Adiabatic shear bands formation in polycrystalline aluminium

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Abstract. The microstructure and texture evolutions during dynamic (or high strain rate) deformation of commercially pure AA1050 alloy have been studied to elucidate the mechanisms of shear band (SB) formation and its propagation across the flattened grains arranged in layers with special attention to texture evolution. The hat-shaped samples were compressed using drop hammer at a strain rate of 5.3 x 10² s⁻¹. Detailed microstructure and texture characterizations were carried out using SEM equipped with EBSD facility. The shear displacement moves the upper part of the hat-shaped sample towards the bottom one leading to a strain localization in the form of SB. It was found that the strain localization occurs via deflection of flattened grains within narrow areas leading to kink-type bands that are considered to be the precursors of SB. The strain-induced crystal lattice rotation led to the formation of new texture components, different from those identified inside the undeformed matrix. It also facilitates slip propagation across the grain boundaries. The mechanism of macro-SB formation is essentially crystallographic in nature since in all grains of the sheared zone, the crystal lattice rotated in such a way that one of the {111} slip planes became nearly parallel to the shear plane and the <011> direction became parallel to the direction of maximum shear.

1. Introduction

The occurrence of shear bands (SB) and their compact clusters called macroscopic shear bands (macro-SB) constitutes an important scientific and technological problem concerning metal forming, e.g. [1-3]. Nevertheless, fundamental studies regarding SB formation in materials deformed at high strain rates are still relatively limited. The so-called adiabatic shear bands (ASB) are areas of localized intense shear that appear very quickly during deformation especially for metals of lower density and lower heat capacity, e.g. Ti and its alloys [4]. A widespread conviction persists that the competition between strain rate hardening and thermal softening determines the onset of SB, as it has been initially proposed by Zenner and Hollomon [5]. When thermal softening prevails over strain hardening, an abrupt loss of homogeneity of plastic deformation occurs and then material falls in the shear banding deformation mode. Rittel et al. [6] revealed that the specimen temperature rise prior to strain localization has almost no effect on the onset of ASB formation. Therefore, Rittel et al. [6] and Ossovski et al. [7] modified the Zenner-Hollomon concept, assuming a microstructural softening due to dynamic recrystallization, where the dynamically stored energy of cold work is used as a quantitative criterion of the onset of ASB.

The descriptions of initial stage of strain localization at high strain rates are to a large extent based on models which do not account for the crystallographic nature of metals or include the crystallographic determinations only partially, e.g. [8]. This is despite the residual microstructure and crystallographic texture have been identified as important and related aspects to shear localization during deformation at
conventional strain rates [3,9-13]. In this context there is lack of clear answer to the fundamental question - what are the conditions for initiating an ASB? It indicates that the real elementary mechanism of band nucleation, its thickening and propagation across the boundaries in polycrystalline aggregate, is not fully understood. The same applies to the formation of deformation texture where strain localization leads to local softening and specific texture components.

In face centred cubic (fcc) metals or alloys ASB has been rarely reported, even though the formation of SB in such metals is one of the most commonly observed phenomenon accompanying the plastic deformation. Generally, two groups of SBs deformed at ‘conventional strain rates’ can be distinguished in fcc metals according to the stacking fault energy (SFE). In metals with a low SFE, brass-type SB can be readily observed after medium deformations, e.g. [3,9-11]. They develop within the microstructures characterized by thin twin-matrix lamellae. In metals of a medium-to-high SFE, copper-type SBs are formed against the background of band-like dislocation structures of elongated cells (microbands), e.g. [12,13]. Shear banding tends to become increasingly difficult in metals of very high SFE under plane strain deformation, and therefore have not always been observed in high purity aluminium. However, SB formation in Al and Al-based alloys is feasible when using hat-shaped samples.

The present work was undertaken to establish the effect of structure of flattened grains on the macro-SB formation during impact loading in technical purity aluminium (AA1050 alloy). The boundaries between grains of initial materials were used as markers of morphological changes during band formation. The deformation of hat-shaped samples leads quickly to well-defined macro-SB and corresponding texture changes. This opens the possibility to track carefully the crystal lattice reorientation within particular grains and the slip propagation across the grain boundaries (GB). Since shear banding strongly affects texture, electron backscattered diffraction (EBSD) in a scanning electron microscopy (SEM) was used.

2. Experimental
Hat-shaped specimen presented in Figs. 1a, d were machined from the hot-rolled sheet maintaining the axis of cylindrical sample perpendicular to the rolling plane of the initial sheet. The samples were deformed using a drop-hammer to logarithmic strains of ~ 0.61, under an applied energy of 600 J and an initial strain rate close to 530 s⁻¹. The grain structure of initial sheet and the as-deformed samples were characterised by SEM/EBSD. In the case of hot-rolled material the orientation maps were measured along the ND-RD and ND/TD sections, where: ND, RD and TD are normal, rolling and transverse directions, respectively. The microstructure of samples after impact deformation were characterized along section parallel to main sample axis (SA) using light microscopy and SEM/EBSD system. In the case of hat-shaped samples the indices {hkl}<uvw> always represent the texture components that have the normal of {hkl} plane perpendicular to SA-RA section, whereas the <uvw> direction parallel to radial axis (RA). (The hat shaped sample section is 60° || SA rotated with respect to ND-RD section of the initial sheet). The SEM orientation maps were measured using a FEI Quanta 3D equipped with a high resolution EBSD facility, typically at 15-20 kV and a working distance of 7 - 10 mm. The step sizes were ranged between 70 and 500 nm. Smaller step sizes (<100 nm) were selected to capture detailed microstructural features, whereas larger ones were used in the analyses of the textural changes during macro-SB formation.

