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CRYSTALLOGRAPHY OF THE COPPER-TYPE SHEAR BANDS

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The microtextural changes within macroscopic shear bands (MSBs) in high purity aluminium and copper single crystals with {112}⟨111⟩ initial orientation, deformed by channel die compression, have been studied in detail. Systematic measurements of single orientations by SEM/EBSD and TEM/CBED clearly show that the investigated crystals are stable only in a global sense. The occurrence of the first set of MSBs is connected with a local lattice rotation towards the {001}⟨110⟩ orientation. In particular, this process directs the (111) slip plane, towards a shear plane and the activation of new, highly stressed {111}⟨101⟩+{111}⟨011⟩→ →CP{111}⟨112⟩ slip systems, is documented. The deformed matrix near MSBs represents a relatively more stable behaviour, and the group of the orientations situated near the C{112}⟨111⟩−D{4 4 11}⟨11 11 8⟩ positions describes it.

W pracy analizowano mikroteksturowe zmiany zachodzące w obszarze makroskopowo obserwowanych pasm ścinania (MSBs), w monokryształach aluminium i miedzi o orientacji wyjściowej {112}⟨111⟩, odkształcanych w próbie nieswobodnego ściskania. Pomiary orientacji lokalnych z wykorzystaniem techniki TEM/CBED a zwłaszcza SEM/EBSD, umożliwiającej „konstruowanie" map orientacji, tzw. Orientation Imaging Maps (OIM) wskazują, że badane kryształy są stabilne jedynie w sensie globalnym. Pojawienie się pierwszej rodziny makroskopowych pasm ścinania związane jest z lokalną zmianą orientacji, obserwowaną na figurze biegunowej jako rozmycie w kierunku orientacji {001}⟨110⟩. Jednym ze skutków rotacji sieci krystalicznej wewnątrz obszaru pasma, jest zmiana nachylenia płaszczyzny poślizgów współpłaszczyznowych CP(111) w kierunku nałożenia się z płaszczyzną ścinania a także aktywacja operujących na tej płaszczyźnie poślizgów typu: {111}⟨101⟩+{111}⟨011⟩→CP{111}⟨112⟩. Zmiany orientacji obserwowane w obszarach poza pasmem są znacznie mniejsze i mogą być opisane ciągiem położeń: C{112}⟨111⟩−D{4 4 11}⟨11 11 8⟩.

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1. Introduction

The shear bands (SBs) with their specific internal microstructure composed of very fine cells, with high angle boundaries, resemble to some extent *nanocrystalline materials*. Therefore, they appear to be interesting objects for the investigations of both, deformation and recrystallization behaviour. The crystallographic aspects of these bands formation have not been so far fully recognized.

Earlier experiments [1–7] revealed a strong tendency of C, that is $\{112\}\langle 111\rangle$ initially oriented f.c.c. single crystals to form SBs. It is also generally accepted that the formation of the structure of elongated cells or deformation twins, determine the occurrence of this type of inhomogeneities. The formation of the 'brass'-type SBs in materials with low stacking fault energy (SFE) is well documented [7–11]. With increasing SFE, the formation of another kind of SBs, so-called 'copper'-type can be expected, but this process becomes increasingly difficult.

In this work a preliminary assumption has been made that the scattering of initial $\{112\}\langle 111\rangle$ orientation at large deformation is mainly related to the local orientation changes within SBs and to their compact clusters called macroscopic shear bands (MSBs). The experimental verification of this hypothesis requires the application of local orientation measurements. Such a technique has been mostly used in the investigations of the microtexture development within recrystallized or hot deformed samples [12–15]. The investigations of the changes of crystal orientation, which results from the SBs formation within cold deformed metals, are rare and have a random character [5, 11, 16–18]. Sometimes, it has been even concluded that it is impossible to define the SBs microtexture components. This statement is based on the observation of a strong disturbance of the internal microstructure of SBs, due to high dislocation density.

