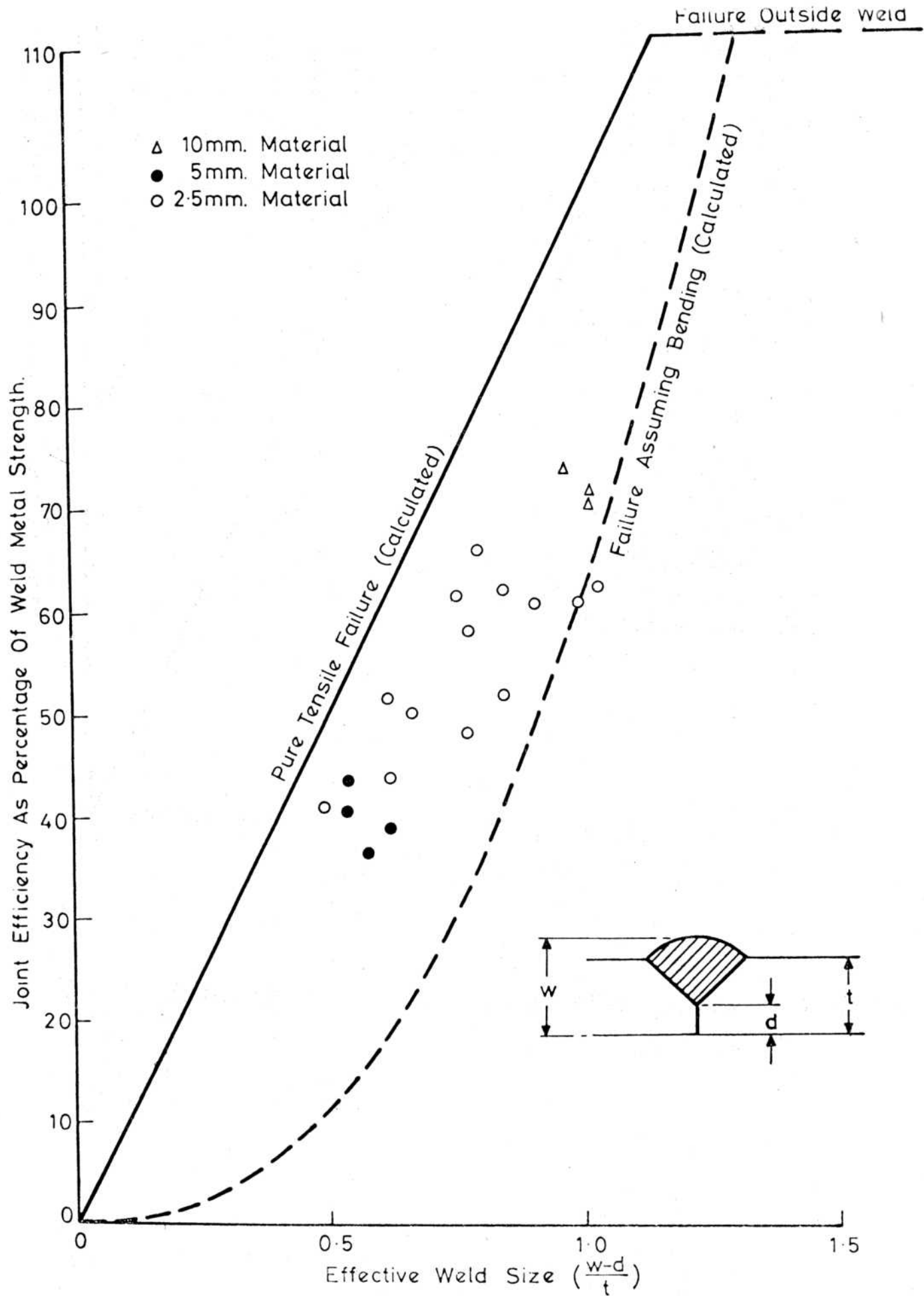


FIG. 4. STRENGTH OF SINGLE-SIDED WELDS WITH LACK OF PENETRATION.

FUSION WELDING OF MAGNESIUM

Several properties of magnesium and its alloys can affect fusion welding characteristics. The oxide film on magnesium-base materials presents problems of removal and protection similar to those with aluminium alloys whilst the high thermal conductivity and coefficient of expansion can lead to distortion, restraint cracking, and residual stresses. These problems may be minimised by attention to technique but stress relief (at 150° - 350° C) is necessary with alloys susceptible to stress corrosion, i.e. aluminium containing alloys. The wide freezing range of some alloys can lead to hot cracking in heavily restrained joints although cracking is not such a problem as with the aluminium alloys, and can generally be avoided.

For many years oxyacetylene welding, using highly active chloride-base fluxes, was the only suitable fusion welding process, until the TIG process was introduced in 1940. At present TIG welding is the most widely used process for magnesium and has almost completely replaced oxyacetylene welding, whilst MIG welding is also used in the United States.

TIG welding was originally developed for magnesium, using helium shielding and direct current, electrode positive. This technique was inefficient since 60% of the heat produced appeared at the tungsten anode, necessitating a large diameter electrode to prevent melting and giving shallow weld penetration, wide weld beads and wide heat affected zones. It was soon found that alternating current gave adequate oxide removal (with helium or argon shielding) and lower power consumption, with better penetration and narrower heat affected zone, and

a.c. with argon, helium or argon-helium mixtures has now completely succeeded the use of d.c., electrode positive. The use of d.c., electrode negative with helium shielding has been reported but Lockwood has recently stated that it is unsuitable, resulting in serious oxide contamination of the weld deposit and arc instability at currents above 170 amps.

Some arc instability is often found during a.c. TIG welding, particularly with argon shielding, and this can cause difficulty in welding. This arc instability which is characterised by a change in arc appearance (from a blue to a green colouration) and sound, has been studied by Kenyon using a high speed camera, oscilloscope and visual examination and a superimposed high frequency current component was detected when the magnesium was in the negative half cycle. The result of this instability, which has been attributed to a variation in electron emission from the magnesium, is surface roughness and occasional surface cracking (Fig.). It had earlier been found empirically that the addition of 0.15% nitrogen to the argon shield eliminated arc instability when welding Magnox AL80 and gave a smooth surface (Fig.), but had apparently no effect on other alloys. It was found to be due to the elimination of the high frequency component in the magnesium negative half cycle and attributed to a change from a plasma emission to a field emission mechanism for producing electrons. Nitrogen in larger quantities and at longer arc lengths was also found to be effective with other magnesium-base alloys and it was presumed that nitrogen was generally effective provided that the shielding atmosphere was largely argon and not magnesium vapour.

Argon shielded MIG welding (direct current, electrode positive) is feasible using spray transfer arcs for thicknesses above 0.190 in. and a short circuiting arc for thicknesses down to 0.025 in. With spray transfer in particular, burn-off rates are much higher than with aluminium (Fig.) requiring high wire feed rates and welding speeds. A problem with both types of metal transfer is that the arcs are stable only over limited current ranges; excessive currents lead to turbulence with high spatter and contamination.

Oxyacetylene welding is now obsolescent but is still used occasionally for the aluminium - $1\frac{1}{2}\%$ manganese and magnesium - 6% aluminium - 1% zinc alloys. In addition to the general problems of a distributed heat source with wide heat affected zone and considerable distortion, the chloride-based fluxes used for oxide removal are highly reactive and can lead to rapid weld corrosion if left on the weld surface. Flux entrapment in zirconium containing alloys cannot be avoided due to interconnecting microporosity and these alloys must not be gas welded. Attempts have been made to modify existing fluxes and to develop new non-corrosive fluoride-based fluxes but these have been unsuccessful, largely due to the high melting points of these non-corrosive flux mixtures.

Weld Defects

A number of defects can occur in magnesium alloy welds, the majority are due to incorrect welding techniques although some are influenced by arc instability or metallurgical factors.

The weld profile can be very difficult to control in manual welding and this can lead to poor appearance and reduced fatigue strength. A poor weld

profile is often accompanied by heavy surface oxidation; this is generally caused by insufficient gas shielding although occasionally heavy arc instability may cause turbulence in the inert gas shield. Oxide and nitride films (Fig.) may also be entrapped in the weld, reducing static and fatigue strength. These films may be caused by insufficient shielding, draughts, dirty weld edges or filler rod; welding technique is also important. The low volume heat capacity of magnesium alloys can make control of weld penetration difficult and even with trained welders it is possible to get uneven penetration or even drop-through if backing is not used. Microporosity is found in several alloys, particularly those containing zirconium or with high aluminium contents, but is not detrimental unless the micropores entrap flux and interconnect with the surface, in which case rapid corrosion can occur. Gas cavities and massive porosity will result from the presence of moisture or oily weld surfaces or filler wire and may be detrimental to weld strength. Because of arc instability, tungsten inclusions occur in TIG welding in magnesium rather more frequently than with many other materials, but they appear to have little effect on weld properties.

Hot cracking can occur in several magnesium alloys and may appear in the weld metal, weld/heat affected zone junction or weld crater. Heavy restraint markedly increases the incidence of cracking and manual welds are more sensitive than machine welds. However the problem of hot cracking should not be over-emphasized and it is less of a problem than with many aluminium alloys. The presence of zinc reduces the resistance to hot cracking in both the aluminium and zirconium containing alloys. Suggested reasons for this reduced resistance are

the increased solidification range and the presence of a grain boundary zinc-rich phase (possibly $MgZn$). The 6% aluminium 1% zinc alloy has good weldability but increase in either zinc or aluminium content markedly increases the crack susceptibility. Zirconium gives a fine grain size and confers some resistance to cracking but this tends to be overshadowed by the detrimental effect of zinc in the magnesium-zirconium-zinc alloys. Thorium or the rare earth elements tend to offset the effect of zinc by producing a crack resistant grain boundary phase and thorium or rare earth additions are matched to the zinc content in many alloys. Since the high temperature magnesium alloys invariably contain thorium or mischmetal they tend to be fully weldable. Minor additions of calcium between 0.04 and 0.2% greatly increase the susceptibility to junction cracking; West attributes the effect of calcium to an increase in the solidification range but Emley considers it is due to low melting point grain boundary films. Early versions of the Magnox AL80 nuclear cladding alloy contained calcium, which was beneficial for oxidation behaviour, but its use had to be discontinued due to severe hot cracking in the end cap sealing welds.

A further form of cracking is stress corrosion cracking which is found in aluminium containing magnesium alloys. It can be prevented by stress relief after welding.

Filler Metal

Filler materials are generally of identical or similar compositions to the parent material. More highly alloyed filler compositions with lower melting points and narrower solidification ranges are sometimes beneficial in reducing

weld cracking and may also give higher strengths. Particular care must be taken to avoid welding zirconium containing alloys with fillers containing aluminium or manganese, since this results in insoluble intermetallics in the weld metal.

Weld deposits generally have a fine grain size (Fig.) and show reasonable strength and ductility. With TIG welds joint efficiencies of 90% are possible in all wrought alloys except magnesium - 1½% manganese and failure generally occurs in the heat affected zone where grain growth has occurred. With the magnesium - 1½% manganese alloy joint efficiencies are reduced to 60% for both TIG and gas welds due to the very coarse grain size, although peening can improve the efficiency to 75% or more. Typical joint properties are shown in Table I.

Postweld heat treatment will improve properties only rarely, but it is generally employed for stress relief and to correct distortion. Postheat treatment is essential for aluminium containing alloys to prevent stress corrosion. With castings, postweld stress relief in the range 250° - 300°C (according to the particular alloy) is mandatory, whilst preheat in the same temperature range is also generally employed.

Copper is a face centred cubic metal, melting at 1083°C , with a pleasant lustre and exceptionally good ductility and malleability.

The principal properties of copper are shown in Table I; the very high electrical and thermal conductivities are of particular interest when assessing weldability. The high electrical conductivity is likely to make resistance welding difficult whilst the high thermal conductivity could cause difficulty in fusion welding by allowing rapid heat dissipation. The coefficient of thermal expansion lies midway between those of steel and aluminium, and combined with the high thermal conductivity can result in distortion and high shrinkage stresses.

Copper and several of its alloys have been known and used since antiquity for both decorative work and more practical applications. At present copper alloys find widespread use in many fields of engineering although its annual tonnage consumption is comparatively small. The uses of copper and its alloys depend largely upon its electrical and thermal conductivities, its corrosion resistance and the bearing properties of some of its alloys. Thus copper or its alloys find extensive use in the fields of electrical, chemical and marine engineering, whilst the attractive appearance of the metal and many of its alloys give a widespread use for coins, medals and general decorative work.

Copper

Copper is used quite extensively in its commercially pure forms or with minor alloying additions.

Tough pitch high conductivity copper contains up to 0.1% oxygen in the form of cuprous oxide, Cu_2O , which forms a copper-cuprous oxide eutectic melting at 1065°C . In wrought form this eutectic is widely dispersed and has little effect on strength but in cast form, or weld metal, the eutectic forms grain boundary films that result in brittleness and hot shortness. In the presence of hydrogen the cuprous oxide is reduced and the resultant steam gives rise to porosity and voids("gassing"). However, the cuprous oxide has virtually no effect on electrical or thermal conductivity and tough pitch copper finds a wide use in the electrical field.

Where "gassing" is likely to be a problem deoxidized or oxygen free copper must be used. Oxygen free high conductivity copper retains the high conductivity of tough pitch copper, but is expensive, and where electrical conductivity is relatively unimportant it is more usual to add a deoxidizer. Phosphorus (in amounts up to 0.03%) is the most commonly used deoxidizing agent, it removes the problem of "gassing" but reduces the electrical conductivity by up to 40%. Arsenical copper (0.3 - 0.5% arsenic)

has slightly higher strength and resistance to scaling but lowers the conductivity even further. Many elements are added to copper in small amounts in order to improve specific properties. Thus silver up to 0.1% and cadmium up to 1% (Cadmium Copper) increase the softening temperature, strength and toughness without decreasing electrical conductivity.

For high strength, chromium (0.4 - 0.8%) or beryllium (1.5 - 2.0%) additions are used, which allow precipitation hardening. In the hardened condition chromium copper has a strength of 35 tsi combined with an electrical conductivity of 80% IACS and beryllium copper strengths of 75 - 100 tsi and conductivity of 24% IACS. These alloys are widely used where high strength combined with some electrical conductivity is required, including spot welding electrodes.

The Alloys of Copper brasses

The alloys of copper and zinc cover a wide field and their applications are extensive. Up to 38% zinc is soluble in copper giving a face centred cubic α solid solution. Above this zinc content a β phase is precipitated from a peritectic reaction, the structure being fully β - phase above 45% zinc. In fact only the α and $\alpha + \beta$ brasses are of commercial interest, containing up to 40% zinc, sometimes with minor additions of aluminium, iron, manganese, nickel, silicon, tin or lead. The α brasses are primarily cold working alloys and the $\alpha + \beta$ brasses hot working and casting alloys.

Additions of 2 - 5% Zinc produce cap copper which is really a zinc deoxidised copper, whilst alloys containing between 5% and 20% zinc are known as gilding metals because of their pleasant golden shades, varying from red to yellow. Both these series of alloys, which have extremely good malleability, are largely used for decorative purposes and the most important α alloy commercially is cartridge brass, containing 70% copper and 30% zinc. This brass is the most malleable of all the α - brasses and is suitable for all cold working operations including deep drawing. Variations are alloys containing 35 - 37% zinc, which are cheaper, and admiralty brass (1% tin) and aluminium brass (2% aluminium) which have improved corrosion resistance. Additions of nickel and aluminium to α brass allows precipitation hardening to be effected, giving an increase in tensile strength from 23 to 36 tsi.

The duplex alloys are based upon the 60% copper 40% zinc composition (yellow metal) and are basically hot working and casting alloys. At room temperature malleability is quite low. The high tensile brasses, or manganese bronzes are duplex alloys containing additions of aluminium, iron, manganese, nickel and tin up to a maximum of 7%. These alloys, which are usually used in cast or extruded form, have tensile strengths of 50 - 45tsi

compared to 24 tsi for yellow metal.

Lead is often added to brasses in amounts up to 3.5% in order to improve machinability. Since lead is immiscible in all copper base alloys, the lead bearing alloys are hot short above the melting point of lead (327°C).

Bronzes and Gunmetals

Bronzes are alloys of copper and tin but other alloying elements i.e. zinc, phosphorus, nickel and lead, are frequently present. Alloys of copper, tin and zinc are known as gunmetals. Additions of phosphorus up to 0.4% - the phosphor bronzes, are also popular, since considerable improvement in mechanical properties can be obtained.

Tin is soluble in copper up to 16% at 550°C and although the solubility falls rapidly with temperature, α solid solutions can be retained readily. The cold working alloys are all α phase alloys and contain less than 10% tin; these alloys are not normally hot worked since the plastic range is very narrow and the hot short range fairly wide. The cast bronzes and gunmetals are all complex structures containing the brittle tin-rich ϵ phase. The composition and properties of several wrought and cast bronzes are given in Table III.

The applications of the bronzes and gunmetals are quite limited: wrought alloys are used generally for small components such as gauge and springs, whereas the cast bronzes are used mainly for bearings and patterns.

Silicon Bronzes

The addition of silicon to copper improves corrosion resistance, tensile strength (23 tsi) and also weldability. The wrought alloys, containing up to 3% silicon and 1% manganese consist solely of the α phase but the casting alloys, which contain 4 - 5% silicon together with zinc, iron and manganese is a duplex structure of α and γ phases. Silicon bronze is used in the chemical engineering field for storage tanks, piping, pickling crates, etc.

Copper - Nickel Alloys

Copper and nickel have mutual solid solubility and form a useful range of hot and cold working alloys. Compositions range from 90% copper, 10% nickel, which is really a toughened copper up to 70% copper, 30% nickel. Increasing nickel content gives good resistance to sea water and the cupro-nickels are used extensively for condenser tubes. Addition of iron to the 5 - 10% nickel alloys also gives an alloy with good resistance to attack by aerated sea water.

Gold Silvers

The nickel silvers, containing 55 - 63% copper, 10 - 30% nickel, balance zinc, are single phase alloys used mainly for decorative purposes.

Aluminium Bronzes

These alloys are based on copper and aluminium but frequently contain iron, nickel and manganese; they are characterised by good oxidation and corrosion resistance imparted by a self healing impervious alumina surface film. The solubility of aluminium in copper extends up to 9.4% and all the cold working alloys are in this range. With greater amounts of aluminium, a duplex $\alpha + \beta$ alloy occurs at higher temperatures, although the β phase can break down to an $\alpha + \gamma_2$ mixture, in which the γ_2 phase is considerably more brittle than the β phase. Fortunately the transformation is sluggish and air cooling will normally result in retention of the β phase.

The single phase alloys contain 4 - 8% aluminium and frequently up to 4% iron, nickel and manganese. These alloys, which can be worked cold or hot worked in the range 790 - 900°C., have a tensile strength of 20 - 30tsi in the annealed condition, but this can be increased to 50tsi by cold work. The alloys are suitable for use up to about 300°C in the work hardened condition and 400°C in the annealed condition. A hot short range occurs between 500° and 700°C. The single phase alloys are largely used for decorative purposes, although in plate and tube form they have been used for heat exchangers.

The duplex alloys contain 8 - 11% aluminium and frequently iron, manganese and nickel. Iron and manganese improve the mechanical properties but have little effect on the structure, but nickel can also confer age hardening characteristics in certain composition ranges. The hot working duplex alloys normally contain up to 5% alloying additions other than aluminium, and, since there is no hot short range, can be worked at all temperatures up to 950°C. Some cold drawing of hot rolled and extruded bars is carried out but cold working is not generally recommended. The casting alloys cover a wider composition range than the hot working alloys. Iron is frequently added to prevent the transformation of the β phase at slow cooling rates and manganese and nickel also refine the structure and give improved mechanical properties. Up to 2% lead may be added to the duplex alloys to improve the machining properties of the hot working alloys and the bearing properties of the casting alloys. As with other copper alloys, lead is immiscible in aluminium bronze and leads to hot shortness.

The duplex alloys are of greater industrial importance than the single phase alloys and are extensively used in marine, chemical and general engineering. However many industrial alloys have wide composition ranges and can have micro-structures that are mainly α phase or $\alpha + \beta$ phases, plus additional iron or nickel rich precipitates. Typical applications are pump casings, valve fittings, propellers, gears and shafts.

Fusion Welding

The fusion welding of copper and its alloys can be affected by a number of properties common to many of these alloys i.e.

thermal conductivity
thermal expansion
oxidation resistance
hot shortness
porosity.

