

TEXTURE DEVELOPMENT IN Al-3%Mg INFLUENCED BY SHEAR BANDS

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ABSTRACT

The rolling and recrystallization texture development of a high purity Al-3wt%Mg alloy was investigated. Al-Mg alloys are well known to exhibit large amounts of deformation inhomogeneities (shear bands). The influence of these shear bands on the texture development is presented and the results are discussed in terms of various theories.

INTRODUCTION

Several solution treated Al alloys deform rather inhomogeneously at higher deformation degrees which has often been analyzed by means of microstructural investigations in poly- and single crystals¹⁻⁶. Especially in solution treated Al-Cu alloys the influence of the changed deformation mode on the texture development has been studied¹. The present paper analyzes the effect of shear band formation on the texture and microstructure development of an Al-3wt%Mg alloy and tries to interpret the underlying mechanisms.

EXPERIMENTAL PROCEDURE AND RESULTS

An Al-3wt%Mg alloy was produced from 99.999% Al and high purity Mg by continuous casting. By several appropriate cold rolling steps and annealing treatments a 20 mm thick specimen was obtained. The solution treated specimen was homogeneously cold rolled to rolling degrees from 50% to 98%. Pole figures were measured by means of an automatic X-ray texture goniometer⁷. For a more detailed analysis the 3-dimensional orientation distribution functions (ODF) were

calculated by the series expansion method⁸ and ghost corrected by the method of Lücke et al.⁹.

In Fig.1 the ghost corrected ODF's of the 75% and 95% cold rolled samples are represented. They show a typical Cu-type rolling texture with the C- $\{112\}\langle 111 \rangle$, S- $\{123\}\langle 634 \rangle$ and B- $\{011\}\langle 211 \rangle$ orientations assembled along the β -fibre. For the different rolling degrees the orientation density along this β -fibre is plotted versus the corresponding φ_2 -angle in Fig.2. At low reduction degrees deformation is rather homogeneous and the texture development is very similar to that of pure Al exhibiting a predominant C-orientation. Differences, however, become evident after 90% reduction. For the Al-Mg alloy the Goss- $\{011\}\langle 100 \rangle$ orientation remains stable even at high deformation degrees, and an orientation near the 45°ND rotated Cube, the so called Q- $\{013\}\langle 331 \rangle$ orientation, appears (Fig.1b). Fig.2 shows the increase of the B-orientation after 90% reduction which even becomes the major component at 98% reduction whereas only a slight further increase of the C-orientation is obtained. Some micrographs of the deformed samples exhibiting strong shear banding are shown in Fig.3.

Subsequently, the cold rolled samples were recrystallized at 310°C and 450°C in a salt bath so as to exclude grain growth. During recrystallisation at 310°C a dominant Cube-orientation $\{001\}\langle 100 \rangle$ with strong RD-scatterings towards Goss $\{011\}\langle 100 \rangle$ as well as TD-scatterings and a minor R- $\{123\}\langle 634 \rangle$ orientation are formed (Fig.4a). The texture sharpness and particularly the R-orientation are increased at higher reduction degrees. Furthermore, the Q-orientation appears. With increasing annealing temperature both the Cube- and the R-orientation exhibit lower intensities. Instead the random part (phon) and the Q-orientation increase significantly. Fig.4b exhibits the strong Q- and Goss-orientation beside the predominant Cube-orientation of the 98% reduced sample recrystallized at 450°C.

DISCUSSION OF THE ROLLING EXPERIMENTS

At lower reduction degrees deformation is rather homogeneous and the texture development is in accordance to the predictions of the RC-Taylor type models¹⁰. With increasing rolling degree the texture development can not be further explained by Taylor type models. This behaviour can be connected to the metallographically observed onset of strong Cu-type shear band formation at these reduction degrees. As can be seen (Fig.3) the formation of shear bands increases with increasing deformation. These grain scale shear bands are mostly inclined by $\sim 35^\circ$ towards the rolling plane.

In literature shear band formation is discussed in different ways, though the exact mechanisms are not satisfactorily understood yet. Shear band formation seems to be

related to a lamellar structure in the deformed microstructure. While in brass this lamellar structure is built by twin-matrix lamellae³, in the present Al-Mg alloy it seems to consist of a lamellar dislocation arrangement⁴.

