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Strain-induced migration of tilt grain boundaries

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Experimental results showing the influence of strain on the migration of ⟨1 1 2⟩ tilt grain boundaries in Al bicrystals are presented. The bicrystals were deformed in a channel-die experiment with different levels of strain and then annealed at different temperatures. The migration of the boundaries was measured in situ by X-ray diffraction. The activation enthalpies were determined from the in situ data and it is shown that the strain level influences the activation parameters for strain-induced grain boundary migration.

Keywords: Grain boundary migration; Al; X-ray diffraction; Strain

One of the most important processes during primary static recrystallization is the interaction between lattice dislocations in the grains and grain boundaries that are able to move. Grain boundaries move into regions with high stored energies, or in other words, high dislocation densities. Behind the grain boundaries regions of very low dislocation density can be found. Although some investigations of the interactions between dislocations and stationary grain boundaries exist [1–3], only a few studies have focussed on the interaction between dislocations and moving grain boundaries [4]. In this study the change of migration rate of curved high-angle grain boundaries in bicrystals was observed when the grain boundary meets a region with higher dislocation density.

The aim of the study presented in this paper is to investigate the influence of strain, or in other words, of dislocations, on the migration of planar grain boundaries in Al bicrystals. For the first time, results of in situ experiments are presented which demonstrate a direct influence of the strain level in the crystal on the activation parameters for grain boundary migration.

Aluminum bicrystals were grown from 99.999% pure material using the Bridgman technique with seed crystals of preselected orientations [5,6]. Bicrystals with symmetric ⟨1 1 2⟩ tilt grain boundaries with 3.6°, 3.7°, 4.3°, 11.0°, 18.1° and 35.7° misorientation across the grain boundary were used for the experiments. The average deviation from the ideal ⟨1 1 2⟩-tilt axis was $1.27° \pm 0.2°$, the average deviation from the ideal symmetric position was $0.65° \pm 0.2°$. Samples about $17 \times 19 \times 4\ \mathrm{mm}^3$ were cut by spark erosion from each bicrystal and deformed with a channel-die set-up [7].

The coordinates for the channel-die deformation are referred to as rolling direction (RD) for the free elongation direction, transverse direction (TD) for the direction constrained by the channel die, and normal direction (ND) for the compression direction. The grain boundary normal was parallel to the compression axis.

Plane strain compression experiments were then conducted at a strain rate of $1.7 \times 10^{-5}\ \mathrm{s}^{-1}$ using a channel-die set-up as described in detail in Refs. [7,8]. Samples with 2%, 4%, 6% and 8% thickness reduction were produced for each single bicrystal. In order to reduce frictional effects the specimens were wrapped in several layers of an 80 μm thick Teflon foil.

After deformation, samples $10 \times 8 \times 2\ \mathrm{mm}^3$ were cut from each deformed sample and then all samples were mechanically ground and electropolished. For comparison, samples were also cut from the undeformed bicrystals.

In order to measure the grain boundary motion continuously, a special X-ray diffraction tracking device was used [5,6]. A small mechanical shear stress was used to activate the grain boundary motion of the planar grain boundaries. The application of the stress, the experimental set-up and the method to measure the grain boundary migration in situ are described in detail in Refs. [5,9,10]. In the experiments the shear stress acted parallel to the grain boundary normal direction, the magnitude

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of the applied shear stress was between 10^{-1} and 10^{-3} MPa, which leads to a driving force on the grain boundary of between 10^{-2} and 10^{-4} MPa. The driving force due to the shear stress can be calculated by using the Peach–Koehler equation [5,11]. The driving force due to the strain can be calculated by using the following equations:

$$\rho(\varepsilon) = \rho_0 + 1.167 \cdot \rho_0 \cdot \varepsilon, \tag{1}$$

where $\rho(\varepsilon)$ is the dislocation density caused by the strain ε, ρ_0 is the dislocation density of the undeformed bicrystal and is approximately $10^9\ \mathrm{cm}^{-2}$ [12]. The factor 1.167 was determined by assuming that the dislocation density caused by a strain ε depends linearly on the strain and by using the value of $0.8 \times 10^{10}\ \mathrm{cm}^{-2}$ for a strain of 6% found in Al single crystals with comparable purity and deformed at a similar strain rate in Refs. [13,14]. The driving force on the grain boundaries is then caused by the difference in the dislocation densities when the grain boundary moves into the deformed region:

$$p_\varepsilon = \frac{1}{2} \cdot Gb^2 \cdot (\rho(\varepsilon) - \rho_0), \tag{2}$$

where G is the shear modulus, which is approximately 25 GPa [15] for aluminum, and b is the Burgers vector, given by $b = 2.86 \times 10^{-10}$ m for Al. The dislocation density ρ_0 is the dislocation density behind the moving grain boundary and is approximately equal to the dislocation density of the undeformed bicrystal.

