RECRYSTALLISATION AND MICROTEXTURE DEVELOPMENT IN AN ALUMINIUM–IRON ALLOY

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ABSTRACT

A high–purity Al–1.41 Fe wt.% alloy has been studied with a view to revealing the role of $FeAl_3$ particles in the development of recrystallisation. Transmission electron microscopy of deformed and partially annealed samples revealed the early stages of recrystallisation at the particles. Crystallographic orientations of individual grains have been measured using electron diffraction methods. Thus, the misorientations of newly recrystallised grains with respect to the surrounding deformed matrix, and also with respect to adjacent recrystallised grains, have been determined. An analysis of these data reveals that the behaviour of matrix grains at the ends of needle shaped precipitates tends to be quite different when compared with the grains at the sides of the precipitates. The implications of the results are discussed in the context of particle stimulated nucleation of recrystallisation.

INTRODUCTION

A major aspect of the physical metallurgy of aluminium alloys is concerned with the development of crystallographic texture during deformation and heat–treatment. The aim of the work presented here is to study some specific effects of precipitate particles in influencing recrystallisation behaviour. Use is made of microtextural data, in which the misorientations between adjacent grains in the vicinity of the precipitate can give valuable information on the early stages of recrystallisation.

EXPERIMENTAL

Slabs (4 cm thick) of a super purity Al–1.41 Fe wt.% cast alloy (< 0.0001 wt.% impurity content) were homogenised by annealing at 400 °C for 1hr, and hot–rolled down to 15mm thickness. After annealing at 380 °C for 1 hr, the samples were cold rolled down to 4 mm, a thickness large enough to extract specimens on a plane normal to the rolling plane, while at the same time including the long transverse direction. Further heat treatments were carried out at 250 °C to study the recrystallisation of the cold rolled microstructure.

Thin foils for transmission electron microscopy (TEM, Philips EM400T at 120 kV) were prepared from the longitudinal section of the rolled billet using electropolishing in a solution consisting of 0.5% nitric acid, 3% perchloric acid and distilled water at -10 °C. Diffraction patterns were obtained without sample tilting whenever measurements were made on a group of some 5–15 grains. Bulk texture was measured using X–ray diffraction on a Siemens diffractometer. Hardness measurements were made on each specimen using a 10 kg load.

AS–DEFORMED MICROSTRUCTURE

The $FeAl_3$ particles were characterised as being about 1–7μm in length and 0.2–0.5μm in width. They were not uniformly dispersed, but occurred in bands approximately 20–40 μm apart in the transverse directions. The mean particle spacings within the bands was found to be about 1–6μm . It is expected that such particles are not shearable. Since they do not change shape during deformation, the matrix around each particle is expected to undergo quite complex deformation, whose character varies with distance from the particle/matrix interface [1].

