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COMPARISON OF SUBGRAINS IN AL AND AL-5MG DEFORMED AT THE SAME STRESS

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Abstract

The evolution of the subgrain structure with strain during deformation at 5 MPa and $\approx 10^{-4}$/s is much faster in pure Al compared to Al-5Mg, in contrast to the quantitative agreement of the steady state properties of the structures in the two materials including the relative contribution of subgrain boundary migration to deformation. The difference in evolution rate is discussed in terms of different degree of heterogeneity of plastic deformation.

Introduction

Al and Al-5Mg represent the two different classes, M (pure metals) and A (strongly solute hardened alloys), of deformation behavior at elevated temperature T. While in the past a qualitative difference between the two classes with regard to subgrain (SG) development has been claimed, it has meanwhile become clear that the steady state properties of the dislocation structures are quite similar [1, 2, 3, 4]. In order to see these similarities, it is, however, necessary to provide the strain which is required for full development of the microstructure [2, 4]. This is, e.g., possible in torsion tests [3, 5]. The aim of the present work is to study the microstructural development in the *transient* deformation range and look for possible differences between the two classes.

Experimental

The experiments were done with polycrystals of pure Al (99.99% purity) and of Al-4.8wt.%Mg with the following impurities (in mass% from spectroscopic analysis): Si: 0.02, Fe: 0.03, Zn: < 0.005, Ti: < 0.01. The Al bar was as cast with 2.8 mm grain size (mean intercept). Al-Mg had a grain size of 400 μm after annealing. Deformation was done in compression at constant stress. The migration of subgrain boundaries (SGB) was investigated by observing migration traces with differential interference contrast (DIC). Experimental details are reported in [6].

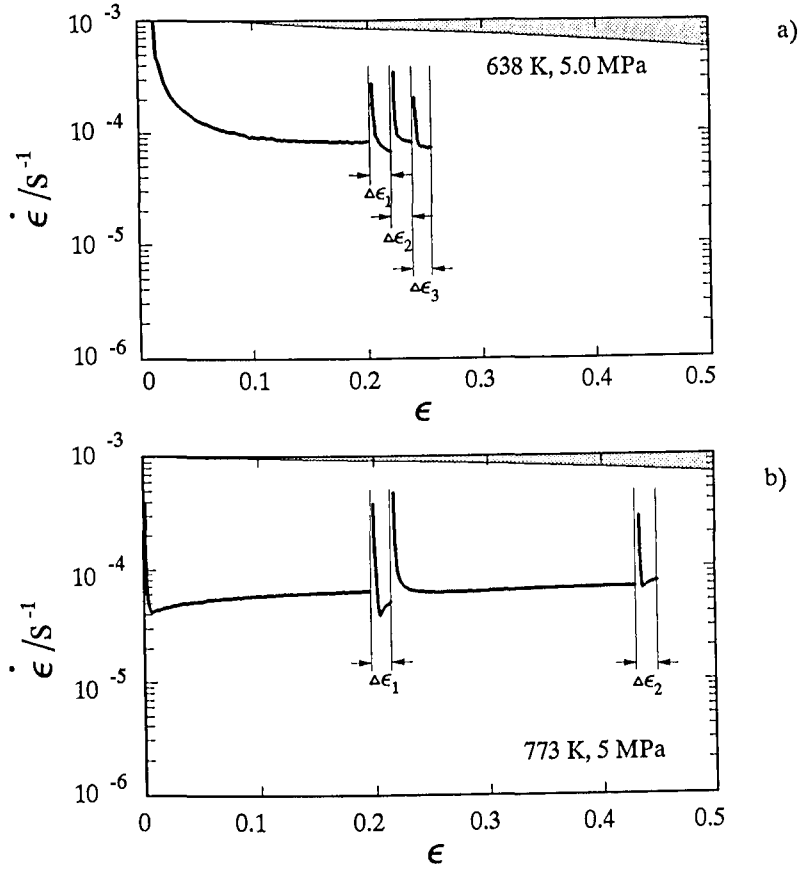
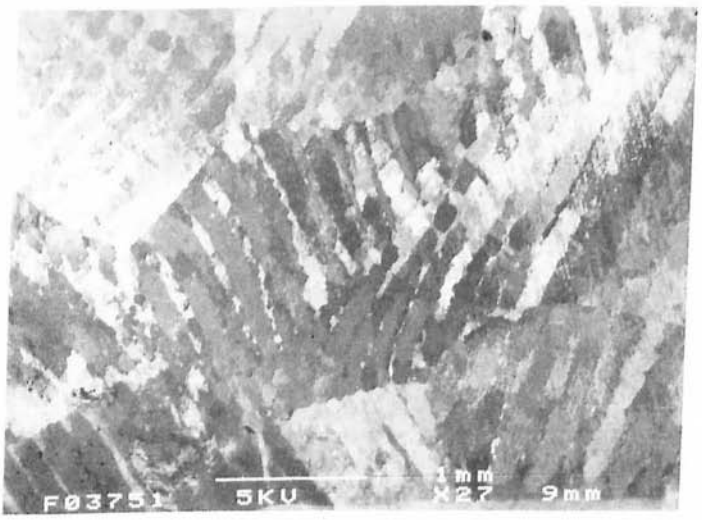


Figure 1: $\dot{\epsilon}$-ϵ-curves at 5 MPa for a) Al and b) Al-Mg.

