

On twinning and shear banding in copper single crystals with rolling texture orientations plane strain compressed at high strain rate

H. Paul^{a,*}, I. Mania^a, R. Chulist^a, S. Puchlerska^b, M. Prażmowski^c

^a Institute of Metallurgy and Materials Science, Polish Academy of Sciences, Krakow, Poland

^b Faculty of Non-Ferrous Metals, AGH University of Science and Technology, Krakow, Poland

^c Opole University of Technology, Mechanical Department, Opole, Poland

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ABSTRACT

Crystal lattice rotations developed in copper single crystals with (110)[1–1–2](Br), (110)[001](G), (112)[11–1](C), and S (346)[63–5](S) orientations have been examined to assess the role of an impact loading on twinning, deformation banding and shear banding. The single-crystalline specimens were channel-die compressed to logarithmic strains of 0.9 with a strain rate of $4.7 \times 10^5 \text{ s}^{-1}$. The microstructures were characterised using scanning (SEM) and transmission (TEM) electron microscope, whereas electron backscattered diffraction (EBSD) in SEM and X-ray diffraction have been applied to evaluate the texture evolution. The results showed that a very high strain rate promoted the formation of deformation twins in all orientations. However, a strong variation in the twin density and number of twin families was observed depending on the orientation of the single crystal. In the Br- and G-oriented crystals, only single needle-shaped twins were observed, whereas in the C- and S-oriented crystals, the formation of compact clusters of deformation twins of two generations on different twinning planes was identified. SEM/EBSD measurements revealed that deformation bands were formed in the crystallites with G and Br orientations, whereas shear bands (SBs) were observed in the C- and S-oriented crystals. This study demonstrates that the rotation of twin and matrix platelets in narrow areas of SBs developed in C- and S-oriented crystallites, combined with deformation twinning in the re-oriented primary matrix platelets can result in the formation of texture components near the $\{110\}<001>$ orientation. Finally, a crystallographic model of the formation of SBs in twinned structures of face-centred cubic metals deformed at high strain rates is proposed.

1. Introduction

The microstructural and textural evolutions with strain and the mechanical response of metals strongly depend on the applied strain rate. For strain rates smaller than 10^4 s^{-1} (static or quasi-static loading), a thermally activated mechanism governs the dislocation motion because an increase in temperature helps the dislocations overcome existing obstacles, thereby reducing the strength of the material [1–6]. Studies conducted on medium-to-high stacking fault energy (SFE) face-centred cubic (fcc) polycrystalline metals under plane-strain deformation at room temperature (RT) [7–9] showed that under quasi-static loading, the dislocation structures create two basic configurations [10]. Dislocation cell blocks and first-generation microbands were observed in low-to-medium logarithmic strains (<0.6) [11]. At logarithmic strains higher than 0.6, the microstructure is dominated by compact clusters of microbands, which gradually became parallel to the

compression or rolling plane as the deformation increased. At large strains, shear bands (SBs) appeared within narrow areas because of the instability of the layered structure of the microbands [12,13]. However, deformation twins independent of the applied deformation path have not been observed in fcc metals with high-to-medium SFE deformed at RT [12,14].

When the strain rate increases above $\sim 10^4 \text{ s}^{-1}$, the strain-rate sensitivity of metals increases rapidly, and the stresses are sufficient for dislocations to overcome obstacles without the contribution of thermal fluctuations [15–18]. In early studies, this increase in strain rate sensitivity was interpreted as a change in the deformation mechanism, from thermally-activated deformation at low strain rates to the dislocation-drag mechanism at high strain rates [15]. However, subsequent works showed that the increased rate sensitivity arises from the rate sensitivity of the structural evolution rather than from any change in the deformation mechanism because the dependence of the material

* Corresponding author at: Institute of Metallurgy and Materials Science, Polish Academy of Sciences, 30-059 Krakow, 25 Reymonta St., Poland.

E-mail address: h.paul@imim.pl (H. Paul).

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response on temperature is considerably smaller than that of the strain rate, especially in the range of large deformation [1]. Yaghoobi and Voyiadjis [18] and Fan et al. [19] revealed that the interaction between moving dislocations is the mechanism governing the strength of the material during impact loading, and the contribution of the thermally activated mechanism can be neglected. In this context, reducing the thermal activation of deformation mechanisms during impact loading can be considered similar to lowering the temperature or decreasing the SFE. Therefore, twinning has been observed in the impact-loaded [20–24] *fcc* metals and alloys with high and medium SFE, which was not the case at a low strain rate deformation [25–27].