The high thermal conductivity of many of the alloys can lead to rapid dissipation of heat and necessitates a high rate of heat input. Combined with the comparatively high coefficient of thermal expansion of many alloys, the high conductivity can lead to severe distortion and shrinkage stresses, and often welder discomfort due to the proximity of large areas of hot metal. The practical solution of such problems may cover

- (i) the use of an intense heat source with a high welding speed to limit the total heat input
- (ii) slow preheating of the complete workpiece (by the use of oxyacetylene torches or coal gas burners)
- (iii) avoidance of restraint
- (iv) slow, uniform cooling after welding.

The high conductivity can also be advantageous where rapid cooling is required to reduce segregation or hot shortness.

Oxide films form rapidly at elevated temperatures in copper and many of its alloys. These films can be removed by use of deoxidants (silicon, zinc, phosphorus) in the filler wire, or a borax flux. Problems are most likely to be encountered in aluminium or beryllium containing alloys which form tenacious oxide films.

Hot short temperature ranges occur in several alloys in which the ductility is extremely low, and in these ranges cracking can occur at low stress levels. Alloys exhibiting hot short ranges include some grades of tough pitch arsenical copper, lead bearing coppers and alloys, tin bronzes and single phase aluminium bronzes. In some cases the hot short range is immediately sub-solidus, for example the narrow hot short range in silicon bronzes ($950^{\circ} - 800^{\circ}\text{C}$) but in other cases it can be much lower.

Occurrence of porosity is erratic, and rarely a problem with copper alloys. It is affected by filler wire composition, welding speed and the type and flow rate of shielding gas. In general well deoxidized fillers are advisable, especially where nitrogen is used as the shielding gas.

The restrictions covered above are not a severe limitation on the welding of copper and its alloys and most conventional processes are affected to some extent. The processes normally encountered are:

shielded metal arc welding
TIG and MIG welding using argon, nitrogen
or argon-nitrogen mixtures
oxyacetylene welding
carbon arc welding with or without a flux.

Copper

Tough pitch copper is difficult to weld because of steam porosity and "gassing", and welds tend to suffer from brittleness, porosity and hot tears. Oxygen contents as low as 0.05% can result in porosity although 0.10 - 0.15% oxygen is necessary to give brittle grain boundary eutectics. No process is ideal but joint efficiencies of up to 80% can be obtained with the argon shielded processes, which give narrow heat affected zones. Filler wires should be heavily loaded with deoxidants. Oxygen free high conductivity copper is readily welded with the inert gas shielded processes, but if the weld needs to retain good conductivity the filler wires should contain only boron, silver, zinc or zirconium as deoxidants.

The deoxidized coppers can be welded quite readily, giving joint efficiencies of 80 - 100% although arsenic containing filler wires should be avoided due to the danger of hot shortness. The oxyacetylene process is still used quite widely although the inert gas shielded arc processes are preferable. Because of the need for high heat input the tungsten arc process should be restricted to sheet.

Hot shortness problems arise with lead containing coppers because of the low melting point of lead (327°C) and alloys containing more than 0.1% lead are virtually unweldable. The oxide film formed on beryllium coppers can also give difficulty although the inert gas shielded metal arc processes can be used.

Helium, argon, nitrogen or their mixtures can be used for the inert gas shielded processes, although helium is not used in the United Kingdom because of its high cost. Nitrogen is attractive because of its lower cost and the greater heat intensity of the arc. For tungsten arc welding nitrogen shielding requires the use of thoriated tungsten electrodes and special filler wires, whilst with metal arc welding transfer characteristics are poor and there is a danger of oxide inclusions. Recent work on the inert gas metal arc process has indicated that argon-nitrogen mixtures can give the advantages of a more intense heat source and lower gas costs together with good metal transfer, and it has been shown that such mixtures enable the elimination or reduction of preheat.

Brasses

In general the brasses, other than leaded brasses, are readily welded by all processes, although shielded metal arc welding utilizes 5% tin bronze electrodes. The inert gas shielded processes also use tin, silicon

or aluminium bronze filler wires. Preheat of 200 - 400°C is recommended, particularly with castings. Joint efficiencies of 100% are attainable.

The most serious welding problem is zinc fumes which obscure the weld and are unpleasant, and even harmful in large amounts to the welder. Respirators may be worn, and in oxyacetylene welding the use of an oxidising flame and suitable fluxes is recommended. The thermal conductivity of the brasses is much less than that of copper and this reduces distortion and residual stress problems, although the welding of cold worked material can give serious distortion. For use in corrosive atmospheres, particularly those containing ammonia, stress relief at 250° - 270°C. should be carried out, to avoid "season-cracking".

Tin bronzes

Fortunately the welding of these alloys is not often required, since the wide freezing range (which increases with tin content) leads to hot shortness and shrinkage voids. Moderate preheat (200 - 300°C) is desirable to avoid shrinkage stresses, and rapid cooling will reduce the danger of hot cracking and shrinkage porosity. For these reasons oxyacetylene welding is not recommended; where it is used an oxidising flame should be employed to avoid steam or hydrogen porosity. The tungsten arc welding is used fairly widely but the shielded metal arc process using electrodes containing 5 - 9% tin and up to 0.5% phosphorus is most popular. A stringer bead multi-run technique gives some refinement of the "as cast" weld metal structure and hot peening below the hot short range also gives grain refinement and reduces distortion.

Silicon Bronze

The silicon bronzes are the easiest of the copper alloys to weld because the silicon acts as a deoxidiser and the thermal conductivity is low. A hot short range does occur, however, just below the solidus, and care should be taken to avoid overheating. Preheat should not be used and interpass temperatures should be limited to 100°C., and for this reason a stringer bead welding technique should be used.

In the as welded condition the joint strength should approach that of the parent material i.e. 20tsi, whilst cold peening can give a further improvement.

All the processes listed above can be used for welding silicon bronze although the inert gas shielded processes are preferred.

Copper-Nickel Alloys

The only problem with these alloys is sulphur embrittlement which can be controlled by magnesium and manganese additions. The tungsten inert gas and oxyacetylene processes are the most used in this country but metal inert gas welding also gives good results.

Nickel Silvers

The nickel silvers are generally oxyacetylene welded, using a high zinc bronze or high zinc nickel silver for colour matching. A suitable flux is essential to dissolve any nickel oxide and avoid porosity.

Aluminium Bronze

With these alloys, welding is largely used for repair work, although some actual welded fabrication is carried out for reaction vessels to hold saline and other corrosive liquors at moderately elevated temperatures. The shielded metal arc and metal arc inert gas processes are the most widely used, but tungsten arc (with alternating current) and oxyacetylene welding is quite satisfactory. The metal arc inert gas process gives the best results with positional welding. The oxide film can be troublesome and careful scratch brushing and degreasing should be carried out immediately before welding. Electrode coatings and gas welding fluxes should contain high proportions of fluorides.

The single phase alloys are more difficult to weld than the duplex alloys because of a hot short range between 700° and 450°C . Welds should be cooled rapidly through this temperature range to avoid hot cracking, and single or double run deposits are preferable to multi-run welds. Hydrogen porosity can also be a problem with single phase aluminium bronzes. Matching weld metal compositions are used satisfactorily in many cases, but where hot cracking and porosity is encountered in the weld, a duplex weld deposit may be used. However, in corrosive media there is a danger of preferential attack on the β phase in the weld metal, known as dealuminification. In these cases it is advisable to seal the weld with capping runs giving a single phase deposit.

The duplex aluminium bronzes can be welded comparatively easily provided that attention is paid to cleanliness. The hot short range is very narrow in these alloys, so that hot cracking is not a problem, whilst hydrogen has sufficient solubility in the β phase to eliminate the danger of porosity. The distribution of β phase is also important and in some alloys a continuous grain boundary network of β phase can lead to hot cracking and grain boundary corrosion.

Brazing and Soldering

Copper and most of its alloys can be readily soldered using tin-lead solders and proprietary fluxes. Intergranular penetration and cracking can occur in high zinc bronzes in the cold worked condition, however, and prior stress relief at $250^{\circ} - 270^{\circ}\text{C}$. is advisable. Difficulty is also experienced with aluminium or beryllium containing alloys due to the formation of tenacious oxide films. Highly active fluxes are necessary and these can lead to corrosion problems if inadequate cleaning is carried out

after soldering.

Brazing, using a straight brass or a silver-copper-zinc alloy (silver solder), is an extremely effective method of joining copper and most of its alloys. Torch and furnace brazing are equally suitable although a flux is necessary in all cases. In the case of the oxygen containing tough pitch coppers brazing is more suitable than any welding process, since gassing is avoided. Similarly, brazing is the most suitable process for the leaded coppers and copper alloys. Aluminium containing alloys can be brazed although an active flux is necessary; the difficulty increases with aluminium content and special techniques are necessary with aluminium bronzes containing 8 - 12% aluminium. Intergranular penetration can occur with several alloys in the cold worked condition; these alloys include the nickel silvers, leaded brasses, silicon bronzes and cupro-nickels, and stress relief should be carried out before brazing.

Resistance Welding

The factors influencing the ease of resistance welding are the electrical and thermal conductivities. The problem is to get sufficient heat generated at the interface in a short time, despite the low interfacial resistance and high rate of heat dissipation and, in general, weldability varies inversely with the electrical and thermal conductivities. The problem is most acute with seam welding and least acute with flash butt welding.

Seam welding is impractical for alloys with an electrical conductivity greater than 20% IACS, and this eliminates all the coppers and most brasses and tin bronzes. The most suitable alloys for seam welding are the cupro-nickels, silicon bronzes, nickel silvers and aluminium bronzes. High quality spot welds can be produced in alloys with conductivities of 30% IACS or less, although less consistent welds can be made with conductivities of 60% IACS. Again the best results are obtained with silicon bronzes and cupro-nickels, followed by aluminium bronze. Leaded coppers and alloys are virtually impossible to weld because of hot cracking in the weld nugget and adjacent parent metal.

Conductivity problems are not so apparent in resistance butt and flash butt welding and these processes can be used for most coppers and alloys. Flash butt welding is particularly useful since it can be used for welding alloys with lead contents up to 1.0%.

Titanium has only become a commercial material in the last few years and even now, its metallurgy is imperfectly understood. Despite its high cost it is an important material in the fields of aeronautical and chemical engineering and its use is likely to increase. From the welding viewpoint it has a further significance: it was the first of the reactive metals to reach commercial importance and many of the problems solved in welding this metal can be applied to the other 'new' metals, for example zirconium, hafnium and niobium.

Properties.¹

Titanium is a comparatively light metal (50% denser than aluminium) with a melting point of 1668°C . Its thermal and electrical conductivities are low, as is its coefficient of thermal expansion. At atmospheric temperatures titanium has excellent corrosion resistance to all except strong reducing agents but at elevated temperatures (above 600°C) the metal is extremely reactive and will absorb oxygen, nitrogen, hydrogen and carbon. The major effect of hydrogen is to reduce the notch toughness, whereas oxygen and nitrogen, and to a lesser extent, carbon, strengthen and embrittle the metal. Thus pure titanium is relatively soft and ductile but with increasing impurity content it becomes much stronger with a strength to weight ratio that compares very favourably with other materials.

Below 882.5°C , titanium has a close packed hexagonal structure with a $\frac{c}{a}$ value of 1.587, which is slightly less than the ideal value for close packing (1.633). Above this temperature however, a body centred cubic form is stable. The lower temperature structure is referred to as α - titanium and the high temperature one as β - titanium. The high temperature structure cannot be retained at lower temperatures by quenching, although high cooling rates result in an acicular or basket weave structure which is believed to result from a martensitic type of transformation. This structure remains soft and ductile and cannot be heat treated to give improved properties.

Alloying of Titanium

Alloying additions to the metal may act as either α or β stabilizers extending the temperature range over which either phase is stable. Most alloying atoms form substitutional solid solutions but oxygen, nitrogen, carbon and hydrogen, form interstitial solutions. The majority of elements tend to stabilize the β phase.

In low alloys rapid cooling will result in the formation of martensitic α , but with higher alloy contents it is possible to retain some or all of the β phase on cooling. The α alloys cannot be heat treated but alloys retaining some β phase are amenable to heat treatment.

The commercial alloys of titanium fall into four groups:-

- (i) Commercially pure titanium, with controlled impurity contents.
- (ii) α stabilized non-heat treatable alloys.
- (iii) β stabilized alloys giving a heat treatable $\alpha + \beta$ structure.
- (iv) β stabilized alloys giving a metastable β structure that is amenable to heat treatment.

The commercially pure titaniums are still the most widely used because of their superior ductility and malleability. However, extensive alloy development is still being carried out and a range of α , $\alpha - \beta$ and β alloys is available.

The mechanical properties of a range of titanium alloys are given in Table III. Commercially pure titanium is reasonably ductile and insensitive to notches, whereas the α alloys have increased strength at the expense of ductility, and are rather more notch sensitive. The $\alpha - \beta$ and β alloys in the heat treated condition exhibit the highest strengths but do tend to exhibit a greater notch sensitivity. Notch sensitivity is particularly influenced by interstitial impurities, with hydrogen being the worst offender, and a hydrogen level of 200 ppm is sufficient to reduce notch toughness.

The strength:weight ratios of titanium alloys are extremely good at low and moderate temperatures but unfortunately, these ratios are not maintained above 500°C. In addition, rapid oxidation occurs above this temperature and alloying has also failed to solve this problem.

Alpha Alloys.

These alloys which were the first to be developed are basically single phase, although small amounts of precipitates, due to impurities, can generally be observed in the microstructure. The α alloys give an improvement in strength compared to the commercially pure material, at the cost of some ductility. Further hardening can only be attained by cold working since quenching from the β phase results in soft acicular α with no tendency to delayed precipitation.

The original alloys gave only moderate strength increases over the commercially pure metal and it appeared that alloy contents were strictly limited by loss in workability. More recently however, 'super - α ' alloys have been developed with increased alloy contents, giving markedly higher strengths and acceptable ductility (Table III). Some of these alloys contain very small traces of β phase, but the amount is generally insufficient to allow any marked precipitation hardening.

Alpha-Beta Alloys

Alloys containing moderate amounts of β stabilizing elements

may exhibit duplex $\alpha + \beta$ structures.

The β phase is metastable and ageing which can occur at fairly low temperatures may result in embrittlement. However, overageing of the β phase will often give alloys with good strength and ductility. The ageing treatment results either in finely dispersed α or in the non-equilibrium ω phase.

Commercial $\alpha - \beta$ alloys fall into two distinct groups:

- (1) those containing only β stabilizing transition elements
- (2) those containing an α stabilizer i.e., aluminium, in addition to β stabilizers.

Into the first group fall the titanium-manganese alloys, which appear to be little used nowadays, and more recently developed alloys containing small quantities of several elements. Alloys in the second group normally contain one or more β stabilizers plus aluminium.

The transition elements are virtually insoluble in α titanium so that the α phase will be unalloyed and soft in alloys containing only these elements. The retained β phase will be metastable and quite low temperatures may be sufficient to cause ageing and embrittlement. It is generally necessary, therefore, to ensure that the structure is overaged if adequate ductility is to be guaranteed. However, the thermal history of the alloys has a marked effect on properties.

The aluminium addition to the second group of $\alpha - \beta$ alloys, results in a marked strengthening of the α phase. Aluminium is also soluble in the β phase and increases the rate of overageing so that a ductile overaged structure occurs even on quite rapid cooling.

Beta Alloys

Larger amounts of β stabilizing elements allow metastable β phase to be retained on rapid cooling to ambient temperatures, which permits subsequent precipitation hardening treatment. These alloys have the great advantage that in β form they are soft and readily workable whilst subsequent precipitation hardening can give tensile strengths of 80 tsi and above.

The ageing process will depend upon the temperatures: at high temperatures the α phase will precipitate first, followed by the intermetallic compound, but at lower temperatures an intermediate non-equilibrium phase, designated ω is formed first.

The fully β stabilized alloys have been introduced in the last few years and only one or two alloys are commercially available. However, the advantages of good formability in the solution treated condition and high strength when heat treated may be expected to lead to considerable usage, and these alloys may well replace the $\alpha - \beta$ titanium alloys.

Fusion Welding.²

The basic problem in the fusion welding of titanium and its alloys is the problem of eliminating atmospheric contamination, as it is with all the reactive metals. This is not to state that no other problems exist, but in the commercially pure material and α alloys at least, other problems are merely secondary. With the heat treatable alloys, the thermal cycle that is unavoidably connected with fusion welding can result in the formation of extremely brittle structures, and most α - β alloys must be considered unweldable. The other properties of titanium tend to favour welding and the low conductivity and thermal expansion of the metal and its alloys cause few problems of extensive heat affected zones or distortion.

The Gas Shielding Problem.³

The high reactivity of titanium and its alloys restrict fusion welding methods to the inert gas shielded processes. The majority of fusion welding utilizes the TIG process, although MIG welding is perfectly feasible and is used for thicker material. However, even the conventional TIG process may not give adequate protection against atmospheric contamination and the use of auxiliary gas shielding or even a closed welding chamber may be necessary.

The effect of atmospheric contamination is given in Fig. 3. The most likely contaminants are oxygen and nitrogen, picked up due to air entrainment, but hydrogen may be absorbed if moisture is present. The maximum limits of tolerance for oxygen and nitrogen in weld metal have been estimated⁴ at 0.3% and 0.15% respectively, whilst hydrogen should not exceed 150 ppm. The carbon level in parent material is restricted to .07%, and further pick up in the weld is unlikely unless surface contamination has occurred; levels likely to cause embrittlement have been estimated at 0.1 - 0.28%, the actual figure being dependent upon other impurities and alloying constituents.

Surface discoloration may give some indication of contamination. Thus, under the best conditions of gas shielding the weld bead and heat affected zone will be bright whereas slight contamination will result in a light straw-coloured appearance. Increasing contamination results in the appearance changing from straw to blue. The absence of surface discoloration does not necessarily mean freedom from contamination: if a second melt run is made over a discoloured weld bead it may result in a silvery bead, but bend tests will show very low ductility. In this case, the surface contamination has been absorbed into the metal.

Slight surface contamination (light straw colour) can generally be tolerated as the fall in ductility is small, but with higher levels of

contamination the joint and heat affected zone must be cut out, and the component rewelded. Where only surface contamination has occurred, removal of the surface layers (mechanically or by pickling) will be sufficient.

With these limitations it is obvious that care must be taken to ensure the cleanliness of parent material and filler wire, and to prevent entrainment of air into the inert gas shroud. Material cleanliness can be readily achieved by careful pre-weld preparation - scratch brushing with a stainless steel brush and degreasing, but adequate inert gas protection during welding may be more difficult. With parent sheet levels of oxygen and nitrogen that may reach 0.2% and 0.05%, and maximum permissible levels of 0.3% and 0.15% in the weld metal, it is clear that practically no contamination is permissible.

Protection against contamination of the weld and heat affected zone may be attained by the use of welding chambers filled with an inert gas or by open air welding using supplementary gas shielding.

The use of a welding chamber offers the greatest opportunity of weld protection since air can be completely excluded from the work piece. The disadvantages of such a chamber are that the chamber size will generally restrict its use to small components and the amount of inert gas required. Types of chamber may vary from a rigid cabinet with facilities for evacuation, to flexible chambers and plastic bags which can be collapsed to remove air prior to filling with the inert gas.

The highest weld quality is obtained with chambers that are evacuated to at least 200 microns (0.2mm mercury) prior to argon or helium filling. The main problem with this type of chamber is to keep vacuum and gas lines to a minimum in order to reduce the danger of air pockets or leaks. With larger chambers where it may be impossible to evacuate to 200 microns or prevent minor leaks, double purging may be used, i.e., evacuating, filling with inert gas and re-evacuating, to reduce the residual air content before finally filling with gas. Final traces of air may be gettered by making trial weld runs on scrap titanium. With these types of chamber a standard TIG torch may be unnecessary and a simple electrode holder with a tungsten electrode, can be used, giving improved accessibility and visibility for the operator. However, if leakage of air into the chamber can occur, it is better to use a standard torch with a separate gas flow to give extra protection.

With simpler chambers, air may be displaced by collapsing the chamber or simply purging with the inert gas. This method cannot be generally recommended since it is virtually impossible to reduce air levels to an acceptable value. Under some conditions contamination may be worse than with normal open air welding. However, for large fabrications or site

work, these methods may prove adequate provided that TIG torches are used with separate inert gas supply.