3. Results and discussion

3.1. Microstructure and texture of initial material
The microstructure of the initial material was characterized in two sections perpendicular to the rolling plane by light microscopy and SEM/EBSD. The microstructure images reveal layers of very large flat grains (Figs. 1a, b) aligned along the rolling plane. The grain size in ND ranged between 20 µm and 60 µm. The grain dimensions in TD and particularly along RD were significantly larger at level of a few hundred of microns. The hot-rolled AA1050 alloy is characterized by some components of the copper-
type rolling texture, i.e. two variants of brass \{110\}<112> and four of S{123}<634>, but no C{112}<111> component was observed, as shown in the \{111\} pole figures (PFs) in Fig. 1c.

![Image](image.png)

**Figure 1.** (a) Schematic showing of the sample sectioning. (b) Initial microstructure in longitudinal section and (c) corresponding texture. (d) Hat shaped sample. Optical microscopy observation and SEM/EBSD measurements with step size of 1 µm.

3.2. **Morphological changes due to shear banding**

The microstructural features of the macro- shear banding were analysed by optical microscopy on the anodized sample sections. The SA - RA section shows formation of two almost symmetrically situated families of macro-SB: positively (macro-SB\(^1\)) and negatively (macro-SB\(^2\)) inclined at 55 - 85° with respect to RA. The axisymmetric SBs are occurring due the hat-shaped geometry of the sample and are in fact the same microstructural elements but mirror-reflected. The very first plastic instability, with which the shear banding starts, was due to their double kinking, as observed in the SA-RA section. The macro-SB were initiated near the sample free surfaces, where the grain boundary kinking was the greatest. In the next step SB propagates towards sample centre. The obtained contrast revealed changes in the grain boundary inclination during macro-SB formation. A rigid rotation of the lamellae by ~ 75 - 85° near the sample free surfaces was observed as well as their relative translation in the areas outside the band (Figs. 2a, b). In the areas near the central part of the sample, the GB inclination angle gradually decreases to ~ 30°. This suggests that the MSB propagated from the sample free surfaces towards the sample centre [12]. At the macro-SB boundary, the traces of GB show diverse inclinations, varying from 0° to about 85° to RA forming very broad transition layers which separate the core of the macro-SB from the matrix, as earlier observed at TEM scale in twinned structures of Cu-Al alloy, e.g. [3, 9-11]. The core region of each macro-SB is surrounded by two transition layers, in which the boundaries between the flattened grains are progressively bent (Fig. 2a). For this specific case of macro-SB formation the transition layers have a significantly larger thickness than in the core region. The GB traces revealed a well-defined rotation of opposite sign within the symmetrically situated bands (Figs. 2a, b). Therefore, it is deduced that the textures in the macro-SB\(^1\) and macro-SB\(^2\) areas are essentially different and stems from two opposite rotations. The sense of the rotation is to increase the inclination of the flattened grains with respect to RA. This locates the traces of the GB closer to the shear plane.
3.3. Orientation changes as a result of macro-SB formation.

The local orientation measurements with a high spatial resolution provide information about the mechanisms of band formation by following the subtle orientation changes in the areas of macro-SB. To a first approximation, the orientation maps confirm that the traces of GBs (in analysed area of given macro-SB) increase their inclination angle with respect to RA via rotation by $\approx 45^\circ$ around the axis perpendicular to the analysed section. However, this simple rotation cannot explain alone the mechanism responsible for ‘transformation’ of the matrix texture components into the one identified inside the band. This leads to the conclusion that an accurate crystal lattice rotation is more complicated than described above rigid body rotation.

![Figure 2](image-url) (a) Optical micrograph showing the structure of MSB on sample scale (sample axial section), and (b) schematic presentation of the rotations inside the MSB; reorientation of the structure elements inside the MSB and displacement of the layers outside the band.

It is apparent that the experimentally observed texture of the matrix was composed of three scattered groups of orientations. Since the scattering of each component group often exceeds 30° each group can be represented by an ‘average’ orientation and described as follows: $M^1$ - (-110)[-1-12], $M^2$ - (05-3)[535] and $M^3$ - (-110)[-5-5-2]. The texture image measured inside each macro-SB is composed of two main groups of components. For macro-SB$^R$ or L the texture components can be defined as: MSB$^{R1}$ - (-110)[-1-13] and MSB$^{R2}$ - (02-3)[-532] (for the right side), whereas for MSB$^{L1}$ - (-110)[115] and MSB$^{L2}$ - (-12-1)[-101] (for the left side).