The thermal softening effect is particularly important in the context of shear bands (SBs) formation. SBs are characterised by large deformations localised in a narrow band and occur intensively during both quasi-static and impact loading. Most previous experimental studies on the shear banding phenomenon at high strain rates have examined microstructural/textural changes using ‘hat-shaped’ polycrystalline samples. This specific axisymmetrical sample geometry forces shearing, even in metals that do not tend to form SBs spontaneously during channel-die compression or rolling. However, drawing conclusions about the effect of high strain rates on deformation texture in this particular case is complex, especially when trying to relate the observed textural changes in hat-shaped specimens to changes occurring in cold-rolled sheets/plates. Thus, the fundamental mechanisms responsible for the nucleation, thickening, and propagation of SBs in deformed structures and their impact on the texture of cold-rolled sheets are still not well understood.

Therefore, the aim of this study is to characterize the evolutions of the microstructure and texture of pure copper deformed at extremely

high strain rates under strictly controlled conditions. Single crystalline samples with typical rolling texture orientations are used in order to avoid the heterogeneous plastic deformation resulting from grain interactions in the polycrystalline aggregate. The deformed microstructures are characterised over a range of scales, ranging from 10^{-3} to 10^4 μm by the combination of transmission electron microscopy (TEM) and scanning electron microscopy (SEM) as well as light microscopy. Microscopic observations are supplemented by texture measurements using X-ray diffraction and SEM equipped with an electron backscattered-diffraction (EBSD) facility. The results are then correlated with the microstructures and textures of rolled or channel-die compressed single- and poly-crystalline samples deformed during quasi-static loading at similar strains or uniaxially compressed at high strain rates.

2. Experimental

Oriented single crystals of high-purity Cu (99.998 %) were prepared by directional (vertical) solidification and were carefully cut with a wire saw into rectangular compression samples with dimensions of $8 \times 10 \times 8$ mm³ (width/height/length). The selected orientations are representative of copper- and brass-type *fcc* rolling-texture components, namely: (346)[63–5] (denoted hereafter as S), which is close to a variant of the S {123}<634> orientation, (112)[11–1] (C), (110)[1–1–2] (Br or brass), and (110)[001] (G or Goss) orientations. The nomenclature used for the crystals is the same as that used for the corresponding *fcc* rolling texture components. Two of the orientations, G and Br, are nominally stable during quasi-static loading in terms of plane-strain deformation to logarithmic strains up to 1.5 [28,29]. The other two (C and S) are unstable