When open air welding is used, attention must be paid to the weld pool during melting, solidification and subsequent cooling, and to those parts of the heat affected zone that are raised above 550°C . Heavy jigging is generally used to reduce the size of the heat affected zone to a minimum and this jigging is often designed to form a baffle and channel the shielding gas along the joint line. The back of the weld is generally protected by a subsidiary inert gas supply, although in some cases adequate protection can be achieved by clamping the parent metal to an ungrooved copper backing strip. The arc is protected by a normal shroud and the cooling weld and heat affected zone by a trailing shroud with its own inert gas supply. This type of system, i.e. use of a trailing shroud, works very well with automatic welding but is cumbersome for a manual operator.

For manual welding it is still quite common to rely on the skill of the operator to shield the weld zone, without using a trailing shield. This method cannot be generally recommended although it often gives good results with a skilled welder.

Borland and Hull⁵ have designed a nozzle suitable for manual TIG welding of sheet up to 0.1 inch thickness. This nozzle is designed to give main shielding to the arc together with outer peripheral shielding and shielding to the rear of the arc. This nozzle was designed from a consideration of:-

- (i) the length of argon coverage required to protect the weld zone during cooling, and
- (ii) the degree of coverage provided by gas flow from a nozzle.

In the first instance the length of coverage can be calculated from the welding speed and the time to cool below, say, 550°C , whilst in the second it is necessary to consider the onset of turbulence (and hence air entrainment) in the inert gas flow from the nozzle.

Commercially Pure Titanium.

Once the gas shielding problem has been solved the fusion welding of the commercially pure metal presents few difficulties. The harder grades of commercial titanium are not generally recommended for welding, however, since oxygen and nitrogen contents are very near the limit for acceptable ductility, and allow no margin for even slight contamination during welding. Transverse tensile tests exhibit failures well away from the weld zone but longitudinal bend tests reveal a lower ductility in the heat and weld affected zone than in the parent sheet, with failure occurring in the heat affected zone. A slight increase in tensile strength with welding speed may be noted.

Commercial Alpha Alloys.

Commercially the welding of the α alloys has been confined to the 5% aluminium 2.5% tin alloy, although all alloys are reported to be weldable. The weld properties match parent sheet properties extremely well except for a slight drop in ductility.³ Porosity has been a problem with these alloys, being particularly associated with poorly cleaned plate and roughly prepared edges; TIG welds have shown a greater tendency towards porosity than MIG welds. The cause of the porosity is not certain although argon entrapment between rough edges of a closing joint gap has been suggested. It is certain that porosity can be reduced by the use of smooth, milled sheet edges, by increased welding speeds and, on occasion, by remelting.

Cracking is rarely a problem with the 5% aluminium 2.5% tin alloy although cases of micro-cracking, hot cracking and delayed cracking, have been reported.⁷ Micro-cracking in TIG welds in 0.125 in. - 0.180 in. sheet has been attributed to high restraint and microembrittlement not detectable by normal testing techniques; rather surprisingly, MIG welds were not susceptible to cracking. The other forms of cracking have been attributed³ to contamination or excessive impurity contents: hot cracking is considered to be caused by high oxygen and nitrogen contents (causing low hot ductility) and delayed cracking to a high hydrogen content (over 350 ppm). Hot cracking and delayed cracking were of particular concern when titanium technology was not so advanced as at present and impurity contents are now unlikely to give these problems.

The super-alpha titanium alloys are also reported⁸ to exhibit good weldability. Ambient temperature properties of welded sheet compare well with parent material properties, although ductility is reduced slightly. In the as welded condition the tensile and yield strengths are higher than the parent material but ductility is generally lower. Response to heat treatment is a little variable but unless ageing occurs it is generally possible to improve ductility at the expense of strength. It appears impossible to fully restore the tensile elongation of welded joints but weld bend ductility can be matched to that of the parent sheet. Good creep stability has been observed at 420°C but some deterioration in tensile elongation at 540°C, particularly if some precipitation occurs. Cryogenic properties of the alloys are also satisfactory.

Alpha-Beta Alloys.

The only α - β alloy that has been welded commercially is the 6% aluminium, 4% vanadium alloy. All other α - β alloys tend to be seriously embrittled in the 'as welded' condition due to breakdown of the β phase on cooling. In some cases the ductility can be restored by postweld heat treatment but such treatment is often impracticable.

The 6 Al 4V alloy can give 100% joint strengths in the 'as welded' condition but some deterioration in tensile elongation and bend ductility does occur (Table VIII) whilst there is also an increase in notch sensitivity.⁹ Tensile strength and elongation are maintained at a satisfactory level up to 520°C. Some improvement in bend ductility can be obtained in TIG welding if autogenous filler wire is replaced by commercially pure titanium filler,¹⁰ but even in this case the improvement is only partial. Postweld heat treatment will give further recovery of ductility if practicable.

Some work has also been done¹¹ on a recently developed alloy containing 6% aluminium, 6% vanadium, 2% tin, 0.3% copper and 0.5% iron. This alloy exhibited serious loss of notch toughness and heat affected zone cracking on welding and notch toughness could not be improved by subsequent heat treatment. It was found however, that a preheat of 260°C prevented cracking and allowed maximum retention of toughness. It is possible that similar levels of preheat may be beneficial for other alloys.

Beta Alloys.

The single phase β alloys are readily welded by TIG and MIG processes; welds in solution treated plate have good ductility and low strength. Welding of precipitation hardened material appears less successful.

The American alloy containing 13% vanadium, 11% chromium and 3% aluminium, can be readily welded¹² in the solution treated condition, but a complex double ageing treatment ('low-high' treatment) is necessary to obtain parent metal strength. Such a treatment may not be possible in commercial practice.

The 15% molybdenum β alloy is also readily weldable⁶ in the solution treated condition. After ageing tensile strengths of 78.5 - 81 tsi and an elongation of 12% on 1 inch can be obtained, compared to 65.7 - 89 tsi and 9 - 4.5% on 2 inches for parent sheet. The microstructure in the 'as welded' condition shows grain growth with a little α precipitate in the heat affected zone and large β grains in the weld. Ageing gives overall α phase precipitation. Welds in the aged material exhibit grain growth and grain boundary precipitation in the heat affected zone, whilst the weld consists of large β grains.

Resistance Welding.

Spot, seam and stitch welding can be readily utilized for titanium and its alloys. The low electrical and thermal conductivities, and low thermal expansion coefficients of these materials allow adequate welds

to be made with little distortion, welding conditions being very similar to those used for stainless steel. The time cycle for welding is short so that inert gas shielding is unnecessary.

Spot welds in all alloys exhibit good tension shear strengths that compare with strengths obtained in stainless steel, but cross tension strengths are quite low indicating poor weld ductility.¹³ This poor ductility may be due to slight contamination of the weld nugget during the welding cycle. The low cross tension strength has led to the suggestion¹² that the ends of spot welded seams should be rivetted to prevent peeling. It is interesting to note that the weld strengths are similar for α and $\alpha - \beta$ alloys, irrespective of condition (Table X) although a post weld heat treatment on the welding machine has been found to increase the cross tension strength of the 6 Al 4 V alloy.¹⁴

Flash butt welding has proved successful¹⁵ for the joining of bars of commercially pure titanium and the 5 Al 2.5 Sn alloy. Argon shielding is essential, however, to prevent contamination and embrittlement of the joint. This welding process, if properly applied, results in a hot worked pressure weld, since all the melted metal is squeezed into the weld fin, and afterwards removed, and exceptionally good joint properties are obtained (Table XI).

Brazing.

The main problems that are encountered in the brazing of titanium and its alloys are:-

- (1) prevention of contamination and subsequent embrittlement.
- (2) selection of brazing alloys that will give sufficient alloying for joint formation without forming brittle intermetallics.
- (3) prevention of the impairment of parent metal properties by the brazing cycle.

The fact that titanium can be brazed by a number of processes, including furnace, induction, resistance and torch brazing, is a fair indication of the success achieved in solving these problems.¹⁶

Prevention of contamination is normally achieved by inert gas shielding, which may be used for induction or furnace brazing, or brazing with a TIG torch. A vacuum can also be used where appropriate. Surprisingly oxy-acetylene torch brazing has been used successfully for titanium although special fluxes are needed to prevent contamination.² These fluxes generally consist of mixtures of chlorides and fluorides, and their exact operating mechanism appears uncertain. In some cases a simple exclusion of air may occur but with other fluxes it appears that electrochemical deposition of a metal film occurs protecting the titanium from oxidation. Some suggestion has been made that the chloride content of these fluxes can

result in stress corrosion cracking if flux entrapment occurs.

Titanium alloys ^{react} readily with most molten brazing alloys, but it is important to choose a composition where excess alloying does not occur, since this will result in base metal undercutting and erosion. The formation of brittle intermetallics must also be avoided. Silver and its alloys are used almost exclusively for brazing titanium,¹⁷ giving a bond that consists largely of a solid solution of silver in titanium. An intermetallic compound Ti Ag, is formed in some circumstances but is reasonably ductile. Optimum brazing temperatures for silver are $990^{\circ} - 1040^{\circ}\text{C}$ for times of 1 minute, which give shear strengths of 21000 - 24000 psi. A brazing temperature of $1150^{\circ} - 1260^{\circ}\text{C}$ for 10-15 minutes gave a much higher shear strength of 35000 psi but slightly impaired parent metal properties. Several silver base alloys have been developed¹⁸ for titanium but in most cases it is doubtful whether they offer any advantages. However, alloys containing aluminium or manganese have been reported as exhibiting good properties up to 205°C and resistance to salt spray corrosion. For the brazing of very thin sections e.g. honeycombs¹⁹, it is preferable to use lower temperatures of $760^{\circ} - 815^{\circ}\text{C}$ to prevent erosion of the thin parent sheet and a number of low melting point silver alloys have been developed, including some alloys containing 2-3% lithium. Joints with adequate strength can be obtained but resistance to oxidation and salt spray corrosion is low.

The temperatures employed in brazing have little effect on the properties of commercially pure titanium and the α phase alloys. With the heat treatable $\alpha - \beta$ or β alloys, however, care must be taken that the parent metal properties are not impaired. Alloys in the annealed condition may be re-annealed after brazing or the cooling rate may be controlled to give an annealed structure, but the problem is more difficult with age hardened alloys. Re-solution treatment and ageing are reported² to be possible with β alloys which exhibit sluggish transformation but are difficult to achieve with many $\alpha - \beta$ alloys which need rapid cooling after solution treatment if precipitation is to be avoided.

A brazing development²⁰ which offers many advantages is vacuum diffusion bonding in which the brazing time is extended to give diffusion of the silver braze metal into the parent material. This method is reported to give tensile and shear properties equal to that of the base metal. The same paper also states that diffusion bonding can be achieved in the same temperature range ($990^{\circ} - 1040^{\circ}\text{C}$) without filler metal, provided that metal to metal fit occurs.

Nickel is a malleable face centred cubic metal melting at 1440°C ; its physical properties are given in Table I. Nickel itself is ferromagnetic, although most of its alloys do not show any strong ferromagnetism, with electrical and thermal conductivities similar to those of steel. Commercially, the attractive properties of nickel are its high resistance to corrosion and oxidation, and these properties are generally found in the nickel-base alloys, often combined with excellent high temperature strength. Nickel is a fairly expensive metal so that it is used sparingly, and generally under rigorous conditions.

Most of the cubic metals are soluble in nickel to a greater or lesser extent. Copper and cobalt exhibit complete solubility in nickel whereas iron, chromium and molybdenum have solubilities of 60%, 35% and 20% by weight, respectively.

Commercial Alloys of Nickel

The commercial nickel alloys are all basically face centred cubic structures in which the alloying additions provide solid solution strengthening. However, very often additions are made to give precipitation hardening characteristics and many of the stronger alloys need artificial ageing to realise their maximum properties. Thus the microstructures of the nickel-base alloys is generally largely single phase, although discrete precipitates are often visible in the grains or grain boundaries as particles or needles. These precipitates may be due to impurities, as in the case of carbides, or they may be precipitation hardening constituents.

Commercially pure nickel is used in chemical plant because of its high degree of resistance to caustic alkalis, certain organic compounds, and to the halogen gases and hydrides. It often contains up to 0.06% carbon and small amounts of manganese, iron, copper and silicon. An age hardenable form contains 94% nickel with small amounts aluminium and titanium.

Monel

The high nickel-copper alloys, known as Monel, contain 58 - 65% nickel and 40 - 30% copper, often with up to 1.5% iron as an impurity. Certain of these alloys also contain aluminium and titanium to make them age hardenable. These alloys again exhibit good corrosion resistance and are used in chemical plant, land and marine power plants, textile and pickling equipment, and in the food industry.

Nickel-Chromium Alloys

These alloys, excluding the Nimonic series, are important for their excellent oxidation and corrosion resistance, which is often combined with good high temperature strength. The Inconel alloys generally

contain 75 - 80% nickel and 15 - 20% chromium, although in some instances iron or cobalt may replace some of the nickel. For the highest strengths, aluminium and titanium are added to give age hardening properties. These alloys are used for heat treatment and furnace equipment, pickling racks and in the food industry. Other alloys may contain up to 33% iron and are generally used for heating elements.

The Nimonic Alloys

The high temperature or super alloys - the Nimonics - are the most important of the nickel-base alloys. Originally developed for aircraft gas turbines, which is still their major use, they have found increasing applications in other fields such as steam turbines for power generation where highly stressed components are required to operate at elevated temperatures.

Development of the Nimonic alloys (Table II) started from a basic 80% nickel 20% chromium alloy, Nimonic 75.

This composition is very nearly that of Inconel, which was known to have good oxidation resistance, whilst the strength was improved by the addition of 0.5% titanium. Nimonic 75 is a single phase alloy, although some carbide precipitates may occur, and is not amenable to precipitation hardening. This alloy was a major improvement on previous high temperature alloys, but it was found that the high temperature properties varied markedly with prior treatment and an alloy with more stable creep properties was desirable. Additions of titanium and aluminium showed that satisfactory creep resistance was obtained when the limit of solid solubility was exceeded by a moderate amount (at the testing temperature), allowing precipitation hardening to occur. This work led to the introduction of Nimonic 80 and Nimonic 80A.

Considerable work was done upon this type of age hardening nickel alloy, which has led to the introduction of Nimonic 90, 95, 100, 105, 110 and 115. This work showed that the high temperature properties were influenced by

- (1) Composition of the base matrix
- (2) Constitution, amount and distribution of precipitated phases
- (3) presence and distribution of primary compounds.

Addition of up to 20% cobalt (Nimonic 90) goes into solution in the γ phase but has considerable influence on the characteristics of precipitation. The alloys in the Nimonic 100 - 115 range have an increased aluminium content and additions of molybdenum, and exhibit improved oxidation resistance and creep properties. However they retain the precipitation hardened face centred cubic structure of the other alloys.

The microstructure of all the Nimonic is similar, exhibiting a face centred cubic matrix (γ) plus precipitated carbides and intermetallics such as the γ phase (Ni_3Ti) and γ' phase ($\text{Ni}_3(\text{Al,Ti})$). The γ' phase is face centred cubic in structure and is isomorphous with the matrix, on metallographic examination it shows as a speckled effect within the grains. The γ phase occurs as lines or needles within the grains. All these alloys contain up to 0.1% or 0.15% carbon and chromium carbide, Cr_7C_3 occurs as discontinuous particles within the grain boundaries. The chromium carbide can be partially replaced by titanium carbide, which occurs as discrete particles within the grains. Additions of cobalt have the beneficial effect of reducing the amount of chromium carbide present in the alloy, and also of aiding the preferential formation of titanium carbide.

The heat treatment of the precipitation hardening alloys has an important influence on creep characteristics. The recommended heat treatments for several alloys are given in Table II; slightly better results can be obtained by more complex treatments but these are industrially impractical. The effect of solution treatment is

- (1) Dissolution of the γ' and γ phases which are then available for precipitation
- (2) Grain growth
- (3) Carbides left in matrix not dissolved

Dissolution is a constant effect above a minimum temperature but grain growth is temperature dependent so that the solution temperature is critical, which shows the effect of temperature on creep life. At lower solution temperatures a fine grain size occurs, and increasing the grain size improves stability, and hence creep life, but above a certain optimum grain size the creep life again falls. Solution of carbides does not occur below 1200°C so that for most of the Nimonic alloys the carbide will remain undissolved. However, Nimonic 110 and 115 have solution treatments of 1200°C and 1190°C respectively and solution of carbides is distinctly probable. The ageing temperature is also important, and variations in temperature can markedly influence creep properties.

A recent development in the Nimonic series is Nimonic PE.7, a heat treatable alloy designed to have good properties in the intermediate temperature range up to 700°C . In this alloy approximately 37% of the nickel is replaced by iron, presumably to reduce the cost of the alloy.

Fusion Welding

The physical properties of nickel and its alloys impose no exceptional welding problems and most alloys are readily weldable. However, many alloys are prone to cracking in the work hardened condition and should be welded in the annealed condition. The non-heat treatable alloys generally

present no difficulty and welding has little effect on the metallographic ^{structure} except some grain growth in the heat affected zone, and joint efficiencies of 100% are obtainable. Cleanliness is also important, oxide films can result in arc instability, whilst oil films, etc., can give porosity and cracking.

The precipitation hardening alloys are best welded in the solution treated condition, and the best joint properties are obtained by re-solution treating and ageing after welding. Fortunately most components in these alloys are small enough to allow complete heat treatment after welding, although distortion may be a problem. It is often possible to weld fully heat treated material but, invariably, joint properties will be well below the properties of the parent material. In these cases the weld should be positioned away from the critical load bearing regions of the component or structure.

Most of the major alloying additions have no effect on weldability, except in so far as they introduce impurity elements. Thus chromium has no effect in itself, but nickel-chromium alloys are more sensitive to hot cracking in the presence of silicon than other nickel alloys. Sulphur has the most serious effect since it results in the formation of a nickel/nickel sulphide eutectic melting at $650^{\circ}\text{C}.$, and so gives rise to serious hot cracking. Lead and phosphorus also lead to hot shortness, but are rarely found in nickel alloys. Zirconium and boron are often added to improve high temperature properties, but severe hot cracking is found if more than .02% zirconium or 0.001% boron is present. Calcium and magnesium are often added to improve casting properties. In small amounts they are not harmful, but above .015% calcium and 0.020% magnesium can lead to hot cracking. Some disagreement exists over the effect of magnesium, since in the presence of sulphur it can form magnesium sulphide, and so reduce hot shortness due to sulphur.

Silicon can also produce hot cracking in nickel alloys, although the threshold varies with the alloy. Chromium bearing alloys are particularly susceptible and the silicon content of the Nimonics and Inconels are carefully controlled. Alternatively, additions of niobium can be made which reduce the influence of silicon. Aluminium and titanium, which are often added to give ageing characteristics, have little effect on weldability, although in large amounts they can increase the sensitivity to hot cracking. Carbon, which is present as an impurity in amounts of 0.01 - 0.20%, has little effect on weldability. In commercially pure nickel, graphite can be formed in the grain boundaries after long exposure at $315 - 760^{\circ}\text{C}.$, leading to rupture, whilst chromium carbide can form in the grain boundaries of

chromium-bearing alloys. This is not detrimental unless fully continuous boundary precipitates are formed or unless chromium depletion impairs corrosion or oxidation resistance.

All the conventional welding processes can be applied to nickel and one or more of its alloys. The choice of process will depend partially on the normal accessibility and economic factors, but the service application is generally of paramount importance in these alloys, which are used under severe corrosive conditions or where oxidation and creep resistance are essential. The most used processes are shielded metal arc and argon shielded welding. For shielded metal arc electrodes the technique required is generally a little more difficult than, say, the use of austenitic stainless steel electrodes, but porosity and hot cracking can be avoided by a skilled welder.

Argon shielded tungsten arc welding is perhaps the most used process, and automatic welding without filler wire is often used on thinner sheets. Porosity can be a problem with this process, although modification of filler wire compositions has largely eliminated it. Alternatively, additions of 5 - 15% hydrogen to the shielding gas will reduce or eliminate porosity, although in excess it will cause severe piping. The advantages of hydrogen additions are only evident in single pass welds or for root runs in heavy sections. Porosity may also be reduced by increasing the speed of welding, although this is only practicable in automatic welding. Argon shielded metal arc welding is also used for nickel, and alloys in the Monel, Inconel and Nimonic series. Wire compositions approximate to parent metal compositions plus additions to control porosity and hot cracking. Metal transfer is by the spray mode, although the transition from globular transfer is not very marked; the use of fine wire welding may be expected to increase.

