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TEXTURE AND PLASTIC ANISOTROPY IN MODEL ALUMINIUM - LITHIUM ALLOYS

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ABSTRACT

The specific roles of Li, Zr and Cu alloying elements in Al-Li alloys have been investigated in terms of their influence on texture evolution during rolling and subsequent plastic anisotropy. Three model alloys; Al-2.3% Li, Al-2.3% Li-0.1% Zr and Al-2.3% Li-1.2% Cu-0.1% Zr were hot and cold rolled to reductions up to 92% and their texture evolution systematically characterised by X-ray pole figures and complete ODFs. Hot rolling of the Zr containing alloys, which do not recrystallize, led to a strong Bs texture component. Cold rolling of all alloys gave the usual Taylor-type ß fibre components, with certain variations according to the precipitation state which can favour or inhibit shear banding.

Pronounced plastic anisotropies, $R(\alpha)$, were measured on the rolled sheets with R_{min} at $\alpha = 0^\circ$ and R_{max} at $\alpha \approx 60^\circ$. Actual values depend upon rolling temperature and strain but typically $R_{min} \approx 0.3$ and $R_{max} \approx 1.5$ to 2. The quantitative texture data was used together with the CMTP and Taylor models to predict the anisotropy; in general the CMTP model gave a reasonable correlation with the experimental values.

1. INTRODUCTION

It is now known that the new high strength Al-Li based alloys going into commercial production often have particularly strong textures; for example hot rolling can develop a pronounced Bs type deformation texture that is difficult to remove since recrystallization does not occur easily in these alloys. The development of this strong hot rolling texture appears to be related to the presence of certain alloying elements but, in complex industrial alloys, the roles of the individual elements are not easily characterised [1,2]. The aim of the present study is therefore to systematically determine the influence of common alloying elements: Li, Zr and Cu in model Al-Li alloys, on hot and cold rolling texture development and subsequent plastic anisotropy.

2. ALLOYS AND EXPERIMENTAL METHODS

The following three high purity Al-Li base alloys have been used in this study: A: Al-2.3% Li; B: Al-2.3% Li -0.1% Zr and C: Al-2.3% Li- 1.2% Cu- 0.1% Zr.

The alloys were received in the form of hot extruded bands of width 100 mm and thickness 13 mm. Samples of these alloys were hot rolled (in 1 direction in the temperature range 420 to 250° C) to reductions of 70, 85 and 92%. After hot rolling and static cooling A was partially recrystallized whereas B and C showed no clear signs of recrystallisation. Other samples, in the as-extruded state, were cold rolled using reverse passes to reductions of 50, 70, 85 and 92%. Metallographic examination revealed shear band formation in B and, to a smaller extent, in A but very few shear bands in C.

Four incomplete pole figures were determined for each sample and used to evaluate complete ODFs by the spherical harmonics method using the Gauss mode functions proposed by Lücke et al. [3]. F(g) sections were also plotted along the important skeleton lines, eg. the ß fibre, and volume fractions of the main texture components evaluated.

Standard tensile tests were carried out on rolled sheet samples to determine the flow stress $\sigma(\alpha)$ and plastic strain rate ratios $R(\alpha)$, where α is the angle of the tensile axis to the rolling direction. The Lankford coefficient R, defined as $\dot{\epsilon}_{width}/\dot{\epsilon}_{thickness}$ at $\epsilon=0$ was determined by back extrapolation to zero strain. Theoretical $R(\alpha)$ values were also calculated from the quantitative texture data by two methods:

(i) the CMTP method which uses a quadratic function to approximate the yield surface whose form depends upon the volume fractions of the texture components [4,5].

(ii) the classical Taylor method [6] which calculates the R value to minimise the internal plastic work rate of slip systems using the C^{l}_{mn} coefficients (even part).