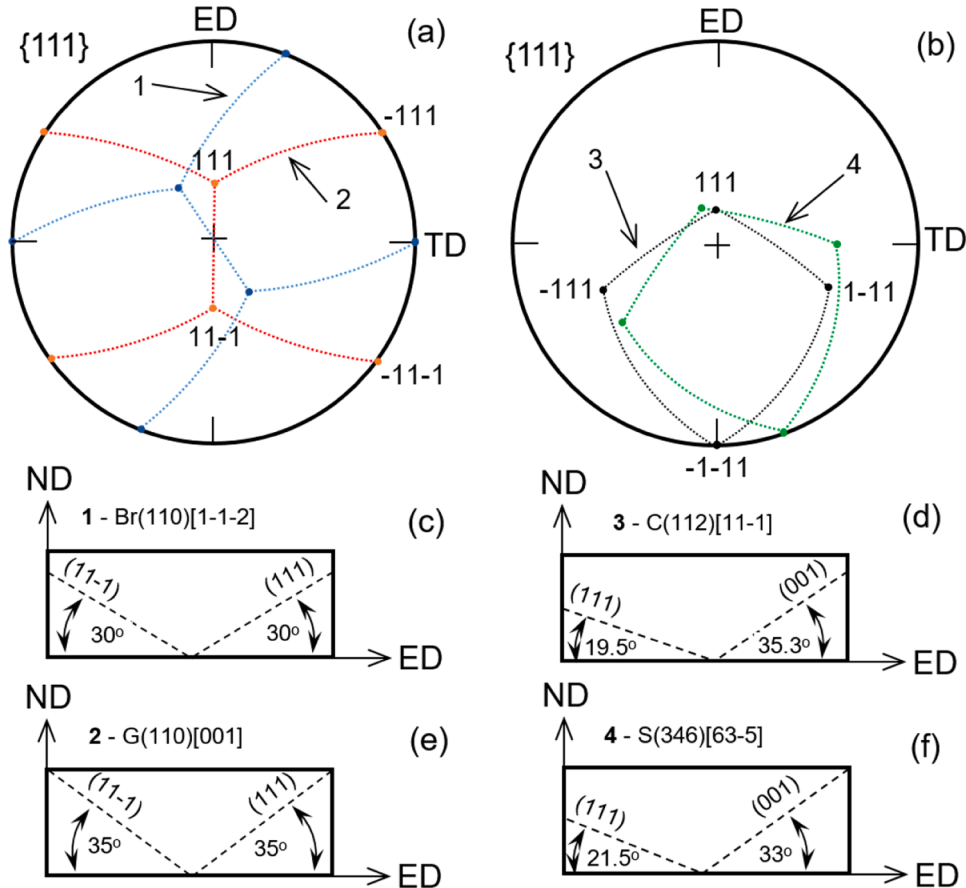


Fig. 1. Stereographic projections showing two groups of orientations which are nominally (a) stable, i.e. Br(110)[1–1–2] and G(110)[001], and (b) unstable, i.e. C(112)[11–1] and S(346)[63–5] orientations, during plane strain deformation. (c–f) Schemes showing the situation of traces of the most active slip planes in the ND/ED section.

[12,13] because starting from moderate strains, they show strong texture scattering owing to SBs formation. Stereographic projections showing the crystal orientation and a schematic representation of the locations of the important slip plane traces observed on the (normal direction)/(extension direction) (ND/ED) plane are presented in Fig. 1. After electropolishing, X-ray-diffraction measurements were performed on three perpendicular planes to determine the exact orientation of the crystals. The orientations of the initial crystals were not completely perfect. However, they only deviated by a few degrees from their reference positions. The initial orientations of the crystals are listed in Table 1 in terms of the Euler angles and approximate Miller indices.

The samples were then subjected to channel-die deformation in a single step up to 60 % (logarithmic strains of ~ 0.9) using explosive energy to propel the punch. The strain is defined as $\varepsilon = \ln(h_0/h_f)$, where h_0 and h_f are the sample height, before and after dynamic channel-die compression, respectively. Teflon films were used as lubricant to minimize the effect of friction. Additionally, the die and punch walls were covered with a thin PTFE spray coating. Fig. 2 schematically illustrates the test setup. A steel driving (flyer) plate with a thickness of 10 mm and spacer blocks with heights of 4 mm in the channel were used for the tests. The distance between the channel-die punch and drive plate was 15 mm. A modified Ammonium Nitrate - Fuel Oil explosive charge was used to propel the flyer plate and initiate an impact velocity of 470 m/s. This means that the single-crystalline samples were subjected to initial strain rate of approximately $4.7 \times 10^5 \text{ s}^{-1}$. A fibre-optic probe (Explo-met-Fo-2000, Kontinitro, Switzerland) was used to measure the vertical velocities of the flyer plate. High-speed deformation of single-crystalline specimens in a channel-die was carried out by ZTW Explomet, Opole (Poland).

After a final strain of 0.9, the crystal orientation was determined using standard X-ray diffraction. Most metallographic analyses were conducted along the ND/ED section. In selected cases, these analyses were supplemented with those performed along the ND/transverse-direction (TD) and/or ED/TD sections. Meso-scale analyses of the microstructural and textural evolutions were performed using a 3D microscope (Quanta, FEI, USA) equipped with an EBSD facility. TSL OIM Analysis 8 software (EDAX Inc., USA) was used to acquire and index diffraction patterns. Orientation maps were measured in beam-scanning mode using step sizes ranging from 50 to 100 nm. The standard procedure for sample preparation was used for microstructure and texture investigations. Deformed single crystalline samples were mechanically ground, polished with 1 μm and 1/4 μm diamond suspension, and then with colloidal silica and electropolished. Nano-scale analyses were made using a TEM microscope (Technai Super TWIN G2 FEG, FEI, USA; 200 kV). Thin foils were cut from the bulk samples, mostly along the ND/ED plane, using a wire saw and then subjected to mechanical polishing and electropolishing. Finally, the samples were prepared using a twin-jet electropolishing method using TenuPol-5 in a standard D2 solution (Struers, Denmark) at 5 °C and 10–12 V. Each thin foil had a large transparent area of 10^3 – $10^4 \mu\text{m}^2$.

