Tensile yielding of Al-Al_3Ni eutectic crystals

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Directionally-solidified single crystals of Al–Al_3Ni eutectic have been grown at differing rates to produce fibre spacings of the Al_3Ni-phase of between 1.26 and 2.26 μm. After heat-treatment to produce a low matrix dislocation density, the tensile yield stress of the various crystals was measured. Comparison of the yield stresses of crystals of differing fibre spacing indicate that the resolved shear stress is lower at smaller fibre spacings, suggesting yield is not by the Orowan mechanism. It is concluded that on yield, dislocations propagate from sources near the fibre–matrix interfaces, and that due to their higher elastic modulus, the fibres tend to repel the matrix dislocations.

1. Introduction

In a recent review [1], Chadwick emphasizes that much needs to be learnt about the fundamental deformation processes in eutectic materials before alloy development can progress beyond the present empirical stage. In particular, he notes that little is known in detail of the micromechanics of the room temperature tensile behaviour of any eutectic composite, although, in the case of Al–Al_3Ni, several single-crystal studies have been made [2, 3].

Cantor and Chadwick [3] report that the deformation behaviour of these eutectic crystals is controlled by the very high density of dislocations in the as-grown eutectics which arise from differential thermal contraction effects. Gould and Martin [2] examined crystals in which this high dislocation density (2 to 3×10^9 lines cm^{-2}) was reduced to a value of approximately $10^7 cm^{-2}$ by annealing heat-treatments, and although their electron microscopic examination of deformed crystals suggested that an Orowan process had occurred, no systematic examination of yield stress with fibre spacing was undertaken to establish if yielding was in fact controlled by an Orowan process.

The object of the present work was to study the effect of varying the fibre spacing on the yield stress of such specimens.

2. Experimental

Master alloys of composition 6.2 wt% Ni in Al were melted from 99.99 wt% Al and 99.995 wt% Ni in alumina crucibles under a pressure of 500 mm Hg. After scalping, the billets were swaged to 4.4 mm diameter rods, suitable for crystal growth.

Single crystals of 4.7 mm diameter and 180 mm length were grown from the melt under a vacuum of $< 10^{-5}$ mm HG using a modified Bridgeman technique and drawing rates of between 80 and 150 mm h^{-1}, the temperature gradient at the solid/liquid interface being $6°$ C mm^{-1}. 50 mm long tensile specimens were cut from each crystal and were then annealed in order to reduce the density of grown-in dislocations [2]. This involved annealing the specimens at 225° C for 5 h in oil, and then cooling them at $5°$ C h^{-1} to room temperature.

Before testing, specimens were electropolished as previously described [2], and tests were conducted using a Polanyi-type frame and friction grips on an Instron TT machine. The strain was measured using a 10 mm gauge length extensometer at a magnification of $\times$ 2500 where a strain of 1% (corresponding to a displacement of 0.1 mm on the specimen) showed as 250 mm or chart movement. All tests were conducted at room temperature at a strain-rate of 5.5×10^{-4} sec^{-1}.

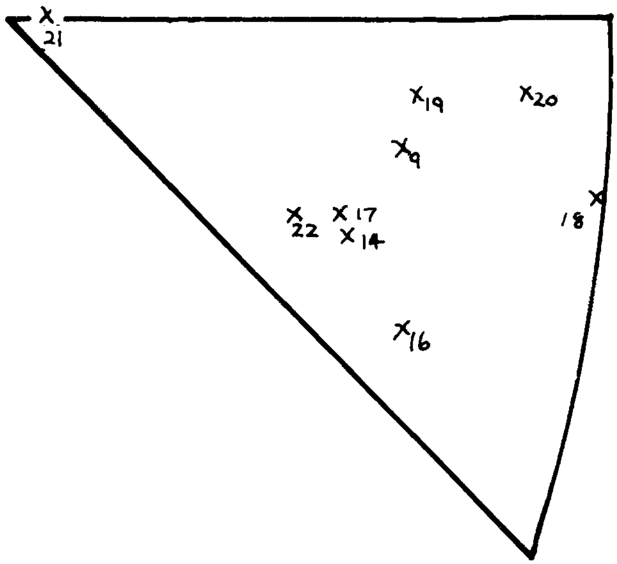

Figure 1 The orientations of the tensile axis with respect to the Al matrix.

3. Results

Two tensile specimens were produced from each of the nine crystals prepared, and the matrix orientations, determined by a back-reflection Laue technique, are shown in Fig. 1. The angles between the tensile axes and the slip directions and plane normals, λ and ϕ respectively, were measured and are listed in Table I, together with the Schmidt factor, $\cos\lambda \cos\phi$. From electron micrographs of transverse sections of the crystals, measurements were made of the fibre–fibre spacings (D_s) and fibre diameters (d_s), and these values are also shown in Table I.

Yield stresses were determined from the stress–strain curves by extrapolation from the slopes of the curve in the elastic and plastic stages, and the values obtained are listed in Table I.