3. RESULTS AND DISCUSSION

3.1 Textures

The <u>initial texture</u> in all three hot extruded alloys were inhomogeneous both in the width and thickness directions. The texture gradients in the width are consistent with the different deformation modes experienced by different parts of the material during hot extrusion from cylinder to slab - fig. 1. Thus in the centre the usual plane strain compression texture components are found, ie. Bs {110}<112>, Cu {112}<111> and S {123}<634> (with Bs exceptionally strong in A but relatively weak in B and C). Towards the edge the <111> + <100> fibre texture are observed; a strong <111> fibre in alloy B and mixed <111> and <100> components in A and C.

The thickness variations of the initial texture were quite complicated; in most alloys a rolling texture, basically Cu+S occurs in mid thickness, Bs increases at about 1/4 thickness and finally some asymmetric components are found close to the surface.

The texture gradients decrease substantially after rolling reductions > 50% but are not completely eliminated. Mid-thickness textures are described in the following.

Some typical textures developed after <u>hot rolling</u> are illustrated by the ODFs of figure 2. Alloy A exhibits rapid changes in texture as a consequence of hot deformation and partial recrystallisation cycles. Thus 70% reduction leads to a rolling texture but recrystallization occurs after 85% leading to a very strong cube texture (+Goss component). Further hot rolling then decreases the cube and increases the rolling components. Alloys B and C do not recrystallize during hot rolling due to the strong grain boundary pinning effect of Zr. Large rolling reductions are possible with relatively homogeneous deformation leading to very strong rolling textures (f_{max} ≈20 to 30). At all deformations > 50% the S component dominates in both B and C. At higher strains there is a clear increase in the Bs component and a steady decrease of Cu. Alloy C tends to develop a less pronouced Bs component than B probably because of the coarse Cu-rich precipitates in C which tend to homogenise the deformation.

The <u>cold rolling</u> experiments also reveal the effects of precipitates and grain size. The cold rolling texture changes are best illustrated by the ß-fibre skeleton lines, figure 3. The general tendency of all alloys is for f(g) of all components between Cu and Bs to increase during cold rolling, although there are some significant differences [7]. Alloy A shows a strong increase of the S component while the initially strong Bs first decreases then increases after 50% reduction. Alloy B develops a component between S and Cu while Bs increases probably as a consequence of extensive shear banding. Alloy C very rapidly develops a strong S component while Bs remains stable. These texture evolutions are generally consistent with the usual Taylor type relaxed constraints model predictions, the detailed differences can be qualitatively correlated with the tendency to shear banding which promotes the Bs component. At room temperature all alloys contain fine shearable δ' Al_3Li precipitates and thus can develop shear bands. These are reduced in alloy C by the Cu rich particles and initially in alloy A by the relatively fine recristallized grains and so shear banding only occurs extensively in alloy B.

3.2 Plastic Anisotropy

The measured values of the Lankford coefficient $R(\alpha)$ of the 92% cold rolled samples are shown in figure 4a. Large variations of R occur between ≈ 0.3 in RD to a maximum at ≈ 60° (1.4 for A, 1.9 for C) and then down to ≈ 0.9 in TD. The calculated $R(\alpha)$ plots are shown in figs 4b and c for the CMTP and Taylor methods respectively. Both methods give roughly symmetric $R(\alpha)$ plots with a maximum at 45°. However the peak values are about 4 to 6 using the Taylor method and only 1.4 using CMTP; the latter is close to the experimental values.

Generally similar behaviour is observed in the hot rolled samples, figures 5 and 6 (70% and 85% respectively). After 70% rolling $R(\alpha)$ exhibits a maximum of ≈ 1.2 at 45° with very low values at 0 and 90°. The CMTP calculations predict the correct shape of the curve and R values only slightly too high. However the Taylor calculations predict asymmetric $R(\alpha)$ curves and R_{max} values of 3 to 4. Figure 6, corresponding to the alloys rolled 85%, reveals the particularly interesting behaviour of alloy A (which at this strain has recrystallized and has a very strong cube texture). Whereas B and C have strong rolling textures (Bs+Cu+S) and R_{max} at ≈ 45°, A exhibits a min R value (0.2) at 45° and a maximum of ≈ 1 at 90° (this is of course expected of a cube texture). The CMTP and Taylor calculations predict the general shapes of the $R(\alpha)$ plots but, yet again, the Taylor method predict excessively large $R(\alpha)$ variations.