Table 1
The initial orientations of the single crystals measured using X-ray diffraction.

Measured orientation		Ideal orientation	
Miller indices	Euler angles	Miller indices	Euler angles
(112)[11–1]	271.5, 34.6, 42.2	C(112)[11–1]	270, 35, 45
(346)[63–5]	291.0, 39.8, 36.9	S(123)[63–4]	301, 37, 27
(110)[3–3–4]	316.7, 90.0, 45.0	Br(110)[1–1–2]	305, 90, 45
(110)[001]	91.0, 90.0, 44.5	G(110)[001]	90, 90, 45

3. Results

3.1. Microstructure evolution at the macro-scale. Shape changes of the sample

Fig. 3 shows the optical micrographs documenting the shape change of all single crystals after high-speed deformation up to 60 %. These macro-observations were performed in the ND/ED (longitudinal) and ED/TD (compression) sections. A closer look at the shape changes shows that they are much less pronounced (if they occur at all) than for specimens deformed under semi-static conditions [13,14]. The most probable reason is the effect of structural inhomogeneities on the ability to accommodate deformation. In samples with nominally stable orientations, i.e., G and Br, this manifests itself in the formation of deformation bands (Fig. 3a, b), while in unstable crystallites with C and S orientations by the formation of various twin families and shear banding (Fig. 3c, d). As expected, a Br(110)[1–1–2] oriented crystallite undergoes macroscopic $\varepsilon_{(ED/ND)}$ shear. This weakly marked shape distortion observed on the ED/TD plane is also good in line with the standard crystal plasticity which predicts this shear as a consequence of slip on the (111)[10–1] and (11–1)[011] systems operating on planes inclined at 30° to ED. For the ideal Br orientation, these two systems have equal amplitudes and their lattice rotations balance out to zero rotation up to large strains [28]. In the case of a C-oriented crystallite, a slightly marked shape distortion is observed on the ND/ED plane due to $\varepsilon_{(ED/TD)}$ shears. This shear may result from massive twinning, predominantly on the (111) plane, and then intense shear banding [13]. Both shape distortions are typically observed during the quasi-static deformation of C and Br-oriented crystallites. The other two orientations (G and S) do not show significant shape changes. Goss crystal orientation under plane strain compression exhibits an orthotropic symmetry, so no significant shape distortions are expected. Such behaviour is determined by the activity of two pairs of highly favoured (co-planar) slip systems operating on the (111) and (11–1) planes located parallel to TD and inclined at 35° to ED. However, it can be observed even at this macro-scale, that G-oriented crystallites tend to undergo a macroscopic flow along arms inclined at about 45° to ED. In the S-oriented single crystals, significant macroscopic $\varepsilon_{(ED/ND)}$ and $\varepsilon_{(ED/TD)}$ shears occur after semi-static deformation. These shears are not observed in dynamically deformed crystals, despite the intense formation of twins and shear bands.

3.2. Global orientation changes

Schmid factor analysis has been proven to be the most efficient method for predicting slip and twinning system activity [30–32]. In this work, the Schmid factors were calculated for all slip and twinning systems of {111}<011> and {111}<112>-type, respectively. By carrying out this analysis, it was clarified whether the dominant slip and twinning systems have always the largest Schmid factors. Theoretical predictions of the deformation behaviour and the evolution of the initial grain orientation are strongly dependent on the applied state of stress or deformation. In Tucker's approach [30] the state of stress for the channel-die compression is described as a superposition of uniaxial tension and compression, for which: $\sigma_{ED/ED} = -\sigma_{ND/ND}$, and $\sigma_{TD/TD} = \sigma_{RD/TD} = \sigma_{TD/ND} = \sigma_{ED/ND} = 0$. According to this assumption, the value of Schmid's factor for all possible slip systems and all considered orientations was calculated using equation: $w_s = \tau/\sigma = \cos\alpha_c \cos\beta_c - \cos\alpha_e \cos\beta_e$, where: α_c, β_c - angles between the vector normal to the slip plane and the axes of compression and tension, and α_e, β_e - angles between the vector of the slip direction and the axes of compression and tension.

