with a ⟨1 1 1⟩ axis in several cubic system substrates. He showed that those deviations could be accounted for quantitatively by a simple misfit dislocation model. Thus, it is clear that small deviations from parallel epitaxial alignment occur in a number of cases and it appears generally that these phenomena can be understood in misfit dislocation terms.

Acknowledgements

Thanks are due to Professor J.G. Ball for the provision of research facilities. One of us (FHG) wishes to thank the Governments of India and of Kamataka State for the provision of a study grant.

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Received 9 December 1977
and accepted 2 February 1978.

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Strain-induced continuous recrystallization in Zr-bearing aluminium alloys

It has recently been demonstrated that a series of aluminium alloys can be made superplastic by the addition of about 0.5% Zr [1–4]. Although superplastic elongation requires a stable fine-grained structure, a salient structural aspect regarding these alloys is that, prior to superplastic deformation, they generally are not recrystallized. A fine-grained structure is obtained during hot forming after an initial 10 to 50% straining, apparently due to a continuous recrystallization process [3]. The stability of the as recrystallized structure during continued plastic flow relies on the inhibition of grain growth by finely-dispersed Al_3Zr-particles. The mechanism, however, by which the fine-grained structure is obtained cannot be considered as established.

This note reports some preliminary results on the structural stability of a Zr-bearing Al–Mn alloy (Al–0.9 Mn–0.4 Zr) during hot deformation using strain rates typical of superplastic forming operations. The alloy, which was of commercial purity, was chill cast. Following casting the alloy was cold rolled about 85% and then given a pre-treatment of 4 h at 440° C which produced a subgrain structure stabilized by Zr- and MnFeSi-rich particles. Fig. 1 illustrates the fine dispersion of the metastable cubic Al_3Zr particles. The distribution of the MnFeSi-rich dispersoids is similar to that reported in [5]. The stability of this structure has been followed during isothermal annealing at 500° C and during plastic straining at 480° C. The tensile specimens had a gauge length of about 10 mm, a width of 5.5 mm and thickness about 1 mm. The specimens were deformed at a constant velocity of 0.2 cm min^{-1}, i.e. at an initial strain rate of 3.3×10^{-3} sec^{-1}.

The pre-heat treated alloy has been annealed for several hours at 500° C without any signs of recrystallization. During such heat treatments only minor subgrain growth occurred. During hot deformation, however, the subgrain size was found to increase with increasing strain as illustrated by the diagram in Fig. 2. The TEM micrographs in Figs. 3a, b and c show the undeformed subgrain structure, and those after 40% and 180% deformation, respectively.

Figure 1 TEM dark field micrograph showing the distribution of Al_3Zr particles.

The stability of the subgrain structure during isothermal annealing at 500° C has been attributed to the restraining effect of the Al_3Zr particles, as previous work [5, 6] has demonstrated that the MnFeSi-rich dispersoids are not capable of stabilizing the subgrain structure above 450° C. A model for the annealing behaviour of a cold-worked substructure (in a dispersoid-containing alloy) has recently been presented by the present author [6, 7]. According to this analysis the nucleation rate in static recrystallization will rapidly approach zero as the average subgrain size, $\bar{\rho}$, approaches a limiting size, $\bar{\rho}_{\text{lim}}$ given by

$$\bar{\rho}_{\text{lim}} = \frac{4}{3\alpha}\frac{r}{f},$$

where r is the particle radius, f the particle volume fraction and α is a constant of order unity. If we assume that the present thermal pretreatment resulted in a subgrain size $\bar{\rho} \simeq \bar{\rho}_{\text{lim}}$ (r and f refer to the Al_3Zr particles) this would then account for the high temperature subgrain stability.