It is clear therefore that the large plastic anisotropies of these strongly textured materials are qualitatively consistent with the textures. The CMTP method which in principle is only approximate (using volume fractions estimated by simple Gaussian functions and quadratic functions for the yield surfaces) does in fact give a much better prediction of the $R(\alpha)$ curves than the Taylor method using 5 imposed strain rate components. Royer et al. [6] have shown that less pronounced $R(\alpha)$ variations are obtained using relaxed constrains methods; this may well by the case of strongly textured Al-Li alloys.

4. CONCLUSIONS

The results of this study of texture evolution in model Al-Li alloys can be summarised as follows:

- hot rolling develops very strong S and Bs texture components in Zr containing Al-Li alloys (B and C) since Zr inhibits static or dynamic recrystallization and thus allows the formation of very long thin grains at high strains.
- very strong cube textures can be developed at intermediate hot rolling strains in the binary alloy A by recrystallization
- the presence of Cu-rich particles in alloy C tends to redistribute slip more homogeneously thereby reducing shear banding particularly during cold rolling
- the quantitative texture data (C^{l}_{mn} coefficients and volume fractions of texture components) can be used to predict the plastic anisotropy. In these strongly textured alloys the CMTP method gives better agreement with experiment than the conventional Taylor method

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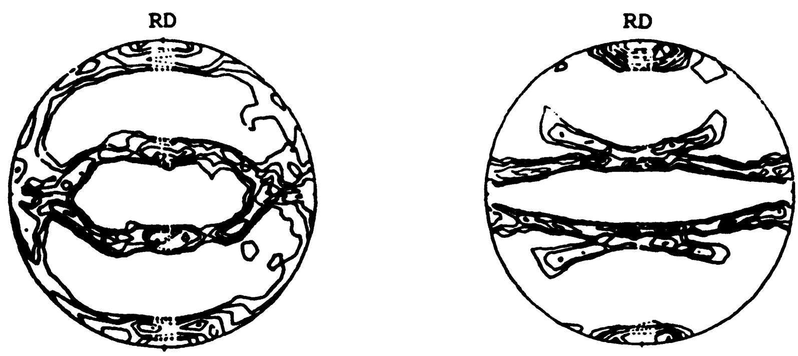
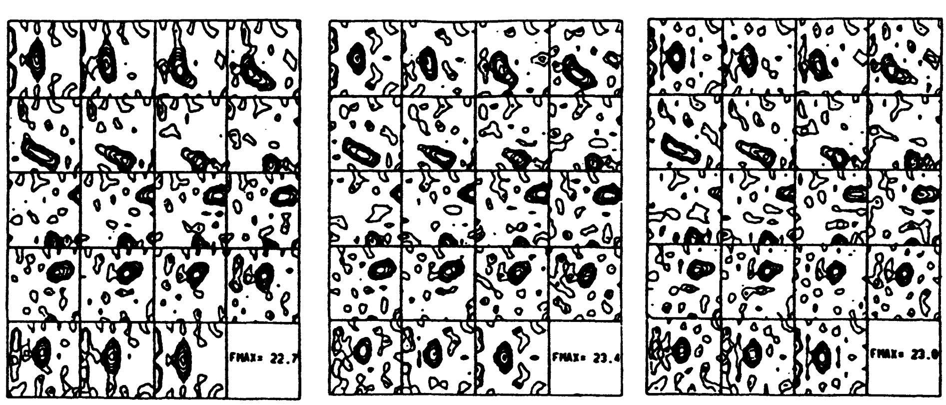


a) middle of the bar (max. 19) b) outer part of the bar (max. 32)

Figure 1. {111} pole figures of the C samples rotated around RD (levels 0.5-1-2-4-7-10-15-20)



a) A (70%) b) B (70%) c) C (70%)