In the present study particular attention has been paid to the experimentally observed fact of SBs occurrence *in non-crystallographic positions with respect to the orientation of the deformed matrix outside the band.* The trace of the shear band plane deviates from the (111) octahedric slip plane by about 22° toward the compression direction, around the transverse direction (TD) [5]. Therefore, to clarify this, a detailed qualitative investigations of main microtexture components formed within 'copper'-type SBs and the neighbouring matrix were performed. The aim of the present work is to enlighten the nature of both the crystal re-orientations during shear banding and the crystallographic relations between 'copper-type' shear bands and the surrounding matrix in medium (copper) and high (aluminium) SFE single crystals deformed in channel-die compression tests. The problem '*of the mechanism by which strain is accommodated within shear bands*' appears to be crucial in this respect. A detailed analysis of the evolution of microtexture components within and outside MSBs has therefore been undertaken using the local orientation measurements.

2. Experimental

High purity aluminium (99.998%) and copper (99.995%) single crystals with initial C(112)[11$\bar{1}$] orientation, obtained by B r i d g m a n method, were subjected to investigations. Cubic samples of $10 \times 10 \times 10$ mm^3 covered with teflon foils, were plain strain compressed (PSC) in multi stage channel-die tests at a nominal strain rate $\sim 10^{-5}s^{-1}$, retaining the ratio $l/h = 1$ (where: l – sample length, h – sample height) at the beginning of each stage. The applied deformation temperatures were 77 K and 293 K for Al and Cu, respectively and they ensured the formation of broad MSBs in sample scale, clearly visible in the plane perpendicular to TD.

The detection of the MSBs was mainly based on longitudinal sections containing normal (ND) and elongation (ED) directions. The deformation microstructures of PSC samples were characterized using a transmission electron microscopy (TEM), scanning electron microscopy (SEM) and optical metallography. An etching technique, as developed by Köhlhoff et al. [19] was used on Cu samples, while observations in polarized light on slightly anodized sections served for the Al ones. The microtexture determination within the particular areas of the deformed samples by convergent beam electron diffraction (CBED) on TEM and electron back scattered diffraction (EBSD) on SEM, were applied to identify the slip systems operating within MSBs areas,

The local orientations were investigated using a 200 kV PHILIPS CM20 (TEM) with on-line technique of the Kikuchi diffraction patterns analysis and JOEL 6400 (SEM) equipped with automatic HKL™ Technology EBSD software. The microstructures of the deformed samples were examined by TEM on thin foils prepared from a section perpendicular to the TD.

3. Results and discussion

3.1. Global deformation structures

Optical metallography of both investigated materials revealed a clearly defined MSBs. Typical examples of these bands in sample scale are shown in Fig. 1a and b. The longitudinal sections presents the whole thickness of samples. Early stages of the shear banding are manifested when only one set of MSBs parallel to TD can be visible. In the presented cases they run across the whole thickness of the sample and occupy planes inclined at ~35° and 40–45° to ED (Figs. 1a and b), for Cu and Al, respectively.

In the case of Cu deformed up to 37%, application of the K ö h l h o f f etching develops the {111} facets, which intersect along the ⟨110⟩ directions, so that the etched surface exhibits lines as projections along the ⟨110⟩ directions. For copper or copper based alloys this technique gives the same kind of extensive local texture data as TEM/CBED and SEM/EBSD. It is clearly shown in Fig. 1c, that for the deformed areas outside the bands, the etch patterns indicate orientations close to D(4 4 11)[11 11 $\bar{8}$]. One can see parallel lines of ⟨110⟩ directions inclined at ~27° to ED. In the narrow SBs