The crystal orientation rotations were analysed in the {111} pole figures measured on the longitudinal (ND-ED) (Fig. 4a-d) and compression (ED-TD) (Fig. 4e-l) planes. For all analysed crystallites, the pole figures show a wide orientation spread, typical of unstable orientations. A strong 'scattering' of initial orientation applies not only to unstable crystallites with C and S orientations, but also crystallites with

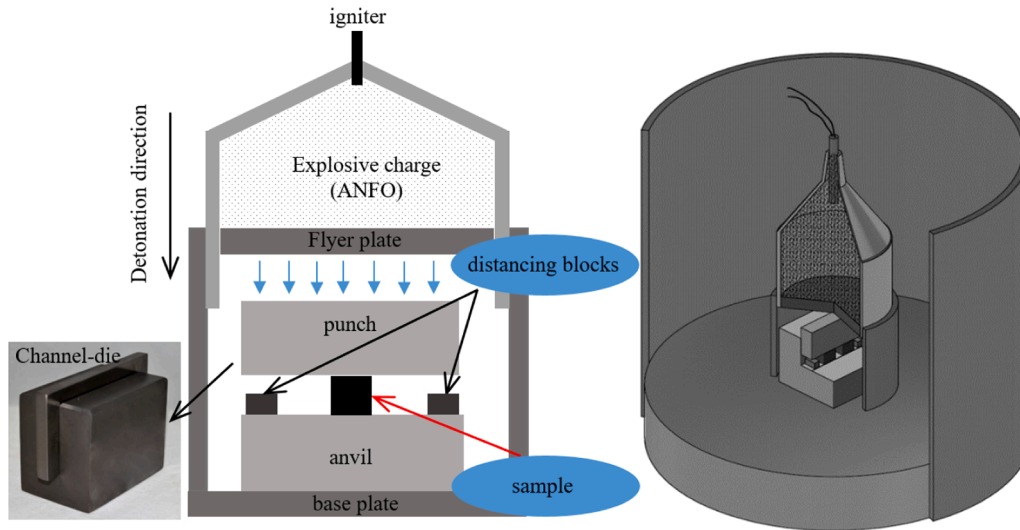


Fig. 2. Schematic presentation of the experimental set up for deformation of single crystalline samples in a channel-die using explosive energy.

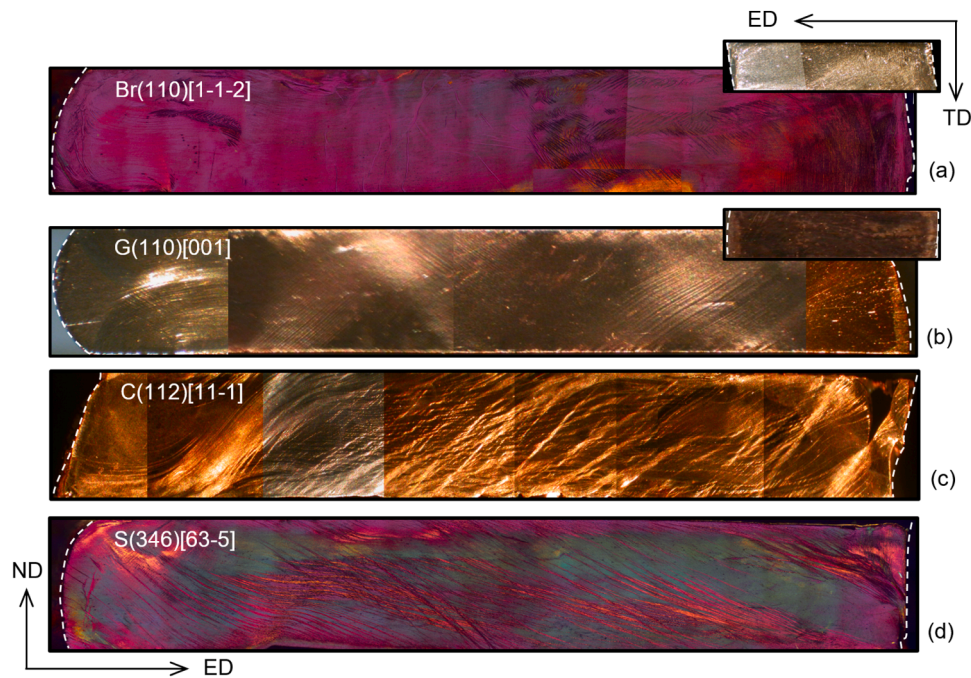


Fig. 3. Optical microscopy micrographs showing the shape changes and microstructures in longitudinal sections, developed in: (a) Br, (b) G, (c) C and (d) S-oriented samples continuously deformed up to 60 %. The insets in Figures (a) and (b) show the shape changes observed in the ED-TD plane characteristic of the Br-oriented crystallite and the other three orientations, respectively.

