Textures and Microstructures, 1988, Vols. 8 & 9, pp. 233–246
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Printed in the United Kingdom

Texture Changes in an Al–Li Alloy Sheet Superplastically Deformed to High Strains

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(Received 25 August 1987; in final form 23 September, 1987)

Dedicated to the memory of Professor Günter Wassermann

Changes in the texture and grain shape of an Al–Li alloy sheet superplastically deformed into the shape of a cone have been measured as a function of strain. The starting microstructure was characterised by strong {110}⟨112⟩ and {123}⟨634⟩ orientations, together with a number of minor orientations, and an elongated grain shape. Increases and decreases in the intensities of the individual orientations occurred at low values of strain (<0.5) as the elongated grains became equiaxed. Once all the grains were equiaxed, however, there was a continuous decrease in the intensities of all orientations, except for those of some minor orientations which remained essentially constant. The results are explained in terms of slip occurring in the early stages of deformation, because elongated grains will experience difficulty in sliding and rotating, but becoming negligible when grains are equiaxed and are free to slide and rotate. There was some evidence of dynamic recrystallisation but only whilst slip occurred.

KEY WORDS: Al–Li alloys, superplastic deformation, pole figures, ODF analysis.

INTRODUCTION

The role of texture and slip during superplastic deformation continues to be a contentious issue, as witnessed by recent correspondence (Padmanabhan, 1980; Melton and Edington, 1983). Padmanabhan (1980) argues that published data on texture

changes are explicable entirely by grain boundary flow processes, whilst Melton and Edington (1983) maintain that slip, activated as a direct consequence of the applied stress and favoured by a high Schmid factor and a strong texture, makes a small but positive contribution to the overall strain. In this paper results for a single-phase Al–Li alloy sheet superplastically deformed to high strains are presented, which indicate that texture can exert a positive influence on slip but that it is restricted to a well-defined strain range since its contribution would appear to be related directly to grain shape.

EXPERIMENTAL

A sheet of the 8090 Al–Li alloy (composition in wt%: Al-2.39 Li-1.21 Cu-0.64 Mg-0.12 Zr) was superplastically deformed under biaxial loading into the shape of a cone (Figure 1). Note that in this, and subsequent, figures RD refers to the original rolling direction in the sheet, TD to the transverse direction and ND to the direction normal to the cone surface. The forming conditions used are listed in Table 1.

Specimens were then cut from the undeformed flange (to represent the thermally-cycled condition) and from various posi-

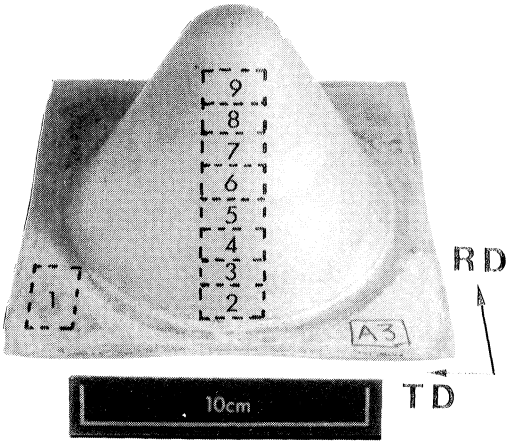


Figure 1 Cone superplastically formed from an Al–Li sheet.

Table 1

Forming temperature (°C)	Strain rate† (s^{-1})	Forming pressure MPa	Hydrostatic pressure‡ MPa
515	2.7×10^{-4}	0.138	3.45

† By the use of a pressure-time cycle calculated to give this strain rate.
‡ To prevent cavitation (see Ridley and Pilling (1985) for a recent review).

tions up the face of the cone. The positions of these specimens (1–9) are indicated in Figure 1. The initial thickness in the flange was 2.35 mm, which tapered to 0.22 mm just below the top of the cone. Values of the average superplastic true strain (calculated from the reduction in average thickness $e = \ln t_0/t_f$, where t_0 and t_f are the original and final thicknesses) indicated that the thickness of 0.22 mm corresponded to a strain of 2.4 (~1100% elongation). These values of strain are listed in Table 2. To minimise problems due to texture gradients (Partridge *et al.* 1986) all specimens were polished to half-thickness prior to texture analysis.

