Shear bands formation after a strain path change for AA1050 alloy pre-deformed by ECAP and subsequently plane strain compressed

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Abstract. Crystal lattice rotations induced by shear bands developed in an AA1050 aluminium alloy have been examined in order to investigate the influence of the fine-grained structure on the slip propagation across the grain boundaries and the resulting texture evolution. Samples of the AA1050 alloy were pre-deformed in ECAP up to 6 passes via route C, then machined and further compressed in a channel-die up to ~25% at room temperature. The microstructure and texture were characterized by SEM equipped with a high resolution EBSD facility.

The ECAP-processing leads to the formation of a fine grained structure. The grains were grouped into nearly complementarily oriented layers. During the secondary straining in the channel-die, the layers of fine grains, initially situated almost parallel to the compression plane, undergo deflection within some narrow areas. This is the beginning stage of the macroscopic shear band (MSB) formation. In all the deformed grains examined (within MSB) a strong tendency for strain-induced re-orientation could be observed. The SEM orientation mapping shows how the layers of flattened grains are incorporated into the MSB area, and what kinds of mechanisms are responsible for the strain accommodation at the macro-scale. Finally, a crystallographic description of the mechanism of MSB formation in AA1050 aluminium alloy is proposed based on the local lattice re-orientations due to localized kinking.

Keywords: Shear bands, Strain path change, Orientation mapping, Texture, AA1050 alloy

1. Introduction

The microstructure and texture evolutions in polycrystalline materials with a face centred cubic lattice (fcc) and medium-to-high stacking fault energy (SFE) has been the subject of intense research studies in the last few decades, e.g. [1-8]. The latter have been focused on a detailed description of the macro-/micro-structure development during different shaping processes and an identification of the correlations that appear between the initial grain orientation and the (micro)structure and texture of the deformed state for a wide range of strains. The formation of the structure of macro-/micro- layers (flattened grains or elongated dislocation cells) and their further instability is the most characteristic feature observed after the large deformation of medium-to-high SFE fcc metals under the terms of a plane state of strains, e.g. [3-5]. From the microscopic observations it is widely accepted, e.g. [9] that the thickness of the layers of the disoriented dislocation cells significantly decreases as the strains increases, and this microstructural changes coincide with the strong increase of disorientation between them. In the range of extremely large deformations, these flat dislocation arrangements are usually situated nearly parallel to the rolling (or compression) plane and they are nearly complementarily oriented, e.g. [1-5].

The formation of a layered structure is strongly influenced by the crystallography and SFE. The boundaries between layers are thought to be an effective barrier for dislocation motion. This causes a high stress (and dislocation) concentration at the boundaries up to a point when dislocations destroy a
layered structure and create localized slip in the form of shear bands (SB) [1]. Thus, shear banding during continuous straining is preceded by the formation of essentially planar obstacles to homogeneous dislocation glide across the boundaries. Two types of shear bands can be distinguished according to the SFE. In metals with low SFE brass-type SB can be readily observed after medium deformations, e.g. [10-15]. They develop within a highly twinned microstructure characterized by thin twin-matrix lamellae. With increasing SFE (metals of medium-to-high SFE) copper-type SB are formed against the background of band-like dislocation structures of elongated cells, e.g. [1-9]. This process tends to become increasingly difficult in the metals of very high SFE so that SBs have not always been observed in high purity aluminium. It is evident that for both cases (twinned or non-twinned metals) the different morphology and texture components of the shear band interior are expected to form in the crystallites of initially similar orientation. The SB are not observed at the early stages of deformation of the fcc metals, i.e. within the range of deformations when the microstructure is ‘dominated’ by equiaxed dislocation cells and the metals still show the properties close to isotropic. During continuous deformation, the structure of the equiaxed cells is systematically ‘transformed’ into the structure of the microbands (elongated cells) and this transformation coincides with the increase of anisotropy [16]. The conclusion that the layered structure formation is one of the most important factors necessary for the occurrence of shear bands seems to be valid independently on the SFE of the deformed metals. Moreover, there is no essential difference between the metals with the low and the medium-to-high SFE, i.e. twinned or non-twinned structure, respectively. For both groups of metals, the rotation-induced mechanical instability within narrow areas of the anisotropic structure of elongated cells or twin-matrix layers leads, first of all, to the formation of kink-type bands and next (after shearing along the shear plane) to the SB occurrence [11-14].

