THE DEVELOPMENT OF HIERARCHICAL DISLOCATION SUBSTRUCTURE IN ALUMINIUM DEFORMED AT AN ELEVATED TEMPERATURE

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Abstract The microstructure evolution during tensile deformation of a strongly cube-textured aluminium alloy 5005 at 250°C has been studied in detail using X-ray texture goniometry and transmission electron microscopy. Major emphasis was focused on the evaluation of misorientation vectors across planar dense dislocation walls (DDWs) that developed gradually on the background of tangled dislocation cells. The results showed that the initial cube texture remained stable during straining. The DDWs gradually developed a pronounced hierarchical character and the corresponding mean misorientation angles increased progressively with strain. The results obtained were compared to those available for similar deformation conditions at room temperature and interpreted using both the classical dislocation and dislocation/disclination models of deformation microstructure evolution. The latter model appeared to have somewhat higher predictive ability than the former concept. Neither morphological nor crystallographic characteristics of the observed dislocation arrangements supported predictions of the self-screening "checkerboard" arrangement model, suggested by the theory of low-energy dislocation structures (LEDS), despite high mobility of dislocations and availability of a wide range of corresponding Burgers vectors.

Keywords: *deformation microstructure, texture, dislocation boundary, partial disclination*

1. Introduction

It has been well established experimentally that, during plastic deformation of a polycrystal at room temperature, individual grains are gradually subdivided into mutually misoriented regions separated by dislocation rotation boundaries [1]. At small strains, slightly misoriented "ordinary dislocation cells" are created that are characterised by tangled dislocation boundaries. According to the theory of low-energy dislocation structures (LEDS) [2], these boundaries result from statistical mutual trapping of glide dislocations into low-energy configurations and are, therefore, termed "incidental dislocation boundaries" (IDBs). With increasing strain, planar "dense dislocation walls" (DDWs) develop gradually on the background of ordinary cells and, thus, divide the grain into "cell blocks" (CBs) [1,3]. DDWs form as geometrical necessity and are, therefore, often called "geometrically necessary boundaries" (GNBs). At the present time, there are two major schools aiming to account for the observed grain fragmentation during plastic deformation. Hansen et al. [1,3] have proposed a concept using essentially the classical dislocation theory framework incorporating some principles of the LEDS theory. Rybin et al. [4,5] have introduced a dislocation/disclination approach, where the grain subdivision into CBs is described via the multiplication, motion and branching of partial disclinations, leaving DDWs behind them. Both the above models similarly suggest that the number of operating slip systems in individual grains is less than five required by Taylor criterion of grain compatibility [6]. This is supposed to be energetically favourable and, in turn, brings about the grain fragmentation [1-5]. On the other hand, these concepts somewhat differ in their predictions as a result of different ways of theoretical modelling.

The aim of the present study was to undertake a detailed investigation of the microstructure evolution during warm tensile deformation of aluminium and attempt to interpret the results using both Hansen's and Rybin's models. Experimental parameters, allowing a high degree of freedom in dislocation rearrangement during plastic deformation, were chosen in order to test the validity of the self-screening "checkerboard" dislocation arrangement proposed by the LEDS theory [2].

2. Experimental

Commercial aluminium alloy AA5005 was used in the present study with the chemical composition of 0.14mass% Si, 0.63% Fe, 0.06% Cu, 0.61% Mg, 0.01% Zn, 0.03% Mn and the balance Al. The starting material was in a sheet form with a mean recrystallised grain size of about 60 μm, characterised by a strong cube texture. The alloy was subjected to uniaxial tensile deformation at 250°C at an initial strain rate of $2x10^{-4}$ s^{-1} to true strains of 0.1, 0.3 and 0.5. Crystallographic texture was determined at midthickness of the tensile sheet specimens using the Schulz reflection technique [7]. The transmission electron microscopy investigation of thin foils was performed using a Philips CM 20 microscope operated at 200 kV. Convergent beam electron diffraction was used to study local crystallographic orientations and misorientations [8].