Table 2

Specimen no.	1	2	3	4	5	6	7	8	9
Average superplastic true strain (e)	0	0.25	0.5	0.75	1.0	1.25	1.5	2.0	2.4

Complete (111), (200) and (220) pole figures were measured on an automated Siemens texture goniometer using CuK_α radiation and the results were processed to produce three-dimensional orientation distribution function (ODF) data, uncorrected for ghosts (Lücke *et al.*, 1981; Lee *et al.*, 1987).

Changes in microstructure, particularly grain shape and size, were determined on polished cross-sections of samples cut from different positions up the face of the cone.

RESULTS

The starting texture was similar to that reported for 1.6 mm 8090 sheet (Partridge *et al.*, 1986), consisting predominantly of a very

strong $\{110\}\langle 112\rangle$ texture, with some $\{123\}\langle 634\rangle$ (Figure 2a). The initial grain shape is shown in Figure 2b.

Selected results showing changes in pole figures and grain shape as a function of superplastic strain are shown in Figures 3–5. Little change in the pole figure is apparent at $e = 0.5$, after which the pole intensities decrease with some rotation of poles apparent at a strain of 2.4. Note that for Figure 5(a), however, the specimen thickness of $\sim 100\,\mu$m means that the pole figure is an average of the centre and edge ($\{112\}\langle 111\rangle$ (Partridge *et al.*, 1986)) texture in this alloy and is not strictly comparable to the other pole figures. (Relative changes in the strengths of individual orientations as a function of component shape, specimen position, thickness and superplastic strain are currently being studied). The initially elongated grain shape was broken down gradually by the superplastic deformation. After a strain of 0.4 there were still some elongated grains (Figure 3b) but after a strain of 1.0 all elongated grains had been removed (Figure 4b). Growth of equiaxed grains was also apparent with increasing strain (see Figures 2b–5b).

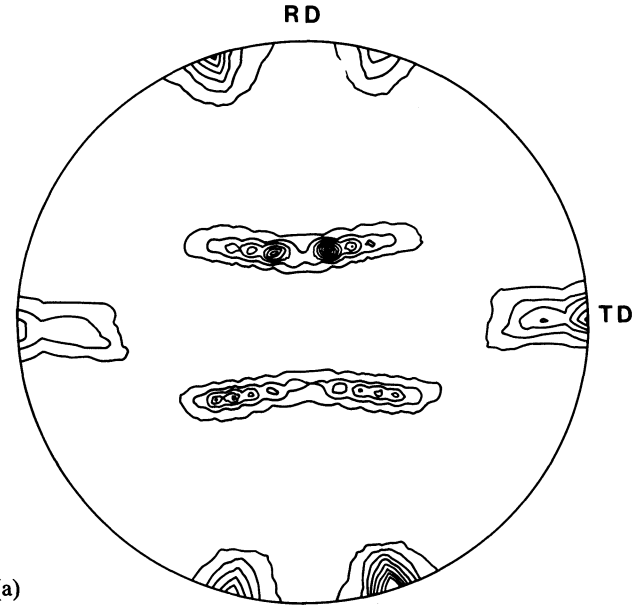
More detailed analysis at smaller strain increments, using the ODF data, showed that there were six identifiable orientations, the intensities of which did not decrease gradually but exhibited perturbations in intensities at low strain values (Figure 6). These changes are summarized below:

– the intensity of the $\{110\}\langle 112\rangle$ orientation decreased and then increased in the strain interval to 0.5, after which it decreased continuously until it had reached an intensity of ~ 2 at $e = 2.4$.

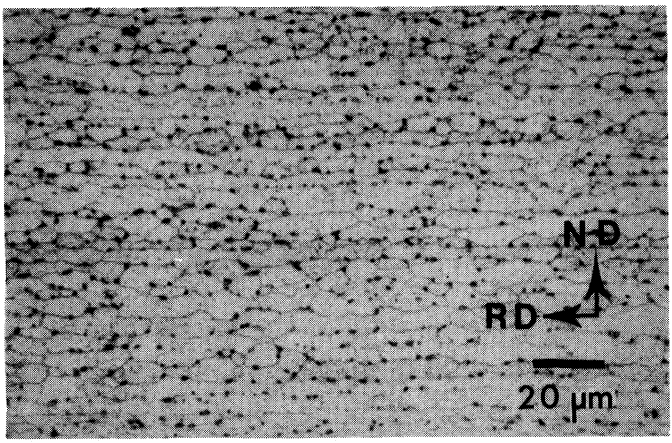
– the intensity of the $\{123\}\langle 634\rangle$ orientation increased markedly at low strain values, almost doubling in intensity at $e = 0.5$. At larger values of strain it followed the trend of the $\{110\}\langle 112\rangle$ orientation and decreased gradually until its intensity was also about ~ 2 at $e = 2.4$.