Oxyacetylene welding is still used quite extensively, although it is losing ground to the tungsten arc process. It can be used for all the non-heat treatable alloys but is generally confined to nickel, Monel, Inconel and Nimonic 75. Carbon pick-up can reduce corrosion resistance. No flux is used for welding nickel but special boron-free fluxes are used for the other alloys.

Specific filler materials are available for most but not all of the nickel-base alloys. The filler is generally chosen to give matching corrosion or oxidation resistance rather than properties, and matching of weld and parent metal compositions is rare. In most cases the filler wire will contain additions to reduce, or eliminate, porosity and hot cracking.

The Nimonic Alloys

The wrought Nimonic alloys, and the equivalent casting alloys in the Nimocast series were originally developed for small components that did not require welding. However, the usefulness of these alloys in high temperature applications has extended their use so that the welding of Nimonic sheet, and more recently of heavier gauges has become necessary.

Nimonic 75 (and also Nimocast 75) is basically a single phase alloy, although some chromium carbide is generally found at the grain boundaries, or within the grains. Welding is straightforward and the only change in microstructure in the heat affected zone is a slight grain coarsening, accompanied by a small decrease in hardness. Shielded metal arc or inert gas shielded welding may be used; the recommended filler metal composition is 80% nickel, 20% chromium which gives weld metal creep strength above that of the parent metal. With heavier sections of Nimonic 75 up to 1 inch thickness shielded metal arc welding giving a deposit of 75% nickel, 15% chromium, 10% iron was found to be satisfactory, but with thicker sections weld metal cracking occurred. With these thicker sections a weld deposit containing 5 - 10% manganese and 1 - 2.5% niobium appears to be the most suitable.

With the heat treatable alloys i.e. Nimonic 80A and 90, a serious drop in strength occurs in the weld metal and heat affected zone, joint efficiency falling well below 50%. In welds in these alloys the heat affected zone is single phase with all precipitates dissolved, including carbides, whilst the weld structure is columnar with very few precipitated particles. Improvement in weld strength can only be obtained by a complete solution treatment and ageing treatment, which will give joint strengths equivalent to those of the parent materials. Intergranular cracking in the heat affected zone is a problem with welds in thicker sections of Nimonic 80A and 90. This cracking has been attributed to low melting point grain boundary eutectics formed by zirconium and boron, which are present in these alloys. Weldable grades of Nimonic 80A and 90 have been developed which contain less than 0.02% zirconium and 0.001% boron, although a slight loss in creep strength results. The equivalent grades of casting alloys, Nimocast 80 and 90, exhibit the same sensitivity to zirconium and boron. The casting alloys also frequently contain residual amounts of the deoxidants calcium and magnesium which have also been shown to cause hot cracking in amounts greater than 0.015% and 0.020% respectively. Fortunately, the residual percentages of these elements are generally much less. With thick sections, inert gas metal arc welding of thick sections of the modified Nimonic 90 has been found to give fine intercrystalline cracks, although tungsten arc welds were quite satisfactory. This has been overcome by the use of a short

circuiting arc technique which reduces the heat input to the weld.

The more recently developed alloys Nimonic 100 - 115 are very similar to the earlier precipitation hardening alloys in microstructure, consisting of a γ phase matrix together with precipitated γ' phase and complex carbides. Thus welding will again result in solution of the γ' and carbide phases in the weld metal and heat affected zone, and to develop full weld strength a complete solution and ageing treatment is necessary after welding. With Nimonic 100 sheet, micro-cracking has been found in the weld heat affected zone after solution treatment. It was established that cracking was influenced by the time and temperature of heat treatment and that 800 - 950°C. constituted the critical range. By heating and cooling rapidly through this range it was possible to avoid cracking.

The lower nickel content of Nimonic PE.7 is beneficial for welding and no heat affected zone cracking has been found in alloys containing the normal amounts of zirconium and boron. Again, however, a full solution and ageing treatment is required after welding to develop high joint strengths.

Brazing

Brazing and silver soldering can be used for joining nickel and its alloys, although, since these materials are generally used for corrosion or heat resistance, it must be accepted that there may be a deterioration in these properties at the joint. Torch brazing and silver soldering can be used, as can salt bath and dip brazing, but the best results are obtained with electric furnace brazing in vacuo or a dry hydrogen atmosphere, using high melting point brazing alloys (Table IV). In all these processes cleanliness is very important, and care must be taken to remove all possible sources of lead, sulphur and phosphorus.

High temperature brazing has the advantage that it avoids the extreme thermal effects produced by welding, although the high temperature properties of the joint are inferior to those of the parent material. In general the brazing temperatures correspond to the solution treatment temperatures of the Nimonic alloys and, in these cases, brazing and solution treatment may be carried out simultaneously. The tenacious oxide films that are formed on the surface of the Nimonic alloys demand considerable care in the exclusion of moisture and air from the furnace atmosphere, and vacuum brazing is normally preferred.

Some success has been achieved with the Nimonic alloys using a pressure brazing technique, combining brazing and solid phase welding. This process employs a nickel-manganese-palladium filler and, after a diffusion treatment at 1200°C. the interface almost completely disappears. After an ageing treatment the stress-rupture properties of the joint were similar to those of the parent material.

Resistance Welding

All the resistance welding processes can be readily applied to the nickel base alloys, although spot and seam welding are the most used. Projection welding is used for cross wire welding of meshes and flash butt welding is used for the joining of bar sections, particularly for gas turbine components.

The electrical and thermal conductivities of the nickel-base alloys are similar to steel, so that power requirements for welding are similar. However, the hot strength of many alloys is high and more closely controlled pressure cycles are essential. The relevant factors governing resistant welding conditions are:

- (1) Surface cleanliness,
- (2) Current,
- (3) Weld time,
- (4) Pressure.

Cleanliness is more important than for most other metals.

In particular all sulphur and lead containing contaminants must be removed or serious hot cracking will result. Surface oxide is also important as it can greatly affect interface resistance, and inadequate removal will give inconsistent results. Pickling is generally sufficient for nickel and Monel, but should be followed by mechanical abrasion in the case of the other alloys. The welding current is somewhat higher and time shorter than with steels, to prevent sticking of the electrodes. With thinner sheets weld times can be as low as 2 - 3 cycles and phaseshift heat control is essential. For adequate forging, pressures of 20,000 - 40,000 p.s.i. are required, depending on the alloy, and electrodes should be made from the high strength copper alloys e.g. beryllium - copper.

The importance of the Nimonic alloys for use under rigorous conditions has again led to considerable work being done on the resistance welding of these alloys, principally by the British Welding Research Association. Satisfactory welds can only be obtained over a narrow range of welding conditions, although the current range can be extended by using larger electrode tip diameters for any given nugget size.

In the case of the non-heat treatable alloy, Nimonic 75, weld properties are satisfactory although post-weld stress relief has been claimed to enhance the subsequent performance of the welded component. Spot welds in Nimonic 75 often contain large internal cavities; it is debatable whether these cavities are detrimental to weld quality, but they can be greatly reduced by the use of high post-heat after welding.

Spot welds in the age hardening Nimonic alloys, Nimonic 80A and 90 are susceptible to hot cracking, which is believed to be due to the imposition of high cooling stresses on weld nuggets containing low strength

segregates. Any means of consolidating the weld nugget and reducing the cooling rate will reduce the likelihood of internal cracking and increased electrode pressures and longer welding times have been found particularly successful. The use of post-weld heating and forge pressures have also been found to be advantageous. Weld nuggets in these alloys often appear to have peripheral cracks which have been sealed with molten metal. It is considered that this phenomenon has no effect on weld properties or service life. The problem of seam welding the heat treatable Nimonics is very similar to that of spot welding. Particular attention must be paid to correct nugget overlap if cracking is to be avoided, whilst slower wheel speeds and longer on/off times are also helpful. Good results have been obtained with continuous current seam welding, which reduces heating and cooling rates. Post weld solution and ageing treatments are required to obtain optimum properties.

Residual stresses

Part I

by T. R. Gurney, M.A., C.Eng., A.M.I.C.E.

It is not very many years since any failure that occurred in a welded structure was almost automatically ascribed to the effect of residual stresses. Although recent work has tended to show that this view was misplaced, there still seems to be doubt in the minds of many people as to what exactly is the role of residual stresses in the behaviour of welded structures. This is the first of a series of articles in which we shall explore this subject.

What are residual stresses?

Although stresses are often thought of as being the intensities of the forces set up in a body in order to balance the external loads, stresses can nevertheless exist in the absence of any external load. These are normally called residual stresses. In fact residual stresses can be defined as the stresses existing in a body, or part of a body, in the absence of any externally applied load.

Strictly speaking residual stresses are of two distinct kinds. The first kind of stresses, which have on occasions been termed 'local residual stresses', can be defined more accurately as the stresses existing in a **member** in the absence of any external load, which are at the same time self balancing **within that member**. Thus, in the context of welded structures, they are confined to the neighbourhood of a weld, and decrease rapidly on each side of it. The second type, which is normally referred to as 'reaction' stresses, are stresses in a member which are balanced by the reaction of other members in the system, but are not self balancing within the member.

These definitions can perhaps be clarified by considering, by means of simple examples, how the two types are formed.

The formation of residual stresses

As an example of the formation of 'local' residual stresses consider a solid 'anchor ring' of material initially in a stress free state. If a small section of the ring were to be cut out (Fig. 1) so that a gap was formed in the ring, this could be closed by applying suitable external forces, and as a result, stresses would be produced in the ring. If the two ends of the ring (A and B) were then joined together, the closing forces could be removed, but the stresses would remain in the ring as a system of local residual stresses.

Equally, if a greater amount of fresh material were to be inserted in the gap than was just required to fill it, the ring would be expanded in

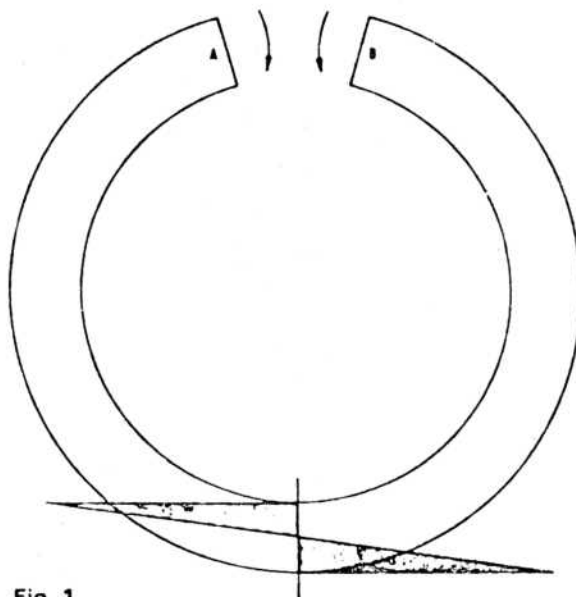


Fig. 1

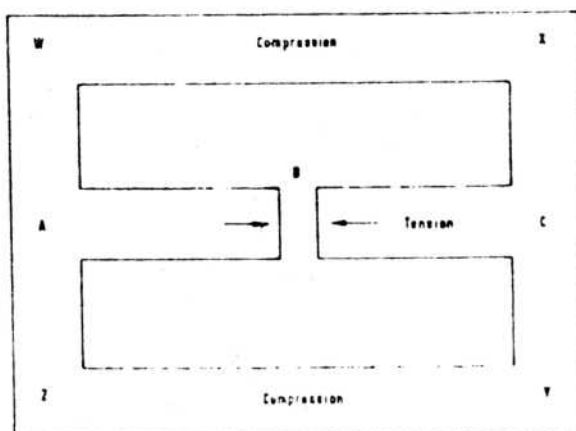
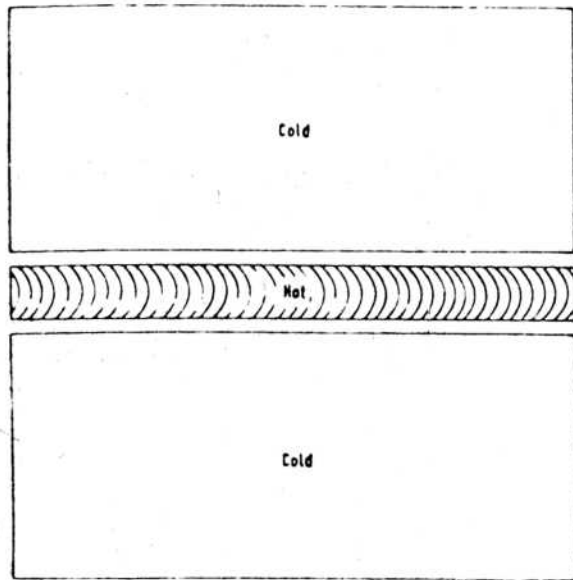


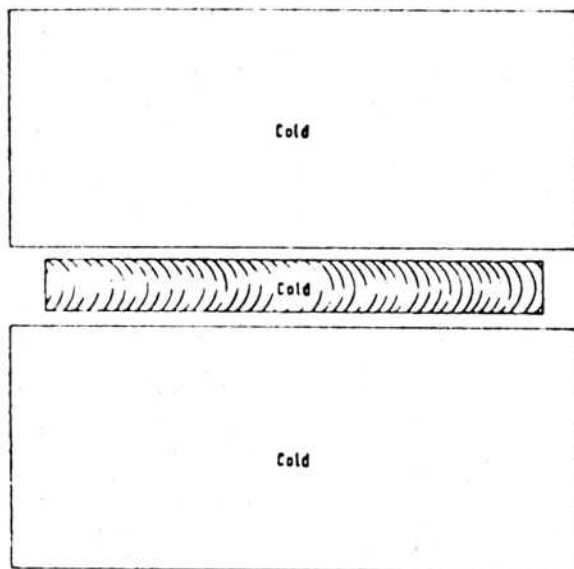
Fig. 2

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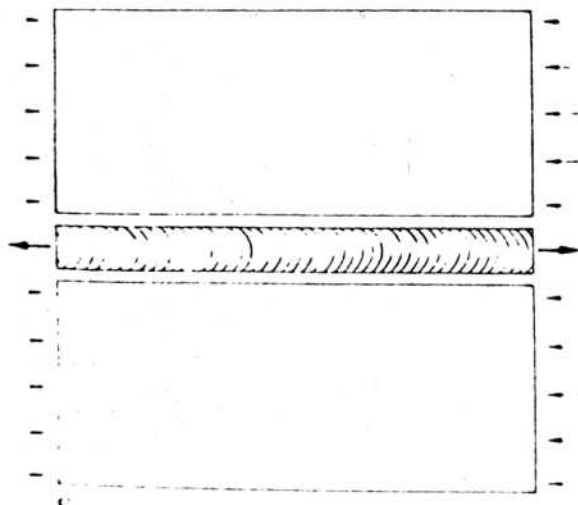
size. In this case, therefore, a residual stress system of opposite sign to that produced previously would be created.



a)



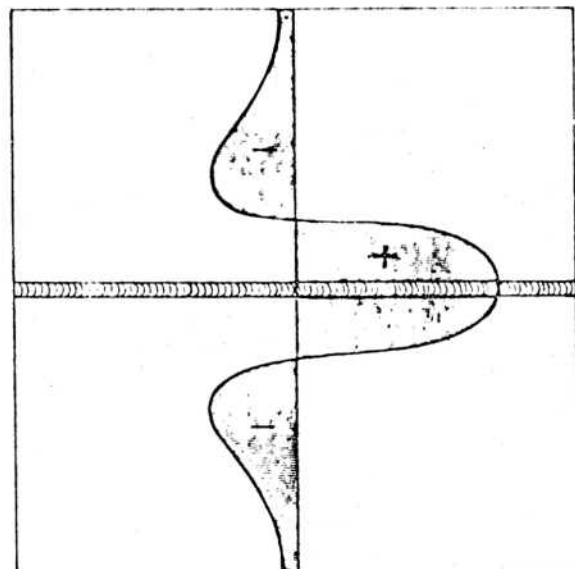
b)



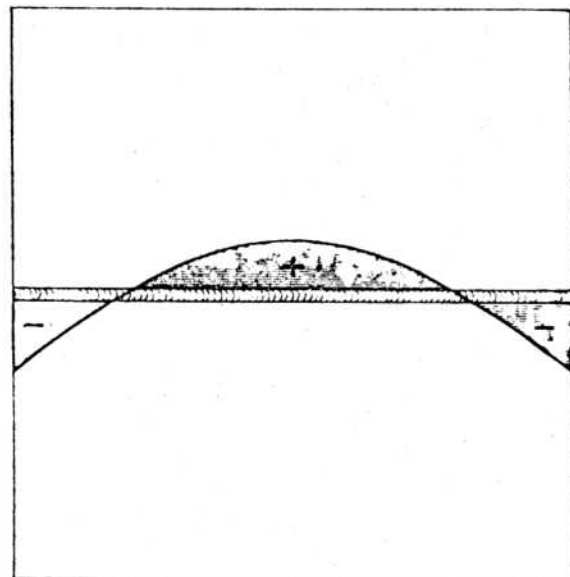
c)

As far as the formation of residual stresses is concerned, the essential feature of these operations is that material was either removed from or added to the ring, thereby causing a permanent alteration in the original body. This principle is universal. Residual stresses can only be produced by processes which involve a definite and permanent alteration of the body as compared with its form in the stress free state. In this way a geometrically incompatible system of deformations is produced, and it is these which are the origin of the stresses. The usual way in which this occurs, of course, is not by cutting and joining, but by plastic flow. Thus, in the case of the ring, if a small part of the circumference were plastically deformed so that it became shorter (in the circumferential direction) the net result would be the same as cutting out a part of the ring and then rejoining the ends. In other words, residual stresses would be created.

Consider now, as another idealised example, the case of the rigid frame WXYZ (Fig. 2) into



d)



e)

Fig. 3

which is to be fixed a central cross-member ABC which consists of two halves, AB and BC, attached to the frame at A and C respectively. Suppose that at this stage there is a slight gap at B. This means that, in order to make the joint at B, tensile loads would have to be applied to the members AB and BC, and these loads would be reacted by compressive **reaction** stresses in the members WX and YZ. Note that in this case the stresses, although self balancing within the structure as a whole, are not self balancing within each individual member.

It is important to grasp this general principle. 'Locked-in stresses' (a term covering both local residual stresses and reaction stresses) are subject to exactly the same laws of equilibrium as apply to ordinary stresses produced by external loads. The only difference is that the external loads are zero, so that the locked-in stresses must be self balancing within the structure.

Residual stresses caused by welding

It is often thought that the residual stresses produced by welding are 'special', in the sense that they are a form of thermal stress. This is not correct. The locked-in stresses of interest are those that exist after the structure has cooled down to ambient temperature after welding, and since it can be shown without difficulty that no thermal stresses can exist in the absence of a temperature gradient in the body, it is immediately obvious that the residual stresses existing after welding are not thermal stresses.

Once again, they are, in fact, the result of plastic deformation. During welding severe thermal stresses must arise because the temperature distribution in the body will be far from uniform, and it is these thermal stresses which cause the relevant plastic deformations. While the body is hot a compatible system will exist between the plastic and thermal deformations, but in the course of cooling the thermal strains disappear and the plastic deformations remain. By themselves, however, these plastic deformations will not satisfy the compatibility conditions so that a residual stress system is formed.

Let us now consider, in a simplified manner, the way in which residual stresses are formed around a butt welded joint. Immediately after

welding, the weld and the region adjacent to it will be hot, but the remainder of the plates will be cold. Thus, if we consider the weld and the two cold plates as three separate elements, all will be of the same length, as shown in Fig. 3a. Now, as the weld cools it will decrease in length, so that when its temperature is the same as that of the plates the situation will be as shown in Fig. 3b. From this it will be seen that the weld is now shorter than the two plates. In fact, however, the weld and the plates are the same length, but this can only be achieved by applying a tensile load to the weld and compressive loads to the two plates (Fig. 3c). In order to maintain equilibrium the tensile and compressive loads must, of course, balance. These forces are effectively the longitudinal stresses in and around the joint. In other words, the weld is a region of high tensile residual stress and this is balanced by compressive residual stresses in the surrounding material.

The real situation in a welded joint is a little more complicated, since the weld is made by progressive deposition of metal rather than instantaneous deposition of a strip. Nevertheless the principles described above are generally valid. The simplification of considering the joint as three elements would result in a residual stress distribution with sudden steps in it. In fact, of course, this does not happen and the real residual stresses acting in a direction parallel to the weld are as shown in Fig. 3d. In addition, there will be residual stresses transverse to the weld and these will be tensile near the centre and compressive near the ends, as shown in Fig. 3e. Normally the residual stresses will reach yield point in tension parallel to the weld and quite often in the transverse direction.

