PLASTIC PROPERTIES OF F.C.C. ORDERED SOLID SOLUTIONS+

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+Progress Report

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N O T E

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Contents

I. INTRODUCTION

A number of theories have been proposed to explain the plastic behaviour of ordered solid solutions. Most of them try to explain the observed data on yield stress, but very few deal with the work hardening behaviour of such alloys.

The present review is intended to correlate the different theoretical and experimental data published up to now, and to comment on future research on both the yield stress and work hardening behaviour of some ordered f.c.c. lattices.

II. THE YIELD STRESS

In a solid solution where Long Range Order (LRO) exists there are, in general, equivalent lattice points occupied by the constituent atoms of it. For example, in β brass (body centered cubic) the (000) and $(\frac{1}{2}\frac{1}{2}\frac{1}{2})$ crystallographic sites are either occupied the former by Cu and the latter by Zn, or viceversa. The symmetry of the lattice is reduced, then, to simple cubic. There is a possibility that two regions - one with the Zn atoms in (000) sites and the other in $(\frac{1}{2}\frac{1}{2}\frac{1}{2})$ sites - coexist within the same ordered crystal. The two are then out of phase in what in respect to the position occupied by atoms of the same species. The limit between the regions being one where the bonding takes place between like atoms, is therefore a region of high energy. Each type of ordered region is known as a <u>antiphase domain</u> (APD) and the limits between them <u>antiphase domain boundaries</u> (APB). It is generally assumed that the LRO takes place by the growing of the APD, the APB's diminishing

progressively. A well annealed crystal shall have, as we will see below, APB's only in association with the existent dislocations. These domains have been observed in several ordered solid solutions, by means of electron microscopy by Marcinkowski and co-workers (1-6) and Pashley and Presland (6,7).

The movement of a normal dislocation in a well ordered lattice produce a change in bonding pairs creating APB's in its wake. Taking .016 e.V. as an approximate value for the energy involved in the change of bonding, the surface tension γ associated with the APB is of the order of 70 dynas/cm. The shear stress necessary for the movement of the dislocation is therefore:

$$\tau = \frac{\gamma}{b} \sim 3 \times 10^9 \text{ dynas/cm}^2$$

where b is the Burgers vector.

Apparently, Koehler and Seitz (8) were the first to suggest that the energy of the system could be lowered if the movement of dislocations, in the ordered lattice, occurred in pairs in the same slip plane. The second dislocations being able to restore the order to the region of the slip plane where the first one has produced disordered in its wake. An equilibrium situation is thus created between the elastic strain forces of the two dislocations involved and the surface force of the APB between them. This configuration is called a "superdislocation" (9) and its components "superpartials" (10), to differenciate them from normal partials of extended dislocations in the f.c.c. lattice. "Superdislocations" have been observed by Marcinkowski and co-workers in ordered Cu_3Au (2), (4), (5), CuAu (4), Ni_3Mn (1), Fe_3Al (1), (3). The equilibrium separation of superpar

tials has been calculated by: Hermann and Brown for β brass (10), Flinn for f.c.c. and b.c.c. structures (11) and by Marcinkowski et al. in the above mentioned papers.

But the introduction of the "superdislocations" did not explain the observed behaviour of the yield stress. Cottrell (12) pointed out that the existence of APB's can produce an increase in yield strength under certain conditions. For a given annealed domain size, the movement of superdislocations produces an increase in the quantity of APB's by slip intersection. This will make more difficult a further movement of the "superdislocations", and as a result the yield strength increases. As domains grow bigger by annealing, the probability for intersection diminishes, therefore, decreasing the yield strength after having passed throught a maximum. According to Cottrell the yield stress τ is related to the mean diameter of the domains ℓ by the equation

$$\tau = \frac{\gamma}{\ell}\left(1 - \frac{\alpha a}{\ell}\right)$$

with a maximum yield stress for $\ell = 2a\alpha$

$$\tau_m = \frac{\gamma}{2\alpha a} \qquad (\ell \sim 40\ \text{Å})$$

Here γ is the surface tension associated with the APB, a is the interatomic specing and α a constant ($\sim$6) which depends on the geometry of the domains.

Logie (13) has made a more detailed calculation of the structural change produced by slip, obtaining similar results to those of Cottrell. Experimentally, the above mentioned predictions have been tested by Biggs and Broom (14) and by Ardley (15). A maximum in the yield strength was found by the former authors for domain size of approximately 50 Å in

diameter. On the basis of the results of Biggs and Broom Logie's calculation gives about two atomic distances for the width of APB.

The results of Ardley for single crystals also show again a maximum yield stress in the region of 15-30 Å domain diameter. Similar characteristics were observed in β brass by Ardley and Cottrell (9).

Finally several other models have been proposed to explain the yield stress behaviour in some particular alloys. These are:

a) Change in the type of crystallographic structure

Jaumot and Sawatzki (16) studied the Cu-Pd system, where a transformation from f.c.c. to tetragonal lattice takes place or ordering.

In this case the observed increase in hardness is explained as a result of the internal stresses set up by the misfit produced by the tetragonal phase in the disordered f.c.c. phase. These internal stresses interfere with the movement of dislocations. Extreme cases would be the f.c.c. or tetragonal pure phases, a maximum in hardness would correspond to maximum misfit in an mixed structure.

b) Change in lattice parameter

Ordering decreases the distance between nearest neighbors, increasing the number of unlike atom at the same time.