– the $\{112\}\langle 111\rangle$ orientation changed its intensity very little with strain, remaining at ~ 2 after a strain of 2.4.

– the $\{111\}\langle 112\rangle$ orientation increased in intensity at $e = 0.5$ but had disappeared at a strain of 0.75. It is unclear at the present time whether this orientation, equivalent to 19° or 90° rotation of $\{112\}\langle 111\rangle$ about the transverse direction, is a ghost.

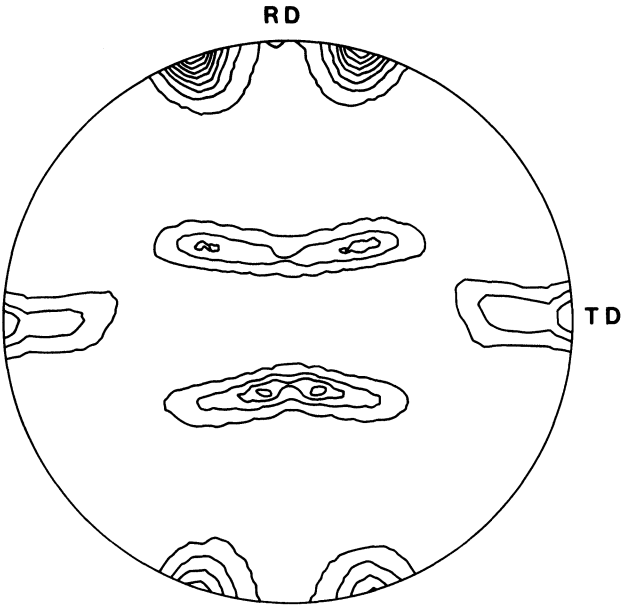


(a)

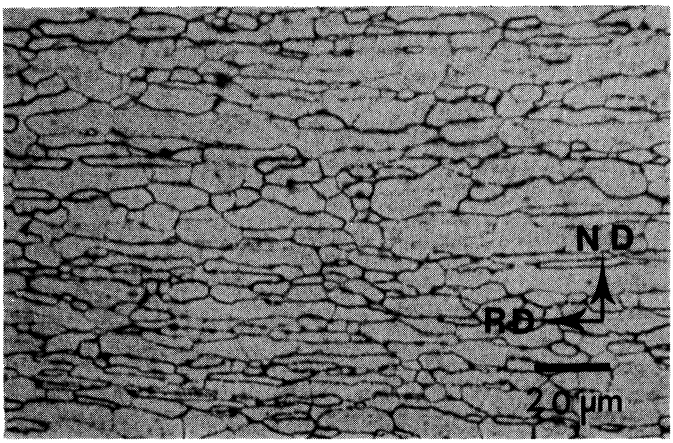


(b)

Figure 2 (a) (111) pole figure from the flange (thermally-cycled) region of the cone (contours × 5 random). (b) Microstructure in flange region.

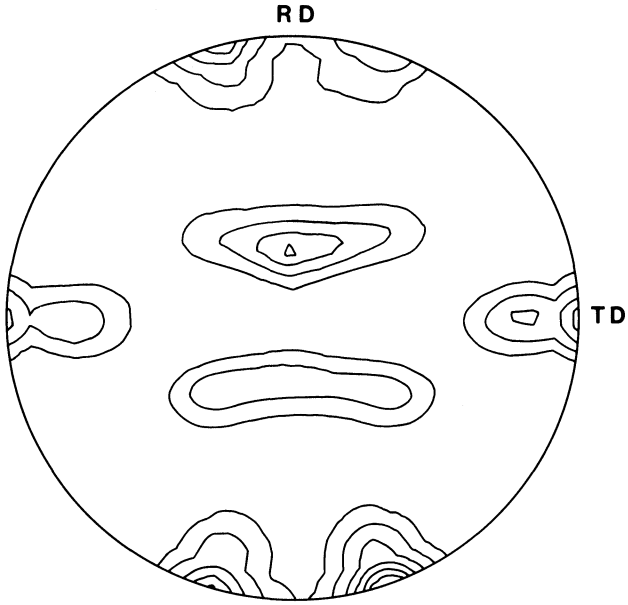


(a)