The breakdown of structure of the dislocation barriers can be ‘created’ either during monotonic straining or earlier, in processes realized with the change of the deformation path [17, 18]. It is clear that the pre-existing dislocation structure is still the same during the sample re-loading. However, their orientation with respect to the external axes is quite different.

The present work was undertaken to establish the effect of the strain path change on the crystallographic aspects of the copper-type SB formation in a fine grained structure of the AA1050 alloy pre-deformed in equal channel angular pressing (ECAP) and, subsequently, plane strain compressed (PSC) in a channel-die [19-21]. Since a very large amount of deformation by ECAP (>6 passes) usually leads to a dynamic recrystallization of this alloy (as observed earlier in [22] for pure Al), the pre-deformation up to 6 passes was the choice for further analysis of shear band formation during the PSC. Combining ECAP with the subsequent channel-die deformation offers an opportunity for texture control. Since shear banding is closely related to textures, e.g. [12, 13], computer-automated electron backscattered diffraction (EBSD) in scanning electron microscopy equipped with a field emission gun (FEGSEM) becomes a particularly suitable method of investigating the phenomenon.

2. Experimental

The investigated material was the commercial AA1050 aluminium alloy. The as-cast ingot was homogenized and then hot rolled on an industrial reversing mill to a 22 mm thick sheet. Samples with the cross section of 10×10 mm² and length of 70 mm were machined parallel to the rolling direction (RD). The samples were then extruded via ECAP at room temperature, with the use of a 10×10mm² die with the angle of 90° and the outer corner radius of 5mm. This gave the nominal (von Mises) effective strain of εM = 1.1 per pass. The samples were processed with a constant frame velocity of 5 mm/min, up to 10 passes through the route C, in which the sample underwent 180° rotation between the consecutive passes. The billet and tooling were well-lubricated with MoS₂-containing grease. After each pass, the die was opened, and the sample was removed from the die, re-polished and subjected to the next pass. After 2, 6 and 10 passes, some tensile specimens with the cross-section of 1.5 mm (thick) × 2 mm (width) and the gauge length of 10 mm were machined from the ECAPed samples with their tensile axes parallel to normal direction. The tensile experiment was conducted at room temperature with the use of the MTS/ADAMEL DY30 testing machine operated at a constant rate of cross-head displacement with a strain rate of about 10^{-3}s^{-1}. Three tensile experiments were tested (up to the fracture) to check the repeatability of the results. The results were compared with the strength properties of the initial material.
As for the plane strain compression (PSC) test in channel-die, the ECAPed samples were sectioned perpendicular to ED (i.e. along the ND-TD plane) in the middle of the billets. ND, ED (||RD) and TD denote the normal, extrusion and transverse directions, respectively. The cubic samples of 10 x 10 x 10 mm$^3$ were inserted into another die that applied plane state of strain. The surfaces were carefully marked so that the elongation direction during the ECAP processing and PSC was parallel to the rolling direction. The PSC tests were performed with the use of an Instron machine with the constant speed of 0.5 mm/min. In order to reduce the frictional constraint between the specimen and the compression rig, 50 µm thick Teflon$^\text{TM}$ lubricant was used during testing. The samples were deformed up to the final (logarithmic) strains of ~0.3 (25%), with the use of two-stage tests. At each step (~0.15 strain), the Teflon$^\text{TM}$ ribbon was changed, so that the friction was very low. The ratio between the height and the length of the test pieces was restored to 1 at the beginning of each deformation step. A thickness reduction of about 0.3 was selected for a detailed analysis of the influence of the internal MSB substructure on the slip propagation across the grain boundaries and the texture changes. The microstructure and texture of the initial material, as well as the ECAP-processed and channel-die compressed samples were characterized by the SEM/EBSD system in two sections: ND-ED and ND-TD, i.e. perpendicular to TD and ED, respectively. The SEM-based orientation mappings were performed with the use of FEI Quanta 3D equipped with a high resolution EBSD facility. The voltage of 20kV, the working distance of 10 mm and the specimen tilt angle of 70$^\circ$ were used. The microscope control, pattern acquisition and indexing were performed with the use of the TSL OIM Analysis 5 software. The mappings were carried out in the beam-scanning mode with the step size of 100 nm or 500nm.