In contrast to pure Al which develops a cell or sub-grain structure, Al-Mg alloys exhibit heavy microband formation due to a change in recovery by the pinning effect of the solute Mg-atoms^{4,5}. The microbands - lying parallel to the corresponding slip system and being separated by dislocation walls from the adjacent matrix - form the lamellar structure⁴. By further deformation the microbands accumulate in microband cluster and will form the shear bands by rotation and fragmentation⁴.

Single crystal experiments revealed a strong orientation dependence of shear band formation⁶. Especially the C-orientation exhibits strong shear bands, which is explained by the asymmetric slip geometry of this orientation. The C-orientation activates two coplanar (CP) and two codirectional (CD) slip systems, with the CP-systems bearing twice the amount of slip than the CD-systems. This asymmetric slip distribution enhances the microband formation on the CP-systems and therewith the shear band formation⁶. These microbands reduce the dislocation mobility on the CD-systems since the CD-dislocations would have to cross the microband walls formed by dislocations. This causes an additional enhancement of slip on the CP-systems and the C-orientation will rotate around TD towards the Goss-orientation¹¹. As a metastable orientation the Goss-orientation will rotate around ND towards the B-orientation with increasing deformation. In the present Al-Mg alloy a strong C-orientation develops at low reduction degrees by homogeneous deformation. Hence, the orientation dependence also influences the shear band formation in polycrystals.

By this model the stability of the Goss-orientation and the increase of the B-orientation at high reduction degrees (Fig.1 and 2) is explained. It is important to note that the enhanced activation of the CP-slip systems in the microbands will cause an $+\epsilon_{NR}$ shear (Fig.5). In the neighbouring regions a compensating $-\epsilon_{NR}$ shear (Fig.5) has to be realized by preferred activation of the CD-systems leading to a rotation of the C-orientation towards the $45^\circ ND$ rotated Cube-orientation¹¹. This is in accordance with observations of Nakayama et al.⁴ who found a preferred activation of the CD-slip systems between intense shear bands in C-oriented cold rolled Al-3%Mg single crystals. The Q-orientation is situated about 10° away from the $45^\circ ND$ -rotated Cube-orientation and exhibits a certain stability especially in Taylor type models with relaxed ϵ_{NR} -shear¹². It is assumed that the Q-orientation is generated in the scatterings of the $45^\circ ND$ -rotated Cube-orientation, formed by the mechanism of compensating shear.

DISCUSSION OF THE RECRYSTALLIZATION EXPERIMENTS

Concerning the recrystallization texture the observed dominant Cube-orientation with strong RD- and TD-scatterings and the R-orientation all have pronounced growth capacities by a more or less exact $40^\circ\langle 111 \rangle$ orientation relationship to the deformed matrix¹². The nuclei for the Cube-orientation and its scatterings are assumed to be formed according to the Dillamore-Katoch mechanism¹³. The nucleation of the R-orientation is supposed to take place at grain boundaries formed by two S-variants¹². At higher recrystallization temperatures the growth preference of the Cube- and R-oriented nuclei is decreasing and thus the random part and particularly the Q-orientation are significantly increasing. The Q-orientation already appears in the rolling texture and is supposed to be formed in the neighbouring regions of the shear bands as discussed above. With increasing deformation degree also the Goss-orientation increases (Fig.4b) which is explained by nucleation at shear bands².

Concerning recrystallisation nucleation, the microstructure of the partly recrystallized 95% cold rolled sample reveals nucleation mostly at grain boundaries, but grains nucleated at shear bands prove that nucleation also takes place in this strongly deformed regions (Fig.6). However, these grains seem to grow to slow to be identified at higher recrystallisation states.

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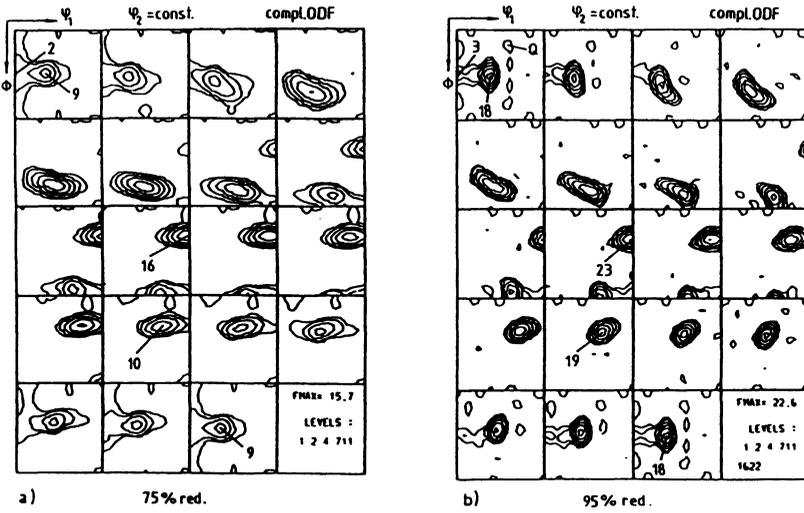