Sumino (17) calculates the interaction between the elastic stress field of dislocations and ordering. He points out that in places where the elastic field

tends to increase the atomic distance, ordering decreases, and additional energy is necessary for the disorder produced by the dislocation movement. A maximum in yield strength is predicted at temperatures near the critical ordering temperature. The movement of screw dislocations is easier than that of edge dislocations in the ordered lattice, because of the difference in the strain energy.

c) Different types of APB

Brown (18), (19) has studied in detail the effect of APB's on deformation. He argues that APB produced by slip are different from those produced at thermal equilibrium, the former being sharp, while the latter are diffuse. Because of this difference the second "superpartial" fail to reorder completely the disorder produced by the first. This diffused boundary will became sharp at O °K. The stress to move the "superdislocation" in the β brass type superlattice is calculated by Brown as:

$$\tau = NbTS_o^2/4b$$

where N is the density of atoms in the APB, K is the Boltzmann's constant, T is the absolute temperature and S_o is the initial degree of order.

The maximum stress is found at an intermediate state of order, because for $T > Tc$ there is no LRO, and at $T = 0$ there is no difference between both types of APB.

III. EFFECTS OF ORDER ON THE WORK HARDENING

Various investigators have reported an enhanced work hardening rate in the ordered alloys (11), (20) - (23).

Flinn (11) has made a detailed calculation of the energy associated with APB's in both f.c.c. and b.c.c. lattices. According to his results, the configuration of minimum energy in general is such, that the common plane of the superpartials is not a slip plane. In the case of ordered $L1_2(Cu_3Au)$ type of structure he concludes:

i) For pure screw dislocations, the common plane changes from (111) to (001), lowering therefore the APB energy because of vanishing neighbors energy.

ii) For dislocations in the $[2\bar{1}\bar{1}]$ direction maximum reduction occurs if the common plane changes from (111) to $(1\bar{1}3)$. As the elastic term appearing for this orientation has a minimum in (011), the equilibrium position will be between $(1\bar{1}3)$ and (011) planes.

iii) Pure edge dislocations in $[11\bar{2}]$ direction have a minimum in surface tension in (111) but the corresponding elastic energy has a maximum there. This is a position of metastable equilibrium for a hump is found between the relative minimum in energy at (111) and the absolute minimum in $(1\bar{1}0)$.

Similar results were obtained for B_2 (β brass) type of structure.

On the basis of the above considerations Flinn analyzes three cases for deformation:

a) Low temperature (no diffusion)

Before deformation takes place, dislocations are either at stable or metastable positions, because those in unstable positions dissapear during ordering, while diffusion is possible. Only in orientations near to 112 (pure edge) will both "superpartials" be on the same slip plane. This "superdislocations" or portions of them with this orientation will be preferred to move and will became sources by the Frank-Read mechanism.

The principal obstacle to the movement of dislocation in this model of deformation are the APB.

b) Intermediate temperature

Portion of dislocations will now be able, by diffusion, to change to a more favourable configuration. The creation of new APB makes further deformation difficult. The flow stress should increase with temperature in this range.

c) High temperature

Again "superdislocations" will be in stable or metastable position. The flow stress will be controlled by the movement of "superdislocations" which do not have their common plane on the slip plane, creating a viscous flow.

The rapid work hardening was attributed by Flinn to the reduction of the antiphase domains by slip intersection, the newly created APB making further deformation more difficult. In this case the work hardening rate should be dependent on the initial domain size.

Marcinkowski and Miller (22) have observed in Ni_3Mn that ordering produces an increase in the stacking fault energy. The immediate consequence of this is that cross-slip and double cross-slip should be easier in the ordered than in the disordered alloy. Further discussion on this work is given below.

Strain aging tests were performed by Vidoz, Lazarevic and Cahn (23), particularly in Ni_3Fe alloys. They found not only an increased rate of work hardening for the ordered phase, but also that the work hardening is greater, the greater the previous deformation.

Taoka et al. (24), (25), using electron microscopy and Vidoz et al. (23) in optical (phase contrast) microscopy, have shown in a series of alloys that cross-slip, both, simple and double, occurs in the disordered structure. The ordered alloy exhibits evidence of fine slip, with a regular spacing but no cross slip. Taoka and co-workers concluded that this type of slip is characteristic of the bulk specimen, being homogeneous, that is, with deformation localized in each band.

In order to explain the differences in work hardening between ordered and disordered alloys, Vidoz and Brown (10) developed the following model:

The "superdislocations" will became jogged on its intersection with forest "superdislocations". These jogs will, in general, not be aligned. If a jogged "superdislocation" moves, it will produce tubes of APB. An additional energy will be needed, therefore for the tube for

mation. The energy associated with the APB ribbon is

$$E_{ribb} = (\lambda/n)\,\tau\, b^2 \quad ; \quad \lambda = \text{const.}$$

independent of the degree of order and the domain size. If the same jog density is present in both the ordered and disordered lattices the difference $\Delta\tau$ in the flow stress of both types of lattices at a given strain should be proportional to $\varepsilon^{\frac{1}{2}}$, which is in agreement with experimental results.

The higher the prestrain, the higher the number of forest dislocations, therefore increasing the number of intersections per unit strain and therefore the work hardening. Based on this model Pampillo (26) has been able to stablish the relation

$$R = \Theta_{II}\text{order.}/\Theta_{II}\text{disor.}$$

between the work hardening coefficients of the ordered and disordered alloys. The constant values of R found in the polycrystalline case would indicate that the same process is responsible for the differences in work hardening in all the alloys investigated.