Br and G orientations, which are nominally stable during plane state deformation under quasi-static loading. A detailed analysis confirms that regardless of the crystallite orientation, texture scattering is mainly related to the nucleation of deformation twins. However, a strong variation in the intensity of twinning is observed, depending on the crystal orientation. This strong variability concerns the number of twin families, i.e. the occurrence of twins on different $\{111\}$ planes, as well as the twin volume fraction and order of successive generations of twins, which implies twins inside previously twinned areas.

As shown in the $\{111\}$ pole figure in Fig. 4a and e, the dominant scattering tendency of the initially $Br(110)[1-1-2]$ -oriented crystals is associated with the rotation of the crystal lattice by an angle of $\sim 40^\circ$ around the axes located near the TD (in the ideal Br orientation, the TD coincides with the $\langle 111 \rangle$ -type crystallographic direction). A strong

tendency for twin orientations to appear on all four of the $\{111\}$ planes is also observed, including those associated with the twinning plane normal, situated parallel to the TD.

In the $G(110)[001]$ -oriented crystallites, texture measurements (Fig. 4b, f) reveal a large variation in the texture image depending on the location of the analysis. Near areas characterised by strong texture stability, areas of high orientation variability, mainly associated with the formation of twin orientation(s), can be identified. The $\{111\}$ pole figure measured along the longitudinal section (Fig. 4f) reveals a dominant twinning system on the (111) plane. However, traces of the twin transformations on the (11-1) plane can also be observed in the scattering limit of the $\langle 111 \rangle$ poles. This indicates the occurrence of twin transformations on the two planes on which the most favoured slip systems operate. In both cases of twinning, that is, on the (111) and