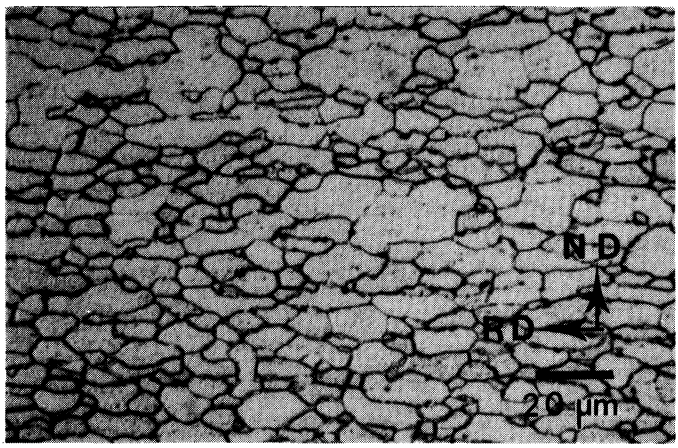


(b)

Figure 3 (a) (111) pole figure from wall of cone after a strain of 0.5 (contours × 5 random). (b) Microstructure in wall of cone after a strain of 0.4.

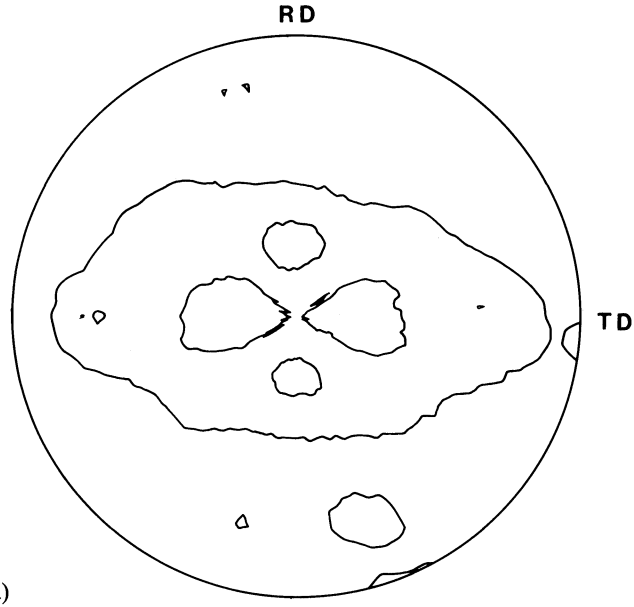


(a)

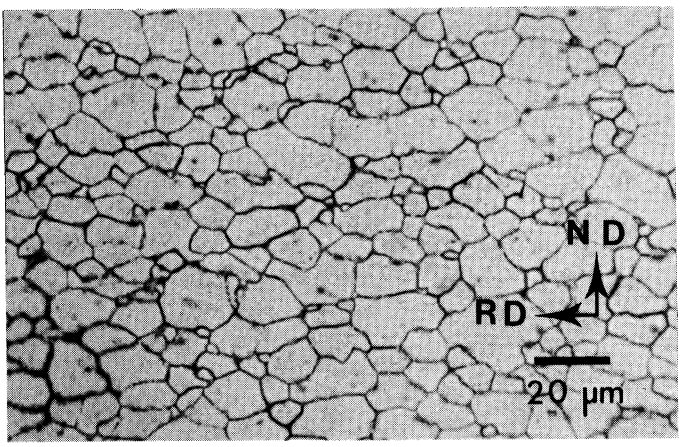


(b)

Figure 4 (a) (111) pole figure from wall of cone after a strain of 1.0 (contours × 3 random). (b) Microstructure in wall of cone after a strain of 1.0.



(a)



(b)

Figure 5 (a) (111) pole figure from wall of cone after a strain of 2.4 (contours × 3 random). (b) Microstructure in wall of cone after a strain of 2.4.