3. Results

The initial cube texture was found to remain stable during straining and to undergo only a small increase in scatter. The deformation microstructure was composed of ordinary dislocation cells with superimposed DDWs. The characteristics of ordinary dislocation cells, that did not seem to undergo significant changes during straining, have already been reported elsewhere [9]. Thus, only the development of DDWs will be discussed in detail in the following.

At a strain of 0.1, the overall density of DDWs was quite low and these walls were largely situated in the vicinity of the grain boundaries, as illustrated in Fig. 1a. The corresponding misorientation axis vectors, expressed either in the sample reference frame or in the crystal lattice coordinates, were mostly scattered around a large area of the respective pole figures, Figs. 1c and 1d. With increasing strain, the density of DDWs gradually increased and, consequently, CBs delineated by these boundaries shrank in size. Morphologically, CBs remained mostly roughly equiaxed throughout the strain interval used. The misorientation angles across DDWs increased progressively during straining and so did the hierarchical character of their arrangement.

A typical example of the configuration of DDWs formed in the grain interiors at a strain of 0.5 is shown in Fig. 2a. A large increase in misorientations between adjacent CBs, caused by plastic deformation to this strain level, was also manifest by the increased scatter in the corresponding lattice orientations, compare Figs. 2b and 1b. Some CB orientations were occasionally found to rotate by significant angles away from the rest of the grain, as illustrated by the orientation group labeled A in Fig. 2b. These particular orientations correspond to the crosshatched small region in the misorientation map in Fig. 2a situated next to the grain boundary. Such areas, occasionally observed to develop predominantly in the grain boundary regions and in the boundary triple junctions, essentially represent fine deformation bands developed gradually from CBs [10] and their occurrence became more noticeable at a strain of 0.5. Also, at this strain level, the overall distribution of misorientation axis vectors still displayed quite a large scatter, similar to the lower strains (see Figs. 1c and 1d). However, there was a clear tendency for the larger-angle misorientation axes to orient themselves roughly perpendicular to the tensile axis direction, Fig. 2c. Such an alignment with respect to the macroscopic specimen geometry was not reflected by the corresponding distribution of these misorientation axis vectors in the crystal lattice coordinates which remained quite random. Fig. 2d shows the cumulative misorientations calculated for the misorientation map in Fig. 2a along the line parallel to the tensile axis direction. Similar "wavy" cumulative misorientation patterns, characterised by a steady increase in misorientation angles over significant distances, were observed throughout the entire strain interval studied. Fig. 3 shows the evolution of the mean CB characteristics during straining and, thus, illustrates quantitatively the observed progressive increase in misorientation angles across DDWs and concurrent decrease in CB size with increasing strain. The gradual increase in the hierarchical character of DDWs, mentioned above, is clearly documented by an enhanced scatter in the misorientation angle values corresponding to higher strain levels.

The results of a trace analysis performed for DDWs at a strain of 0.5 are summarised in Fig.4. They illustrate well the general trends observed throughout the entire strain interval studied. DDWs did not seem to be preferentially aligned (within ± 5 deg) along any of the {111}, {110} or {100} planes, Fig. 4a. There was some tendency for DDWs to orient themselves within an angular range of about 30 to 60 degrees with respect to the tensile axis direction, Fig. 4b. Dislocation walls with large tilt components were found much more frequently than those having large twist components, but there was always a significant amount of DDWs with a general character, Fig. 4c.