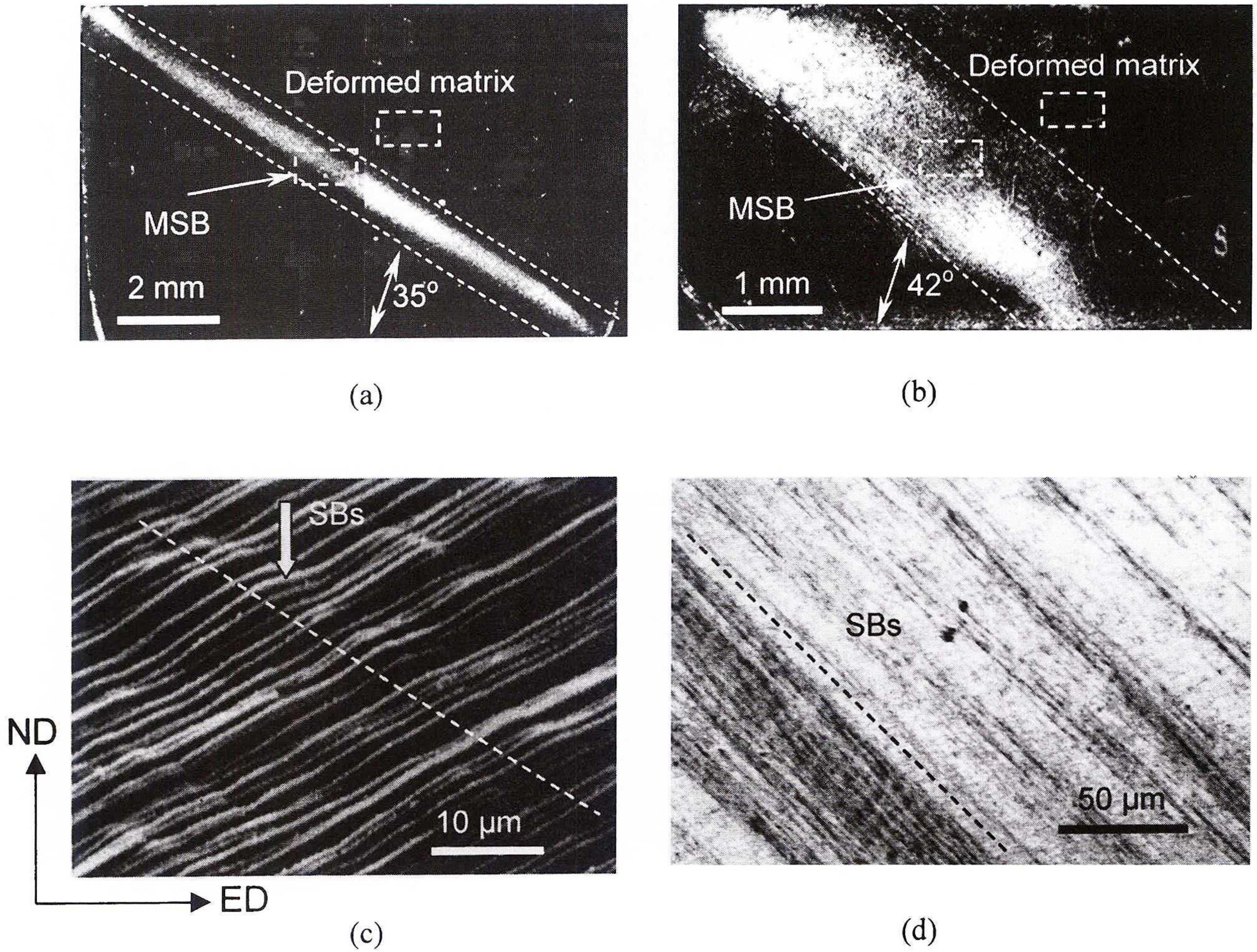

Fig 1. MSBs formation in copper deformed 37% at room temperature (a) and aluminium deformed 62% at 77 K (b), and the corresponding internal optical microscopy microstructures of the bands for cooper etched in nitric acid (c) and aluminium after anodic oxidation (d). The arrow in (c) shows a kink of the $\langle 110 \rangle$ direction

region, the inclination of these directions decreases up to 0–10° with respect to ED, corresponding to the orientation close to (100)[011].

In the case of Al samples, the higher applied deformation (62%) caused that MSBs observed in the lateral plane are significantly broader (of about 2 mm in width). Higher magnification shows (Fig. 1d) that these macroscopic shear bands are composed of single SBs.

3.2. Local orientation measurements

3.2.1 SEM/EBSD method

The following results of microtexture investigation within MSBs have been obtained from samples deformed up to 37% and 62% for Cu and Al, respectively. They represent early and advanced stages of the MSBs formation.

Figures 2a and 3a show orientation imaging maps (OIM) of Cu (37%) and Al (62%) performed in so-called band contrast, i.e. as a function of the *Kikuchi band quality factor* (white for high quality, black for low quality typically for heavily deformed areas). It turned out that the as-deformed shear band orientations are rather difficult to resolve by the orientation contrast. For this reason, very careful sample preparation was necessary to enable SEM/EBSD investigations. In the Cu and Al single crystals the macroscopically observed bands (at the sample scale) are composed of single, highly localised bands, showing large lattice rotations. Thus the darker, longitudinal areas (low image quality factor) characterise the shear band positions observed by optical microscopy at a more global scale (Figs. 1a and b).