4. Discussion

In order to examine the possibility of the operation of an Orowan mechanism at yeild, we may employ the approach of Kelly [4], who considered initial flow in a single crystal containing continuous fibres of radius r lying at an angle ϕ to the normal to the slip plane, and defined the Orowan stress to be

$$\tau_\phi = \frac{2T}{b}\left(\frac{1}{D_{s\phi} - d_{s\phi}} + \frac{2V_f}{\pi r}\right) \quad (1)$$

where T is the line tension of the dislocations, b is Burgers vector, $D_{s\phi}$ is the mean planar spacing of the obstacles in the slip plane, $d_{s\phi}$ is the mean diameter of the obstacles in the slip plane, V_f is the volume fraction of the fibres (taken as 0.1 in the present system [5]).

TABLE I Structural and tensile data of the crystals

Crystal ref. no.	λ	ϕ	Schmidt factor	D_s μm	d_s μm	Yield stress (kg mm^{-2})	τ_ϕ (kg mm^{-2})
9	44	47	0.49	2.16	0.80	1.77	0.87
						1.70	0.85
14	37	54	0.47	1.26	0.45	1.67	0.78
						1.61	0.76
16	34	58	0.44	1.56	0.52	2.05	0.90
						2.04	0.90
17	37	52	0.49	1.88	0.79	1.61	0.79
						1.69	0.83
18	50	46	0.45	2.26	0.69	2.82	1.27
						3.08	1.39
19	48	43	0.49	1.90	0.68	2.42	1.19
20	52	41	0.46	1.77	0.56	2.45	1.13
21	45	52	0.44	1.74	0.62	2.55	1.12
22	36	53	0.49	1.76	0.57	2.20	1.08

We may write

$$D_{s\phi} = \frac{D_s}{\cos \phi^{1/2}}$$

and

$$d_{s\phi} = \frac{d_s(1 + \sec \phi)}{2},$$

putting

$$T = \frac{Gb^2}{2\pi(1-\nu)^{1/2}}$$

where G = shear modulus of the matrix, and ν = Poisson's ratio.

We can also apply a factor which takes into account the "dipole effect" of dislocations bending at obstacles, i.e. $\ln (d_{s\phi}/r_0)$, where r_0 is the effective core radius, taken as $5\boldsymbol{b}$. Thus the resulting expression is

$$\tau_\phi = \frac{Gb}{\pi(1-\nu)^{1/2}}\left[\left(\frac{D_s}{\cos \phi^{1/2}} - \frac{d_s(1+\sec\phi)}{2}\right)^{-1} + \frac{2V_f}{\pi r}\right] \cdot \ln\left(\frac{d_s(1+\sec\phi)}{2r_0}\right) \quad (2)$$

and in Fig. 2 a graph is plotted of

$$\tau_\phi \text{ versus } \left[\left(\frac{D_s}{\cos \phi^{1/2}} - \frac{d_s(1+\sec\phi)}{2}\right)^{-1} \frac{2V_f}{\pi r}\right] \ln\left[\frac{d_s(1+\sec\phi)}{2r_0}\right]$$

the theoretical slope being indicated by a broken line.

It is obvious that the trend of these results seems to be towards a lower resolved shear stress at small fibre spacings, which is the opposite to that which would be expected if yield is by an Orowan mechanism. There are many observations in the literature (e.g. [6, 7]) of a Hall–Petch relationship between the yield stress and the inter-fibre spacing in as-grown eutectic crystals, but this is likely to arise from the high dislocation density in these materials being in the form of sub-cells whose scale will be dependent upon the inter-fibre spacing. This is not possible in the present case where care has been taken to reduce the initial dislocation density.

Lawson and Kerr [8] have in fact reported a fall in yield stress at very fine (~ 0.2 μm) fibre spacings of Al–Al_3Ni eutectics grown at very high rates, but this was accounted for by the observation that the perfection of the structure had deteriorated – the fibres being arrayed in colonies containing fine fibres in their interior and coarse ones at the boundaries. Such heterogeneities were not present in the crystals prepared in this study, so that the data cannot be accounted for on this basis.

Cantor *et al.* [3] suggest that matrix yielding in this system is initiated close to the interphase boundaries and that arrays of dislocations propagate into the matrix and eventually interact to form tangled dislocation networks. Following this argument, if the fibre/matrix interface is a likely source of dislocations, so increasing dislocation density at yield will arise from decreasing the fibre spacing (hence from increasing the area of fibre/matrix interface per unit volume). The