According to the present results this substructure is no longer stable during hot deformation. A possible explanation for this is that the imposed straining affects the stability of the restraining Al_3Zr particles. To check such a hypothesis the distribution of Al_3Zr particles in a series of specimens given various amounts of deformation has been examined by TEM. This examination demonstrated that the density of Al_3Zr particles did decrease with increasing strain. This effect is

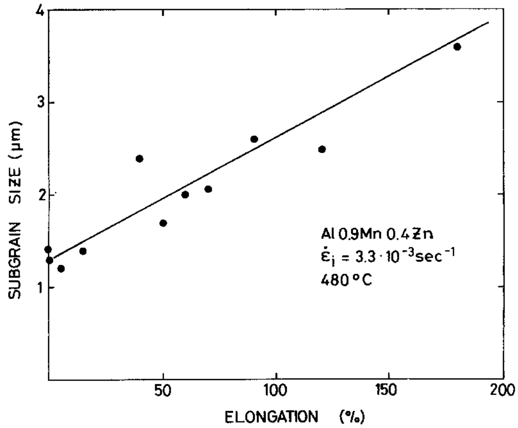

Figure 2 Subgrain size as a function of strain.

illustrated by the TEM micrograph in Fig. 4, taken from a specimen deformed 70%. The strong particle pinning action on the boundary separating the grains marked A and B clearly demonstrates that this boundary has been moving in a direction as outlined by the arrow. The most interesting aspect, however, revealed by this micrograph is that while the grain B contains a high density of Al_3Zr particles (vaguely visible), only a few such particles could be detected in the grain marked A. The Al_3Zr particles appear to have been dissolved by the moving boundary.

Before discussing the effect of this discontinuous particle dissolution reaction on the subgrain stability, the driving force propelling the boundary in Fig. 4 will be briefly considered. As confirmed by the TEM micrographs the subgrains retain their equiaxed shape during hot deformation. This will be energetically favourably when the increase in sub-boundary area due to the deformation exceeds the area fraction occupied by the restraining Al_3Zr particles. To achieve such a shape conservation during plastic flow, subgrain boundary migration must occur, and it follows from simple geometry considerations that the average migration rate will be proportional to the applied strain rate and the average subgrain size.

Why these straining induced migrating boundaries will destabilize the Al_3Zr particles can be interpreted as follows: The Al_3Zr particles are metastable, they have a cubic structure with a lattice parameter close to the aluminium matrix (the equilibrium phase is tetragonal) and they are formed during the pre-treatment with a cube-to-cube particle–matrix orientation relationship [8, 9]. As the majority of these particles are fully coherent with a diameter less than a 100 Å, it may be energetically favourable for the system that they dissolve rather than being forced into a incoherent state by the passing of the boundary. Whether the dissolved particles are being left behind as a supersaturated solid solution or are