On the other hand, Hordon (27) presented some data for Cu_3Au that would indicate that domain size does have an influence on the strain hardening. But the domain sizes given by the author (at which the effect is observed) seem to be much too big, even thought his samples were annealed for very long time. Marcinkowski (1) observed Cu_3Au domain sizes of the order of 500 Å in contrast to the values of $2{,}5.10^3$ - 10^5 Å given by Hordon.

IV. SOME PREVIOUS RESULTS WITH THE ALLOYS TO BE USED IN THE PRESENT INVESTIGATION.

1) Ni_3Fe

A superlattice of the $L1_2$ type (Cu_3Au) exists the Ni/Fe alloy at compositions near to Ni_3Fe. The critical temperature for ordering is 503-506 °C. Maximum domain size is given as 30 Å by Lyashenko et al. from neutrondiffraction data while the data of Wakelin and Yates (28) seems to indicate a domain size of a few hundred angstroms.

The ordering kinetics have been studied in detail by Iida (30), (31) using calorimetric analysis. His results indicate that the process of ordering takes place first as Short Range Ordering (SRO) for annealing times up to 10 hours, LRO develops for longer annealing times to up 60 hours at 490 °C. No changes in internal energy are found to take place for longer annealing periods, although structure dependent properties still show changes for annealing periods well over the 60 hours. The author assumes that another type of mechanism is responsible for such changes.

Creep tests were performed on both the ordered and disordered alloys by Suzuki and Yamamoto (32). They observed an increase in the activation energy with increasing degree of order, a discontinuity taking place in the creep rate versus 1/T

curve near the critical temperature. Creep rate is accelerated below Tc for both the ordered and disordered states. From their results the authors concluded that a climb mechanism is rate controlling, the acceleration of creep being due to an enhanced climb rate because of ordering. At the same time, vacancy diffusion is more difficult in the ordered phase, being preferential in the slip bands. This process, in the authors point of view, accounts for the observed dependence on stress.

The above conclusions are disputed by Davies (33) who concluded, from similar evidence (log. creep rate versus log. stress curves) that the nonconservative motion of jogs is rate controlling. The discrepancies between the above mentioned results are due, according to Davies, to the fact that Suzuka and Yamamoto were not measuring steady state creep, but rather a state of transient creep.

Kornilov (34) reported maxima of hardness and yield strength near the Ni_3Fe composition. For samples homogenized at 950 °C and annealed at 450 °C and 650 °C creep tests in a centrifugal creep machine showed maxima in creep resistance at the Ni_3Fe composition, being the peak sharp for the first annealing temperature and flattened by the 650 °C anneal. According to Kornilov, the first peak is due to the presence of Short Range Order.

Josso and Wache (35) compared the hardness of Ni_3Fe samples which had been ordered, after quenching from a high temperature with the hardness of samples highly cold worked and ordered. They found that in the latter case, the hardening was much more pronounced than in the former.

That is, the hardening of the ordered phase depends on the way in which the initial disordered state is set in. This observation prompted an extensive investigation by Vidoz et al. (23) on the strain aging phenomena in this alloy. In measurements of hardness and tensile properties, two effects were fundamentally examined

a) Short annealing periods (up to 120 minutes). Samples annealed at below and above Tc (43. °C to 525 °C) showed an increase in hardness during the first two minutes of annealing and which is higher, the higher the previous cold work.

In the case of tensile tests it was found that the increment in yield stress increases with prestrain and that the work hardening after aging was only slightly changed. The effect was dependent on grain size and seemed to disappear for coarse grain size crystals (.02 mm. initial grain diameter to .8 mm. final diameter). A high temperature quenching or a previous deformation were both necessary conditions for the effect to be present.

b) Long term hardening (up to 120 hours annealing time). In this case, an increase in hardness was found for the alloys which develope Long Range Order when annealed. In the tensile tests it was observed that the ordered alloys work hardened more rapidly than the disordered ones and that a prestrain before ordering enhanced the work hardening rate, being the effect more pronounced the higher the prestrain. Repeated strain aging treat-

repeatedly enhanced the work hardening.

Another interesting effect found was that disordered, cold worked Ni_3Fe will order but not recrystallize on annealing. Only if the annealing temperature was above Tc, would the alloy recrystallize. This effect seems to be related to the fact that the new grains growing in the cold worked matrix must at the same time, become ordered. Since this requires very complicated atomic rearrangement, the ordering seems to inhibit grain growth.

None of the proposed explanations for the short term hardening (SRO, segregation of solute to dislocations, configurations of quenched out vacancies or antiferromagnetic spin ordering) can fully account for all the experimental observations.

On the other hand the experimental results on the long term hardening can be explained in terms of the theory of Vidoz and Brown (10) already outlined in seccion B.

Recently Fayard and Behot have pointed out, as a consequence of their results in transmission electron microscopy, that both the staking fault and APB energies are low for this alloy. They have also found, in contradiction to the observations of Vidoz et al., that ordering does not impede recrystallization, even though they do not mention the composition of the alloy at which the effect was observed. It should be pointed out here, that Vidoz et al. found recrystallization at temperature of 480 °C and 470 °C but at compositions out of stoichiometry, where the critical temperature becames lower.

2) Ni_3Mn

This alloys apparently behaves very similarly to Ni_3Fe. Marcinkowski and Brown (37) found that the ordered phase consists of LRO in a matrix of SRO. For well annealed samples below Tc (540 °C) the size of the antiphase domains is approximately 500 Å.