Concluding note

In this introductory article an explanation has been given of what residual stresses are and how they are formed. In subsequent articles such subjects as the effect of residual stresses on structures subjected to static and fatigue loading, their role in brittle fracture, their measurement and some aspects of stress relieving will be considered. In this way it is hoped that a comprehensive review of the subject will be presented.

Applications for membership



Overseas Membership

Construction Metalliques de Province, Route de Farasson, France.

Års Aktieförslag, Sweden

The following organisations have applied for membership or associate membership since the February, 1968 issue of the *Bulletin*.

Concrete Utilities Ltd., Great Amwell, Ware, Herts.

National Galvanizers Ltd., Sunderland

Residual stresses

by M. G. Dawes

Part 2 Stresses and strains: measurement

Stress is a concept used to describe the intensity of forces acting within bodies, and cannot be measured directly. What we mean when we talk about measuring stresses is the measurement of displacements over finite gauge lengths, from which we calculate strains and then stresses. In most cases the strain measurements can only be made conveniently on a free surface. Whether or not we are justified in inferring stresses from strain measurements needs careful consideration.

Beating in mind that **residual stresses** are usually the result of plastic deformation (see Part 1, *BWRA Bulletin*, March, 1968, it is clear that displacements based on measurements before and after stresses have been introduced in a body, may be partly elastic and partly plastic. **Residual welding stresses** or **residual flame cutting stresses** result from plastic deformations caused by the different heating and cooling cycles experienced by adjacent volumes of material. These effects are complicated by the fact that the material yield strength decreases with increasing temperature. Also, in welded joints, adjacent volumes of material may have different yield strengths at the same temperature.

In addition to residual stresses the plastic deformations may also lead to distortion on cooling to room temperature. In these circumstances the calculation of stress from strain may bring about completely misleading results. The only reliable way to get an indication of residual stress is to make measurements before and after relaxing the residual stresses. This is known as the relaxation technique and relies on the elastic recovery of stressed material when it is removed from or relieved of the restraint imposed on it by the surrounding material. A more detailed description of the various relaxation techniques is given later on.

It can be concluded from this that there are two procedures for measuring residual strains, one relevant to elastic/plastic strains (distortion) and one relevant to elastic strains and residual stresses. If both procedures are used we can obtain the purely plastic strains also. Examples of all three conditions of strain, in the flange of a welded plate girder, are shown in Fig. 1. In this figure (a) shows the distortion or the longitudinal elastic/plastic strains calculated from measurements before and after welding. Figure 1(b) shows the elastic strains. These were calculated from measurements before and after

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measurements have been found to agree with those obtained by mechanical methods, the same agreement has not been found where plastic strains occur. Other factors known to affect the accuracy of X-ray measurements are grain size and the presence of two or more phases, e.g., austenite and martensite in steel.

Some investigators²³ have suggested that the X-ray determination of residual stresses is feasible only for pure materials. The biggest drawback to this method is probably the fact that it involves point by point measurements; it is time consuming and can only be used conveniently in a laboratory. Some examples of X-ray measurements in welded structures were given recently by Hughes²³.

Relaxation techniques

As defined earlier, the relaxation techniques allow the elastic recovery of stressed material when it is relieved of the restraint imposed on it by surrounding material. More specifically, these techniques can be defined as those which relieve stressed material by separating it from its surrounding material. Thus, the relaxation techniques either damage or completely destroy the structure concerned.

For convenience the completely destructive methods will be described first. These include Stablien's⁸ method, machining from one side, and Sachs' boring out technique, which have been mentioned previously, and what are known generally as the dividing methods. Some examples of the dividing methods are shown in Fig. 3. In this figure, (a) illustrates the method of Campus⁹ in which the specimen is reduced to square elements, each containing a strain rosette. For the special case where the principal stress directions are known, the author²⁴ used the technique for overlapping gauge lengths shown in Fig 3(b).

When using the dividing methods for thick plates and shells it is usually necessary to relax the stresses in the layers of material beneath the plate surfaces. This is normally carried out as shown in Fig. 3(c).

A less destructive method, and one which is ideal for both laboratory and field use, is the trepanning technique developed by Gunnert.⁵ In this technique a hollow drill is used to make a circular groove around the strain rosette. For thin plates a circular disc containing the strain rosettes can either be drilled out of the plate or left attached to the plate by a thin annulus of material. Gunnert⁵ suggests that, to relax the surface stresses in plates greater than 10 mm thick, it is only necessary to drill to a depth of 6–8 mm (approximately 0.3 in.). It should be noted, however, that the 6–8 mm applies to 9 mm gauge lengths on a 15–16 mm dia. disc, and deeper drilling is needed for discs of greater diameter. An example of the trepanning technique applied to a welded girder is shown in Fig. 4. A similar method to Gunnert's⁵ was developed by Meriam, De Garmo and Jonassen²⁵. Instead of a hollow drill, these investigators used overlapping $\frac{1}{4}$ in. dia. holes on a $2\frac{1}{2}$ in. dia. pitch circle.

The hole drilling method developed by Mather²⁶ is probably the least destructive of the relaxation techniques. This method is based on the fact that a drilled hole will relieve stresses in radial directions close to the hole. It has the disadvantage that the stresses tangential to the hole are the product of the stress concentration factor for the hole and the residual stresses, and they can be expected to influence the values of strain relaxed. Notable adaptations of Mather's original technique were made by Campus⁹, for use with mechanical gauge lengths, and by Soete²⁷ for use with electric resistance strain gauges; two examples are shown in Fig. 5.

With all the techniques described great care is necessary to ensure that the machining operations do not introduce further residual stresses. Such additional stresses can arise from the generation of heat, the material work hardening, and in soft materials (e.g., annealed copper), by clamping the specimen to a machine base. Obviously, these effects can be minimised by using adequate amounts of coolant and sharp cutting tools. In very soft materials acid etching and spark erosion techniques²⁸ have been used to overcome some of the difficulties described above. The extent of stresses caused by machining can be easily determined by taking measurements on stress free material before and after machining.

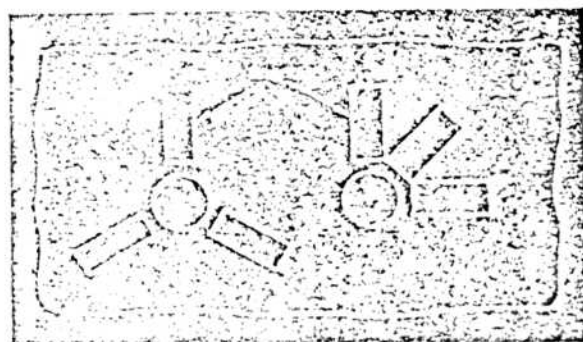
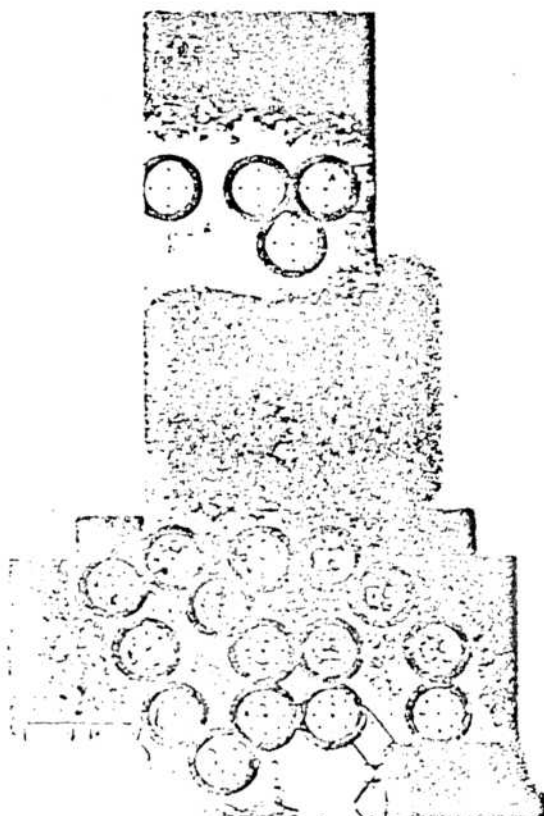


FIG. 3(a) The dividing technique applied to a welded girder.

stress relaxation. In this case stress relaxation is carried out by cutting the specimen into small blocks, each containing a gauge length. Clearly, the distortion gives a misleading indication of the elastic strains. Figure 1(c) shows the plastic or permanent strain. This is the difference between (a) the elastic plastic strain and (b) the purely elastic strain. It is worth noting when analysing elastic plastic measurements that a contraction indicates a compressive strain, whereas a contraction when elastic strains are relaxed indicates that a tensile strain was previously present. Thus, great care is necessary with the sign conventions used for strain measurements to give distortion and residual stresses.

Having obtained the elastic residual strains the calculations of residual stresses are made using Hooke's law. In cases where the strain measurements are made over gauge lengths on free surface, the stress normal to the surface must be zero, and a biaxial state of stress exists. In thick sections, the stress normal to the surface can build up away from the free surface and residual stress measurements for this component stress need special techniques. To determine completely the stresses at a point on a given surface it is necessary to have strain measurements in three directions, or to assume the two principal stress directions and align the gauge lengths with these. When the principal stress directions are known, the stress magnitude can be calculated from the following equations:

$$\sigma_1 = E \frac{(\epsilon_1 - \nu \epsilon_2)}{1 - \nu^2}, \quad \sigma_2 = E \frac{(\epsilon_2 - \nu \epsilon_1)}{1 - \nu^2}$$

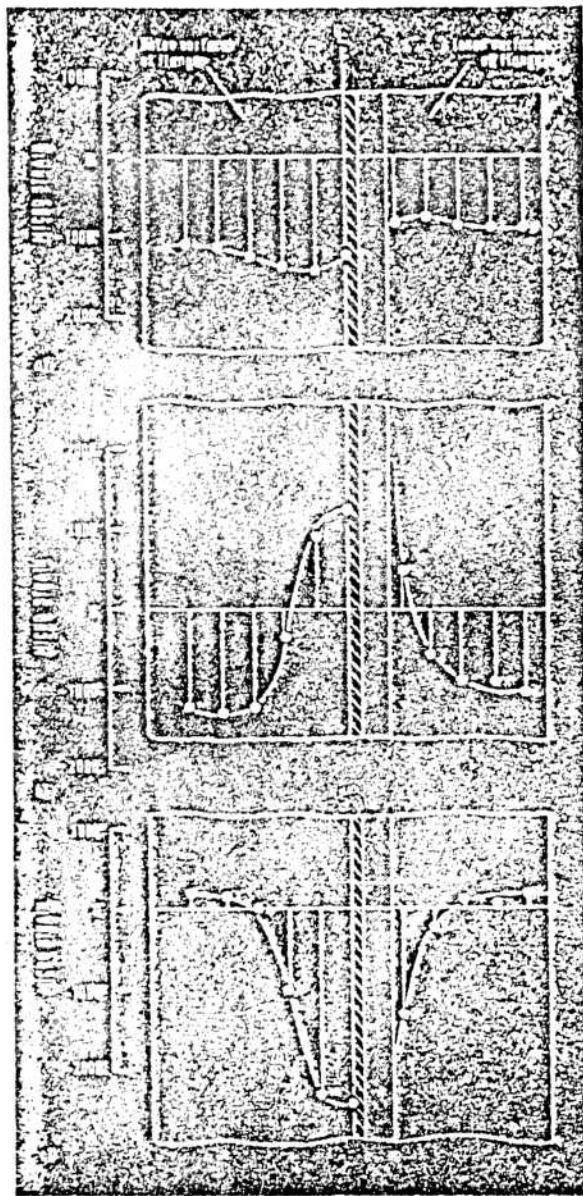
where σ_1, σ_2 are the principal stresses corresponding to principal strains ϵ_1 and ϵ_2 .

Finally, it must be pointed out that residual strain measurements on specimens representative of structural parts may not give a true indication of all the strains in the same detail, if these specimens are integral parts of a structure. In other words, measurements on specimens will give only the local strains causing self balancing stresses, and not the sum of the local and long range reaction strains which might exist in a complete structure. Where it is possible to extract a specimen from a structure, stress relaxation can be considered a two stage process. In this case measurements made before and after extracting the specimens will allow the reaction stresses to be determined.

Choice of measuring technique

By now it will be clear that the choice of a measuring technique will depend upon what is known about the state of stress in the body to be measured. For example, it may be necessary to measure the strains in several directions about a point. Also, the gauge lengths must be small enough to give a true indication of strain in zones where steep strain gradients are present.

The analysis of stresses in actual structures may necessitate using only a semi-destructive relaxation technique, the results of which can be left or easily repaired. Some methods can only be used under laboratory conditions, while some may be impracticable for other reasons, e.g., they may be laborious or time consuming.



1 Residual strains in the flange of a welded plate girder.

- (a) Longitudinal elastic/plastic strains (Measurements before and after welding).
- (b) Longitudinal elastic stresses (Measurements before and after relaxation).
- (c) Longitudinal plastic strain. (By subtraction of strains).

Different methods of measurement

There are many techniques and types of apparatus available for measuring residual strains and many of them have been the subject of critical reviews¹⁻⁶.

In this article it is sufficient to give a simple description of the principles and techniques, together with selected literature references for more detailed information.

Although the measurement of changes in overall dimensions may indicate residual stresses, such measurements cannot normally be used for obtaining stress magnitudes. A notable exception, however, is Sachs⁷ boring technique used for circular cylinders and thick walled tubes. In this method the triaxial stresses are calculated from the incremental changes in diameter and length on boring out in steps.

For bars and rods uniaxially stressed, the deflections accompanying machining from one side can be used to calculate the residual stresses from simple beam formulae. This technique is often referred to as Stablein's⁸ method.

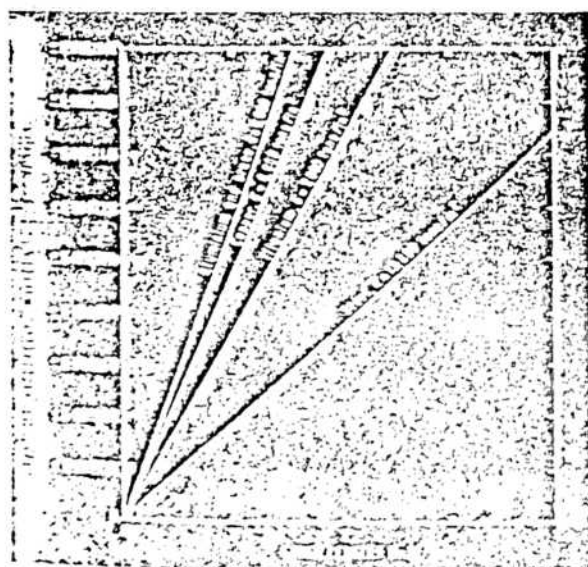
Probably the most widely used methods of residual stress analysis are those which measure the changes in length over surface gauge lengths. These methods usually involve electric resistance

strain gauges or special measuring gauges and extensometers. The choice of gauge lengths is usually a compromise between being short enough to be unaffected by strain gradients and long enough to enable the changes in length to be measured with sufficient accuracy using the equipment available. Figure 2 shows the accuracy required of measuring equipment, using different gauge lengths to give stresses within ± 0.5 tons/in², for several materials.

It is clear that, for materials with a high modulus of elasticity, the use of small gauge lengths calls for extremely accurate measuring equipment. While electric resistance strain gauges are undoubtedly more accurate than the best dial gauges and extensometers, the mechanical methods might be preferred under certain conditions, e.g., where damage to the electric resistance strain gauges could occur between measurements made before and after straining or between initial measurements and measurements after relaxation.

For extreme conditions, such as those close to welds during welding, special mechanical methods have been developed in which the gauge lengths are marked by drilled holes^{5,9} or steel balls^{10,11}. Some investigators have even used gauge lengths marked by hardness indentations or simply by scratches. With all these methods, the measuring instruments are not fixed to the testpiece during straining, but can be adjusted to register on the gauge lengths when required. Because severe strain gradients exist near welds the shortest gauge lengths are used consistent with measuring accuracy. Gunnert⁵ for example, developed an extremely sensitive dial gauge for making measurements over 9 mm gauge lengths.

In this instrument it is claimed that friction is eliminated to such an extent that repeated readings can be taken to within ± 0.3 microns (± 0.000012 in.). The other methods^{5,10,11} are generally used with gauge lengths of between 10 and 20 mm. In addition to friction, other factors which influence the results when using such short gauge lengths are cleanliness, the skill of the operator and temperature. It is obviously easier to keep gauge lengths marked with steel balls cleaner than those marked with holes.



Temperature is particularly important. A 10 deg C change in temperature, for example, will cause a change in gauge length equivalent to a stress of 1.5 tons/in² in steel and 1.0 tons/in² in aluminium. The effect of changes in temperature can be eliminated if the measurements are referred to similar measurements on gauge lengths of the same nominal length on unstressed material, at the same temperature. Suppose that measurements are made before and after the relaxation of residual stress. If L_1 and R_1 are the initial measurements of the testpiece and reference gauge length respectively, and L_2 and R_2 are the corresponding measurements after relaxation when the ambient temperature has changed by ΔT , then the differences are:

$$D_1 = L_1 - R_1 \text{ and } D_2 = L_2 - R_2$$

From these equations it can be shown that the change in length of L_1 :

$$\Delta L_1 = D_1 - \frac{D_2}{1 \pm \alpha \Delta T}$$

where α is the coefficient of linear expansion and ΔT is the temperature change. Because coefficients of expansion for most engineering materials are less than 0.0003 per deg C then, for practical purposes, the value of $\alpha \Delta T$ will be small compared with 1.0 and may be neglected.

Therefore:

$$\Delta L_1 = D_1 - D_2$$

and the engineering strain, $\epsilon = \frac{\Delta L_1}{L_1} = \frac{D_1 - D_2}{L_1}$

As mentioned previously, in all the methods which measure strains over surface gauge lengths, unless the principal strain directions are known, it is necessary to measure strains in at least three directions about a point. It is preferable however to measure strains in four directions using a four gauge 45° rosette because this allows a check on the assumptions concerning strain gradients. Thus, if a, b, c and d are the axes of the gauge lengths, then the strains along them will be compatible when $\epsilon_a + \epsilon_c = \epsilon_b + \epsilon_d$. In many cases it is possible to reduce the error caused by strain gradients by overlapping or staggering the gauge lengths (Fig. 3b).

While the use of surface measurements is strictly applicable only to uniaxial and biaxial stress states such measurements have been used in conjunction with other measurements in attempts to determine triaxial states of stress. Apart from the special case of Sachs¹⁷ boring out method, and the improvements on this by Bühler¹², the methods developed by Rosenthal and Norton¹³, Gunnert¹⁴ and others^{15,16} for thick plates are worth mentioning.

The two main disadvantages of the stress analysis methods which measure changes in length over individual surface gauge lengths are first, that even with the smallest gauge lengths, errors may arise as a result of strain gradients and second, in complicated structures, it is necessary to carry out a large number of measurements to obtain even an approximate indication of the principal stress trajectories.

Two methods which can be used to overcome these disadvantages are the photo-elastic coating technique and Moiré fringe technique. Both of these were described by Hardy¹¹ in a recent *Bulletin*.

Some applications of the photo-elastic coating technique to welded structures are described by Zandman¹². In this technique the surface to be measured is coated with a layer of transparent material which is double refracting when subjected to stresses. When polarised light passes through the layer, and it is viewed through a polarising element, a coloured image is obtained from which the residual stresses can be determined.

A limitation of the method is that auxiliary information is required to separate the principal stresses at positions remote from boundaries. Also, because the photo-elastic coating is sensitive to temperature variations, it cannot be used to measure the strains close to welds during welding. Apart from the fact that the coating material might melt or burn under these conditions, a secondary pattern of fringes may arise as a result of the difference in expansion between the testpiece and the plastic coating.

The Moiré technique is free of these disadvantages. The Moiré fringe method makes use of two grids of closely spaced fine lines. The model grid is fixed to the specimen and distorts with the specimen and the other grid, the master, is fixed to a transparent plate such as glass. When the distorted model grid and the undistorted master grid are superimposed, either directly or photographically, fringe patterns are formed. These patterns are the loci of points of equal relative surface displacement, and allow the calculation of strain perpendicular to the grid lines. While early attempts to use the Moiré technique of strain analysis were limited to strains greater than approximately 0.15%, greater than the elastic strains in many materials, recent refinements in grid reproduction and measuring techniques now make it possible to detect strains of both elastic and elastic plastic magnitudes. By using grid densities of up to 2000 lines/in., and mismatch techniques of measurement, such strains can be computed from the line density and the distance between pairs of fringes or directly from the fringe inclinations with respect to the co-ordinate axes.