Results

Figure 1 shows the (true) strain rate $\dot{\epsilon}$-strain ϵ-curves of the two materials at constant stress σ. The shaded area at the top of the figures indicates the magnitude by which $\log \dot{\epsilon}$ has been increased at a given ϵ to correct for the effect of friction between the flat compression specimens and the machine [6]; this correction ensures that $\dot{\epsilon}$ becomes constant with ϵ in the steady state of deformation. As expected, the transient response is normal for Al ($\dot{\epsilon}$ decreases with increasing dislocation density) and inverted for Al-Mg. At ϵ=0.2 deformation was interrupted to observe the microstructure and prepare the surface for DIC observation of traces of SGB migration after additional deformation.

Figures 2 and 3 exhibit the scanning electron microscopic pictures of the grain structures in channeling contrast from back scattered electrons (SEM-BE). In aluminum all the grains have well developed subgrains (Fig. 2). Some anisotropy of the SG structure shows that the



a)

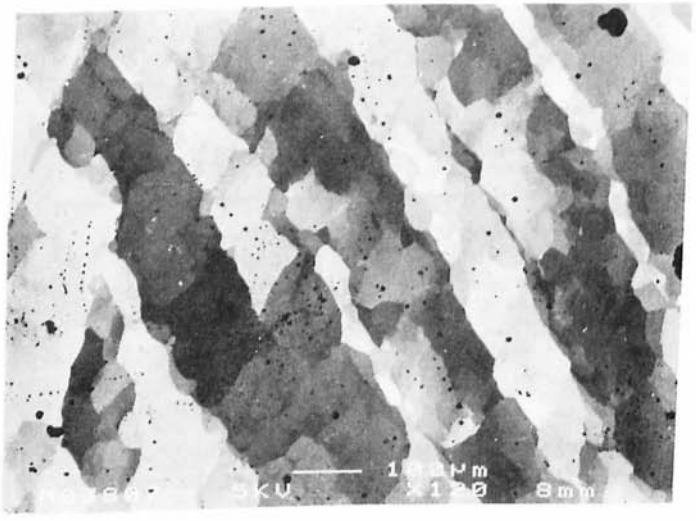


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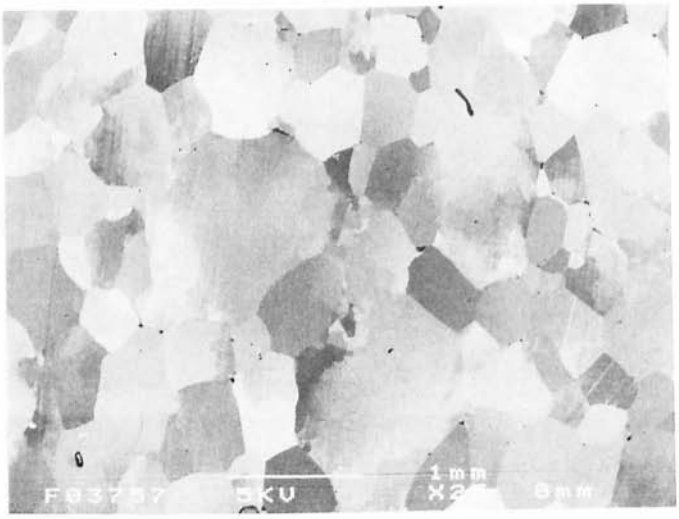
Figure 2: SEM-BE pictures of the grain structure in pure Al deformed as shown in fig. 1a to $\epsilon = 0.2$. a) Grains of several mm size with subgrains, b) subgrains in a single grain.