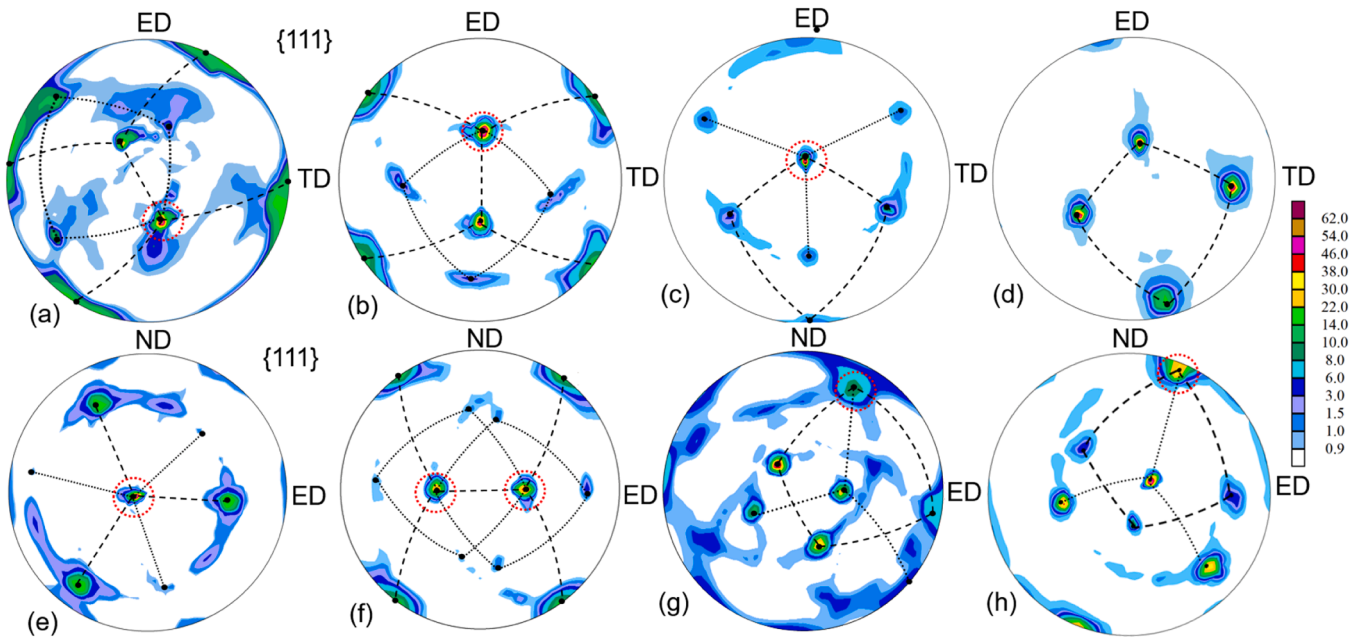


Fig. 4. The $\{111\}$ pole figures showing the scattering of initial orientations of (a) and (e) Br, (b) and (f) G, (c) and (g) C, and (d) and (h) S-oriented single crystals after 60 % deformation. The red dotted circles indicate the normal to the twinning planes. X-ray diffraction on (a–d) ED-TD and (e–h) ND-ED planes. Identified twinning plane normals are marked by red dashed circles.

(11–1) planes, a clear tendency for additional scattering of the texture image via rotation around the axes close to the TD is observed.

In crystallites with initial $C(112)[11-1]$ and $S(346)[63-5]$ orientations, the most important textural changes are related to the twin transformation on the (111) plane (Fig. 4d, h). Other twinning systems can also be identified, but at much less intensity. In both orientations, the scattering of the texture due to twinning is also associated with an additional crystal lattice rotation around the axes located in the vicinity

of the TD.

3.3. Identification of the twinning systems

The light microscopy microstructure observed on the ND/ED section of the crystallite with an initial Br (110) $[1-1-2]$ orientation shows only small areas in which twins are formed during deformation, whereas the rest of the section shows a homogeneous, non-twinned microstructure.