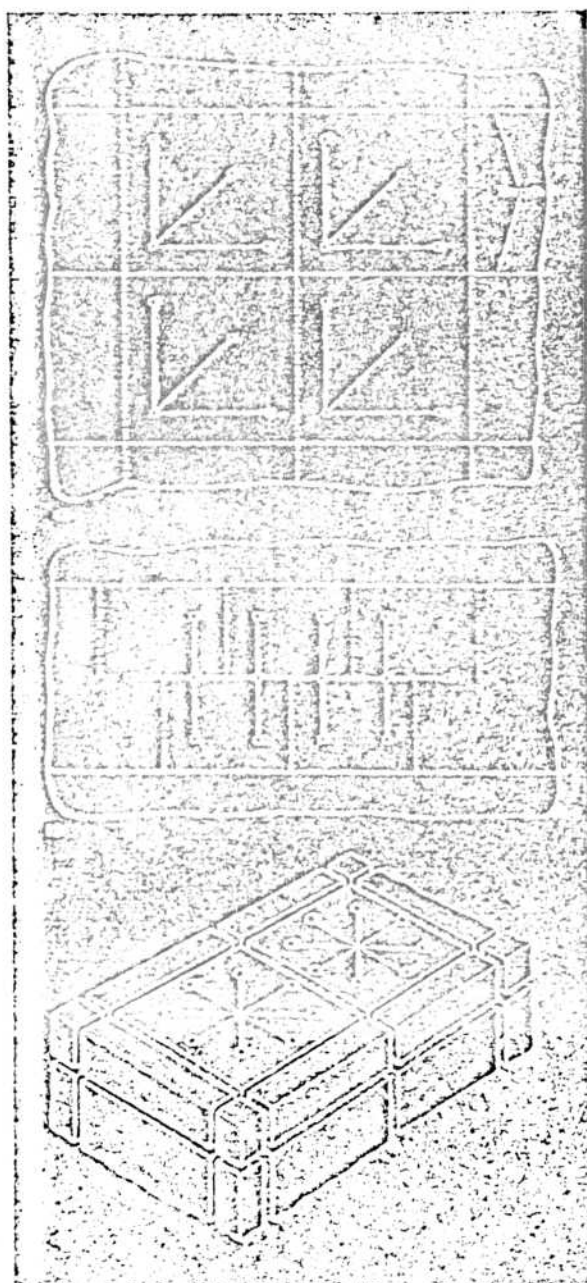
The Moiré grids can be considered equivalent to an infinite number of strain gauges attached to the specimen surface. Provided precautions are taken to protect the grids, measurements can be made close to welds during welding. The method has been used with great success by Sorel *et al.*¹³ to record the transient strains occurring at notches near welds, and also by Riley and Dinelli¹⁴ for the determination of transient stresses in impact loaded models. Because of the versatility of the Moiré technique it will undoubtedly be increasingly used in the determination of surface strains.

In all the techniques for strain analysis described so far the evaluation of the elastic strains and the inference of residual stresses depend on the partial or complete destruction of the speci-

men. This disadvantage is overcome by the X-ray diffraction technique. The measurement of strain by X-ray diffraction consists of evaluating the changes in the atomic distances within metal crystals. Because the interatomic distances are affected in a manner proportional to stress they can be considered to be small gauge lengths and their changes in length as strain. In theory, therefore, the X-ray technique will allow the detection of peak stresses. Either X-ray back reflection photographs or X-ray diffractometers can be used to measure the interplanar spacing, but diffractometers are more accurate for the diffuse diffraction patterns yielded by most materials of practical use.

Any surface strain component at a point can be obtained from two irradiations, both taken in the strained state. One is taken perpendicular to the surface and one obliquely. To determine the principal strains, one irradiation is made perpendicular to the surface and three obliquely.

A clear description of theory of X-ray stress measurement is given in a standard work by Cullity²². Although in the elastic range X-ray



3 Dividing methods of stress relaxation.

- (a) For 3 gauge length 45° rosettes.
- (b) Overlapping 2 gauge length 90° rosettes.
- (c) 4 gauge length 45° rosettes.

measurements have been found to agree with those obtained by mechanical methods, the same agreement has not been found where plastic strains occur. Other factors known to affect the accuracy of X-ray measurements are grain size and the presence of two or more phases, e.g., austenite and martensite in steel.

Some investigators²⁵ have suggested that the X-ray determination of residual stresses is feasible only for pure materials. The biggest drawback to this method is probably the fact that it involves point by point measurements; it is time consuming and can only be used conveniently in a laboratory. Some examples of X-ray measurements in welded structures were given recently by Hughes²⁶.

Relaxation techniques

As defined earlier, the relaxation techniques allow the elastic recovery of stressed material when it is relieved of the restraint imposed on it by surrounding material. More specifically, these techniques can be defined as those which relieve stressed material by separating it from its surrounding material. Thus, the relaxation techniques either damage or completely destroy the structure concerned.

For convenience the completely destructive methods will be described first. These include Stablien's²⁷ method, machining from one side, and Sachs' boring out technique, which have been mentioned previously, and what are known generally as the dividing methods. Some examples of the dividing methods are shown in Fig. 3. In this figure, (a) illustrates the method of Campus⁹ in which the specimen is reduced to square elements, each containing a strain rosette. For the special case where the principal stress directions are known, the author²⁸ used the technique for overlapping gauge lengths shown in Fig 3(b).

When using the dividing methods for thick plates and shells it is usually necessary to relax the stresses in the layers of material beneath the plate surfaces. This is normally carried out as shown in Fig. 3(c).

A less destructive method, and one which is ideal for both laboratory and field use, is the trepanning technique developed by Gunnert.⁵ In this technique a hollow drill is used to make a circular groove around the strain rosette. For thin plates a circular disc containing the strain rosettes can either be drilled out of the plate or left attached to the plate by a thin annulus of material. Gunnert⁵ suggests that, to relax the surface stresses in plates greater than 10 mm thick, it is only necessary to drill to a depth of 6–8 mm (approximately 0.3 in.). It should be noted, however, that the 6–8 mm applies to 9 mm gauge lengths on a 15–16 mm dia. disc, and deeper drilling is needed for discs of greater diameter. An example of the trepanning technique applied to a welded girder is shown in Fig. 4. A similar method to Gunnert's⁵ was developed by Meriam, De Garmo and Jonassen²⁵. Instead of a hollow drill, these investigators used overlapping $\frac{1}{4}$ in. dia. holes on a $2\frac{1}{2}$ in. dia. pitch circle.

The hole drilling method developed by Mather²⁹ is probably the least destructive of the relaxation techniques. This method is based on the fact that a drilled hole will relieve stresses in radial directions close to the hole. It has the disadvantage that the stresses tangential to the hole are the product of the stress concentration factor for the hole and the residual stresses, and they can be expected to influence the values of strain relaxed. Notable adaptations of Mather's original technique were made by Campus⁹, for use with mechanical gauge lengths, and by Soete²⁷ for use with electric resistance strain gauges; two examples are shown in Fig. 5.

With all the techniques described great care is necessary to ensure that the machining operations do not introduce further residual stresses. Such additional stresses can arise from the generation of heat, the material work hardening, and in soft materials (e.g., annealed copper), by clamping the specimen to a machine base. Obviously, these effects can be minimised by using adequate amounts of coolant and sharp cutting tools. In very soft materials acid etching and spark erosion techniques²⁸ have been used to overcome some of the difficulties described above. The extent of stresses caused by machining can be easily determined by taking measurements on stress free material before and after machining.

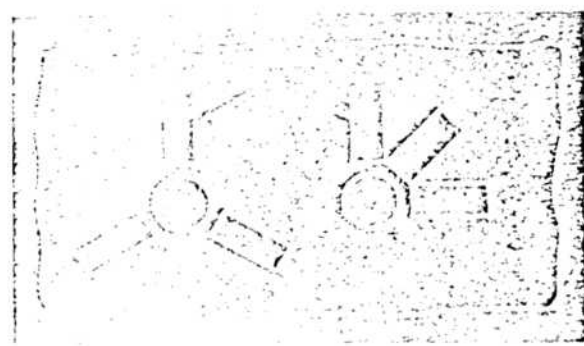
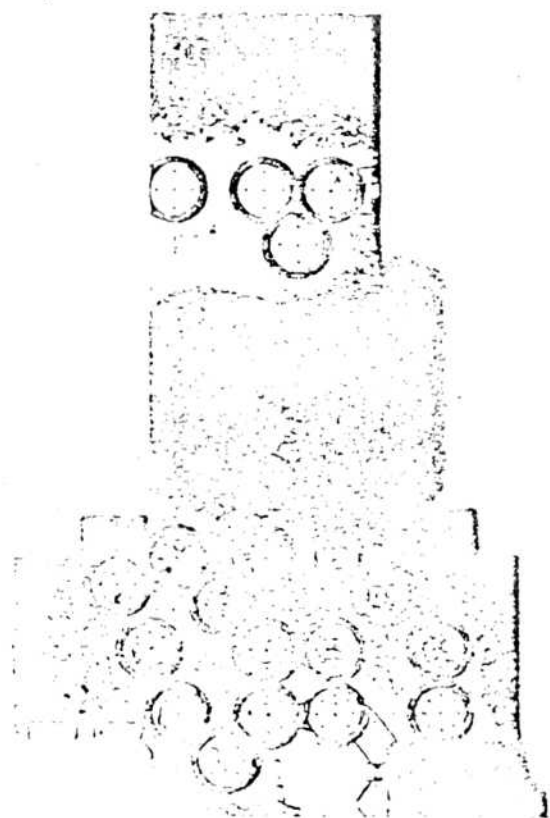


Fig. 3. (a) The dividing technique applied to a rectangular plate.

The present article summarises strain measuring techniques based on changes in overall dimensions, deflections, displacements over surface gauge lengths, photo-elastic coatings, Moiré fringes and X-ray diffraction. All these techniques are in regular use throughout the world. The application of the different methods to the measurement of elastic and elastic plastic strains is considered for both static and transient states of strain. It is concluded that no one method of residual strain measurement is suitable for all applications. Probably the nearest to an ideal method is the Moiré fringe technique. But even with the most recent developments in this technique it cannot measure elastic strains with the degree of accuracy obtained with most of the gauge length measuring methods.

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Residual stresses

by G. G. Saunders, B.A.

Part 3 Thermal stress relief and associated metallurgical phenomena

The first two articles in this series described how residual stresses arise and how they may be measured. The present article deals with the removal of residual stress by thermal treatment and will be followed by an article on mechanical stress relief. Since post weld thermal treatments are used mainly on ferritic steels, illustrations based on these materials will be used, except where necessary to do otherwise. It should be emphasised that post weld heat treatments are expensive, and that they should be used only in cases where the need for heat treatment and the subsequent effects can be clearly identified.

It was demonstrated in an earlier article¹ that welding residual stresses arise from the plastic deformation (produced by thermal strains) of the material in and around the weld, while it is restrained by the cooler surrounding material. It is sometimes necessary to remove these residual stresses, which can attain yield point magnitude^{2,3}, since they may impair the subsequent performance of the weldment in service. For instance, if yield point stresses are present, the addition of service stresses immediately causes plastic deformation. If defects or strain concentrators are present in the region of residual stress, it is possible for the residual stresses to cause crack extension by one of several modes. In the case of structural steels, this could give rise to a brittle fracture.

The presence of residual stresses also increases the likelihood of stress corrosion cracking for certain material and environment combinations. The presence of compressive residual stresses can in certain cases improve the properties of a structure under fatigue conditions, but the removal of tensile residual stresses has not been observed to improve fatigue resistance, except where part of the fatigue cycle is compressive.

A further problem that may arise is that relaxation of residual stresses during machining operations, in which the restraining bulk of material is removed, can cause elastic distortion to occur. If high machining tolerances are required, then the residual stresses must first be relaxed by a thermal treatment before machining operations are carried out.

THE EFFECT OF POST WELD HEAT TREATMENT ON RESIDUAL STRESS

If residual stresses are considered detrimental to

the service performance of a particular weldment, the problem of removing them must be faced. The most commonly accepted method is that of thermal stress relief. This will not only have the effect of relaxing residual stresses, but can also bring about beneficial metallurgical changes, as we shall see later. The thermal treatment is applied in order to raise the temperature of the weldment to a level where the yield stress of the material is lowered considerably.

Figure 1 shows typical figures for the variation of 0.2% proof stress with increasing temperature for three types of steel. If we consider a weld in C: Mn steel, residual stresses of at least 18 tons/in² will be developed when it is cooled to room temperature. If the weld is uniformly heated to 500°C, the yield stress of the material around the weld will be lowered to about 9 tons/in² and it will be unable to support the initial level of the residual stresses, which will be relaxed by plastic deformation to the level of 9 tons/in². In addition,



1 The variation of 0.2% proof stress with temperature for three types of steel

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creep will take place at high temperature and strain will occur by a diffusion mechanism, relaxing the residual stresses further. If the weld is then uniformly cooled to room temperature, the level of residual stress present will remain at some level below 9 tons/in².

The extent to which residual stresses are relaxed will depend on temperature for any given material, and on material for any given temperature, as can be seen from Fig. 1. For example, low alloy steel welds usually require higher temperatures than mild steel welds to reduce their residual stresses to the same level, for reasons which will become clear later.

The use of uniform temperature distribution has been emphasised, because temperature gradients cause thermal stresses which, if they exceed the yield point of the material at the temperature at which the stress is developed, can give rise on cooling to a new set of residual stresses. This is particularly important in the application of local stress relief, which is used when stress relief of the total structure is impractical. For instance, a large pressure vessel which is too big for available furnace capacity may be constructed in two halves, each of which is stress relieved in a furnace before being welded to form the complete vessel. It is then necessary to stress relieve the final weld by a local treatment before putting the vessel into service. However, since the weld is restrained by the two halves of the vessel, the temperature distribution of the area heated must be such that gross yielding caused by thermal stresses does not occur.

In particular, the thermal stresses in areas above 500°C where the material has very low yield stress, must be kept low, especially during cooling. If care is not taken to ensure strict control of the temperature distribution of local heat treatments, the thermal stresses generated at high temperatures may give rise to a residual stress system which, although less severe than the welding residual stresses, may be sufficient to cause failure in service. In the limit, the addition of further runs of weld metal may be considered as a local heat treatment, but the associated temperature distribution during cooling creates new residual stresses, and multirun welds require the same degree of stress relief as single run welds.

In addition, thick plate requires longer times at temperature than thin plate in order to ensure equalisation of temperature through the thickness, since it is important for the whole of the structure to be at the required temperature for the correct time.

It will be seen that residual stresses can be removed by raising material to a temperature at which its yield stress is very low, while maintaining a temperature gradient low enough to prevent gross plastic strain at high temperatures. The residual stress level remaining when the material is cooled to room temperature will be somewhat lower than the yield stress at the temperature of stress relief as a result of further relaxation of stresses by creep processes at elevated temperatures.

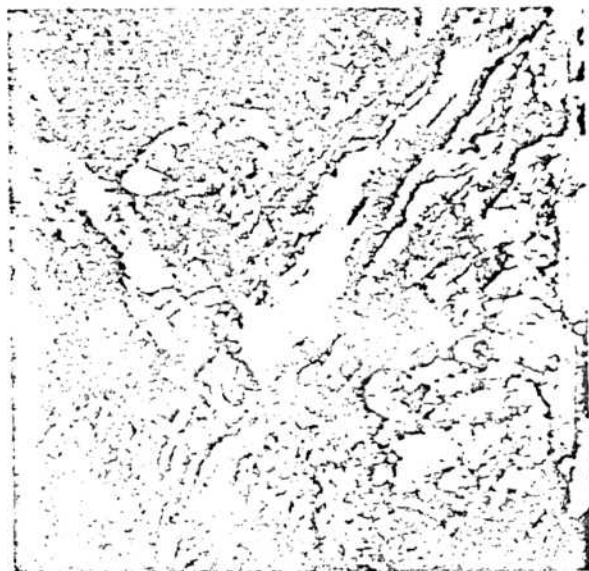
METALLURGICAL CHANGES OCCURRING DURING WELDING

When a ferritic steel is heated above its critical temperature it transforms into a high temperature form of iron known as austenite. Normally, austenite is not stable at room temperature and, on cooling, it transforms to other, more complex microstructures. The exact microstructure formed depends largely on the cooling rate, once the composition of the steel has been fixed. In general, during the heating process, various interstitial and alloying elements which formerly existed as carbides or nitrides, are taken into solid solution. In particular, carbon and some manganese in C-Mn steels, and in addition to these two, chromium, molybdenum, vanadium and niobium in alloy steels, become dissolved in the austenite. If the subsequent cooling rate of the austenite is not slow enough to allow time for these elements to come out of solution to form carbides and nitrides, as in the original condition of the material, they form a supersaturated solid solution in the iron matrix. This puts the iron lattice under high internal strain which, coupled with transformation, usually increases dislocation density as illustrated in Fig. 2, thereby raising yield strength and hardness.

In the particular case of welding, the material adjacent to the weld metal—i.e., the heat affected zone (HAZ)—undergoes a very rapid heating and cooling cycle which, depending on the severity of the quench and the composition of the steel, can give rise to different microstructures, ranging from twinned martensite, auto-tempered martensite and lower bainite in very hardenable steels, to upper bainite and pearlite in less hardenable steels. In general, the greater the quantity of alloying elements a steel contains, the more hardenable it will be. Since the temperature gradient around the weld will vary with distance from the weld pool, the cooling rate will also vary and so give rise to differences in microstructure across the HAZ. However, these microstructural variations across the HAZ will, in general, be quite small. Thus, the result of welding is to transform a volume of metal around the weld into material with properties which can be very different from those of the original steel.

In addition to the transformed HAZ, there are two other regions of importance in the vicinity of

DISLOCATION
STRUCTURE
TRANSFORMED
OUT OF MILD
1250°C



a weld. The first is the weld metal itself, which consists of the filler material from the electrode or welding rod diluted by melted parent material, the dilution being greatest adjacent to the fusion boundary. As it cools and solidifies, the weld metal will become austenitic in the appropriate temperature range. As with the HAZ, most of the elements present in the weld metal will be in solid solution in the iron matrix, and it will undergo a quenching rate similar to that of the HAZ. It is normally found that weld metal with mechanical properties matching those of the parent plate has a chemical composition different from the parent plate. It is thus to be expected that microstructures occurring in weld metals will differ from those of the HAZ, and it has been noted that the microstructures developed in weld metals cannot be classified in terms of the common structures to be found in the HAZ^{4, 5}.

The second region outside the transformed HAZ is that which is heated to temperatures above 100°C, but below the lower critical temperature. In this temperature range, dynamic strain-ageing can occur as a result of the combined thermal and strain cycle. If a notch, crack, or similar strain concentrator is present in this region of the HAZ, the thermal strain is concentrated at its tip and a high dislocation density may be generated locally. This effect is illustrated in Fig. 3. Such regions at the tip of a strain concentrator have been shown to have greatly reduced resistance to fracture initiation^{6, 7}.

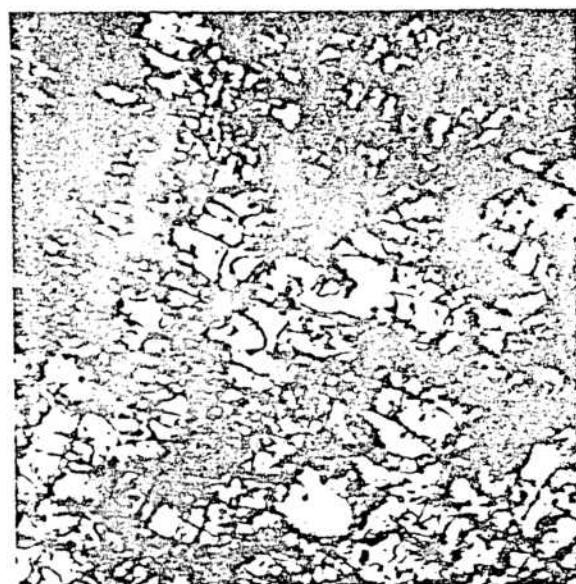
METALLURGICAL CHANGES OCCURRING DURING POST WELD HEAT TREATMENT

Dislocation movement

When the temperature of ferritic material is raised, the reason for the decrease of the yield stress of the iron matrix is that the increased mobility of the iron atoms allows easier movement of dislocations. If the material contains residual stress, raising the temperature allows the dislocations to move more easily, lowers the yield point and causes the material to strain plastically, thereby relaxing the residual stresses. At higher temperatures, diffusion mechanisms become prominent and the process of recovery, involving the annihilation of dislocations and the formation of small cells or sub-grains within the larger grains, is accelerated. All these processes may be grouped under the general heading of creep, although this term is usually reserved for processes involving a diffusion mechanism. It is the ability of creep processes to effect plastic strain below the flow stress at temperature, that enables residual stresses to be reduced below this level.