Fig.1 - Ghost corrected ODF's of the a) 75% and b) 95% cold rolled samples of solution treated Al-3%Mg

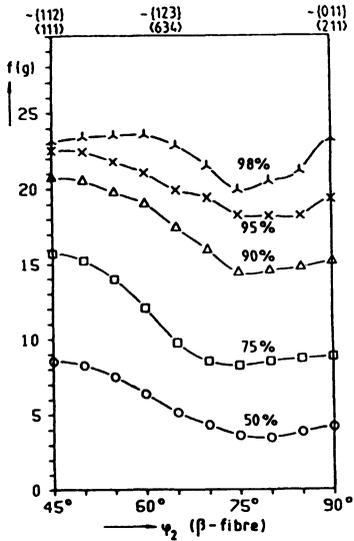
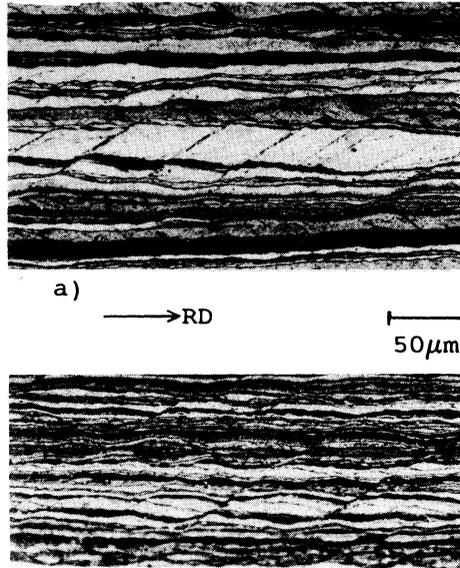


Fig.2 - β -fibre of the cold rolled samples for various rolling reductions



b)

Fig.3 - Micrographs of the a) 95% and b) 98% cold rolled samples (anodically oxidized, X 200)

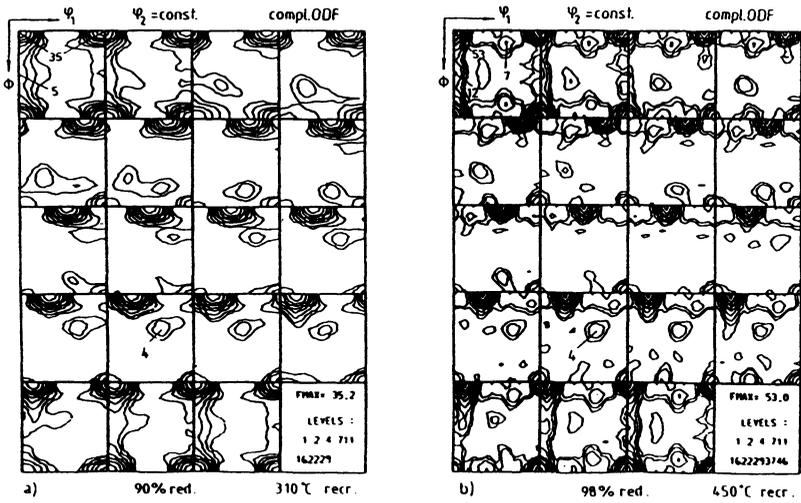


Fig.4 - Ghost corrected ODF's of recrystallized samples:
 a) 90% cold rolled, recrystallized at 310°C,
 b) 98% cold rolled, recrystallized at 450°C

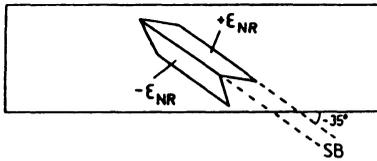


Fig.5 - Schematic representation of the shear component in a shear band and the compensating shear in the neighbouring region

Fig.6 - Micrograph of the 95% cold rolled and partly recrystallized sample (10s at 310°C), (anodically oxidized, X 200)