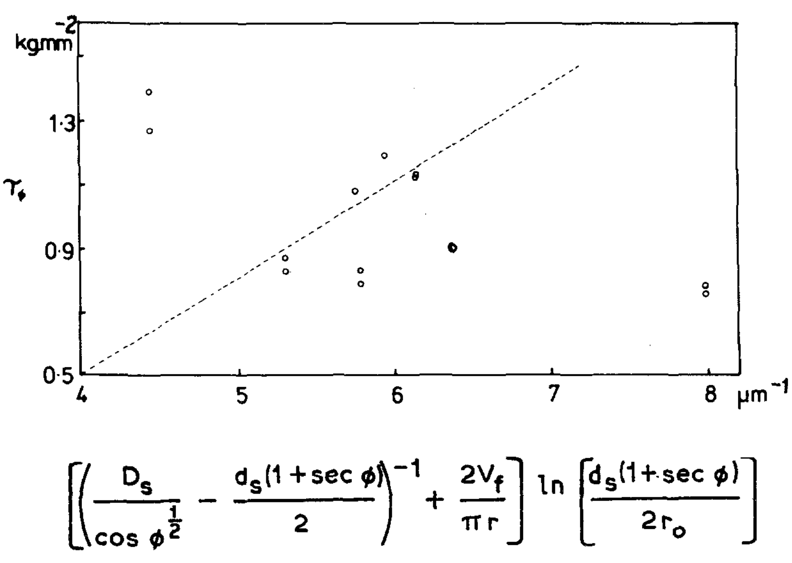

Figure 2 Plot of data according to the Orowan relationship.

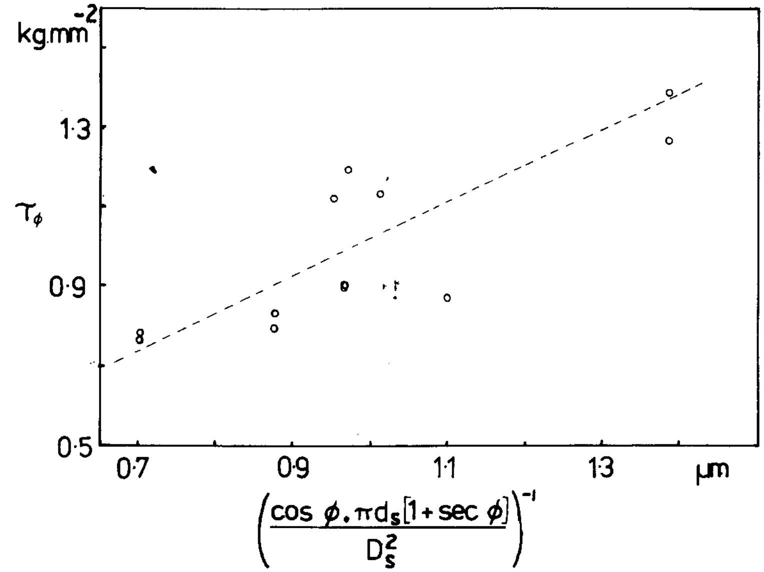

Figure 3 Shear yield stress plotted versus reciprocal of fibre/matrix interfacial length per unit area of slip plane.

average slip distance for a given strain will, therefore be reduced with finer fibre spacings, which implies that, for a given small yield strain, the dislocations remain closer to their fibre interfaces of origin than those in crystals with coarser fibre spacings.

Since the modulus of the fibres is greater than that of the matrix, the energy of the dislocations is higher when their strain fields interact with the fibres, hence they will experience an image force. This effect has been considered by Cline and Stein [8] for the case of lamellar eutectic structures, who observe that only in the limit of very small interphase spacings (comparable to the Burgers vector) will second nearest boundary interactions become significant, so that the magnitude of the force is essentially independent of the inter-fibre spacing, and will depend only upon the difference in shear modulus between the matrix (G_1) and fibre (G_2) and the distance (x) of the dislocation from the fibre/matrix interface:

$$\tau_i b = \left(\frac{G_2 - G_1}{G_2 + G_1}\right)\frac{G_1 b^2}{4\pi x}$$

where τ_i is the image stress, and b the Burgers vector.

As the fibre repulsion force falls off as $1/x$, longer glide distances would require greater external work than would be necessary for shorter glide distances. Thus for a given strain at yield, a lower applied stress might be expected for finer fibre spacings.

We may thus write:

$$\tau_\phi \propto x$$

and since $x \propto 1/\rho$, where ρ is the mobile dislocation density, then

$$\tau_\phi \propto 1/\rho.$$

Assuming that the length of each fibre/matrix interface in the slip plane to be $\pi d_s(1 + \sec\phi)$,

$$\begin{aligned}\rho &\propto \pi d_s(1+\sec\phi) \times \text{no. of fibres/unit area of slip plane}\\ &\propto \pi d_s(1+\sec\phi)\cdot 1/D_\phi^2\\ &\propto \frac{\pi d_s(1+\sec\phi)\cos\phi}{D_s^2}.\end{aligned}$$

The yield stress should, therefore, be proportional to $(\pi d_s(1 + \sec\phi)\cos\phi/D_s^2)^{-1}$ and Fig. 3 illustrates this relationship, where it is seen that the data are consistent with this model, in that a monotonic relationship appears to exist, and a notional straight line is shown on the graph as a broken line.

5. Conclusions

(1) The variation in yield stress with inter-fibre spacing is not consistent with the operation of an Orowan mechanism.

(2) The observed data are consistent with the propagation of dislocations from sources near the fibre/matrix interface at yield, and that the fibres tend to repel these dislocations leading to a lower

observed yield stress when the length of fibre/matrix interface per unit area of slip-plane is increased.

Acknowledgements

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