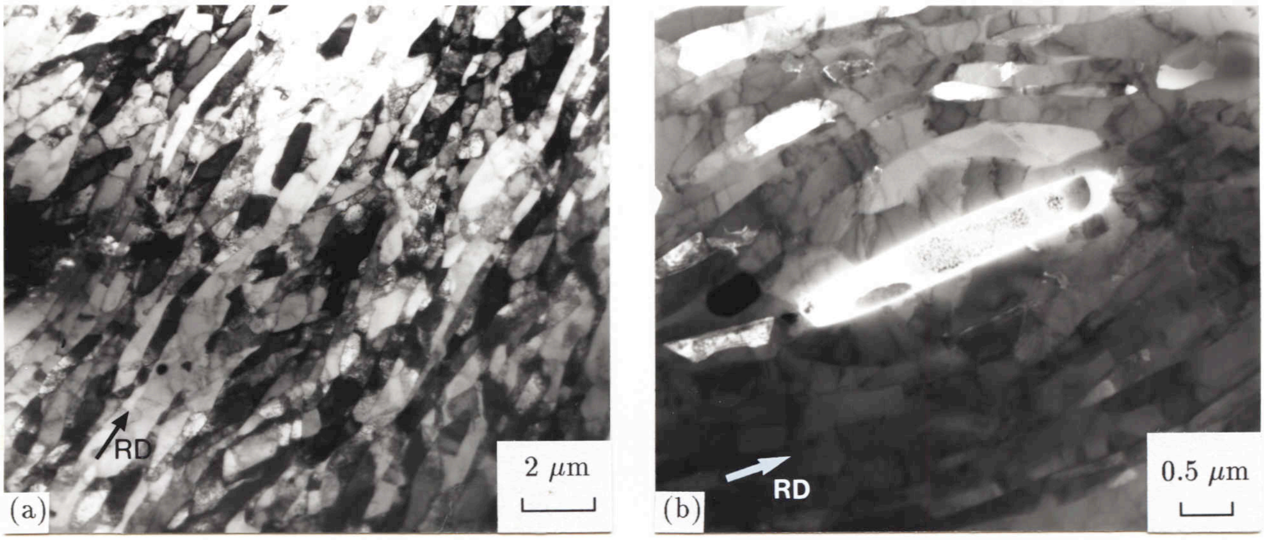

Figure 1: Transmission electron micrographs taken in the long transverse section showing the microstructure of Al–1.41 wt.%Fe Alloy. (a) In the as–rolled condition away from particles, (b) microstructure in the proximity of a particle of $FeAl_3$. Note that the particle has in fact dropped out of the foil.

Areas free of large precipitates, illustrating elongated subgrains and grains together with some light shear banding are illustrated in Fig. 1a. The region surrounding a $FeAl_3$ precipitate is shown in Fig. 1b; there is some indication that the microstructure has more intense deformation near the ends of the elongated precipitate when compared with the regions adjacent to its longer dimension. There will, in general, be a deformation and crystallographic orientation gradient in a direction away from the particle/matrix interface. To assess the effect of the particle in the development of such gradients (and hence on recrystallisation behaviour), it was deemed necessary to measure the orientation of the grains adjacent to the particle relative to those located in the particle–free matrix some 1–2μm from the particle/matrix interface. Such measurements are henceforth denoted "remote orientations". Fig. 2a shows a histogram of experimentally measured remote orientations, expressed in the form of a coincidence–site lattice parameter Σ. Note that the values of Σ do not imply exact orientations, but are intended to represent the nearest value to the actual orientations. This method allows a ready classification in terms of the energy of boundaries. It is evident that the misorientations are quite large. For comparison purposes, Fig. 2b shows the misorientations between the adjacent elongated subgrains in the remote matrix; since all these measurements indicate very small misorientations, the elongated grains are true subgrains, and a comparison of Figs. 2a and 2b demonstrates the major perturbation caused by the presence of particles.

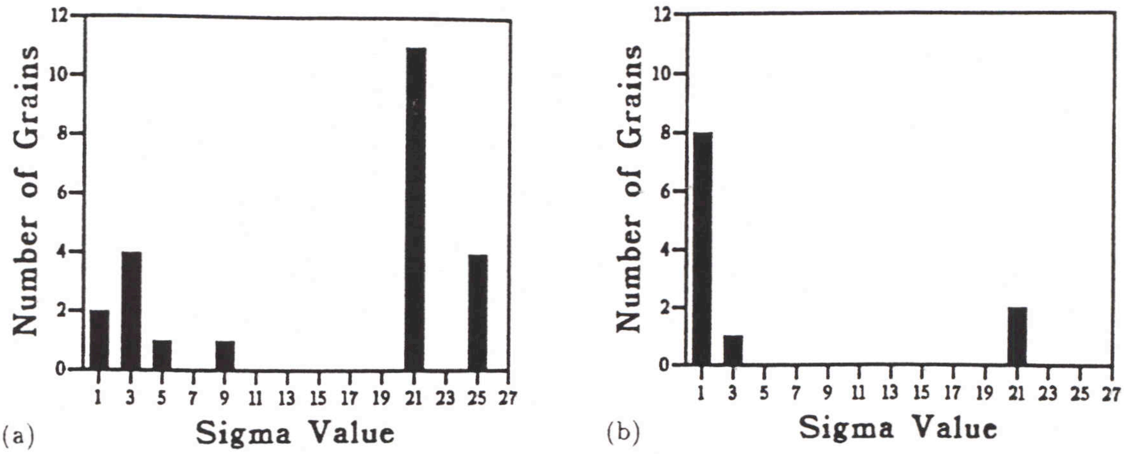

Figure 2: (a) Misorientations between grains adjacent to the particles and remote grains in the unperturbed matrix. (b) Misorientations between adjacent elongated subgrains in the matrix remote from large particles.