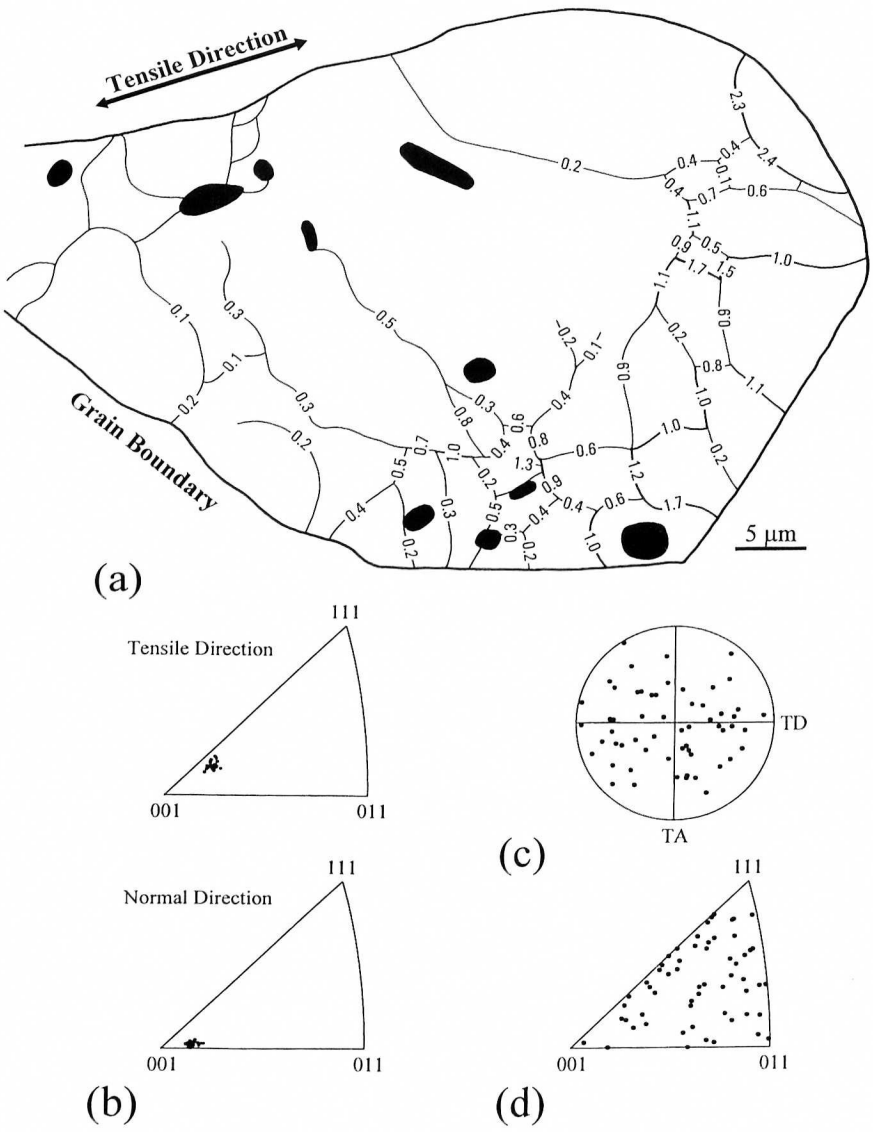

Fig. 1. Characteristics of the DDWs in the interior of a deformed grain at a strain of 0.1: (a) misorientation map; (b) crystallite orientations in inverse pole figures; (c) misorientation axis vectors in the sample coordinates; (d) same vectors in the crystal lattice coordinates.