Marcinkowski and Miller (22) have investigated the plastic properties of this alloy. They calculated the separation between "superpartials" as a function of both the stacking fault and APB energies, and the degree of order. Their values for the separation of "superpartials" in the fully ordered alloy are bigger by 25% than the observed ones in electron microscopy (170 Å). A high density of stacking faults was observed. Ordering produces, as it was already mentioned, an increase in the stacking fault energy. This is due to the surface energy associated with the APB between "superpartial". They have calculated for the APB, the stacking fault energy as a function of the degree of order as:

$$E_F^{(s)} = (E_F + S^2 E_{OR}/2)$$

where E_F is the stacking fault energy for the disordered lattice and E_{OR} is the APB energy. The higher value for the stacking fault energy increases the probability for simple and double cross-slip. A higher yield point was observed for the ordered than for disordered alloys.

The higher work hardening rate observed for the ordered alloy was thought to be due to a decrease of domain

size with deformation. Plots of the flow stress versus quenching temperature for different strains, show a maximum near Tc. As the domain diameters are at least 500 Å for the annealing treatments used, Cottrell's mechanism or a similar domain mechanism, are not applicable. The existence of SRO coexisting with LRO and introducing resistance to the movement of "superdislocations" was thought to be responsible for the effect. A similar explanation was given for a small broad maximum observed in flow stress versus annealing time (at 400 °C) for approximately 200 hours of annealing. It should be pointed out regarding to this explanation that, according to Cottrell (12) and Fisher (38) the energy per unit area of the slip plane associated with SRO, is less than γ, the surface tension of the APB. For an alloy with SRO the stress needed for the movement of dislocation is $\sim 10^8$ dynas/cm^2 being $\sim 10^9$ dynas/cm^2 for the fully ordered. A possible pile-up in the same slip plane would reduce this difference, but still the LRO should be rate determining for the deformation process.

For annealing at 455 °C a broad maximum in flow stress was found after 8 hours of annealing. According to the authors it was due to the difficulty to the dislocation motion produced by S.R.O.

Pile-up were observed (1) in the ordered phase. They cross several domains, indicating, therefore, that at least in the stage II of the work hardening the domain structure is not important for the deformation process.

3) Cu_3Au

As already discussed in section II, the results of Bigge and Brown (14) and Ardley (15) showed maxima in yield stress versus domain size in Cu_3Au. These were explained on the basis of Cottrell's theory. In the work of Ardley (15) on Cu_3Au single crystals, a second maximum was found at temperature above Tc, which was dependent on strain rate.

The results of Kuczynski, Doyama and Fine (21) show again an increase in yield stress at Tc (395 °C) in specimens annealed at various temperatures and quenched to room temperature. Slight discontinuities are also found at 590 °C and at 350 °C. The discontinuity at 590 °C is attributed to Fisher's SRO mechanism. An increase in the work hardening rate was also observed at temperatures near Tc and somewhere between 500 °C and 600 °C.

The influence of cold-work on this alloy has been studied in detail by Cohen and Bever (39) and Roessler and Bever (40). In the first paper the influence of room temperature cold work on different physical properties was studied. Particularly they observed that having similar initial hardness for the ordered and disordered alloys, the hardness increase with cold work was greater for the ordered case. This is an indication of the higher rate of work hardening of the ordered alloy. Similar effects were observed in yield stress. In the second paper the deformation and observation of various properties were made at 78 °K. Measurements made in this way were compared to those observed at room temperature. Observations of stored energy at 78 °K would indicated the presence of defects at 78 °K that could

not been annealed out at that temperature. In the case of ordered specimens, curves of stored energy versus resistivity were plotted, for equal deformations at 78 °K and room temperature. It was found that both curves were completely coincident, indicating that the mechanism of deformation was not dependent on the temperature. Such a mechanism could have been the reduction of the antiphase domain size by slip intersection. No superposition in the curves for 78 °K and room temperature was found for the disordered alloy.

Marcinkowski et al. (2) showed in this alloy for the first time the existence of APB's and "superdislocations". The separation between superpartials was found to be 130 Å, in good agreement with the theoretical calculation. Further observations by Marcinkowski and Zwell (4) on near stoicheometric Cu_3Au composition have shown the existence of two types of ordering and preponderance a of screw dislocation in the ordered alloy.

4) Fe_3Al

Even though the ordering characteristics of this alloy are not very simple, a study of its tensile properties could show interesting differences with other ordered systems. Marcinkowski and Brown (3) found that the APB energy of Fe_3Al is low, and therefore the separation between "superpartials" large. They have shown by electron microscopy that dislocations in ordered Fe_3Al do not seem to behave as "superdislocations". Again a preponderance of screw dislocation was observed in the ordered phase.

