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THE TEMPERATURE DEPENDENCE OF WORK-HARDENING IN Cu_3Au AND Ni_3Fe

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There is still some discussion about the nature of work-hardening in superlattices. Among the mechanisms which have been proposed (1) the most widely accepted ones in alloys of the LI_2 structure are the cross-slip of superdislocations (2) and the production of antiphase boundary tubes (APB tubes) (3).

The cross-slip of superdislocations from the glide plane of type {111} into a plane {100} is energetically favorable since the APB energy in the {100} plane is small (4). In this way sessile superdislocations can be formed (2,5,6). Davies and Stoloff (2) proposed that this mechanism is responsible for the work-hardening and used it to explain the temperature dependence of the work-hardening coefficient θ in ordered Cu_3Au which increases from a value of $\theta/G = 2.10^{-3}$ at 77°K to a maximum of about twice this value at $T_o(Cu_3Au) \approx 350$°K and then decreases again at higher temperatures: Since cross-slip is a thermally activated process its probability should increase with increasing temperature thus contributing to a higher work-hardening rate. At high enough temperature cross-slip should however occur so readily that the superdislocations in {100} can cross-slip back into the {111} glide planes which should lead to a reduction of θ at high temperatures.

The production of APB tubes by non aligned jogs in superdislocations has been first discussed by Vidoz and Brown (3) and Vidoz (7) and a detailed quantitative theory of work-hardening based on this mechanism has been worked out by Schoeck (8). It is based on an analysis of the stability of APB tubes (9) which shows that jogs in screw dislocations align easily by glide. In dislocations of edge type however at low temperatures the jogs will not align during the time the dislocation

moves. The work-hardening in this theory results from a very small activity of secondary slip systems which cause the formation of APB tubes in the primary system. Since the secondary slip systems must intersect the primary one by thermal activation, their activity will increase somewhat at higher temperatures thus leading to a higher work-hardening rate. At even higher temperatures however the jogs in edge superdislocations can align by climb thus eliminating APB tubes and reducing θ. For Cu_3Au the temperature above which this takes place was estimated to be about 400°K, (9).

The main objection against the cross-slip theory is a quantitative one (8). The activation energy U_{cr} for cross-slip depends essentially on the dimensionless parameters Gb/γ and τ/G (10,11,12) (where G = shear modulus, b = magnitude of Burgers vector, γ = stacking fault energy, τ = resolved shear stress). What is of importance here is that U_{cr} increases (weakly) with decreasing stress τ. Since in Cu_3Au the flow stress τ decreases slightly with decreasing temperature (2) we would expect U_{cr} to increase towards lower temperature. For our argument we assume the favorable case that U_{cr} does not change with temperature.

If now cross-slip is to be observed at all around room temperature its activation energy should be somewhere between .5 eV and 2 eV. The ratio of cross-slip probability at T_1 = 350°K and T_2 = 77°K is then given by

$$r = \exp \{U_{cr}(T_1 - T_2) / k\, T_1\, T_2\}$$

which with the most favorable value of U_{cr} is about $r \approx 10^{19}$. Since such a ratio is completely unreasonable, cross-slip can not occur over the whole temperature range, and hence it can not be responsible for work-hardening.

Since the work-hardening mechanism should apply generally the validity of the theory can be checked by comparing the behaviour of different alloys. For this reason it was decided to measure the temperture dependence of the work-hardening in Ni_3Fe which also has the LI_2 structure. The single crystals were of the same batch used by Victoria and Vidoz (13). The crystals had been annealed for 120 hrs at 480°C and their long range order parameter was probably close to S = .8. The crystals were extended for strain intervals about 10% to 20% at one

temperature in order that θ could be determined and then the temperature was changed. The results are shown in Fig. 1. Although there is some scattering between different crystals all the curves show a broad maximum around the temperature $T_o(Ni_3Fe) \approx 520°K$.

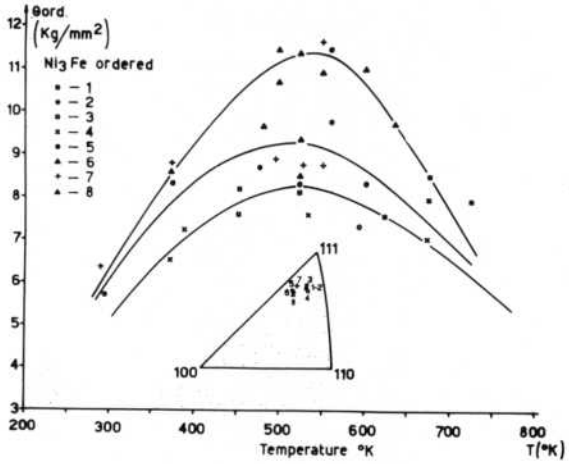

Fig. 1

Work-hardening coefficient θ of ordered Ni_3Fe single crystals with indicated orientation as function of temperature

According to the APB tube mechanism the temperature of the maximum in θ should be proportional to the energy U_v for formation of a vacancy (9). Unfortunately this energy is not known in Cu_3Au and Ni_3Fe and direct diffusion experiments can not be made in reasonable time below the ordering temperature. However from the ordering kinetics in Cu_3Au Feder et al (14) derived a diffusion energy of about 2.03 eV which agrees with that for self-diffusion in pure Cu (15) and hence we may reasonably assume that the energy for formation of a vacancy in ordered Cu_3Au is the same as in pure Cu and hence $U_v(Cu_3Au) \approx 1$ eV. For ordered Ni_3Fe there are no data but its ordering kinetic is much more sluggish than the one in Cu_3Au. The activation energy for creep in the disordered alloy at temperatures approaching the ordering temperature is of the order of 2.9 eV (16), which agrees with the self-diffusion energy of Ni (15). By analogy we may assume that a reasonable value for the formation energy of a vacancy is the same as in pure Ni (15) and hence $U_v(Ni_3Fe) \approx 1.4$ eV.

Since the geometry of deformation in the both alloys should be similar the

APB tube theory would predict (quite independent of finer details) that the temperature $T_o(th)$ where θ in Ni_3Fe reaches the maximum should be $T_o(th) = T_o(Cu_3Au)\ U_v(Ni_3Fe) / U_v(Cu_3Au) = 490°K$. The agreement with the experimental value of $T_o(Ni_3Fe) = 520°K$ is as close as can be expected considering the uncertainty of the values. Hence we may conclude that the APB tube model of work-hardening (8) gives a reasonable explanation of the experimental observations whereas the cross-slip model encounters serious difficulties. It would be interesting to check work-hardening in other superlattices of LI_2 type for their agreement with the theory.

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