PARTIAL RECRYSTALLISATION

Annealing the cold–deformed microstructure produced the usual recovery effects, together with the early stages of recrystallisation. Figure 3a shows a growing recrystallised grain located at the tip of a $FeAl_3$ particle. It is apparent from the data presented in Figs. 3b & 3c that the recrystallised grains have more or less the same remote orientations as the deformed grains adjacent to the particles. Consistent with a lot of published work [2,3], this implies that recrystallisation involves strain–induced subgrain boundary migration, rather than the formation of entirely new grains. Thus, any differences in the development of crystallographic texture during recrystallisation in particle–free and particle–containing alloys of this kind must be attributed to the deformation gradients caused by the presence of particles. It would therefore be of interest in future work to compare microtexture development in super–pure aluminium without the particles.

Figure 4a shows the changes in hardness as a function of annealing time at 250 °C; it is evident that only small changes in hardness occur, consistent with the fact that only partial recrystallisation and recovery is observed. Note that the fully annealed alloy has a hardness of about 30 Hv. In fact, recovery is the dominant process for time periods less than about 30 min. The bulk texture measurements (Fig. 4b and Table 1) also show little change, partly because the degree of recrystallisation is rather small. In addition, the mechanism of recrystallisation appears to be strain–induced grain boundary migration, in which case, much of the original orientations in the deformed microstructure are expected to be preserved, especially at the early stages of recrystallisation.

It was often found that for a given particle, the Σ values (corresponding to the remote orientations) for clusters of grains at the particle tended to be similar. This might be expected since the deformation gradient may not vary much along the length of the particles. Each cluster was identified by the dominant Σ value for the cluster, irrespective of whether it contained recrystallised or deformed grains. A distinction was made between the clusters located at the ends of the particle and those along the particle length.

The end clusters were always found to give large Σ values (Fig. 5a), indicating relatively intense deformation at the particle ends. This is consistent with previous calculations of the strain field around particles in deformed matrices [3], and with the common observation that the ends of particles often are preferential sites for the nucleation of recrystallisation [4].

Annealing has the effect of eliminating many of the large Σ misorientations present in the deformed microstructure at the particle ends (figures 5b and c compared to figure 5a). This is presumably because the large misorientation grains are the first to grow into large recrystallised grains, and hence are eliminated from the measurements, where only on those particles where recrystallisation is at an early stage are focussed upon.