The {111} pole figures obtained for the deformed matrix and MSBs regions are presented in Figs. 2b and c for Cu and in Figs. 3b and c for Al, respectively. The deformed matrix shows a relatively stable behaviour and it can be described by orientations grouped near the D position. In the both analysed cases the clearly defined tendency of the crystal lattice rotation towards (100)[011] orientation within MSBs, can be observed. This fact is manifested in the pole figures by a wide set of positions deviated from D towards the (118)[44$\bar{1}$] orientation.

The micro shear bands exhibit a typical periodic array with a wide range of misorientations characterized by the distribution of misorientation axes between the neighbouring measurement points. They show a great tendency to group near the ⟨110⟩ axis (Figs. 2d and 3d), with the misorientations angles up to 20° and 30° for Cu and Al, respectively. The largest ones roughly correspond to the dark areas seen in the orientation imaging maps. In agreement with earlier systemathic investigations [20, 21], the large misorientations do occur along the edges of the bands.

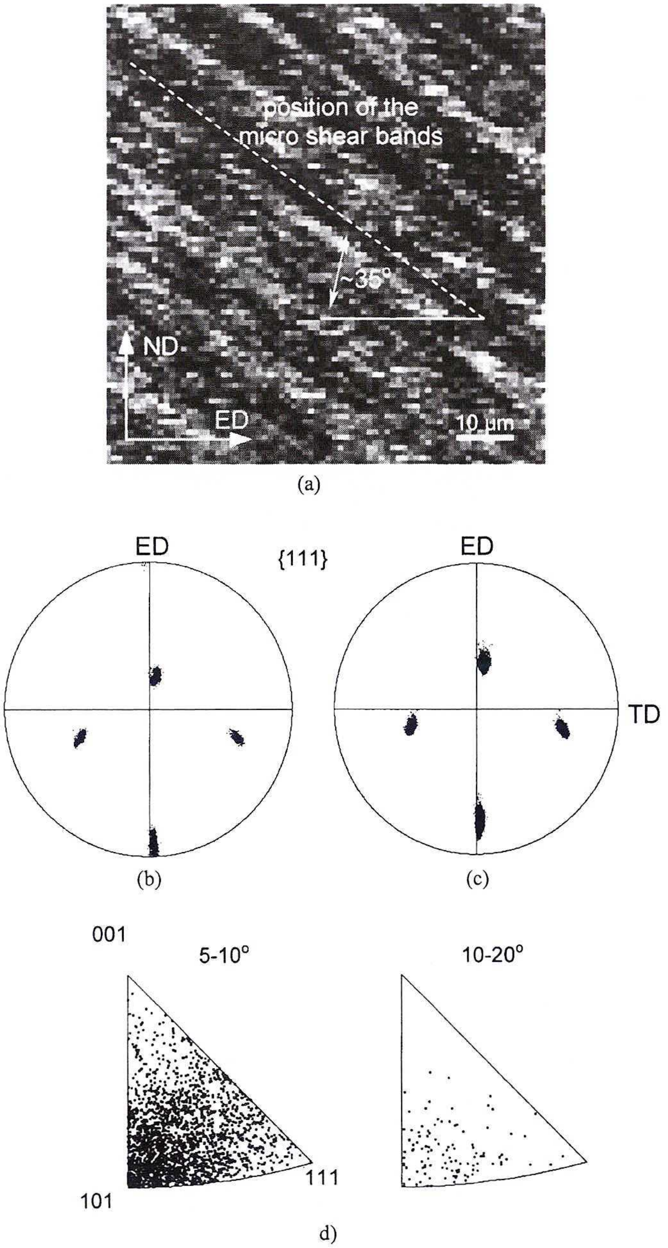

Fig. 2. (a) OIM from area of MSBs in Cu, deformed 37% at room temperature and {111} pole figures from (b) deformed matrix, (c) MSBs area, and (d) distribution of misorientation axes between neighbouring points obtained in SEM/EBSD measurements