Figure 2. Complete ODFs of the hot rolled samples (levels 1-2-4-7-11-16)

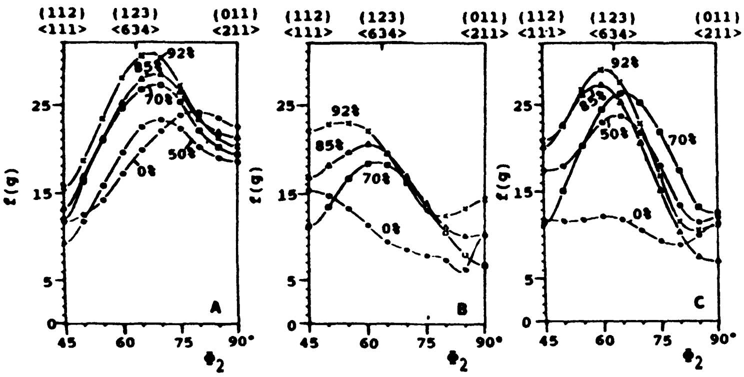


Figure 3. ß-fibre skeleton lines of the cold rolled samples.

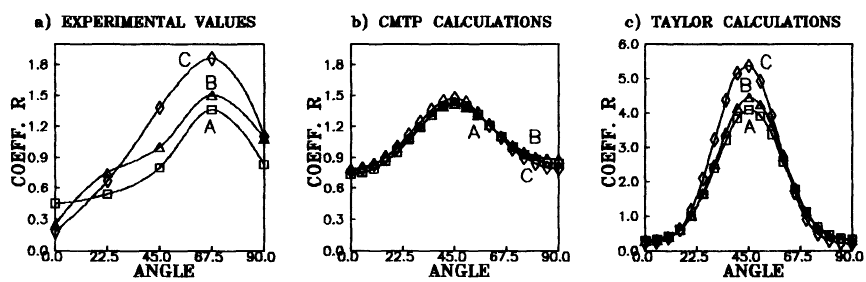


Figure 4. Lankford coefficient R(α) of alloys A, B and C cold rolled to 92%.

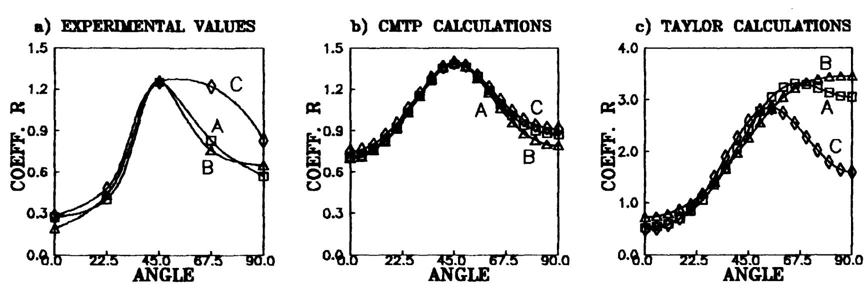


Figure 5. Lankford coefficient R(α) of alloys A, B and C hot rolled to 70%.

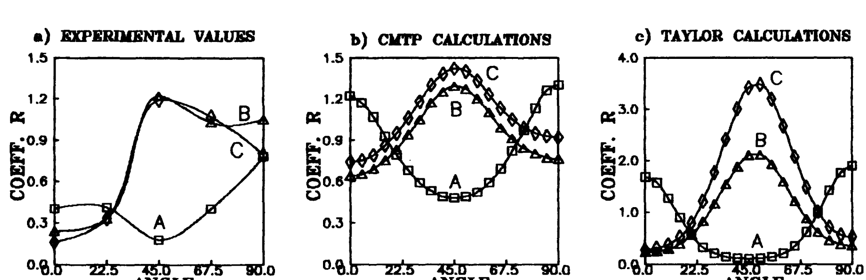


Figure 6. Lankford coefficient R(α) of alloys A, B and C hot rolled to 85%.