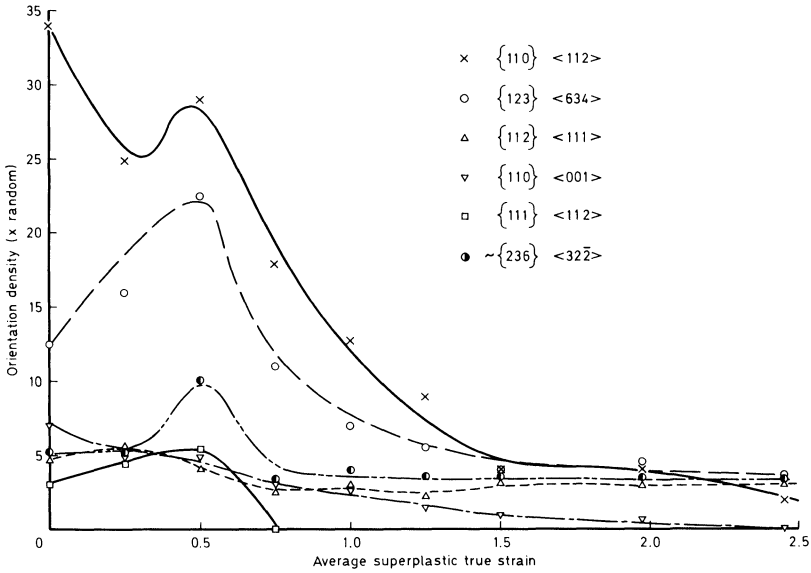


Figure 6 Orientation density as a function of average superplastic true strain for some ideal orientations.

– the $\{236\}\langle 32\bar{2}\rangle$ orientation doubled in intensity at $e = 0.5$, after which it decreased to ~3 and remained at that level to $e = 2.4$.

– the $\{110\}\langle 001\rangle$ orientation decreased continuously with strain until it was zero at $e = 2.4$; although there was a small perturbation at $e = 0.5$.

DISCUSSION

The results in Figure 6 show clearly that texture intensity decreases smoothly with increasing superplastic strain only at strains greater than 0.5. At lower values of strain there are distinct increases and decreases in the intensities of many of the individual orientations. Since these intensity changes occur over the same strain range as the removal of the elongated grains, it is pertinent to enquire if there is a casual relationship between these two trends.

All models for the mechanisms of superplastic deformation involve the accommodation of the imposed strain by the sliding and

rotating of equiaxed grains; with minor contributions locally from slip and diffusion in order to maintain material continuity (see Kashyap and Mukherjee, 1985 for a recent review). In these circumstances any texture present would be expected to decrease gradually in intensity. Note that slip as an accommodation mechanism would not be expected to alter texture since its net effect would be zero (Bowen, 1987, Edington, 1982).

When grains are elongated, however, three effects occur:

a) There is a reduction in the proportion of grain boundary area subjected to a high shear stress.

(b) The displacement strains required by adjacent grains increase markedly.

c) The constraint due to the elongated grains should relax the requirement for five independent slip systems.

These three factors will reduce, either directly or indirectly, the ease of grain boundary sliding with the result that an elongated-grain microstructure, assuming it does not fracture, will seek an alternative means of accommodating the imposed strain. Slip will be the most likely mechanism, and should be identifiable by texture analysis (Bowen, 1987). It must be remembered, however, that elongated grains become equiaxed with increasing strain so that these factors will diminish in importance with strain. Thus the contribution of slip as an alternative mechanism for superplastic deformation would be expected to be a maximum in the initial stages and to decrease in significance until it was negligible when the grains were fully equiaxed.

If this reasoning is applied to the results on the alloy studied here, then the initial microstructure is clearly non-ideal from a superplastic deformation viewpoint. As the strain is applied the grains would not be expected to slide readily. From the changes in texture at low strains the increase in the $\{123\}\langle 634\rangle$ orientation, which is a deformation texture, indicates clearly that slip is occurring in some grains, which must rotate into this orientation as a result of single slip. But since the $\langle 634\rangle$ axis is not stable grain rotation will continue (towards the slip direction) until $\langle 112\rangle$ is reached. This would then explain the increase in intensity of the $\{110\}\langle 112\rangle$ orientation at $e = 0.5$. The increase in the $\{112\}\langle 111\rangle$ orientation at

low strain values could also be claimed to be evidence for slip although the change is small. The overall trend for the $\{110\}\langle 112\rangle$ orientation is to decrease, in marked contrast to the results for flat-bottomed boxed formed from the same alloy (Bowen and Hirsch, 1987). This difference is most likely due to the large out-of-plane displacements involved in forming the cone, which force the elongated grains out of their initial orientations. The remaining significant change at low strain values—in the $\{236\}\langle 32\bar{2}\rangle$ orientation—is indicative of some recrystallisation and may reflect the nucleation of new equiaxed grains due to slip and/or grain boundary motion. It is tempting to conclude that the increase in intensity of the $\{236\}\langle 32\bar{2}\rangle$ orientation is a reflection of the increase in the intensity of $\{123\}\langle 634\rangle$ (with dislocation motion providing the driving force for recrystallisation) and that as soon as dislocation motion begins to reduce at $e = 0.5$ recrystallisation ceases and the intensity of $\{236\}\langle 32\bar{2}\rangle$ then decreases because the grains in this orientation rotate away as a result of grain boundary sliding and rotation. Such recrystallisation would be classed as dynamic. At strains >0.5 most grains are equiaxed and the alloy is then almost ideal for superplasticity. All orientations consequently decrease in intensity, although not dropping below ~ 3 random (Figure 6).