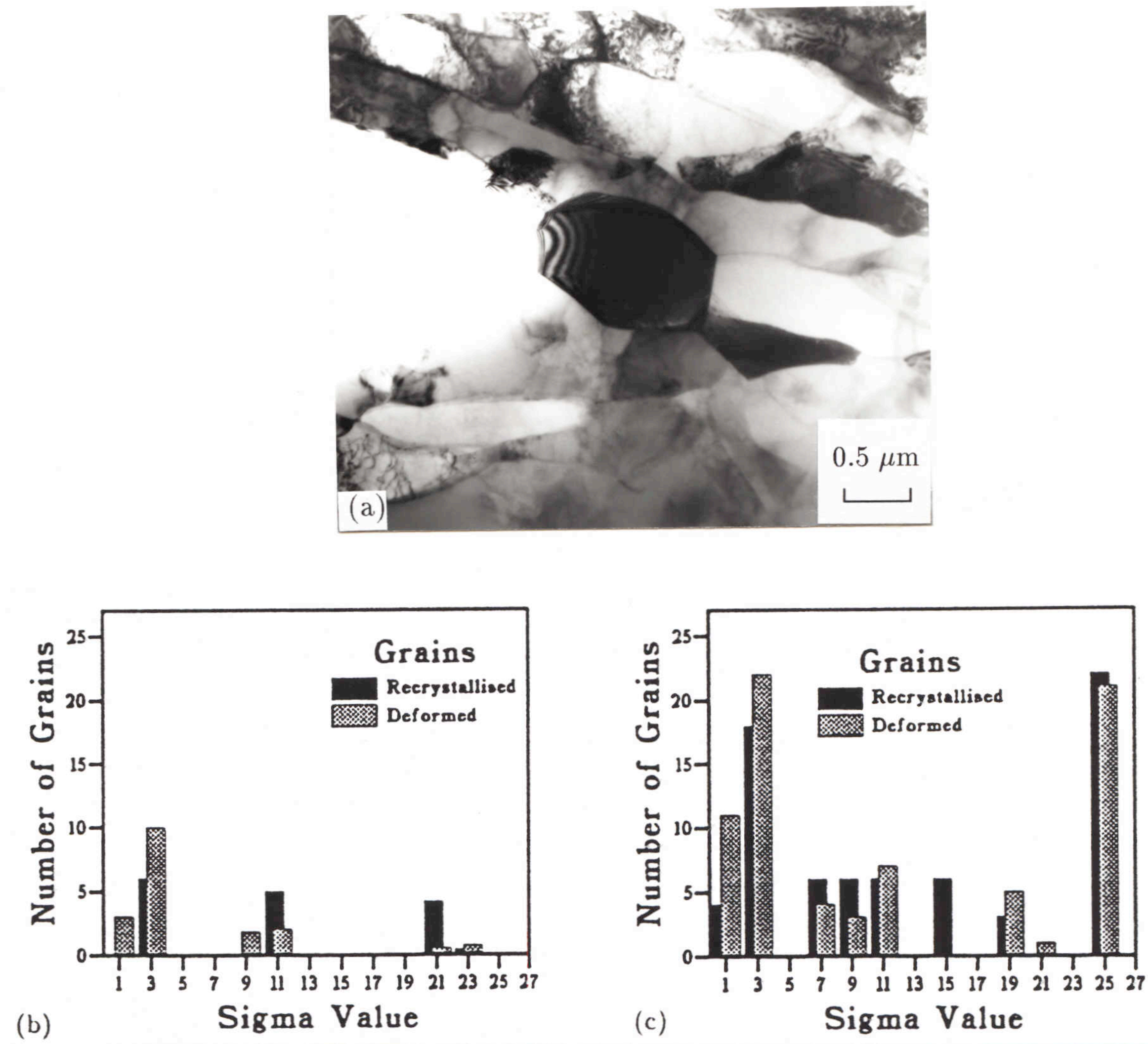

Figure 3: (a) Transmission electron micrograph taken in the long transverse section showing a growing recrystallised grain at the tip of a particle, during annealing at 250 $°C$ for 60 min. (b) Histogram showing the remote misorientations for the recrystallised and unrecrystallised grains adjacent to a particle, after annealing at 250 $°C$ for 30 min. (c) Similar data for 60 min at 250 $°C$.

CONCLUSIONS

In the Al–Fe alloy studied, the presence of large, needle shaped, non–deforming particles of $FeAl_3$ influence the development of crystallographic texture during recrystallisation by causing inhomogeneous deformation on a microscopic scale. The particle shape also has consequences, in that the particle ends are associated with the development of regions of larger misorientation relative to the remote matrix, when compared with the matrix along the particle length.

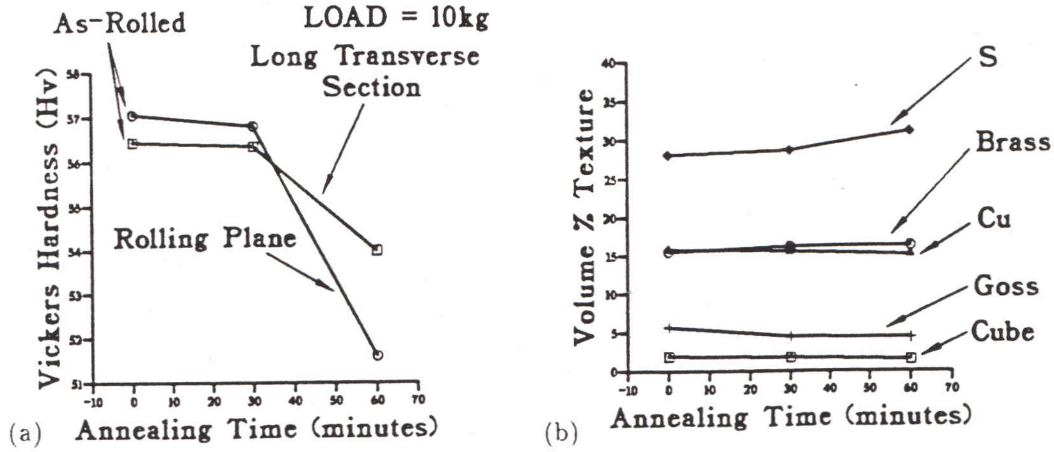

Figure 4: Changes in (a) hardness and (b) bulk texture, during annealing at 250°C.

