On the Mechanism of Plastic Flow in an Overaged Al–2% Ge Alloy

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Investigations on the yield behaviour of an overaged Al–2% Ge alloy have been carried out. Tensile deformation coupled with transmission electron microscopy has been used to establish the dislocation-precipitate interaction. That the glide dislocations avoid the precipitate obstacles and bow out by the Orowan mechanism has been clearly established. The experimental yield stress and the calculated yield values based on the Orowan model of dispersion hardening agree fairly well. The slip line traces also seem to corroborate these facts by showing distinct wavy traces for the overaged alloy, the features of which have been interpreted in terms of the inability of the part of glide dislocations to push through the well-grown and widely-spaced precipitate particles.

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I. Introduction

An understanding of the dislocation-precipitate interactions in age-hardenable alloys has led to the development of newer and stronger alloy systems. As the ageing of a supersaturated solid solution proceeds, a gradual change from coherent clusters with small interparticle spacing to large, widely-spaced and mostly incoherent particles is noticed. The moving dislocations will have either to shear through such precipitates (cutting mechanism) or avoid them, say, by means of Orowan bowing (by-pass mechanism). The energetics generally favour the cutting process in the early stages of ageing, while the by-pass mechanism takes over later on when the overaged state develops. Thus, in the overaged Al–Cu(2)Ge(3), Al–Ag(4), Al–Si(5), Cu–Be(6), and Mg–Mn(7) alloys, it has been shown that dislocations avoid the precipitate particles. Obviously, the large interparticle spacing and the high surface energy associated with such particles make dislocation penetration into the precipitate barriers very unlikely. However, cutting of overaged Mg–Mn precipitate by glide dislocations has been reported by Chun and Byrne(8). This suggests that there is need to examine carefully the dislocation-precipitate interaction in the overaged condition in newer and relatively less-explored alloy systems. In the present work, the dislocation-precipitate (germanium) interaction has been examined in an overaged Al–2%Ge alloy employing such techniques as tensile deformation, optical metallography and transmission electron microscopy.

II. Experimental

The alloy was made by melting together the weighed quantities of superpure aluminium (99.997% pure) and germanium (semiconductor grade) in an argon atmosphere. The ingot was lightly deformed by hammering and then annealed for two days at 673 K. The homogenised rod was flat-rolled with intermediate anneals at 623 K to yield a final thickness of about 1 mm. Chemical analysis showed that the alloy contained 1.98% Ge by weight. Tensile samples cut from the rolled strip were electropolished, solution treated at 723 K for 2 hr and quenched into ice-water mixture. The quenched samples were dipped in acetone, dried and quickly transferred to a paraffin oil bath maintained at 533 K. The samples were aged at this temperature for 41 hr with a view to getting an overaged alloy, as determined by hardness measurements. The specimens were tensile tested in the range 77 to 400°K in a modified Hounsfield tensometer(9). With the help of a pulley arrangement, the 20 mm
gauge length samples were pulled at constant strain rate of $10^{-3}$ sec$^{-1}$. The load on the sample was measured with the help of a resistance strain gauge fixed on to the spring beam. The use of an extensometer fitted with a linear variable differential transformer transducer coil permitted the measurement of strain on the sample.

For the transmission electron microscopy, a Hitachi HU-11E microscope operated at 100 kV was employed. Foils of samples both in undeformed and deformed states, suitable for transmission microscopy were prepared from the aged samples by chemical thinning and electropolishing, using 'window' technique. The solution employed was 90/10 acetic acid/perchloric acid mixture maintained at $-15^\circ$C. For surface slip line studies, the electropolished test samples were solution treated, quenched into water at 0°C, aged in paraffin oil bath, deformed using the tensometer and examined under an optical microscope.

### III. Results

Figure 1 shows the dependence of the 0.2% offset tensile flow stress, $\sigma$, on temperature for the overaged Al–2%Ge alloy; the flow stress values were corrected for the shear modulus variation with temperature assuming the $\mu$–$T$ relationship of aluminium is unaffected by minor additions of germanium. It is clear that $\sigma$ shows only a very small decrease with temperature in the range 77–400 K. This behaviour stands in sharp contrast to that of the quenched alloy$^{(10)}$ or peak-aged alloy$^{(11)}$ in that a much larger temperature dependence of flow stress was noticed. Important information concerning the dislocation-precipitate interaction can be obtained from the work hardening characteristics$^{(3)}$. The rates of work hardening $da/de$ for the quenched as well as overaged samples are given in Table 1 when $T$ is 200 K.

To understand the nature of deformation, slip line patterns are examined optically. In the overaged alloy deformed to 25% strain, the slip traces are seen to be wavy, with shifts from straight paths (Photo. 1). This is, perhaps, an indication of the cross-slip process. Also seen on careful examination of the Photomicrograph is the fragmented slip traces. This feature as well as wavyness of the lines indicates the strong nature of the barriers and the inability of moving dislocation to intersect them$^{(3)(12)}$. By contrast, the micrograph (Photo. 2) taken from the same alloy, aged to less than peak hardness, shows lines that traverse fairly long and straight paths with minimum deviations, if any. Also there is greater spacing between adjacent lines. The stains could perhaps have formed during thermal treatment as the specimens have been electropolished before such treatments. The transmission electron micrograph in Photo. 3 shows well grown and widely spaced germanium precipitates in the

![Photo. 1 Overaged alloy showing fragmented and somewhat wavy slip line traces; $e=25\%_o$ (700 × 7/8)](image)

![Fig. 1 Flow stress-temperature relation in overaged aluminium-2% germanium alloy.](image)
Photo. 2 Straight running, well-defined and widely spaced slip traces on a sample aged at about $440^\circ$K 60 hr and deformed 15% $\varepsilon$, (700 $\times$ 7/8)

Photo. 3 Transmission electron micrograph showing precipitates corresponding to overaged condition.

overaged state. The precipitates have varying shapes including the rod-like ones.