After measuring the grain boundary velocity v at different (absolute) temperatures T, we may use the following equation to determine the activation parameters for grain boundary migration:

$$\frac{v}{p} \equiv m = m_0 \exp\left(-\frac{\Delta H}{kT}\right), \tag{3}$$

where m is the boundary mobility, ΔH and m_0 are the activation enthalpy and the pre-exponential factor for grain boundary migration, and k is Boltzmann's constant. The driving force p acting on the grain boundaries is the sum of the driving forces due to the applied shear stress and due to the strain:

$$p = p_\tau + p_\varepsilon = \tau \cdot \sin\theta + \frac{1}{2} Gb^2 (\rho(\varepsilon) - \rho_0), \tag{4}$$

where τ is the applied shear stress and θ is the misorientation angle of the grain boundary [5,6].

In Figure 1 the grain boundary mobility m of a $\langle 112 \rangle$ low-angle tilt grain boundary with 3.7° misorientation angle is shown as function of reciprocal temperature. One can see a linear dependence on the reciprocal temperature for all samples with different strains. As is obvious from Figure 1, the slope of the curves increases with increasing strain, which means also that the activation enthalpy ΔH increases with increasing strain.

In order to analyze the dependence of the activation enthalpy on the strain, or the dislocation density correlated with the strain, the activation enthalpy is plotted as function of the strain, as shown in Figure 2. It is obvious from Figure 2 that there is a linear dependence between the activation enthalpy and the strain for low-angle grain boundaries, but there seems to be no influence of

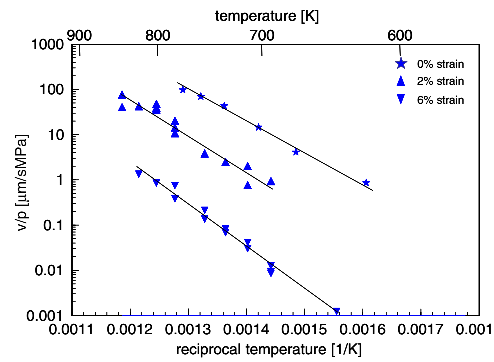

Figure 1. Mobility as function of temperature of a $\langle 112 \rangle$ low-angle tilt grain boundary with 3.7° misorientation angle for samples at different strain levels.

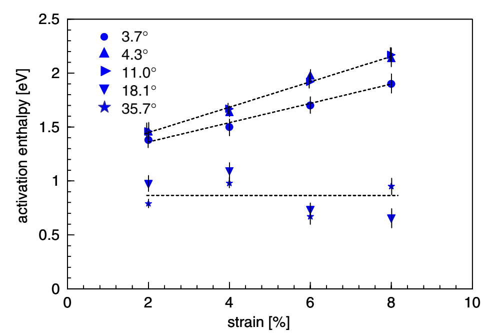

Figure 2. Activation enthalpy as function of the strain for $\langle 112 \rangle$ grain boundaries with different misorientation angles.

the strain on the activation enthalpy of high-angle grain boundaries.

Planar, individual $\langle 112 \rangle$ tilt grain boundaries were subjected to a mechanical shear stress in order to induce a migration of the grain boundaries. Before the migration experiment, the samples were slightly deformed 2%, 4%, 6% and 8%, respectively. The resultant strain leads to the generation of dislocations in the samples. Therefore, two different driving forces act on the moving grain boundary: the applied shear stress and the difference in the dislocation density. Comparison of the magnitudes of the two driving forces shows that the driving force due to the shear stress is of the order of 0.001 MPa whereas the driving force due to the dislocation density is of the order of 0.1 MPa according to Eq. (2), i.e. at least two orders of magnitude higher than the mechanical driving force.

Because of the symmetry of the bicrystals with regard to their crystallographic directions we can assume that the dislocation density is more or less the same on both sides of the planar grain boundary, so that no migration should occur if the sample is only annealed. By application of the shear stress the grain boundary will feel a driving force in a certain direction, so that now the dislocation density in front of the moving grain boundary can be reduced by the moving grain boundary and a dislocation-reduced region is left behind the grain boundary. Therefore, the small driving force due to the shear stress is needed to activate the migration and to give the grain boundary a certain direction to move.