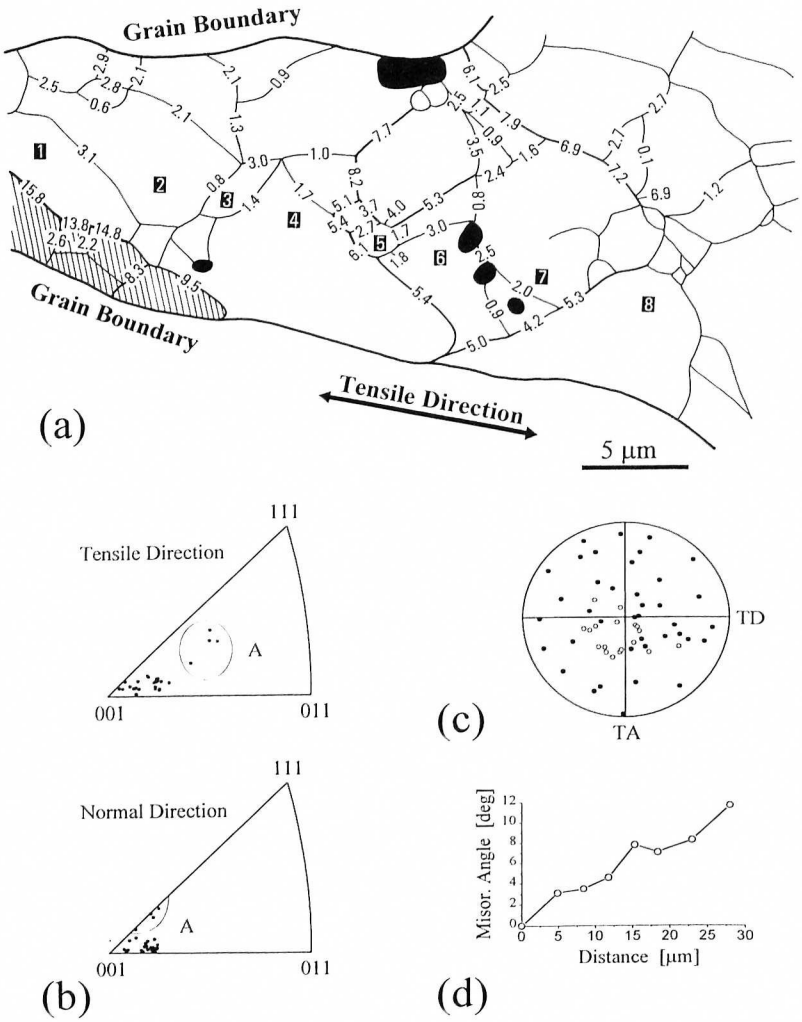

Fig. 2. Characteristics of the DDWs in the interior of a deformed grain at a strain of 0.5: (a) misorientation map; (b) crystallite orientations in inverse pole figures; (c) misorientation axis vectors in the sample coordinates (open and filled circles correspond to the angles above and below 5 deg respectively); (d) cumulative misorientations along the line 1 to 8 indicated in (a).

4. Discussion

The results of the present study show that, under the deformation conditions used, the microstructure evolution inside individual grains was characterised by gradual development of larger-angle planar DDWs on the background of slightly-misoriented tangled dislocation cells. The resulting progressive grain subdivision into mutually rotated CBs, delineated by these walls, during straining is in good correspondence with the predictions of both Hansen's [1,3] and Rybin's [4,5] models. The arrangement of DDWs displayed a hierarchical character that became more pronounced with increasing strain. During the deformation process, "old" DDWs, created first and delineating large regions of a grain, continued to receive dislocations and thus to increase their misorientation angles while new lower-angle walls were being gradually formed inside these

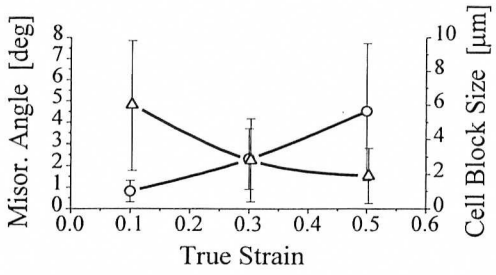

Fig. 3. Evolution of the mean CB characteristics with strain (circles and triangles represent misorientation angles across DDWs and CB diameters respectively).

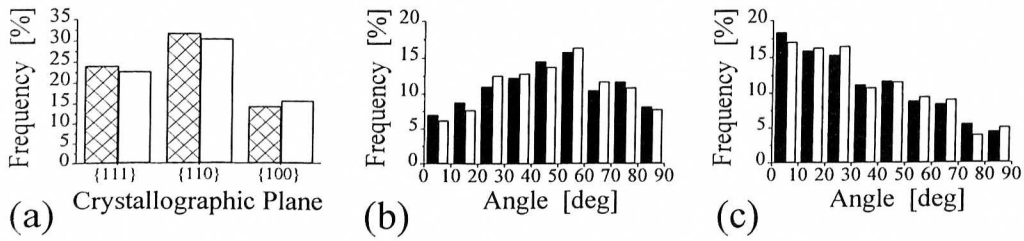

Fig.4. Angles between each DDW trace and: (a) a nearest crystallographic plane trace (within ± 5 deg); (b) the tensile axis trace; (c) a corresponding misorientation axis trace (pattern and black columns represent all the walls analysed, white columns correspond to those with misorientation angles above 5 deg).