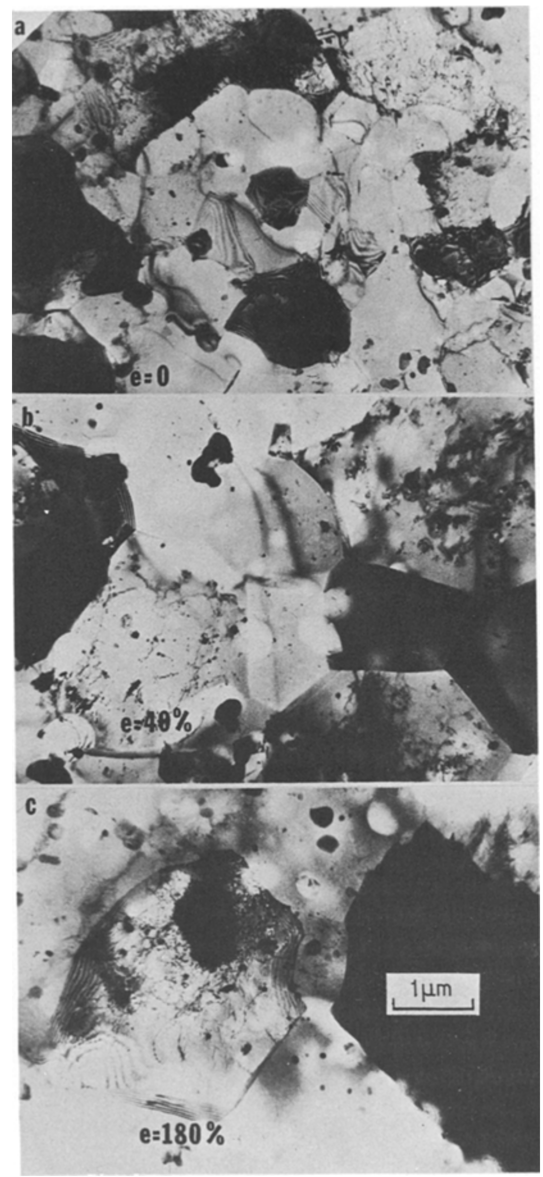

Figure 3 TEM micrographs showing the subgrain structure; (a) in the undeformed state, (b) after 40% and (c) after 180% deformation.

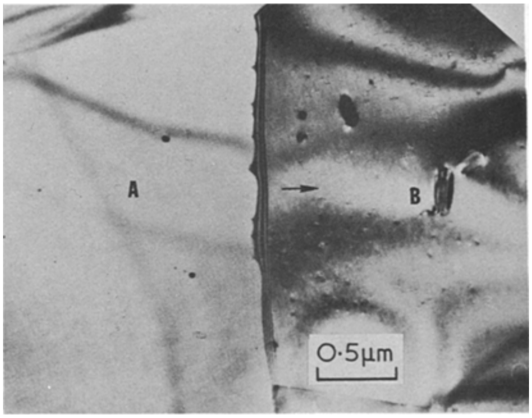

Figure 4 TEM micrograph showing a boundary pinned by particles and moving as outlined by the arrow. Note that grain B contains a high density of Al_3Zr particles while grain A is virtually particle free.

reprecipitate in a much coarser distribution has not been established. In both cases the effective Zener drag force will decrease and the effect on the subgrain structure will be an increase in the particle-stabilized subgrain size with increasing strain. This follows from the simple relation above equating the subgrain size to the particle dispersion, r/f. It is emphasized that the dissolution of particles is the primary effect of the imposed hot deformation, and that this reaction stimulates subgrain growth as a secondary process.

This subgrain growth process will be associated with a gradual increase in orientation difference between neighbouring subgrains and there will be no sharp boundary between what can be characterized as a subgrain structure (Fig. 3a) and a distribution of small grains separated by higher angle boundaries (Fig. 3c). This recrystallization process is analogous to what has been referred to as recrystallization *in situ* or continuous recrystallization [10–12]. Accordingly it is suggested that the present reaction is termed strain-induced continuous recrystallization. It is further suggested that a similar recrystallization reaction accounts for the superplastic behaviour of some of the Zr-bearing aluminium alloys.

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Received 9 December 1977
and accepted 19 January 1978.

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Thermal evidence of overlapping effects of glass transition and crystallization, derived from two different glassy phases in the phase-separated system $Te_{80}Ge_{12.5}Pb_{7.5}$

Interpretation of the calorimetric traces obtained for multicomponent glasses is difficult because of the possibility of a common temperature range for several transformations [1]. In this letter the DSC results obtained for splat-cooled $Te_{80}Ge_{12.5}Pb_{7.5}$ (at.%) glass showing the phase-separation effect are reported. Co-existence of two truly amorphous phases is demonstrated, the glass transition temperature of more stable glass 2 being masked by the crystallization peak of glass 1. Reading of T_{g_2} is possible only after careful pre-heating of the initial sample, which allows crystallization of glassy phase 1.

Fig. 1a shows the DSC-trace for an as-quenched sample. In this case, only one T_g (384.5 K) can be found, and the next two peaks are due to the