3. Results and discussion

3.1. The morphology and texture of initial material

The orientation maps made for the initial material in all three sections are presented in Fig. 1 as IPF color code maps. They show a layered structure composed of very large flattened grains. The traces of grain boundaries were parallel to the ED and TD. (This indicates that flat grains are situated almost perfectly along the rolling plane). The grain thickness, i.e. the distance between the large angle grain boundaries along the ND, was mostly ranged between 40µm and 80µm, and varied slightly in different areas of the sample. Moreover, the grain length along the TD is lower than that along the ED. This initial state of the AA1050 alloy is described by well-developed copper-type rolling texture.

![Fig. 1. Initial microstructure and texture of AA1050 alloy in three perpendicular sections. Orientation maps and corresponding {111} pole figures.](image-url)
components, i.e. brass \(\{1\text{10}\}<1\text{12}\>\), \(S\{1\text{23}\}<6\text{34}\>\) and \(C\{1\text{12}\}<1\text{11}\>\), typical for fcc metals of medium-to-high SFE, as presented in the \(\{\text{111}\}\) pole figures corresponding the orientation maps.

Fig. 2. Microstructure and texture of ECAPed material after 6 passes along route C. Orientation maps and corresponding \(\{\text{111}\}\) pole figures measured in section perpendicular to TD (a and c) and ED (b and d).

3.2. The microstructure and texture of ECAP-processed samples

The number of the applied passes had a strong impact on the global deformation behaviour and on the intensity of the grain refinement. The most important features of the deformed material can be summarized as follows:

- A cellular structure of the deformed state began to form after the first pass.
- Starting from the second pass, two types of lamellar structures were observed. In meso-scale, the deformed structure is composed of thin flat grains forming layers. In micro-scale, some layers bounded by high-angle boundaries contained an additional substructure of compact clusters of elongated cells/grains. (Below, the terms 'layer' and 'lamella' refer to the structures of the first type).
- The thickness of the separate layers depended on the number of the ECAP passes, e.g. after 6 passes, most of the layers were few microns thick, whereas after 10 passes - below 1 micron. A general observation is that the dimensions of the microstructure elements significantly decrease as the number of passes of the ECAP-processing increases. However, after 10 passes, in some places, new recrystallized grains were observed. This room temperature recrystallization, observed also, e.g. in [22] leads to a decrease of the strength properties.
- The observations made for the sample after 6 passes in the ND-ED and ND-TD sections showed (Fig. 2) that most layers took the position almost parallel to TD and deviated from the ED-TD plane by an angle of \(15-20^\circ\). The disorientations between the layers were very large, and often reached \(45-60^\circ\). Another characteristic feature clearly observed in the ND-ED section are well-marked micro shear bands. They cut the background structure of the layers at an angle of \(50-55^\circ\) to ED, as presented in Figs. 2a and b.
After six passes of the ECAP-processing, the texture of the initial material was strongly changed. The \{111\} pole figures measured in the ND-TD and ND-ED sections showed that the typical copper-type rolling texture components of the initial material were replaced by two dominant, nearly complementarily oriented components with the 'common <111> pole' located near the TD. This is presented in Figs. 2c and d of the sample after 6 passes. During the subsequent deformation in a channel-die, the microstructure 'produced' by the ECAP-processing constituted the background for the development of MSB during the subsequent PSC.

3.3. Stress-strain response after strain path change

Figure 3 compares the values of the tensile strength and the true stress-strain curves obtained in the plane strain compression for the samples that were ECAP-ed to a different number of passes and of the annealed material. Additionally, the values of the tensile strength obtained in [23, 24] for the same material and deformation terms are included. The difference between the maximal values of the true stress after the uniaxial tension and PSC are small. The stress and the strain were calculated with the use of standard definitions, \( \sigma = \frac{F}{S} \) and \( \varepsilon = \ln(l_0/l) \), where: \( F \) and \( S \) and \( l_0 \) and \( l \) are the current compression load, the compression surface and the specimen length (initial and actual), respectively.

For the initial material, the true stress shows a monotonic increase with the growing true strain. This part of the true stress–true strain curve is smooth due to the homogeneous slip distribution in the whole sample during PSC. After the successive stages of ECAP-processing, the tensile strength and the maximal value of the true stress during PSC initially increases significantly (up to 158 MPa after 2 passes) before roughly stabilizing and even slightly decreasing, i.e. the strength achieves 162 MPa and 152 MPa after 6 and 10 passes, respectively. The strength decrease in the range of a very high deformation could be correlated with the intense (room temperature) recrystallization.