regions, which led to their further subdivision into smaller "higher-generation" subregions. The observed development of DDWs in a hierarchy has been particularly well predicted by Rybin's model as a result of branching of partial disclinations during their movement through the grains [5].

As a result of a deformation temperature of 250°C, the microstructure morphology observed was characterised by a random arrangement of CBs having complex shapes and remaining roughly equiaxed throughout the strain interval investigated. There appeared to be little role of crystallography in the deformation microstructure development. Misorientation axis vectors across DDWs mostly displayed a random distribution when expressed in the crystal lattice coordinates irrespective of strain. Furthermore, these walls did not seem to be preferentially aligned along any of the {111}, {110} or {100} planes that may serve as possible slip planes at the elevated deformation temperature. Instead, the microstructure characteristics appeared to be governed, to a certain extent, by the macroscopically imposed deformation geometry. DDWs displayed some tendency to attain a preferred orientation with respect to the tensile axis direction and misorientation axis vectors across larger-angle DDWs were frequently observed to orient themselves roughly perpendicular to this direction. The above trends are, again, well predicted by Rybin's model [4,5]. It proposes that partial disclinations are free to move through a grain on any crystallographic plane and, therefore, DDW characteristics are expected to be almost randomly distributed with respect to the crystal lattice coordinates. Conversely, these characteristics should be largely governed by a macroscopic deformation geometry [4,5].

It was interesting to compare the present results with those obtained in the case of tensile deformation of identical experimental material at room temperature, reported in [11]. The room temperature deformation microstructure mostly contained one or two systems of long parallel DDWs superimposed on the background of tangled dislocation cells. The two intersecting systems of walls frequently displayed a rectangular "checkerboard" pattern. Although no systematic crystallographic investigation was undertaken in [11], the microstructure characteristics reported indicated that there was a slightly larger role of crystallography in the deformation microstructure development, DDWs being occasionally aligned along the {111} slip planes. Nevertheless, this development appeared to be governed predominantly by the macroscopically imposed deformation geometry, as DDWs were preferentially aligned with the direction of maximum applied shear stress. Misorientations across these walls locally displayed alternating plus-minus-plus rotations about approximately parallel axes and, as a result, misorientation angles did not cumulate over longer distances in the corresponding regions. Occasional presence of such self-accommodating DDW arrangements has been used by the LEDS theory [2] as evidence that dislocations tend to arrange themselves during straining into a simple "checkerboard" pattern, where they can mutually screen effectively their stress fields. According to this theory, the experimental conditions used in the present study, characterised by increased mobility of dislocations as well as availability of a wide range of Burgers vectors, should enhance the tendency for the well-organised self-screening checkerboard arrangement to be formed during straining. However, on the contrary, the increased deformation temperature led to the formation of quite random microstructure having complex crystallographic characteristics, where misorientations across DDWs were frequently strictly cumulative over significant distances. This indicates that DDW arrangement during plastic deformation might be governed more strongly by strain-accommodation requirements than by energy-minimisation principles [12], which has also been reflected in Rybin's theory [4,5].

5. Conclusions

A detailed, systematic study of the evolution of DDWs during tensile deformation of a strongly cube-textured aluminium alloy 5005 at 250°C has been undertaken in the present work. These walls were observed to develop a pronounced hierarchical character and the corresponding misorientation angles increased progressively with strain. The complex DDW characteristics appeared to be well predicted by the dislocation/disclination model of deformation microstructure evolution. Conversely, the predictions of the self-screening "checkerboard" dislocation arrangement made by the LEDS theory, associated with the classical dislocation model, were not confirmed.

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