IV. Discussion

The main aim of the present work was to evaluate the rate controlling dislocation mechanism for plastic flow of the overaged Al–2Ge alloy at low-temperatures. As can be seen from Fig. 1, temperature has only a negligible effect on the 0.2% strain offset flow stresses for the alloy. Much larger variations of $\sigma$ with temperature have been reported in the case of Al–Cu$^{(13)(14)}$, Al–Zn$^{(15)}$, Al–Ag$^{(15)}$, Mg–Zn$^{(8)}$ and Mg–Mn$^{(7)}$ alloys when the precipitate particles lying in the glide plane are found to be sheared by the dislocations. Since the temperature-dependence of flow-stress in the present alloy is rather very small (Fig. 1), such thermally activated dislocation mechanisms like cutting can obviously be ruled out. This conclusion is consistent with the optical micrographs of Photos. 1 and 2. When precipitates are small and spaced closely, fairly straight slip lines that are long traversing and continuous are observed (Photo. 2), indicating that precipitates are sheared$^{(16)}$; these features are absent for the overaged Al–2Ge alloy (Photo. 1). Also, shearing of the large incoherent precipitates should cause an excessive increase in interfacial energy, because of creation of two new surfaces. It can thus be concluded that the glide dislocations avoid the precipitates most probably by the ‘Orowan bowing’. Evidence for such by-passing is presented in Photo. 4, where the transmission electron micrograph shows a dislocation segment expanding between two grown precipitate particles. The Orowan loops, left behind by the glide dislocation around the precipitates after by-passing them, are also observed (Photo. 5). Further, the Orowan mechanism of yielding gives rise to temperature-independent flow stress. This indeed is what is observed in the present case (Fig. 1). The small variation in flow stress could be attributed to the matrix, containing some amount of solute dissolved in them.

Additional evidence for the by-pass mechanism comes from a consideration of the work

Photo. 4 A transmission electron micrograph illustrating the bowing of dislocation between two precipitate particles by the Orowan method.
hardening rates. As suggested by Fisher et al. (17), the accumulation of loops around precipitates (Photo. 5) leads to a high work hardening rate. In accordance with this view, a high value of \( \mu/10 \) for \( d\sigma/d\varepsilon \) has been observed (3) for \( \theta \) precipitates in Al–Cu, corresponding to Orowan bowing, as against \( \mu/500 \) of G.P. I and G.P. II for the same alloy when tested at 4.2 K. The present results, although obtained on polycrystals, suggest that the values of \( d\sigma/d\varepsilon \) are, in general, higher in the overaged state than in the quenched state particularly at low strains (Table 1). At larger strains, the rates decrease, perhaps because of the increased tendency for cross-slip to occur with an increase in the number of residual loops around precipitate particles. That cross-slip is, in fact, predominant in the heavily deformed alloy can be inferred from Photo. 1, where, the slip lines have a distinct step-like appearance.

It is of interest to compare the experimentally measured flow stress with that predicted by the Orowan formulation; it has been shown (18) that:

\[
\tau = \tau_0 + \frac{\mu b}{4\pi} U \ln \left( \frac{\lambda}{2b} \right) \cdot \frac{1}{\lambda/2}
\]

where first and second terms on the right hand side are the contributions to flow stress from matrix and dispersed precipitates respectively. \( U \) is the mean of the line tension factor for edge and screw dislocations and \( \lambda \) is the mean planar interparticle spacing. Substituting for \( \lambda \) from analysis of electron micrographs, and taking \( \mu \) as that for pure aluminium, the bowing stress is computed to be 1.3 kg/mm\(^2\). It is difficult to assess the contribution from matrix which has, as per equilibrium diagram, about 0.3 at\% germanium. The value of \( \tau_0 \) has to lie between those of pure aluminium and quenched supersaturated alloy containing 0.75 at\% Ge (10); these have been found to be 0.55 and 1.2 kg/mm\(^2\) respectively. Corresponding to these two limits, the flow shear stress works out to be 1.85 kg/mm\(^2\) and 2.5 kg/mm\(^2\) or 3.7 and 5.0 kg/mm\(^2\) for the tensile flow stress, \( \sigma \). The experimentally observed value is 3.8 kg/mm\(^2\). The theoretical limits fall within a factor of about 1.3 of the experimental results and hence the agreement between the two sets of values can be considered close, again supporting the validity of the Orowan process in the present case.

In conclusion, it can be said that the slip line traces suggest that dislocations fail to proceed through the precipitates and transmission electron microscopy confirms this point by showing overaged germanium precipitate in an aluminium–2% germanium alloy being avoided by glide dislocations by bowing between them. Residual loops around precipitates as a result of such bowing are also seen. These as well as the flow stress data, indicate that the yield stress in this alloy is controlled by the Orowan mechanism.

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