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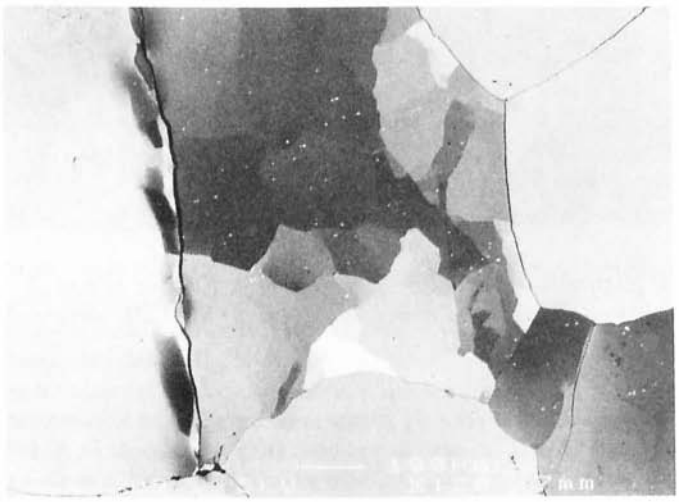

b)

Figure 3: SEM-BE pictures of the grain structure in Al-Mg deformed as shown in fig. 1b to $\epsilon = 0.2$. a) Grains 400 μm in size with weak SG structure contrast in some grains, b) subgrains in a grain at higher magnification.

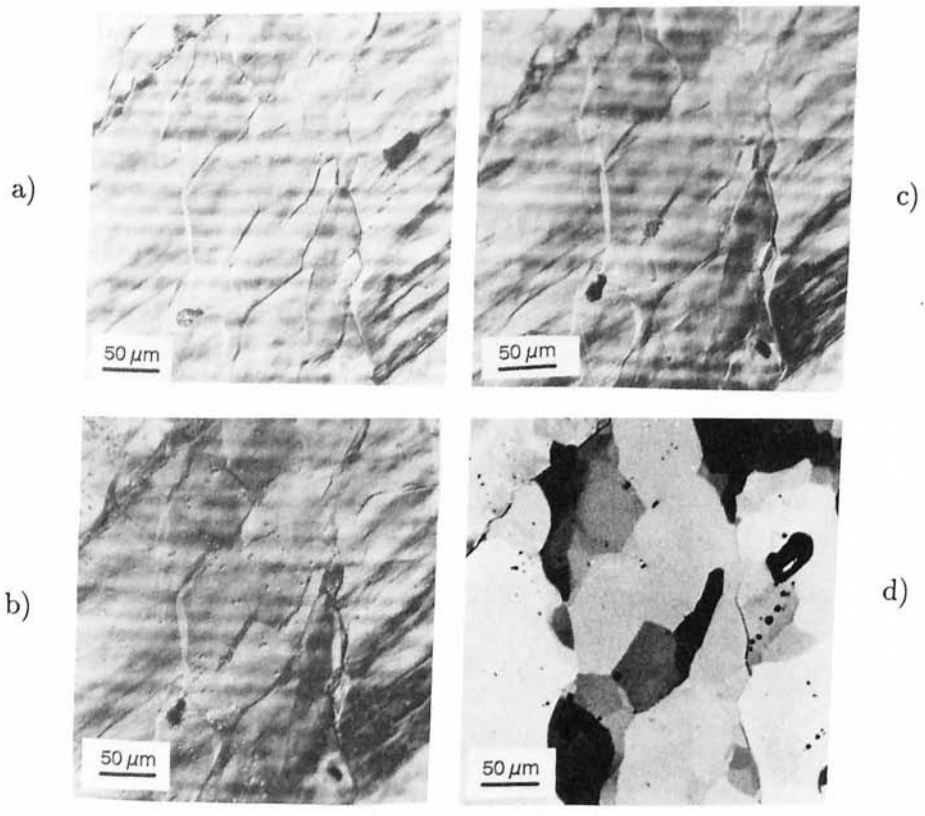

Figure 4: DIC pictures of traces of SGB migration in Al after a) first, b) second, c) third interval of additional deformation shown in fig. 1a. d) SEM-BE picture of the area shown in a) to c).

have well developed subgrains (Fig. 2). Some anisotropy of the SG structure shows that the equiaxed steady state structure has not yet been fully developed. In Al-Mg the state of SG development is quite different. Fig. 3 shows mainly grains (with large angle boundaries). Only few grains have a mottled appearance which is indicative of SG formation. There is a preference for SG formation near grain boundaries (GB).

Consistent with the different degree of development of SG, the traces of SGB migration at the surfaces of the samples after additional deformation are quite different. For Al-Mg such traces are virtually absent. For Al, on the other hand, they are quite pronounced. Figure 4 displays the growth of the traces as the additional deformation increases in three steps (Fig. 1a). The SGB migration traces can easily be correlated with the SG structure of the observed area as depicted by SEM-BE in fig. 4d. The correlation confirms that the traces

are due to migrating SGB.