The mechanical properties of this alloys have been investigated by several authors:

Lawley, Coli and Cahn (41) found that the creep behaviour is critically dependent on the type of ordered structure. Alloys in the Fe_3Al range showed, at temperature near Tc an inverse transient, creep being accelerated during test. They also found a marked change in the activation energy at Tc. The inverse transient is explained by the authors on the basis of directional SRO induced by stress or by the superposition of it with the LRO. The necessary change from right (A-B) bonds to the wrong ones (A-A or B-B) produces an impedment to dislocation movemen and leads to viscous flow. The same process was held to be the mechanism that is rate determining for creep. On the basis of Weertman expression for viscous flow

$$\dot{\epsilon} = \tau^3 b^2 / \mu A B$$

where A is a quantity (temperature dependent) which is determined by the viscous flow; τ is the stress, μ the shear modulus and $B = \mu b^2 / 2\pi(1 - \nu)$. Lawley et al. obtain:

$$\dot{\epsilon} = 0.35\, \tau^3\, D / \mu^2 \gamma b$$

γ being the APB energy, D the diffusion coefficient of either the solute or the solvent. As γ decreases with increasing temperature, it would produce an extra contribution to the activation energy and would explain the increase in activation energy at temperature near the critical one. This explanation is objected by Davies (42) who observes that the reasons of Lawley et al. for disregarding the climb process in their discussion of their results was ba-

sed on a difference of $\sim 10^5$ in the coefficient of diffusion of pure Fe and the Fe-Al alloys and the one obtained from the observed creep tests. Davies found that an extrapolation of creep rates at lower stress gives the correct order of magnitude for both pure iron and Fe-Al alloys. On the basis of his experimental results he proposed therefore that climb is rate determining for temperature below 500 °C, and that the higher activation energies observed in the ordered state are due to the increased activation energy for diffusion in ordered alloys. Dislocation climb in these alloys should be easier because of the observed low APB energy that produces a very large splitting between "superpartials", which behave therefore as "single" dislocations.

At temperature above 500 °C the non-conservative motion of jogs seems to be rate determining. This is confirmed by the measurements of the activation energies and activation volumes, which indicated the correct order of magnitude for a jog mechanism.

Finally, Lawley, Vidoz and Cahn (43) have investigated the room temperature yield stress as function of quenching temperature. They found a maximum in the yield stress for samples quenched from temperatures near Tc, for annealing times of 10 minutes or longer. An explanation for this peak is given on the basis of the S.R.O. mechanism proposed by Sumino and Fisher, the latter being predominant for quenching temperatures higher than Tc.

V. EXPERIMENTAL APPARATUS TO BE USED FOR THE PRESENT INVESTIGATION.

1) The tensile machine. A hard tensile machine (Polanyi type) has been constructed. It is shown in Fig. 1 and 2. The screw (A) of a very accurately cut thread, is held centrally in the framework by ball bearings (A_1, A_2). It has a hole along its complete length. A worm wheel (B) locked at the top end of the screw, is used to transmit the movement to the screw. The fixed framework consists of two steel plates (C_1) (C_2) held together by six parallel steel bars (D_1 to D_6). The turning of the Lead screw produces the linear displacement of the cross head (E). The cross head has only one degree of freedom allowed by constraining it to move along bars D_1 and D_2 of the framework. The rods (F_1, F_2) are attached to the cross-head and run through holes with brass bushing situated in the lower plate of the framework. The tie rods are connected by a small plate (G) which carries the lower specimen grip.

The power is supplied by a constant speed motor with a reduction gear box (H_1, H_2) and transmited by using chains and sprockets (1). The lower sprocket I_1 is connected axially with a screw shaft, which run in ball races, (I_2, I_3), and transmites power to the locked worm wheel (B).

The load is measured by means of a strain gauge bridge (J) attached to the framework at one end and at the other end connected to the tie rod (K) which runs through

the hole in the lead screw and has on the lower end the top specimen grip (G_2).

Strain is measured with linear transducer (L) between the cross head and the framework.

Both signals, the one of the transducer and of the strain gauge bridge are fed into an X-Y Speedomax Recorder in order to obtain directly the stress-strain curve on the recorder chart. On figure 3 are shown the X-Y recorder, the 1 Kc power source for the transducer (constructed at our Electronics Division Laboratories) and the controls of the strain gauge bridge (according to circuits shown).

The tensile machine has been tested up to 500 Kg. load showing complete linearity in this range. The constant strain-rate is 10^{-4} mm/mm/min.

2) The vacuum furnace. The oxidation of Ni/fe alloys makes necessary the use of good vacuum or a protective atmosphere during annealing. Two furnaces were constructed to make the heat treatments:

a) Figure 5 shows a furnace designed for temperature up to 1.600 °C which will be used for the growing of single crystals of Ni-Fe alloys. It is an electrical resistance furnace, with Mo windings. The power supply is a 220/45 Volts, 5 KW transformer. The vacuum is better than 10^{-5} mm. Hg.

b) Figure 6 and 8 show a quenching furnace, designed so to be able to quench the samples from a vacuum high temperature treatment (up to 1000 °C) into different

quenching substance. The samples are placed in tube A, the vacuum chamber, and they are held by means of a thin metal wire. When the desired heat treatment is complete, a current is passed through this wire, cutting it, the samples being quenched through a thin mylar membrane (B) placed at the botton of the furnace tube into the quenching medium.

VI. FUTURE RESEARCH PROGRAM

1) Detailed investigation of the short term hardening in Ni_3Fe alloys.

A study will be made of the influence of different factors on the effect, for prestrains of 15 - 17%. The effect of different annealing temperatures and annealing times will be observed. For the values for which the above variables have the maximum effect, hardness and tensile tests will be made at temperature. Under the same conditions the effect of grain size also will be studied.

It is expected that the data obtained should give enough information to permit a better understanding of the characteristic of the effect.

The possibility of magnetic annealing will be also taken into account. The possible influence of interaction between ferromagnetic domain walls and point defects on the deformation process will be investigated.