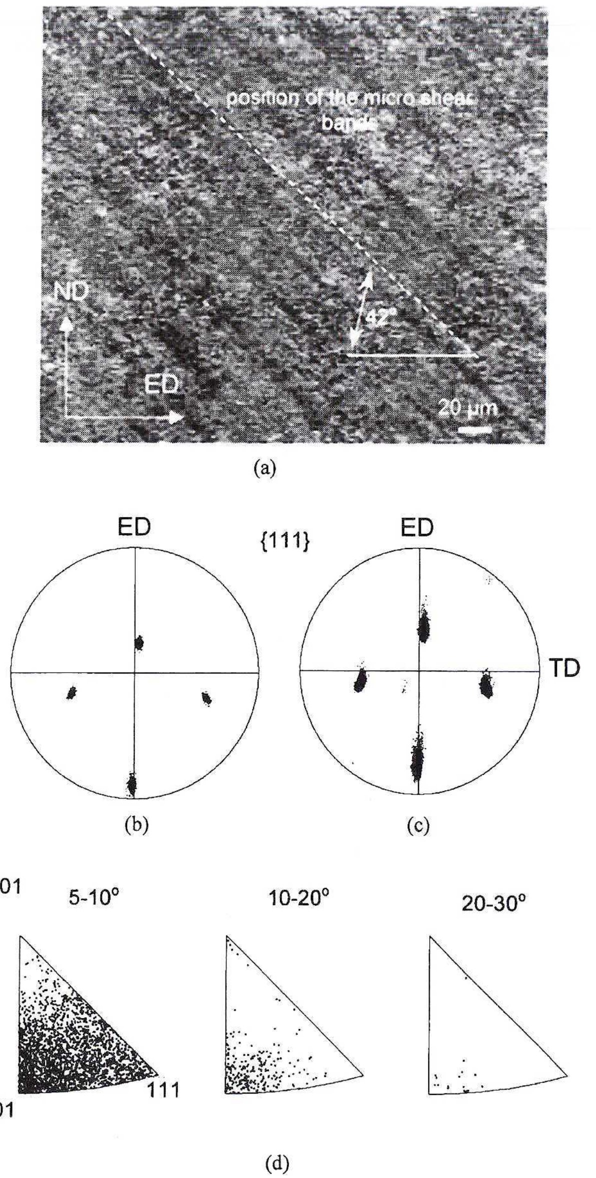

Fig. 3. (a) OIM from area of MSBs in Al deformed 62% at 77 K and {111} pole figures from (b) deformed matrix, (c) MSBs area, and (d) distribution of misorientation axes between neighbouring points obtained in SEM/EBSD measurements

3.2.2. TEM/CBED method

The type of the microstructure and the dominance of a slip system strongly depend on the place where the TEM sample was cut (within or outside the MSBs area).

It should be recalled that for the $\{112\}\langle 111\rangle$ oriented fcc single crystals, two pairs of slip systems operate in the early stages of the deformation process [2, 5, 16]. One of them form a co-directional pair (CD) and may be represented as: $(\bar{1}11)[110]+(1\bar{1}1)[110]\rightarrow(001)[110]$. The second form a co-planar pair (CP) and may be described as: $(111)[\bar{1}01]+(111)[0\bar{1}1]\rightarrow(111)[\bar{1}\bar{1}2]$. The dominance of CD slip systems in nearly equiaxed cell structure within the deformed areas outside MSBs is well documented at low strains. At higher deformation degrees this structure has a great tendency to form compact clusters of elongated cells. This effect, connected with the dominance of CP slip systems, is observed clearly within MSBs areas and tends to dominate as the crystal lattice rotates towards the (001)[110] orientation (Figs. 4a and c). Positive or (+) rotations will lead towards (001)[110] whereas negative or (-) rotations towards Goss (110)[001]. As it is shown in Figs. 4b and d, the macroscopically sheared areas rotate [by (+) TD$\|\langle 110\rangle$ rotation], towards the orientation group usually situated between $(4\ 4\ 11)[11\ 11\ \bar{8}]-[116)[33\bar{1}]$ positions. According to this tendency, the angle between the (111) plane and the compression plane increases from 19.5° at the $C(112)[11\bar{1}]$ initial orientation, through about 40° at $(116)[33\bar{1}]$ up to about 50° at (100)[011].