Discussion

At the deformation conditions applied in the present work the deformation characteristics of the two materials are clearly different: Al has a steady state σ-exponent of $\dot{\epsilon}$ of $n \approx 4.5$ [2] and normal transient response, while Al-Mg has $n = 3.3$ [6] and inverted transient response. These macroscopic differences are, however, not related to a qualitative microstructural difference: Steady state SG structures develop in both cases. To a first approximation, their characteristic parameters, namely the average spacing of free dislocations in the SG interior, $\rho_f^{-0.5}$, the SG size w, and the spacing s of dislocations in the SGB, depend on the stress σ normalized by the elastic shear modulus G [2, 5]: $\rho_f^{-0.5}$ is of the order of bG/σ (b: length of Burgers vector), w is about 30 times larger, and s is about 100 b corresponding to SG misorientations of 0.6°. The relative contribution of SGB migration to strain is also similar, of the order of 5 to 10% [6].

Regardless of the similarity of the steady state dislocation structures, their evolution is quite different. This has become clear in the present work, where essentially similar steady state dislocation structures had to be formed in the two materials, because the stress was the same. The difference in evolution cannot be related to the differences in temperature or initial grain size. The higher T and the smaller grain size in Al-Mg should favour SG formation rather than retard it.

The observed sluggishness of substructural evolution in Al-Mg is consistent with the literature. It explains why it is often reported that SG are absent in the grains even under conditions of steady state deformation defined by constant $\dot{\epsilon}$ at constant σ (or vice versa). The same has been found in this work: As can be seen from fig. 1b, $\dot{\epsilon}$ has nearly reached its steady state value, while the SG structure is still poorly developed.

This result seems to contradict the idea underlying the composite model [7, 8, 9] that SGB constitute hard regions limiting the deformation of the soft SG interior. Quantitative modelling of the deformation kinetics [10] reveals that there is no contradiction. According to the composite model the deforming body forms a system in which the soft and hard regions interact by setting up internal stresses. Whether the soft or the hard region dominates the overall deformation kinetics is not *a priori* clear. It depends on the details of the kinetic laws of deformation in both regions. It turns out that the influence of the effective stress with its strong rate dependence is dominating the behavior of class A materials even if SGB are regions of relatively higher hardness. In effect, both class M and class A materials can be modelled with the composite model. Thus the idea that SGB are hard regions need not be given up for class A materials.

The question is why SG formation is so sluggish in Al-Mg. The answer to this question requires an understanding of SG formation. SG formation is one example of pattern formation. However, its physical basis is only poorly understood in detail. Kratochvíl and Orlová [11] have interpreted SG formation as deformation-induced instability of the internal bending type. Blum et al. [12, 6] have illustrated the decrease in strain energy associated with SG formation. SG formation is equivalent to rotation of a volume relative to its environment.

The driving force for this rotation stems from the fact that the volume has undergone shear, i.e. glide on a *single* slip system leading to internal stresses [12, 6]. Homogeneous compression (by multiple slip) of the specimen in a compression test would yield no driving force for rotation, because there would be no mismatch between neighbouring volume elements.

This consideration gives a clue for understanding the observed difference in SG evolution. In Al dislocations glide fast. Many dislocations are emitted from a local dislocation source creating a sheared slip zone. At the borders of the zone elastic compatibility sets up the stresses which cause SG formation, i.e. cross slip and climb of the dislocations into their right positions in a SGB, accompanied by rotation of the SG [6]. In this way the internal stresses are relieved to some extent. In Al-Mg, on the other hand, the dislocations glide relatively slowly, because the cloud of solute atoms has to diffuse along with them. During this viscous glide there is more opportunity for recombination and annihilation reactions with other dislocations [13]. The effect is that deformation is much more homogeneous on a local scale than in Al. The larger degree of homogeneity diminishes the compatibility stresses between differently deformed volume elements, and therefore reduces the tendency for SG formation. Near the boundaries of grains the gradients in local strain are largest, because of the difference in active slip systems in neighbouring grains. This makes plausible why SG formation starts at GB. Once formed, there is no reason for the SGB to disappear[1]. They take part in deformation as obstacles for free dislocations which have to cut through them and by migrating in the direction prescribed by the stresses acting on the SGB dislocations. The migration allows for a dynamic equilibrium between formation and annihilation or recombination of SGB [15]. Finally an equiaxed SG structure is attained.

If this interpretation is correct, the size of SG should correspond to the extension of glide zones around a dislocation source. A glide zone extension measuring 30 average dislocation spacings appears plausible. and is consistent with observations on slip line lengths [12].

Conclusions

The difference between class M and class A materials with regard to subgrain formation is not qualitative, but quantitative in nature. While the steady state subgrain structures are similar, the rate of substructural evolution with strain is distinctly smaller in Al-5Mg compared to Al at the same stress and strain rate. This difference can be understood by noting that the local heterogeneity of deformation sets up the driving force for subgrain formation: a more homogeneous deformation leads to a smaller driving force for subgrain formation in Al-5Mg compared to Al.

Acknowledgment

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[1]This statement is in opposition to Myshlyaev's claim[14] that SGB are instable and may collapse during creep. To our knowledge such collapse has not been proven.

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