Quenched samples will be also used and hardness measurements should be made with special enphasis on the effect of different quenching velocities.

2) Yield stress versus ordering annealing time will be studied to seek the possible effects of APB interaction. These results should give useful data for the work hardening investigation.

3) An investigation of the work hardening characteristics of single crystals as a function of test temperature and of different strain rates will be started soon. These data, should yield information regarding the validity of Vidoz and Brown's jog model and should give a better picture of the deformation process in ordered alloys.

Tests will be made in other alloys as it is mentioned in section III.

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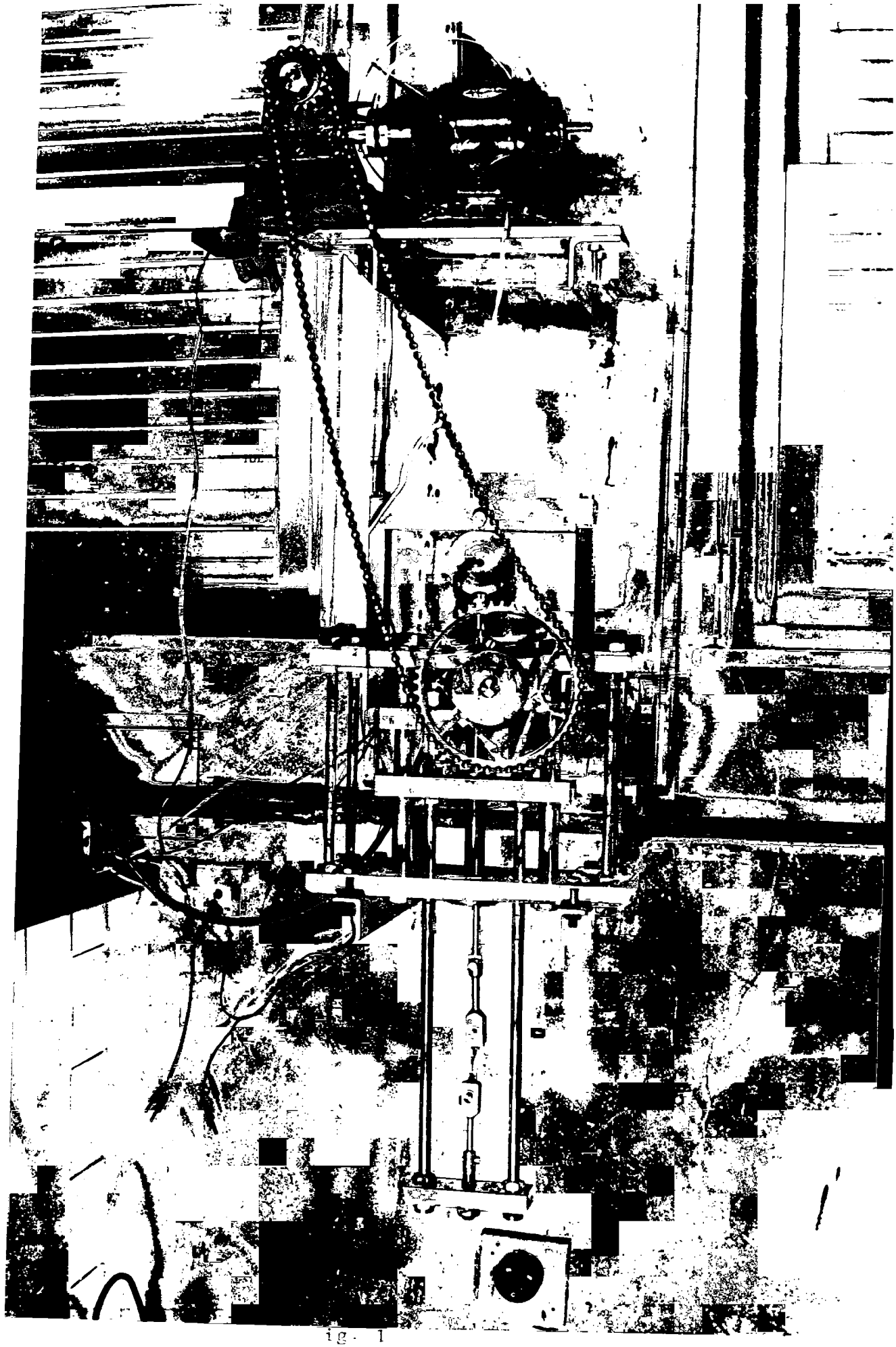