In all analysed cases, formation of the different types of deformation structures is in a good correlation with the preference for the occurrence of active slip systems. These active slip systems show a tendency to produce planar sets of dislocation walls, observed in Fig. 4a and c, for copper and aluminium, respectively. The experimentally observed fact of the crystal lattice reorientation within narrow areas is difficult to explain against the background of the operation of the unit slip systems of $\{111\}\langle 110\rangle$ type. Therefore in this approach, Schmid factor analysis was applied to the possibility of the occurrence of the 'resultant' CD and CP slip systems. In accordance with Tucker's assumption, within the range of orientations of $(112)[11\bar{1}-(4\ 4\ 11)[11\ 11\ \bar{8}]$ the 'resultant' CD system is more privileged than the 'resultant' CP one. In the case of copper, the orientations identified within MSBs are grouped near the $(114)[22\bar{1}]-(116)[33\bar{1}]$ positions (Fig. 4b). These changes of the crystal orientation are more clearly documented within Al substructure (higher deformation degrees), where the compact clusters of the SBs are observed. The rotation from the initial C-position towards (001)[110] facilitates the slip occurrence in the CP systems. The CP(111) slip plane in the $(118)[44\bar{1}]$ orientation (Fig. 4d) is situated parallel to the shear plane inclined at about 45° to ED (see Fig. 1b), and the system $CP(111)[\bar{1}\bar{1}2]$ reveals the highest Schmid factor.

As it has been found, the microstructural and microtextural behaviour of fcc metals with medium SFE is comparable to that of fcc metals with high SFE. In accordance with earlier investigations, the present detailed measurements by TEM/CBED and SEM/EBSD methods have clearly showed a great tendency of the crystal lattice for

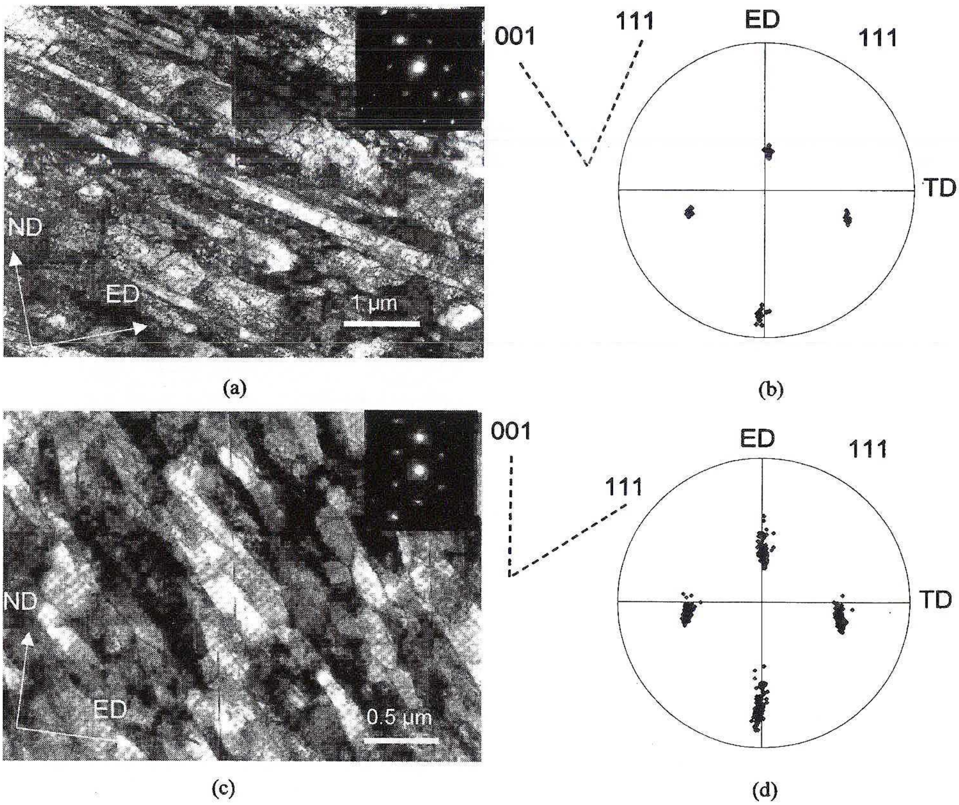

Fig. 4. Microstructures within MSBs areas of Cu deformed 37% at room temperature (a) and Al deformed 62% at 77 K (b), and the corresponding {111} pole figures for Cu (c) and for Al (d). TEM/CBED measurements