It is clearly marked that, despite the new orientation of the dislocation barriers with respect to the external forces, the PSC samples revealed quite similar values of the flow stress with respect to those obtained in the uniaxial tension of the ECAPed samples.

3.4. Morphological and crystallographic aspects of shear band formation after strain path change.

3.4.1 Macro- scale shear band formation

The samples pre-deformed in ECAP up to 2, 6 and 10 passes were further channel-die compressed up to the true strains of ~0.3 to develop MSB. In all the cases, the optical microscopy observations carried out on the longitudinal section (ND-ED plane) showed a clearly marked tendency to the formation of two intersecting and almost symmetrical families of MSB. The width of each set was 1.5-2 mm and they were positively and negatively inclined at 40-45° to ED. A detailed inspection on two sections
perpendicular to ED and TD showed that the MSB planes of both families were parallel to TD. The optical micrographs at the sample scale revealed also small shape distortions in the form of ‘bulging’ on both free surfaces (as clearly visible in the ND-ED section) due to the influence of the friction and strain localization within the MSB (Fig. 4).

More detailed analyses were performed on the re-polished and etched sample after 6 passes. Figure 5a shows the typical structure observed at the sample scale in the ND-ED section. The analysis of the changes of the layer inclination made it possible to track the processes responsible for the layered structure incorporation into the MSB area. At this stage of deformation, the traces of layers (or flat grains) located outside the MSB were nearly parallel to the ED. The very first plastic instability, by which the shear banding began, i.e. the initial distortion of the lamellae within narrow areas, were due to the kinking of the grain boundary traces, as clearly observed in the ND-ED section. This process is initiated near the sample corners. The inclination of the grain boundaries is the largest near the sample corners and systematically decreases near the sample centre and outside the bands. As a consequence of the localized shearing, a rigid body rotation of about 20-25° (Fig. 5a-c) of the material occurs inside the MSB together with the relative translation of the grain boundaries outside the band.

Fig. 4. MSB formation during PSC of ECAPed samples. Optical micrographs showing structure of two crossing MSB. PSC up to logarithmic strains of 0.3. Pre-deformation in ECAP up to: (a) 2, (b) 6 and (c) 10 passes.

The traces of the grain boundaries showed a well-defined rotation in the MSBs area of the opposite sign within each set of bands (Figs. 5b and c). The sense of this rotation is such as to increase the inclination of the traces of the boundaries between the layers (or grains) with respect to the ED, as observed earlier for twinned structures of fcc metals, e.g. [11-14]. A rotation within the MSB area brings the traces of the grain boundaries into a position closer to the shear plane. Moreover, the rigid body rotation of lamellae must lead to significant orientation changes in the MSB area.

3.4.2. Local orientation measurements by SEMFEG/EBSD

The availability of the local orientation measurements with a high spatial resolution created the opportunity to formulate and verify the mechanisms of the shear band formation. This was performed by way of looking into:

- subtle orientation changes within the MSB and area itself, as well as
- at the boundary between the band and the matrix.

The optical microscopy observations reveal the deflection of the flat grain clusters in sparsely spaced narrow zones of the sample. The very first plastic instability, i.e. the initial distortion of the flat grains within narrow areas by which shear banding begins, is similar to that observed in the twinned
structures of low SFE fcc metals described in earlier works, e.g. [11-14]. Since strong rigid body rotations are observed in the MSB areas with respect to the neighboring matrix, it is evident that the textures in the MSB must to be quite different.

The textures within the MSB areas of both families and the neighbouring matrix were measured in the ND-ED section (in places marked in Fig. 5a). The orientations at the points constituting the maps are represented in the \{111\} pole figures (Figs. 5d-f). Independently of the analysed place, two groups of orientations are clearly visible. At first approximation, the \{111\} pole figures show that the textures are positively and negatively rotated around the TD in the areas of MSB\(^1\) and MSB\(^2\), respectively, in respect to the texture ‘image’ of the deformed matrix.

Fig. 5. (a) Microstructure of PSC sample pre-deformed in ECAP up to 6 passes. (b) and (c) Details from the sample top corners showing bending of layers into MSB area. (d) - (f) The \{111\} pole figures showing textures in areas marked in (a). Optical micrographs in ND-ED section after etching and SEM/EBSD local orientation measurements.