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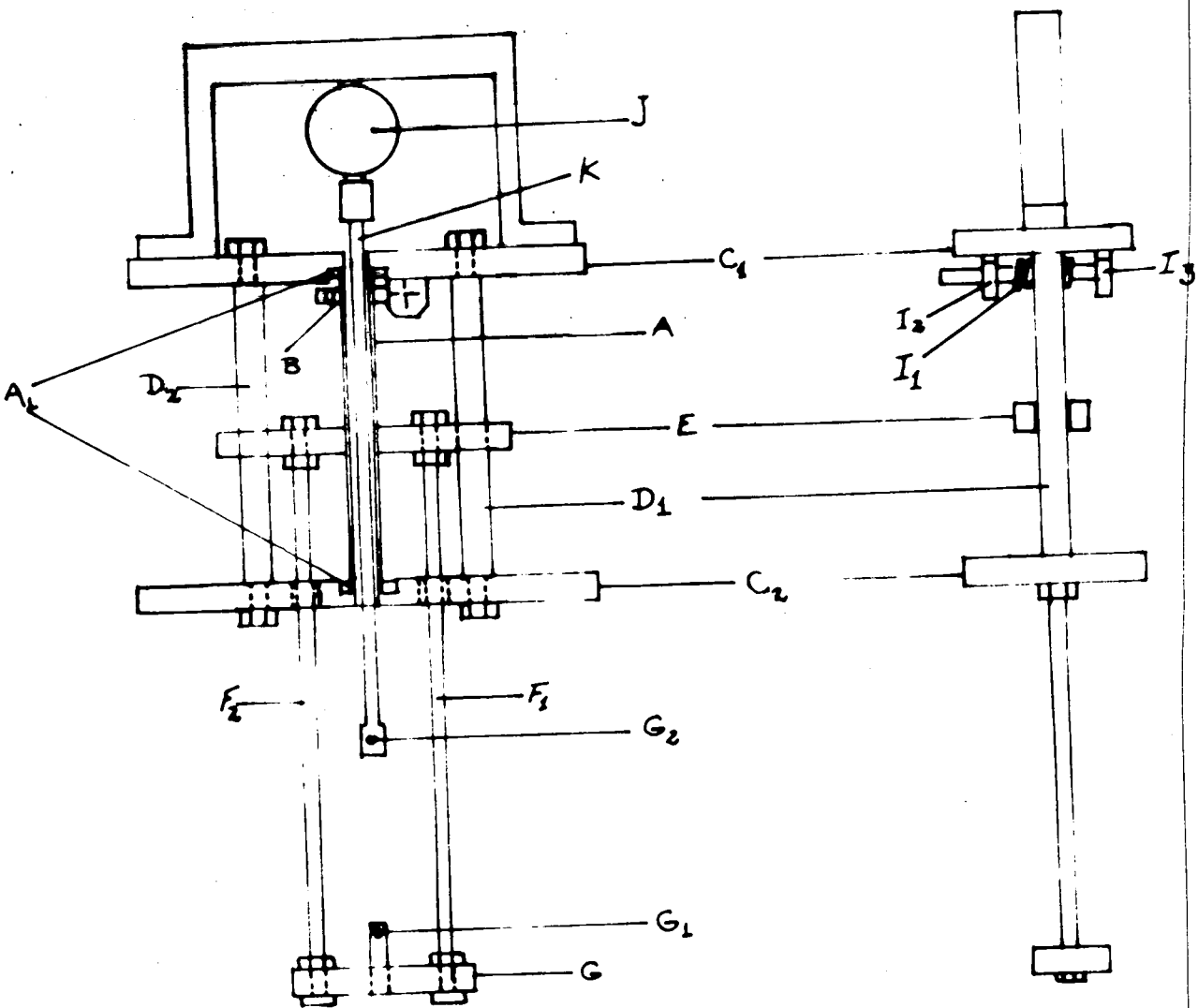