Table 1: Bulk texture changes during annealing at 250°C.

Texture Type	Annealing Time (minutes)		
	0	30	60
S-type	28.09	28.52	30.83
Brass	15.48	16.20	16.22
Cu-type	15.85	15.47	15.4
Goss	5.72	4.47	4.33
Cube	1.95	1.76	1.48

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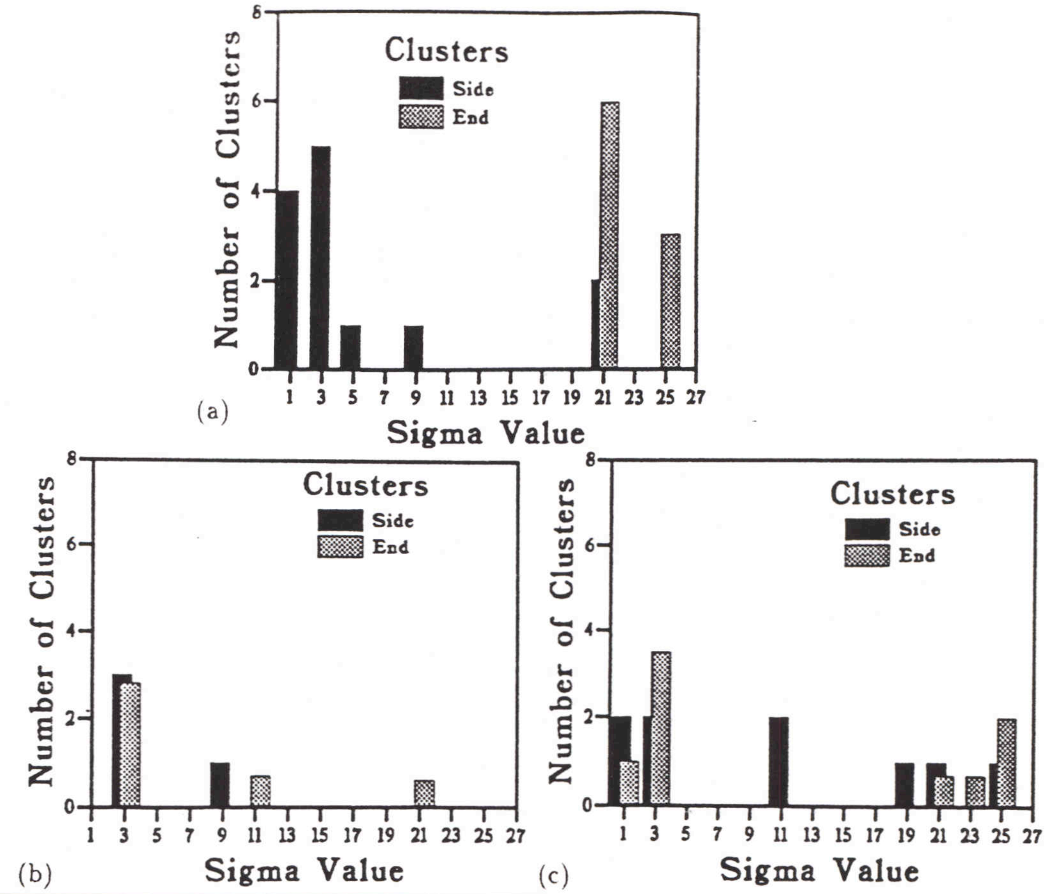

Figure 5: The frequency of grain cluster remote orientations. (a) Deformed sample. (b) Annealed at 250 °C for 30 min. (c) Annealed at 250 °C for 60 min.