The orientations of the matrix undergo different but not accidental rotations during their incorporation into the given macro-SB area (Figs. 3a-c). Therefore, the rotation axes between the matrix and macro-SB provide essential information to understand the mechanisms responsible for SB formation. It is assumed here that the crystal lattice rotation results from the activity of specific slip system(s) and the rotation axis is perpendicular to the active slip plane normal and slip direction. During strain localization, the orientations identified in the matrix undergo rotations around the selected <112> or <011>-type axes due to the dominance of given {111}<011>-type slip system(s). All the identified rotations always align with one of the {111} plane (in each orientation groups), towards position parallel to the SB plane. It is clear that the rotation tendencies described above are also crucial for understand the mechanism of slip propagation across the GB since the {111} slip plane (which is parallel to the shear band plane) is common for neighbouring layers characterized by two main groups of orientations identified inside macro-SB areas.

The proposed scenario of the crystal lattice re-orientation during the matrix structure incorporation to the macro-SB region is as follows. In the early stages of macro-SB formation, the highly stressed slip system leads to the crystal lattice rotation of the grains inside the matrix towards their position inside sheared zone. In the case of all three orientation groups (forming texture of the matrix) the rotation occurs around one of the <112>-type axis (or in the case co-planar slip system around summary the <011>-type axis) leading to formation of two main groups of orientations inside sheared zone with
common the \{111\} plane parallel to the shear band plane. The identified rotations that lead to transformation of matrix components into the MSB\(^R\) one can be defined as follows:

\[
M_1(\text{-110})[\text{-112}] + 15^\circ \text{ rotation around } [-\text{110}] \rightarrow \text{MSB}^{R2}(\text{-110})[\text{-113}]
\]

\[
M_2(\text{04-3})[\text{-534}] + 40^\circ \text{ rotation around } [-\text{21-1}] \rightarrow \text{MSB}^{R(b)}(\text{02-3})[\text{-532}]
\]

\[
M_3(\text{-110})[\text{-5-5-2}] + 30^\circ \text{ rotation around } [-\text{110}] \rightarrow \text{MSB}^{R(a)}(\text{-551})[\text{-4-3-5}]
\]

![Figure 3](image)

**Figure 3.** Orientation changes due to (a) MSB\(^L\), (b) matrix and (c) MSB\(^R\) formation. SEM/EBSD orientation maps measured in SA-RA plane with step size of 70 nm.

Similar rotations can be identified for transformation of matrix components into the MSB\(^L\). If the orientation group of the matrix layers is denoted \(M_2\) only a simple rotation around axis parallel to [-21-1] is needed for the coincidence of one of the \{111\} plane with the shear plane and <110> direction with the close proximity of the shear direction. If the orientation groups of the matrix are denoted \(M_1\) and \(M_3\) a rotation around axis parallel to [-110] is needed (the co-planar slip can take place along a compound <112> direction). Rotation around the [-110] axis is in fact the superposition of two rotations around the axes lying on the same the (11-1) plane, i.e. [-121] and [-21-1]. According to this, the majority of the grain orientations inside the macro-SB are rotated around the <110> axis lying in the shear plane. However, both cases enable shear (on double or single slip system(s)) along the SB plane. These
rotations lead to the coincidence of the shear direction with the <110> direction. In each case the rotations inside macro-SB occur in such a way that one of the {111} plane became (nearly) parallel to the shear plane, despite different initial grain orientations. Additionally, one of the <011>-type directions lying in these planes, systematically tends to coincide with the shear direction. 

In this polycrystalline aggregates the macroscopically observed shear plane, in fact, consists of small segments limited to the particular grains or their fragments (Figs. 3a, c). These parts are only slightly deviated from the macroscopic shear plane. Finally, one of the {111} slip planes (~ common for the all neighbouring layers) becomes nearly parallel to the maximum shear stress plane. The re-orientation of the {111} planes towards the macro-SB plane facilitates further dislocation slip in the <110> shear direction. In this way, the mechanism of band nucleation and development is strictly crystallographic and connected with a single slip or double co-planar slips operation. A natural consequence of this rotation is that the inclinations of the {111} planes within and outside the band are different.

4. Conclusions

This study on AA1050 aluminium alloy clarifies the mechanisms responsible for the microstructure and texture evolution resulting from macro-SB formation during high strain rate deformation. The most important features may be summarized as follows. (i) In the layered structure of flattened grains the shear banding occurs via deflection of flattened grains within some narrow areas leading to kink-type bands that are the precursors of SB. (ii) The mechanism of macro-SB formation is strictly crystallographic in nature since in all the grains of the sheared zone, the crystal lattice rotated in such a way that one of the {111} slip planes became nearly parallel to the shear plane and the <011> direction became parallel to the direction of maximum shear. (iii) This strain-induced crystal lattice rotation led to the formation of new texture components, different from those identified inside non-deformed matrix, that facilitate slip propagation across the grain boundaries. (iv) The macroscopically observed macro-SB plane consists of small segments limited to the particular grains or their fragments. These segments were only slightly deviated from the macroscopic shear plane.

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