Precipitation

If carbon is in supersaturated solid solution in the weld metal or HAZ, carbon atoms will diffuse, during the post weld heat treatment, to suitable nucleation sites within these areas, such as dislocations, and will form carbides. At temperatures below 450°C the carbide formed in C:Mn and low alloy steels is usually a form of iron carbide known as cementite, with some iron atoms being replaced by manganese, or perhaps chromium if present. If the transformation



3 The increase in dislocation density produced by thermal strain concentration in a mild steel

(a) Unaffected parent plate, x 30X

(b) Material within 0.005 in. of the tip of a notch which reached a peak temperature of 550°C x 30X

product is martensite, then in the early stages of tempering (up to 200°C), platelets of a form of iron carbide known as epsilon carbide are produced. Each carbide platelet grows in a particular orientation with respect to the matrix in a crystallographic plane with which it has a nearly identical lattice parameter. Because the lattices do not match exactly, a coherency strain field is set up which takes up the mismatch between the lattice spacings of the particle and the matrix, which are then said to be coherent. This coherency strain field, combined with the small size of the particle, makes it an efficient dislocation barrier.

As a result of this precipitation, dislocation movement is restricted and the yield stress and hardness of the HAZ would be expected to increase. This would be the case, if it were not for the fact that the yield stress and hardness of the matrix are being reduced by the removal of alloying elements from solid solution. The net effect of precipitation is usually to produce a retardation in the softening process. Above 250°C, a second tempering stage sets in and overageing of the carbides occurs. As the size of the carbides increases, coherency (i.e., matching of the matrix and precipitate lattices by the action of a strain field), becomes impossible, and a massive form of



5. (a) Precipitation of vanadium carbide in the transformed HAZ of a weld in 1Cr, 1Mo, 1V steel after tempering $\times 45K$
(b) Precipitation of molybdenum carbide in the transformed HAZ of a 2Cr, 1Mo weld metal after tempering $\times 35K$



cementite is observed. These more massive particles are less efficient dislocation barriers, and so the yield stress and hardness decrease as more carbon is removed from the matrix.

If bainite or pearlite is present in the HAZ, massive cementite is already present and tempering produces further precipitation of cementite, resulting in a decreased overall yield stress. For certain materials and cooling rates, austenite may be retained in the HAZ or weld metal at room temperature and, if present, will decompose to ferrite and iron carbide during the early stages of tempering up to 300°C . As the temperature is raised above 300°C the cementite particles coarsen in size, particularly at grain boundaries, with large particles growing at the expense of smaller ones. The particles also tend to become spherical in shape and large platelets will often break up into spheroids, although this process of spheroidisation is not appreciable below 350°C .

At higher temperatures ($> 450^{\circ}\text{C}$) the presence of alloying elements such as Cr, Mo, V or Nb gives rise to other tempering reactions, and carbides of these elements are nucleated on dislocations or existing carbide particles and grow at the expense of the cementite particles. This process is usually referred to as secondary hardening. Initially, coherent precipitates are formed, and these, by preventing dislocation movement, can give rise to cracking problems, as will be seen later. Higher temperatures (between 600°C and 700°C) cause overageing of these alloy carbides, and softening of the HAZ is effected. The temperature at which this latter process occurs depends critically on the chemical composition of the steel as is shown in Fig. 4. Examples of the type of precipitation that occurs in alloy steels after welding is shown in Fig. 5. It is the occurrence of precipitation and secondary hardening phenomena in low alloy steels which makes it necessary to use higher temperatures, to relieve residual stresses in such steels, than is the case for mild steels.

From the foregoing paragraphs it will be appreciated that welding gives rise to a volume of material which is often inferior to the original steel, because of its greater yield strength and hardness and reduced ductility compared with the latter material. Heating to elevated temperatures allows metallurgical changes to occur which cause softening of this HAZ, but in the case of alloy steels this softening may be retarded by precipitation processes. Hardened structures generally have poor ductility, so that the softening of such structures is a very important concomitant of post weld heat treatments.

PRACTICAL CONSIDERATIONS - WHEN IS THERMAL STRESS RELIEF NECESSARY?

C:Mn steels

Thermal stress relief of C:Mn steels is not always necessary, since HAZ microstructures can often be controlled by heat input and preheat if required, unless the steel is very hardenable, and, provided that the HAZ is adequately ductile, the presence of yield point residual stresses can usually be tolerated in the final weldment unless distortion during machining is a potential pro-

blem. However, stress relief is required in critical applications such as pressure vessel construction, where it can greatly reduce the risk of catastrophic failure by brittle fracture. In particular, thermal stress relief is the only treatment that will soften hardened HAZ structures and eliminate strain ageing embrittlement of the type illustrated in Fig. 3, if present. Stress relief is also necessary when there is a risk of stress corrosion cracking in subsequent service, and thermal stress relief, with its accompanying metallurgical benefits, is the best treatment in this case, because the change in metallurgical structure produced in ferritic steels usually increases the corrosion resistance.

Low alloy steels

In these rather more complex steels, thermal treatments at elevated temperatures can have both beneficial and adverse effects on properties, and if heat treatment is to be applied to them it is important to know which effects operate.

Most low alloy steels are designed to be used in the normalised and tempered, or the quenched and tempered, condition. It can therefore be an economic advantage to weld these steels in the normalised or quenched condition and subject the whole weldment to a combined stress relieving and tempering treatment. This is not always possible or convenient, in which case the weldment is made in fully heat treated material and put into service without further treatment after welding, other than a local stress relief treatment applied to the weld if necessary. However, while heat treatment in the tempering range may be beneficial for the unwelded plate material, it may be very bad for the weld HAZ.

At elevated temperatures there are several processes operating within the HAZ. Firstly there are the processes of slip and creep, which allow dislocations to move and the matrix to strain. These result in a reduction of the matrix yield stress and of the level of residual stresses. Secondly, there is the precipitation of alloy carbides which comes into operation at temperatures above 450°C. In the initial stages of tempering, these carbides resist the movement of dislocations by slip, and the process of stress relaxation is slowed down. Once temperatures greater than 600°C are reached, dislocations are able to climb around the precipitates⁶ and the creep process is allowed to take place.

The balance between these various processes varies with material and temperature, but at temperatures below 600°C the usual effect is to strengthen the HAZ matrix. The residual stresses are still high in this temperature range, so that the strain necessary to reduce the residual stresses is transferred to the prior austenite grain boundaries, which may subsequently crack during treatment. An example of this type of cracking is given in Fig. 6. Figure 7 shows in greater detail the morphology of this type of cracking, produced in a fatigue-cracked laboratory specimen tested at 600°C under decreasing load. This phenomenon is known as post weld heat treatment cracking and is a considerable problem in low alloy steels, in particular the ferritic creep resisting steels, which, by definition, have high resistance to

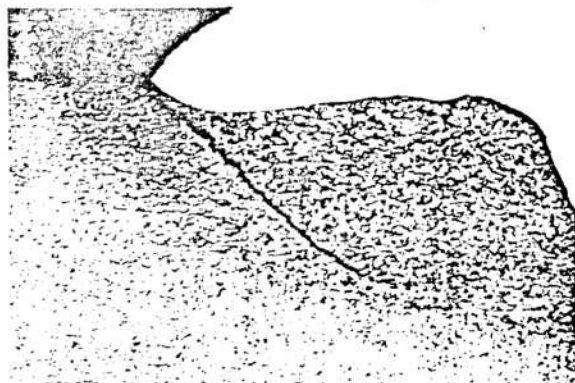
creep at elevated temperatures. It is most commonly found in vanadium bearing steels.

Austenitic materials

In practice it is generally observed that stress relief treatments on this class of material fall into two groups. Low temperature treatments up to 750°C which effect partial stress relief, and high temperature treatments above 950°C, which are designed to give almost complete stress relief.

Low temperature treatments are used when dimensional accuracy is required for a machining operation as discussed earlier. This problem is more common in austenitics than in ferritics, because the former develop much more severe thermal strains adjacent to the weld metal. A low temperature stress relief treatment after welding may be sufficient to avoid this trouble.

In principle, high temperature treatments can be used to eliminate stress corrosion cracking in service, although, in practice, the problem is usually overcome by altering the material or the environment. The main use of such treatments is to avoid reheat cracking in components for high temperature service. This is similar to the stress relief cracking observed in high strength, low alloy ferritic steels. As with the latter materials, cracking can occur during the stress relief treatment, and care in the choice of stress relief temperature, heating rate and temperature distribution must be taken.



6 Intergranular cracking formed during heat treatment of a weld in 3Cr-1Mo-1V steel made with austenitic weld metal. The crack originated at the toe of the weld bead



7 Morphology of reheat cracking produced in a simulated HAZ, tested under conditions of stress relaxation at 600°C x 85

As with the ferritic steels, thermal stress relief can have adverse effects on the properties of austenitic steels as well as beneficial ones. For example, the use of low temperature treatments can increase the susceptibility of stabilised austenitic steels to knife-line attack adjacent to the fusion boundary, while extended treatments at temperatures between 600°C and 880°C can produce sigma phase in weld metals containing ferrite. The sigma phase is brittle at room temperature and could give rise to failure in service. A further problem experienced with unstabilised 18-8 austenitic stainless steels during heat treatment is chromium depletion adjacent to grain boundaries, associated with precipitation of chromium carbide in the boundaries. This can give rise to grain boundary corrosion in service following treatments between 500°C and 900°C.

Weld metals

The difficulty in classifying weld metal microstructures, in terms of those microstructures usually met with in the HAZ, has already been mentioned. However, it would be expected that the same processes of creep and recovery that arise in the HAZ would occur in weld structures when heat treated.

In contrast to the HAZ, where heat treatment produces beneficial effects in the majority of cases, the properties of ferritic weld metals tend to deteriorate when heat treated⁵. In particular, fracture toughness as measured by the Charpy test, is observed to decrease on heat treatment in a large number of weld metals. Care must be taken in such cases to ensure that the heat treatment necessary to obtain adequate HAZ toughness does not give rise to a brittle weld metal.

The problem of heat treatment cracking is not commonly found in weld metals. One of the main reasons for this is that susceptible steels (usually vanadium bearing) are normally welded with non-vanadium bearing weld metals. For example, $\frac{1}{2}\text{Cr}:\frac{1}{2}\text{Mo}:\frac{1}{2}\text{V}$ steels are most frequently welded with $2\frac{1}{2}\text{Cr}:1\text{Mo}$ weld metal, because the use of the latter, instead of creep resistant $\text{Cr}:\text{Mo}:\text{V}$ weld metal, reduces the risk of heat treatment cracking in the HAZ. Apart from a possible deterioration in toughness, weld metals in general do not pose any problems when heat treated, although reheat cracking problems, associated with $2\frac{1}{2}\text{Cr}:1\text{Mo}$ weld metals in the welding of thick plate, have recently been experienced, and $5\text{Cr}:1\text{Mo}$ weld metals do suffer from this problem.

Non-ferrous metals

The use of post weld stress relieving thermal treatments on non-ferrous weldments is very rare. They are used to relieve residual stresses in aluminium, brasses and bronzes, where stress corrosion cracking is likely to be a problem in service. In this case the temperature of treatment is about 250°C. Post weld heat treatments may be necessary in non-ferrous alloys for reasons other than relief of residual stresses, e.g., to develop adequate strength in the joint.

PRACTICAL PROCEDURES

Where post weld heat treatment cracking is a

potential problem, it is most important to choose the correct temperature range for the stress relief/tempering treatment, i.e., that temperature at which maximum relief of stresses is achieved without overageing of the material. However the problem may still remain because the maximum heating rate that can be used is slow enough to cause cracking before the optimum temperature is reached. In such cases it is possible to adjust the heat treatment procedure so as to hold the weldment for as little time as possible in the critical precipitation temperature range. For example, a common procedure is to soak the weldment at a temperature between 400°C and 450°C until an even temperature distribution is achieved, and then heat to the stress relief/tempering temperature as quickly as possible. An additional precaution in critical cases is to use a weld metal of the lowest possible strength, or to butter the weld preparation with low strength weld metal. Since stress relief cracks usually initiate from other defects such as hot tears, the grinding of weld toes before heat treatment can be beneficial, because this will remove a large number of potential initiation points at the position of greatest stress concentration in the weld. The example of a heat treatment crack shown in Fig. 6, which is seen to originate from the toe of the weld, is very typical.

Temperature control during heat treatment is of prime importance, particularly in alloy steels where precipitation processes are critically dependent on temperature. It is very necessary to ensure an even temperature distribution, which must be done by monitoring various remote parts of the weldment during the treatment. In particular, both the top and bottom of the structure must be monitored. It is an established fact that a number of failures in low alloy steels can be attributed to the metallurgical effects of stress relief/tempering treatments that were not appreciated prior to the failure.

Once heat treatment of a welded component is complete, no further welding should be carried out, otherwise the benefit of the treatment may be completely eliminated, especially in respect of brittle fracture. In particular, there must be no further additions, such as ladder attachments, to ease the job of installation. If these are required, they must be added before the treatment and properly designed into the structure, so that they can remain on it during service and do not cause stress concentration.

OTHER POST WELD STRESS RELIEVING HEAT TREATMENTS

Solution treatment

The use of solution treatments after welding followed by annealing, normalising or quenching is very rare, except perhaps on small, easily handled components. Quenching a welded component of any appreciable size will introduce a new residual stress system. Normalising treatments are occasionally used on large steel components welded by the electroslag process. This has the effect of refining the grain size in the HAZ regions adjacent to the fusion boundary and also in the weld metal, thus improving their fracture toughness. Extreme care has to be exercised in

the application of solution treatments, particularly in large steel constructions; at the temperatures involved, the material has very low strength and, with inadequate support, the weight of the component could cause plastic deformation or even collapse. The main aim of all these treatments is to improve the metallurgical condition of the material after welding; stress relief is just a bonus obtained in the process.

CONCLUSIONS

The relief of residual stresses by the use of a thermal treatment is often a necessary adjunct of welding, to ensure adequate service performance. Before being applied, it is important to know the effects of heat treatment on the material used, and equally important to ensure that no further welding or heat treatment is carried out after the main treatment without adequate knowledge of its effects on service behaviour. In no circumstances should heat treatment be applied without established reasons.

In certain instances the use of a thermal stress relieving treatment is unnecessary. In other cases it is inadvisable, in particular when post weld heat treatment cracking is unavoidable or when material is welded in the optimum aged condition and further treatment causes overageing. In such cases the structure is best put into service in the as welded condition, or residual stresses must

be reduced by mechanical means, as will be described in the next article in this series.

ACKNOWLEDGEMENTS

The author would like to thank colleagues at The Welding Institute, in particular Mr. R. E. Dolby and Dr. F. M. Burdekin, for helpful discussions. He is also indebted to Mr. K. P. Bentley for the photographs used in Fig. 5 and 6.

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The force associated with the tensile part of this stress distribution is balanced by an equivalent compressive force which gives rise to a compressive stress distribution of equal area to the tensile stress distribution.

If an external tensile load is now applied, local yielding will occur where residual stresses are of yield point magnitude. Further application of load which gives an increase in stress will result in a local stress distribution as shown in Fig. 1b, in which the areas of elasticity and plasticity are indicated. As the externally applied load is increased, the load carrying capacity of the yielded region will be reached, and the load is shed to the elastically loaded part of the structure. The amount of local strain which occurs will be related to the magnitude of the overload and the post yield characteristics of the material. With the release of the externally applied load the pattern of the stress distribution will remain unchanged and it will return to a condition in which the mean stress level is zero. The release of the applied load results in a stress distribution

which now has a lower maximum value than before, as shown in Fig. 1c. Thus, the high peak values of local residual stress have been redistributed to give a much lower peak value.

From the above description of the principles of mechanical stress relief it can be seen that the amount of stress relief achieved increases as the external load is increased. Hence, the optimum conditions can be obtained by using high external loads, provided that these loads do not impair the service performance of the structure. In general, externally applied loads will be less than the corresponding yield load. To achieve an acceptable level of stress relief the structure has to be subjected to loads which will be in excess of loads likely to be encountered during service and, since nominal stresses will exceed the permitted working stress, this procedure should be applied with some caution. Also, since the stress relief is caused by yielding, resulting from the additive effect of two stress systems, it is only possible to achieve a measure of stress relief when a tensile overload is applied to a tensile residual stress system.

The relief of residual tensile stresses will be an advantage only in applications in which the service load produces tensile stresses, and if they are to be effective, overloads must be applied in the same direction as the working loads.

EFFECT OF A PRE-EXISTING STRAIN CONCENTRATOR

A defect, crack, or geometric discontinuity existing before the overloading will act as a strain concentrator during the process of mechanical stress relief. Larger amounts of local plastic deformation, which are contained by surrounding elastic material, will occur at the tip of this defect. When the external loads are released the elastic stress field will impose a compressive stress around the crack tip as shown in Fig. 2. This can be illustrated by considering the output of a strain gauge attached adjacent to a strain concentrator. If, for example, a pressure component is considered, then the output from the strain gauge will be as shown by the line 1-4 in Fig. 3a. As the pressure is applied the strain increases proportionately up to the point 1, where yielding commences. Continued increases of applied pressure result in a non-linear increase in strain to the point 2, the test or overload pressure. Release of the applied pressure produces elastic unloading along the line 2-3 to give a no load, permanent, strain condition at point 3. Working pressures will result in strain/pressure traces up and down the line 3-4.

If we now consider the local stresses associated with this loading cycle as in Fig. 3b, it can be seen that up to point 1 the stresses increase in proportion to the strain. Because the amount of permanent strain is small it is reasonable to neglect work hardening, and for the process 1-2 there is no increase in stress. The elastic modulus E has not changed and elastic unloading from 2-3 gives the location of the point 3.

Thus, not only has the peak value of stress been reduced, but a stress of opposite sign has been introduced at the crack tip. Moreover, further applications of tensile load will not

Fig. 1a. Compressive stress around a concentrator

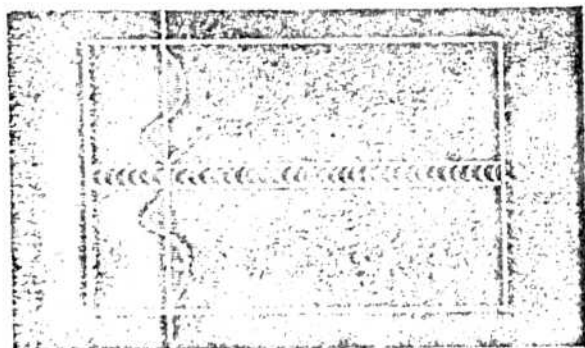


Fig. 2. Local strain distribution for a pressure component

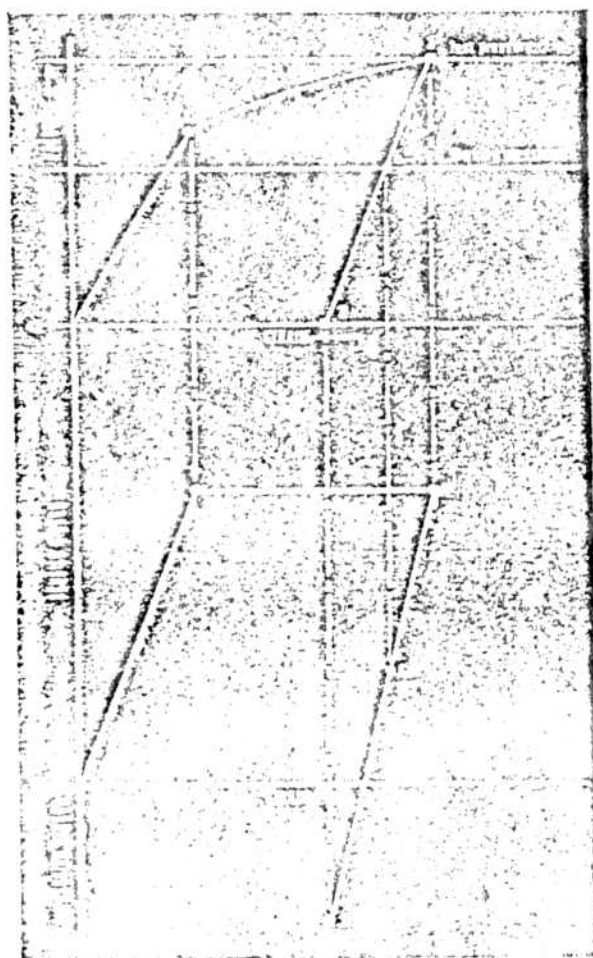


Fig. 3. Strain/pressure relationship

Part 4 Mechanical stress relief

by G. R. Egan, B.E.Mech.(N.Z.)