rotation towards {100}⟨011⟩ orientation within the MSBs area. This local change of orientations suggests that the stability of C-orientation can be considered only in a global sense (in sample scale). The local lattice reorientation within relatively narrow area increases the resolved shear stress on the (111) planes and activates the CP slip systems. In this way, the critically stressed CP(111) slip planes, initially inclined at 19.5° to compression plane, can rotate by increasing this inclination, towards the shear plane. This process is in agreement with previous investigations made by Wagner et al. [2], Köhlhoff et al. [4], Jasieński et al. [5] and Morii and Nakayama [6], who suggested that the local lattice bending leads to the nucleation of shear bands. Therefore, *the position of these bands is non-crystallographic only with respect to initial crystal orientation or the orientation of the deformed matrix outside the band.* It is also important to note that the inclination angle of the band is strongly correlated with the main internal microtexture components identified inside MSBs [22]. However, the interrelation of these two parameters always determines the occurrence of slip in the CP slip systems along the shear plane. This fact is strongly supported by the observations of the second sets of MSBs formation, where the opposite rotation of the crystal lattice towards {110}⟨001⟩ orientation was observed [16].

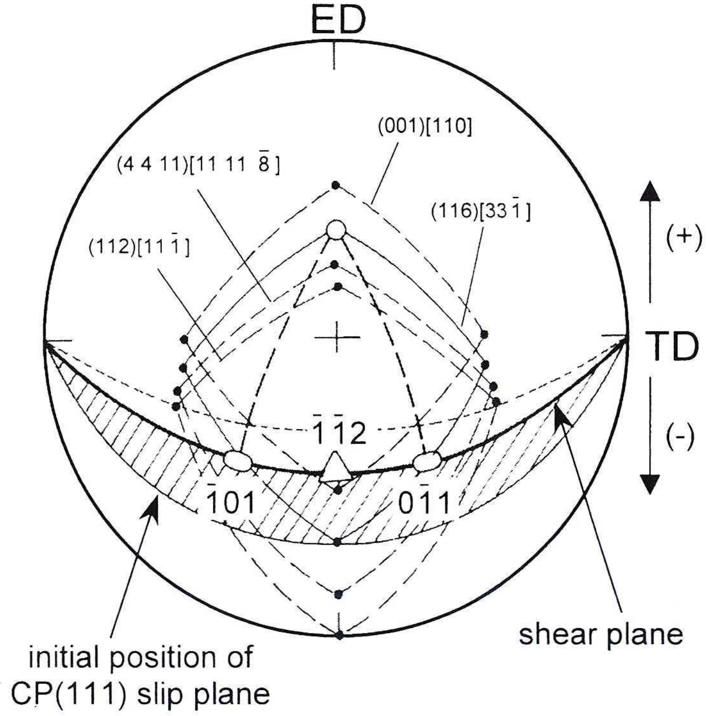

Fig. 5. Stereographic projection showing the change in the crystallographic orientation and the position of the co-planar (111) slip plane as a result of primary macro shear band formation. A possable way by (111) slip plane is marked

The changes of the crystal orientation recorded in TEM/CBED and SEM/EBSD measurements are summarised in stereographic projection in Fig. 5, which shows the primary MSBs formation as a result of the crystal lattice rotation. The passable way for the (111) co-planar slip plane within the band, from the initial to the end position (marked area), corresponds to a (+) 20÷25° TD rotation. In the critical position of the (116)[33$\bar{1}$] orientation, the CP(111) slip plane is inclined at about 42° to ED and is parallel to the material shear plane (marked as a bold line). Therefore, taking into account the observed local orientation changes inside the MSBs, their crystallographic nature seems to be well evidenced.

4. Conclusions

Measurements of the local orientations, in particular within MSBs areas, show that the investigated crystals are stable only in a global sense.

1. The occurrence of MSBs results from the lattice rotation towards the (100)[011] orientation. This rotation directs the CP (111) slip planes towards the shear plane and a $\langle 112 \rangle$ direction close to the shear direction.

2. Within the MSBs areas, the activation of highly stressed $\{111\}\langle 101\rangle +$ $+ \{111\}\langle 011\rangle \rightarrow CP\{111\}\langle 112\rangle$ slip systems, along the shear plane, has been identified. This fact confirms the thesis about the crystallographic nature of shear bands.

3. The deformed matrix near MSBs represents a relatively more stable behaviour, and can be described by the group of orientations situated near the D(4 4 11)[11 11 $\bar{8}$] position.

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