Fig. 2

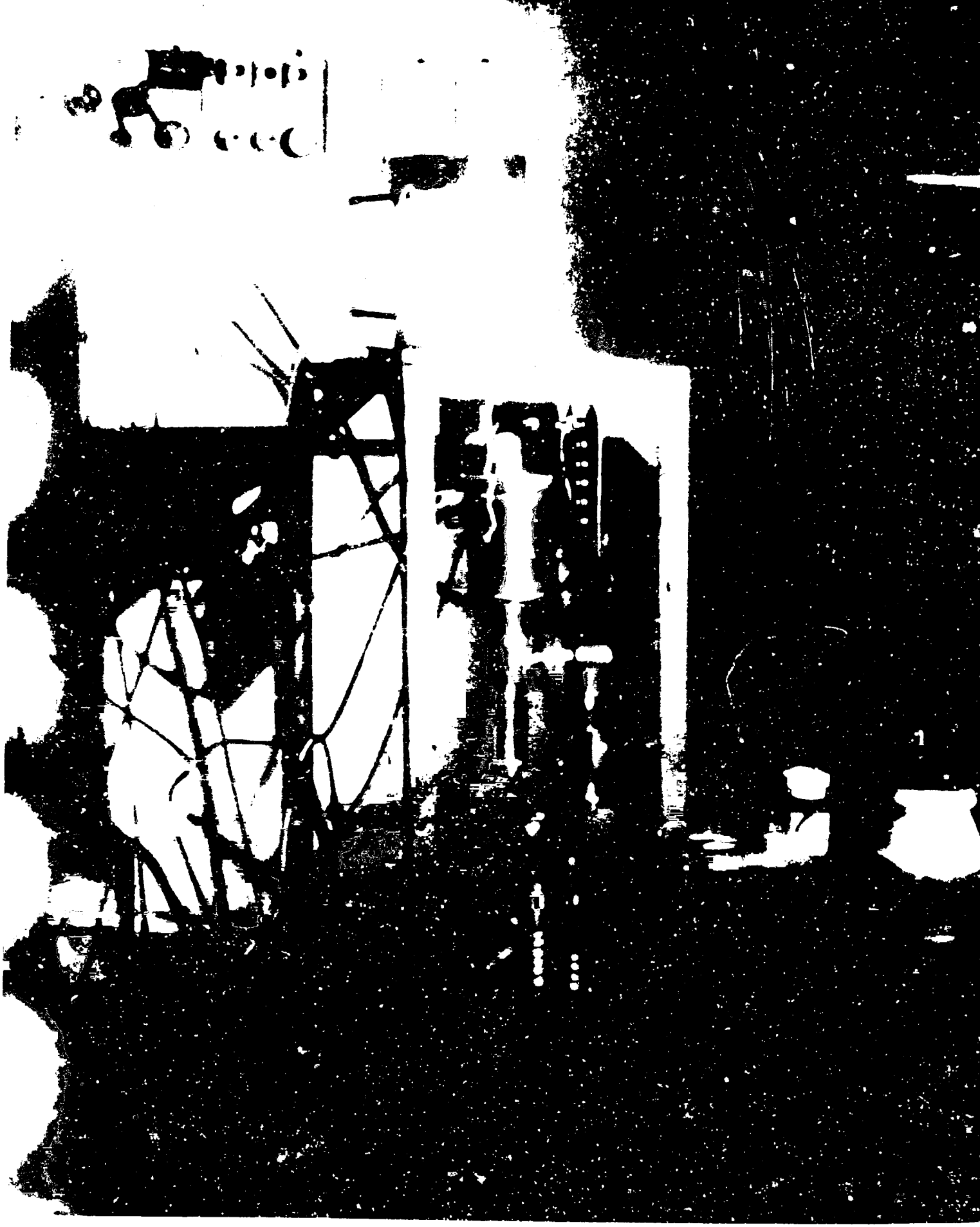

Fig. 4

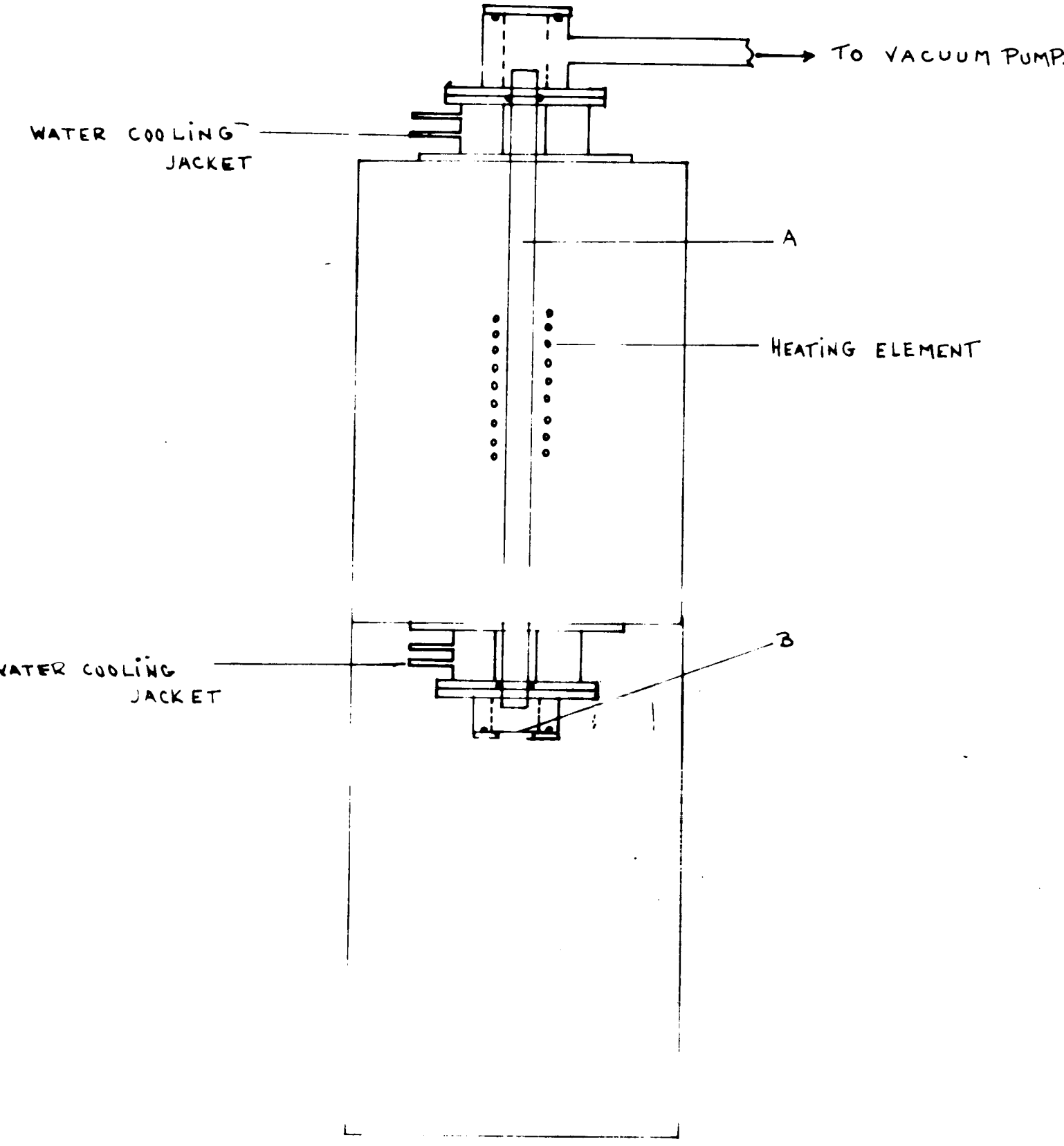
TO VACUUM PUMP
WATER COOLING JACKET
A
HEATING ELEMENT
WATER COOLING JACKET
B

Fig. 5