To get an insight into the mechanisms of the MSB formation, a detailed investigation of the boundary zone between the matrix and the MSB areas was carried out. New SEM/EBSD maps were obtained with the use of the step size of 500nm, to cover a large area (Fig. 6a). The orientation map reproduces the microstructure’s spatial arrangements in the (MSB)/(matrix) boundary zone. Within the MSB, intense slips in the planes inclined at 42° to ED are observed. The change of the colour in the particular matrix layers visualises the gradual orientation rotation caused by the incorporation of the layers into the band. The change is associated with the deflection of the layers, which rotate towards the alignment of the lamellae boundaries with the shear plane. This increases their inclination with respect to the ED. Clearly, the local microtexture within an MSB is different with respect to that outside the
band, so that the \{111\} planes within and outside the bands are quite different. The MSB exhibits a wide range of orientations resulting from the counter-clockwise rotations of the layers. The rotation is reflected on the \{111\} pole figures (Figs. 6b and c) corresponding to the particular areas of the orientation map.

3.5. Orientation changes as a result of shear banding

The accumulation of the SB into bundles and their propagation through the grain boundaries is an important problem in the process of the MSB formation. The crystallographic texture development, observed at the increasing deformation, favours the penetration of the slip in the MSB area through the neighbouring grains. The situation is simple when the neighbouring grains have a similar orientation, and the \{111\} planes coincide with the plane of the maximum shear stress. The slip penetration, however, occurs in the regions of quite different orientations. Nevertheless, from the crystallographic point of view, the existence of a common plane for both areas is required; it is along this plane that the slip can penetrate the boundary [5].

In earlier works based on \{112\}<111>-oriented single crystals of low [10-15] and medium-to-high, e.g. [5-9] SFE metals, it was shown that from the geometrical point of view, the first step of shear banding coincides with the local lattice re-orientation within the area of SB. This process brings the \{111\}-type planes into a position parallel to the shear plane. The second step is shear along the band.

The mechanism of SB formation is strictly crystallographic. In earlier work [13] a model of shear banding is proposed based on the idea of local lattice re-orientation within narrow areas. Kink-type bands are considered to be the precursors of SB leading to a positive (MSB\(^1\)) or negative (MSB\(^2\)) crystal lattice rotation. The present work shows how the mechanism of lattice rotation within emerging SB may lead to a new texture components with large \((\pm 20^\circ)\) scattering. As strain becomes localised a positive or negative rotation occurs in the bands and the \{111\} planes (in the layers) rotate away from the compression plane. Finally, one of the \{111\} slip planes (~ common for neighbouring layers) takes a position nearly parallel to the direction of the maximum shear (poles of these planes are marked in Fig. 6c). The re-orientation of \{111\} planes towards the MSB plane facilitates further dislocation slip in the shear direction.

Fig. 6. (a) Orientation map showing the (micro)structural and textural changes due to MSB formation. (b) and (c) The \{111\} pole figures corresponding to MSB and matrix areas, respectively.

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The neighbouring layers are nearly complementarily oriented, with one of the \{111\} plane inclined at 40-45° to ED. This facilitates the slip propagation across the grain boundaries (and along the shear direction) without any visible variation in the slip direction. Finally, the internal texture of the MSB is significantly different with respect to those of the deformed neighbouring matrix.

4. Conclusions
The optical microscopy and microtexture measurements on the polycrystalline AA1050 alloy samples pre-deformed in ECAP and subsequently channel-die compressed were used to analyse the mechanism of the shear bands formation. The following major conclusions can be drawn.

• The changes in the strain path directly lead to the strain localization in the form of macroscopically visible bands. During the strain path change, the dislocation microstructure, which has been developed, readjusts towards a compatible dislocation arrangement.

• The secondary straining showed similar softening/hardening behaviours.

• The local orientation measurements showed that well-defined crystal lattice re-orientations occurred in some grains situated within the area of the broad MSB although those grains initially had quite different crystallographic orientations. Their crystal lattice rotated in such a way that one of the \{111\} slip planes became nearly parallel to the direction of maximum shear. A natural consequence of this rotation is the formation of specific MSB microtextures which facilitates slip propagation across grain boundaries along the shear direction without any visible variation in the slip direction.

• It was thereby established that shear banding occurred across the grain boundaries by the continuity of the slip direction, although the slip plane did not coincide exactly in the adjacent grains.

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