It is reasonable to conclude, therefore, that there is a causal relationship between grain shape and texture changes during superplastic deformation. This is an important observation, for two reasons:

i) It means that alloys with microstructures non-ideal for superplasticity are capable of being deformed successfully. This has economic significance since special processing to produce small equiaxed grains (Hamilton *et al.*, 1982) may not be essential; although more work needs to be done to check this on alloys of varying grain aspect ratio and texture type and intensity, and an important parameter in these circumstances may be the magnitude of the imposed hydrostatic pressure (Bowen and Hirsch, 1987);

and, more significantly,

ii) It offers a rationale for the divergence of views on the apparently conflicting results on texture changes in superplastically

deformed alloys, since it would now appear that there may not, in fact, be a conflict because different results may be merely a consequence of different initial microstructures—many of the published results, for example, are for extruded materials and metallographic evidence of the initial microstructure is often lacking; see Padmanabhan and Lücke (1986) for a recent review. It is recommended, therefore, that future texture studies should check this link by following changes in texture and grain shape, ideally examining alloys with initially elongated and equiaxed grains but with different texture types and intensities.

Much of the microstructural evaluation of superplastic deformation (see Kashyap *et al.*, 1985 for a recent review), including texture studies (Partridge *et al.*, 1985; Padmanabhan and Lücke, 1986), has been concerned with uniaxially deformed alloys. From the few results that are available on biaxially deformed alloys, however, it is clear that texture behaviour under the two regimes is not the same (Partridge *et al.*, 1986; Bowen and Hirsch, 1987). This is perhaps not surprising when the effect of a second stress axis is considered and this difference should be borne in mind when studying superplasticity as a function of stressing mode. Other related points are:

– strain rate, where it is more difficult to be certain of the exact strain rate during biaxial forming, with the concern being that forming is, in reality, faster than optimum.

– the shape of a component may influence material flow and hence texture changes. The bottoms of flat-bottomed boxes, for instance, are deformed in essentially plane strain conditions (with much smaller out-of-plane displacements than in the cone studied here) and result in very little change in texture in essentially the same microstructure (Bowen and Hirsch, 1987). (Note that this work (Bowen and Hirsch, 1987) has shown that grain boundary sliding and grain rotation produce distinctly different changes in texture).

– the effect of hydrostatic pressure, which will not only suppress cavitation (Ridley and Pilling 1985) but may also influence flow processes during superplastic deformation (Bowen and Hirsch 1987).

– when grains grow during deformation slip may become active because the proportion of grain boundary area becomes insufficient to accommodate sliding. In these cases the distribution of grain sizes (Ghosh and Raj, 1985) will influence the extent of slip locally; although there is no known evidence of changes in texture specific to these circumstances.

Finally, it should be noted that in studying texture changes in two-phase alloys the superplastic deformation characteristics of the individual phases need not be the same, with the softer phase accommodating most of the strain (Heubner *et al.*, 1973, Melton *et al.*, 1974, Cutler *et al.*, 1974, McDarmaid *et al.*, 1985, Bowen *et al.*, 1987).

CONCLUSIONS

A study of the rate of change of microstructure and texture in an Al–Li alloy sheet superplastically deformed into a cone has shown that:

i) Most of the initially elongated grains were equiaxed after a strain of ~0.5.

ii) Some texture components increased in intensity below a strain of ~0.5, but all components became weaker at higher strains.

iii) The increased intensity of these texture components occurred because elongated grains cannot rotate easily with the result that slip had to occur as a temporary means of accommodating the imposed strain. Some dynamic recrystallisation accompanied slip.

iv) Once grains became equiaxed they were free to slide and rotate and all texture components then decreased in intensity with further superplastic strain.

Acknowledgements

The author would like to thank A. Shakesheff for providing the cone and micrographs, and Dr J. Hirsch (RWTH Aachen) for recent discussions.

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