Where it is essential to relieve residual welding stresses, and where thermal stress relief is impractical, it may be possible to stress relieve welded constructions by a mechanical process. Mechanical stress redistribution resulting in favourable stress systems around critical regions of welded structures can also be achieved. This article explains the principles involved in processes commonly referred to as mechanical stress relief treatments.

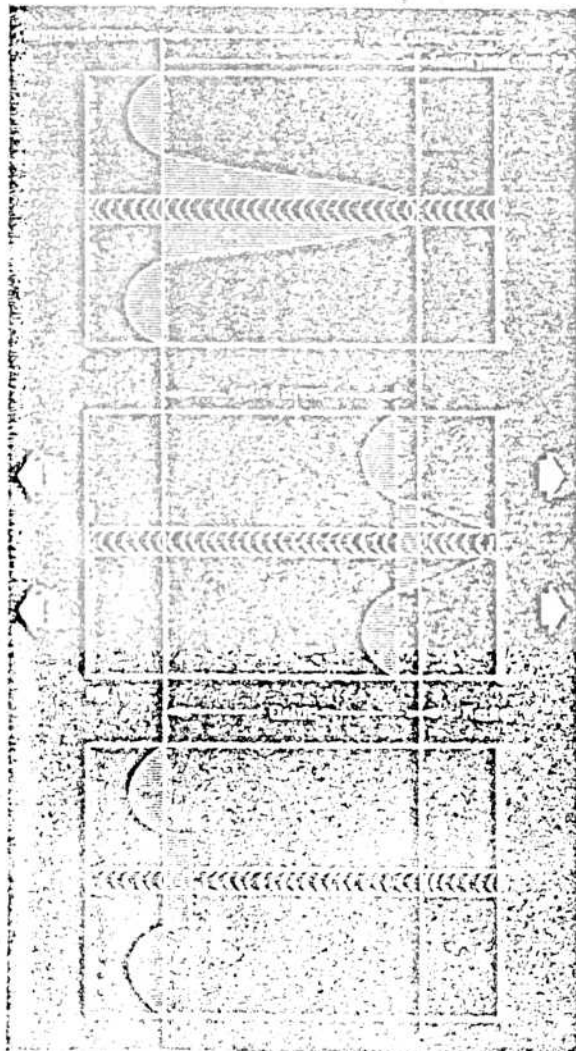
The integrity of any structure or structural member depends on its suitability to sustain the service loads to which it is subjected. For any structure the failure mode is dictated by the types of load which occur during service and the environment present when these loads are applied. For critical applications such as low temperature service, design procedures become meaningless if residual stresses resulting from welding are additionally present. Local residual welding stresses are usually of yield point magnitude¹ and any addition of tensile stress to such a system will cause localised yielding which may lead to premature failure. Residual stresses will influence such failure modes as fatigue, stress corrosion, local buckling and brittle fracture, so that in some critical applications it may be necessary to remove the residual welding stresses. Where stress relief is required and it is not possible to use thermal stress relief (*i.e.*, the structure is too large or facilities are not available on site), it may be possible to stress relieve the structure by a mechanical process.

PRINCIPLES OF MECHANICAL STRESS RELIEF

Mechanical stress relief is a process of stress redistribution during which the high peaks of the existing residual tensile stress distribution are removed by controlled localised yielding. Externally applied loads, in the same direction as the residual stress, will cause yielding in regions where a high residual stress already existed. The release of this externally applied load takes place elastically and results in a reduced peak value of residual stress.

A more detailed explanation of this process can be given by considering the distribution of local residual stresses parallel to a longitudinal butt weld as shown in Fig. 1a. The vertical line on the

right hand side of this figure represents the yield stress of the material, and it will be seen that the maximum tensile stress is of yield point magnitude, as described in the second article of this series².



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(a) The distribution of residual stresses parallel to a longitudinal butt weld (Reference 1)

(b) Stress distribution after the application of an external load

(c) Stress distribution after the release of applied external load

The force associated with the tensile part of this stress distribution is balanced by an equivalent compressive force which gives rise to a compressive stress distribution of equal area to the tensile stress distribution.

If an external tensile load is now applied, local yielding will occur where residual stresses are of yield point magnitude. Further application of load which gives an increase in stress will result in a local stress distribution as shown in Fig. 1b, in which the areas of elasticity and plasticity are indicated. As the externally applied load is increased, the load carrying capacity of the yielded region will be reached, and the load is shed to the elastically loaded part of the structure. The amount of local strain which occurs will be related to the magnitude of the overload and the post yield characteristics of the material. With the release of the externally applied load the pattern of the stress distribution will remain unchanged and it will return to a condition in which the mean stress level is zero. The release of the applied load results in a stress distribution

which now has a lower maximum value than before, as shown in Fig. 1c. Thus, the high peak values of local residual stress have been redistributed to give a much lower peak value.

From the above description of the principles of mechanical stress relief it can be seen that the amount of stress relief achieved increases as the external load is increased. Hence, the optimum conditions can be obtained by using high external loads, provided that these loads do not impair the service performance of the structure. In general, externally applied loads will be less than the corresponding yield load. To achieve an acceptable level of stress relief the structure has to be subjected to loads which will be in excess of loads likely to be encountered during service and, since nominal stresses will exceed the permitted working stress, this procedure should be applied with some caution. Also, since the stress relief is caused by yielding, resulting from the additive effect of two stress systems, it is only possible to achieve a measure of stress relief when a tensile overload is applied to a tensile residual stress system.

The relief of residual tensile stresses will be an advantage only in applications in which the service load produces tensile stresses, and if they are to be effective, overloads must be applied in the same direction as the working loads.

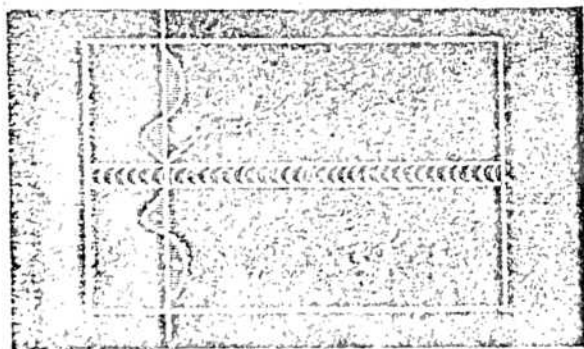
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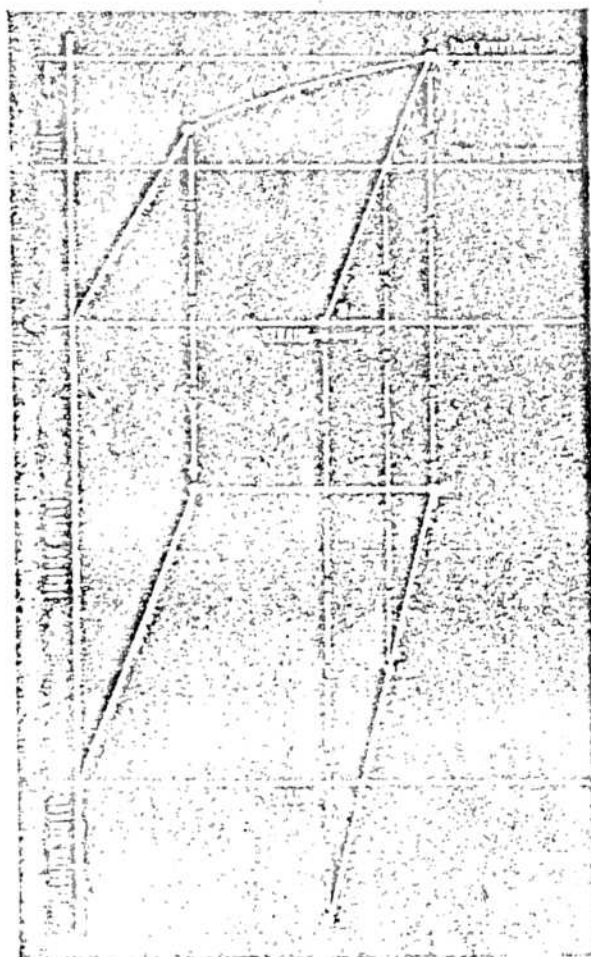
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subject the crack tip to the high tensile strains which were imposed at first, because subsequent tensile loading must first overcome the residual compressive stress. For loads of the same magnitude this also precludes any further growth of the plastic zone at the tip of the crack. This has distinct advantages because favourable stress distributions can be introduced into critical areas of a welded fabrication by the application of an overload which is in the same direction as the service load, but is greater in magnitude. Practical methods of achieving mechanical stress redistribution will be outlined later.

PRACTICAL CONSIDERATIONS

The greatest problem is to decide how much overload is necessary to achieve an acceptable level of stress relief without jeopardising the integrity of the structure. Obviously, the eventualities of buckling and plastic collapse must be avoided in the structure or structural member to which the overload is being applied. It seems reasonable to limit the stress resulting from the overload to some fraction of the nominal yield stress. Also, the prevention of brittle fracture during the overload must be considered. For materials which exhibit a brittle-ductile transition behaviour with temperature, it is best to take advantage of this and ensure that the material behaves in a ductile manner when the overload is applied. For other materials a more detailed analysis of the response of the material to the overload is required.

TREATMENTS

A measure of mechanical stress relief can be achieved simply by overloading a structure above the permitted working loads. A proof test, for example, on a structure which has not been thermally stress relieved will provide some measure of stress relief. Although this is usually regarded as an added benefit from the proof test and not the reason for it, overload tests are carried out for the specific purpose of achieving stress relief². Proof test pressures for internally loaded boiler shells for latest design principles are based on the equivalent design pressure⁴, which gives under these conditions a nominal stress in the shell of about two-thirds of the yield stress. Thus, it seems reasonable to limit external overloads for mechanical stress relief so that nominal stresses are of this order.

Bridge girders in service today have been mechanically stress relieved by jacking apart pairs of girders strapped together at the ends. Mechanical stress relief of this type has also been carried out after construction, in which case the overloads are usually limited to a value which marginally exceeds the working loads⁵. Stress relief of this type will only be effective if the girders are not subjected to reversals of load.

LOW TEMPERATURE STRESS RELIEF

Since the residual stresses adjacent to welds take up, after cooling, the difference in length between the plate material and the weld metal, which has resulted from differential expansion and plastic flow during the welding process, it seems reasonable to assume that the system can be modified

in some way by a further application of heat. This is the basis of a controlled low temperature stress relief treatment in which a zone on each side of the weld is expanded by the application of heat. The rapid heating to a controlled low temperature imposes tensile plastic strain on the weld metal, thus changing the length relationship between the plate material and the weld.

The *mechanical* overload of the residual stress system is achieved by expanding the material adjacent to the system by the application of heat. This process overloads the stress system in a similar manner to that achieved by an applied external load. To obtain an acceptable level of stress relief, the plastic strain induced in the weld metal should be of the same order as the elastic strain set up by the post weld residual stresses. Without restraint, this order of strain can be achieved with a temperature difference of about 125 Centigrade degrees, but when additional restraint is imposed by the material adjacent to the heated zones a wider temperature differential is required. Water sprays are used to cool the weld and the heated zones, but it is still necessary to heat the plate to about 200°C to achieve this temperature differential. Provided that the maximum temperature is kept below 200°C it is unlikely that any detrimental metallurgical effects will occur during this process. Low temperature stress relief has been used for butt welds, intersecting butt welds and fillet welds in ships and tanks⁶, but because it is difficult to establish the correct heating position and conditions, it is limited to relatively simple detail designs.

VIBRATORY TREATMENTS

The use of applied vibrations to obtain resonance in a structure has been described recently⁶, and it is claimed that this technique achieves mechanical stress relief. A mechanical vibrator is attached to the component or structure, and the frequency of the forced vibrations is adjusted until resonance occurs. This is maintained for a time which depends on the weight of the structure being treated. Typical times for structures of 2000 - 10 000 lb would be about 20 - 30 minutes.

No experimental results (residual stress levels after this treatment) have been provided, so the amount of stress relief achieved is unknown. Because a single application of a tensile load to a structure with residual tensile stresses will cause yielding on a local scale, as described previously and shown in Fig. 3, it seems unlikely that any subsequent application of load of the same magnitude will induce further yielding, because of the raised yield point of the material caused by work hardening. Hence, mechanical stress relief will probably be achieved only in the first few tensile half cycles of this process, and subsequent cycles of load will have little effect on the residual stress system. Also, the application of a cyclic load of a magnitude required to achieve an acceptable level of stress relief would probably introduce a fatigue problem⁷.

An earlier investigation⁸ showed that for the loads normally applied during this process the amount of stress relief was small, and even under

Conditions of resonance no marked changes in residual stress distribution were detected. Some measure of dimensional stability will be achieved by this process as a result of the removal of the peaks of residual stress by the application of a cyclic load of unknown magnitude.

DISCUSSION

Unlike thermal stress relief heat treatments, mechanical stress relief treatments do not improve the metallurgical properties of a structure⁹. In fact, material damage may occur during and subsequent to overloading if the material is susceptible to strain ageing embrittlement³. Therefore, the mechanical process should be used when stress relief is required only when it is not possible to use thermal stress relief.

Mechanical stress relief treatments do, however, induce favourable local stress distributions around notches and stress concentrators – an important effect which is not achieved by thermal treatments. This effect requires that overload treatments be applied in the same direction as the service loads, which in some cases may be difficult to define. The benefits of mechanical stress relief can only be relied upon if the working loads do not exceed the applied overload. This requires a comparison of test and service conditions to ensure that unusually high stresses, resulting from such things as supports, attachments and thermal gradients, do not occur.

Local treatments which modify existing stress distributions are useful methods of achieving better structural properties, but they should not be applied without a comprehensive knowledge of the response of the material to such treatments.

CONCLUSIONS

It is often possible to use a mechanical process to relieve residual stresses produced by welding.

Mechanical stress relief may be used, if stress relief is required, when it is not possible to use thermal stress relief treatments. Mechanical treatments cannot be used as a cheap alternative to thermal treatments to obtain full relief of residual stresses and improved metallurgical properties.

The relief of residual welding stresses can only be achieved by the application of a load in the same direction as the residual stresses, and the overload must be in the same direction as the service load. The process will be ineffective if the structure is subjected to reversals of load.

Overloads used to achieve mechanical stress relief should be limited to some fraction of the yield load of the structure or structural member.

The response of the structure should be assessed to ensure that the structure can tolerate the overload without failure.

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Applications for membership



The following organisations have applied for membership since the last list of applications was printed.

Anderson Mavor Ltd., Motherwell.
Avonmouth Engineering Co. Ltd., Avonmouth, Bristol.
Kirkstall Forge Engineering Ltd., Leeds.
Luton College of Technology, Luton.
Sarnum Welding Products, Salisbury.
Metropolitan-Carmell (Holdings) Ltd., Birmingham.
Delaney Gallay Ltd., London.
York Trailer Company Ltd., Corby.

CHEVRON CRACKING - CURRENT EVIDENCE

1. Moisture content of the flux (SAW, MMAW) is a major controlling factor. Also weld metal hydrogen correlates directly with incidence of cracking.
2. Size and incidence of cracks relate directly with alloying content and/or strength. Above a certain level cracking became vertical.
3. From MMAW with different batches of the same electrode, moisture is not the only criterion. One suggestion is that hydrogen availability may be influenced by inclusions acting as hydrogen sinks.
4. Nickel up to 3% did not increase the amount of chevron cracking, but at 4-8% and 8-7% gave large vertical cracks. Some of these may have been hot cracks.
5. In SAW metal chevron cracks were often confined to a single run, apparently stopped by upper and lower reheated regions.
6. In MMAW deposits many chevron cracks occurred in refined and intercritical regions of weld metal. These cracks occurred on or after reheating since they did not show thermal facets. Crack features were generally smaller and more regular than in as deposited weld metal.
7. In as deposited weld metal vertical components were generally associated with proeutectoid ferrite veins, though deviation occurred in many cases. These vertical components were similar to vertical microfissures.

Vertical components mainly quasi-cleavage; horizontal showed dimples.

Vertical components often radiate from a microvoid. In highest strength weld metals the vertical components showed some intergranular failure.

Some thermal facets - these were associated with reheating.

Staircase nature also evident in weld metal cracks in Y-Tekkon tests - SEM showed same fractographic features.

Zapfes planar pressure theory (1947) considered that dissolved H atoms diffuse through the lattice to voids and discontinuities, combining there to form molecular H_2 at high pressure, which resulted in premature failure. Further diffusion would be necessary to maintain these high pressures (i.e stresses), as a crack propagates explaining the time and temperature dependence of HICC. Further refinements have applied dislocation concepts. This theory is still used to explain blistering in steel and has not been completely rejected as an explanation of HICC. It should be noted that the theory runs counter to some ideas that weld metal and HAZ non-metallic inclusions act as 'sinks' for hydrogen and so reduce susceptibility to HICC.

The surface absorption theory (Petch 1952) proposed that absorption of hydrogen on potential crack surfaces lowered the surface energy making propagation easier in front of an array of dislocations piled up against a grain boundary. Dependence on hydrogen diffusion explained the influence of temperature and strainrate.

Troiano's lattice interaction theory suggested hydrogen diffusion to highly stressed regions (stress-induced diffusion), the magnitude depending upon the initial hydrogen concentration, the hydrogen diffusion rate and the time for diffusion. At a critical hydrogen concentration and stress state a crack forms in the lattice at the point of maximum triaxility. The cause is a lowering of cohesive strength of the lattice. Modifications have attempted to show that abnormally large hydrogen concentrations can occur and also the presence of 'non-Hookean' elastic stresses. Crack propagation speed is governed by the rate of transport of hydrogen.

Beacham (1972) found evidence of microscopic plasticity on the surfaces of surfaces suffering hydrogen embrittlement which, combined with existing evidence, there was no single mode of hydrogen cracking, led him to propose that hydrogen diffusing ahead of a crack tip aids whatever deformation process the matrix will allow. Thus hydrogen allows dislocations to multiply or move at reduced stresses rather than locks dislocations. Subsequently, dislocation movement has been suggested as a means for rapid transformation of hydrogen atoms forming 'atmospheres' around dislocations.

Sorag (1978) has used the Beacham approach to put forward a seven point argument for hydrogen assisted cracking based on observations in HY 130 steel.

1. Hydrogen must be present at or above a critical concentration at the crack initiation site or crack tip. This critical concentration is a function of the ratio of the local stress at the crack tip to yield stress and will depend upon microstructure. The local hydrogen concentration may greatly exceed the overall nominal concentration due to segregation at grain boundaries or dislocations (Cottrell atmospheres).
2. The critical stress level for cracking must be exceeded at the crack initiation site or crack tip and will depend upon residual stress, external stress and local stress concentration.
3. Local plastic flow occurs at the crack initiation site (or crack tip) resulting in dislocation pile ups at barriers such as inclusions or grain boundaries which deactivate the dislocation sources.
4. Hydrogen transported to the pile ups by dislocation movement will diffuse across the barrier and initiates a microcrack.
5. Once formed the microcrack absorbs the dislocation pile up and reactivates the source giving additional plastic flow and hydrogen transportation.
6. Hydrogen transportation is more rapid than by diffusion above and aids crack propagation although propagation will stop when hydrogen falls below the critical concentration.
7. At the above stage crack propagation ceases until sufficient hydrogen is transported for further cracking.

It should be noted that the above argument does not explain the process which hydrogen actually assists cracking. However it is implied by Savage that the hydrogen reduces the 'bonding energy' at the interface. If the bonding energy is reduced sufficiently a crack can form without plastic deformation (dislocation movement) but at higher levels deformation will occur before crack formation.