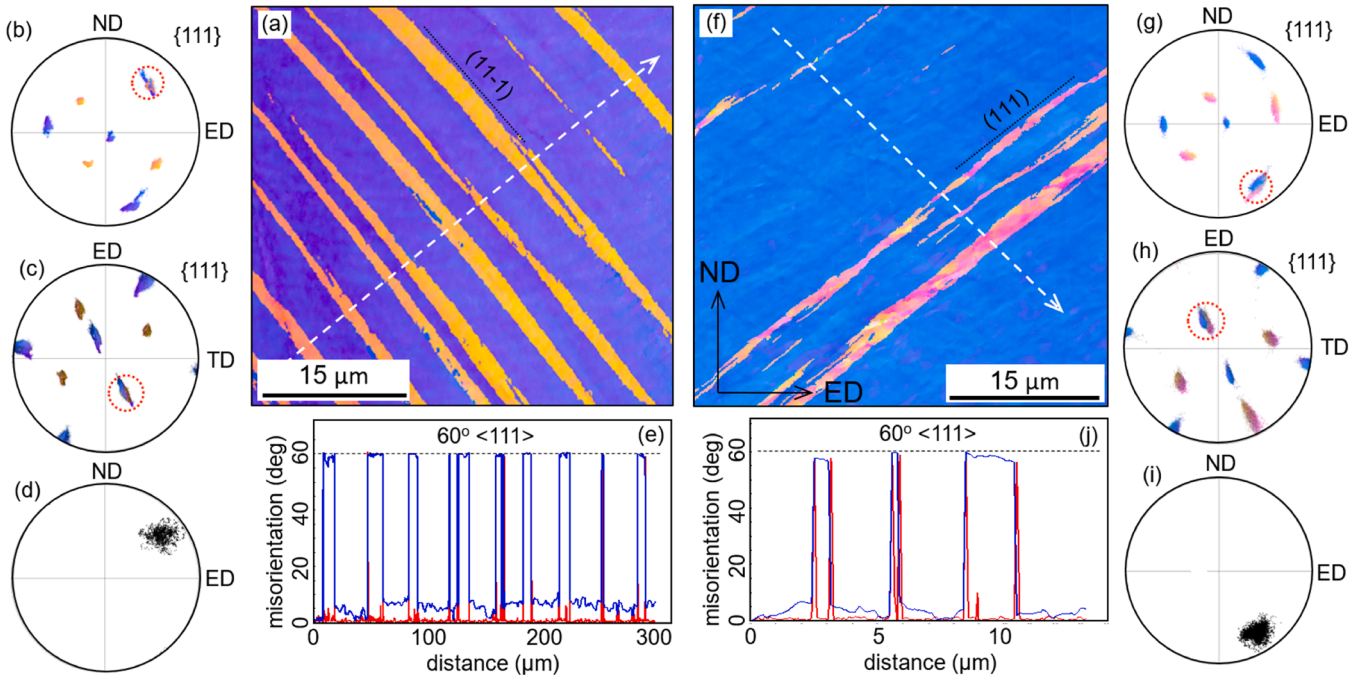


Fig. 5. Deformation twinning in a Br-oriented single crystal. (a, f) Orientation maps showing the thin platelets of twins that result from transformation on the (11–1) and (111) planes and corresponding the $\{111\}$ pole figures on the ND/ED (b, g) and ED/TD (c, h) planes as well as (d, i) the disorientation axes distribution. (e, j) Disorientation line scans across the twin platelets along the dashed lines presented in (a) and (f). SEM/EBSD measurements in the ND-ED section. Step size of 60 nm. Twinning planes are marked by red dashed circles.

SEM/EBSD measurements confirm that, in the homogeneously deformed areas, the microstructure observed in longitudinal section is dominated by two families of microbands, typically with one of them dominant in a given area. The disorientation angles between adjacent points of the orientation map ('pixel to pixel') only occasionally exceed 5° . Consequently, the $\{111\}$ pole figure shows the small scattering of the initial orientation, with the disorientation axes clustered near the TD. This behaviour is similar to that observed in Br-oriented single crystals of medium-to-high SFE metals plane strain deformed under quasi-static loading [12,14,28,29]. The microbands are inclined at an angle of $(+/-) 30^\circ$ with respect to ED. This indicates that the slip traces coincide with the (111) and $(1\bar{1}\bar{1})$ planes, on which the most favoured slip systems operate.

An unusual behaviour that is not observed in Cu deformed under quasi-static loading [12,13,29] is the occurrence of deformation twins. They occur rarely, mainly as very thin platelets with traces of twinning planes located close to the traces of the (111) and $(1\bar{1}\bar{1})$ planes, which are most favoured for dislocation motion (Fig. 5). SEM/EBSD and X-ray diffraction measurements indicate that deformation twins also appear on the $(1\bar{1}\bar{1})$ and $(1\bar{1}1)$ planes, i.e. planes located perpendicular to the compression plane (Fig. 6), which are unfavourable for dislocation slip (and twinning). The normals to the $(1\bar{1}\bar{1})$ and $(1\bar{1}1)$ planes are located in proximity to the ED and TD, respectively. The orientation map shown in Fig. 6a was obtained from an area of thin separate needles with

traces of the twinning planes (observed on the ND/ED section) that deviate by $\sim 10^\circ$ from the ND. In all cases, the disorientation line scans show that needles are at a $60^\circ \langle 111 \rangle$ angle to the surrounding matrix (Figs. 5 and 6). In the SEM/EBSD study, a significant but unsuccessful effort was made to reveal twin transformation on the $(1\bar{1}\bar{1})$ plane. In this case, the normal of the $(1\bar{1}\bar{1})$ plane is parallel to the TD, therefore, the traces of the twinning planes could not be identified on the ND/ED plane. However, thin needles of these twins, were observable on TEM thin foils cut along the compression plane (Fig. 5d).

In crystallites with $G(110)[001]$ orientation, X-ray-diffraction measurements (Fig. 4) reveal two families of twins that appear owing to transformations on the (111) and $(1\bar{1}\bar{1})$ planes, which are the most favoured for dislocation motion. However, the areas of intensive deformation twinning are distributed nonhomogeneously over the ND/ED section. The twin transformation is most pronounced in areas near the free ends of the sample, where a strong rotation of the crystal lattice occurred (Fig. 7a–c). This confirms earlier observations that the probability and intensity of twin nucleation are orientation-dependent, as suggested by Cao et al. [33]. The orientation map measured near one of the free ends of the sample (Fig. 7a) shows that the microstructure is dominated by elongated areas disoriented from each other at angles ranging between 30° and 40° (Fig. 7d), mostly as a result of rotation around the $[1\bar{1}0] \parallel$ TD axis. A more detailed analysis shows that deformation twinning occurs in areas with an orientation close to (552)

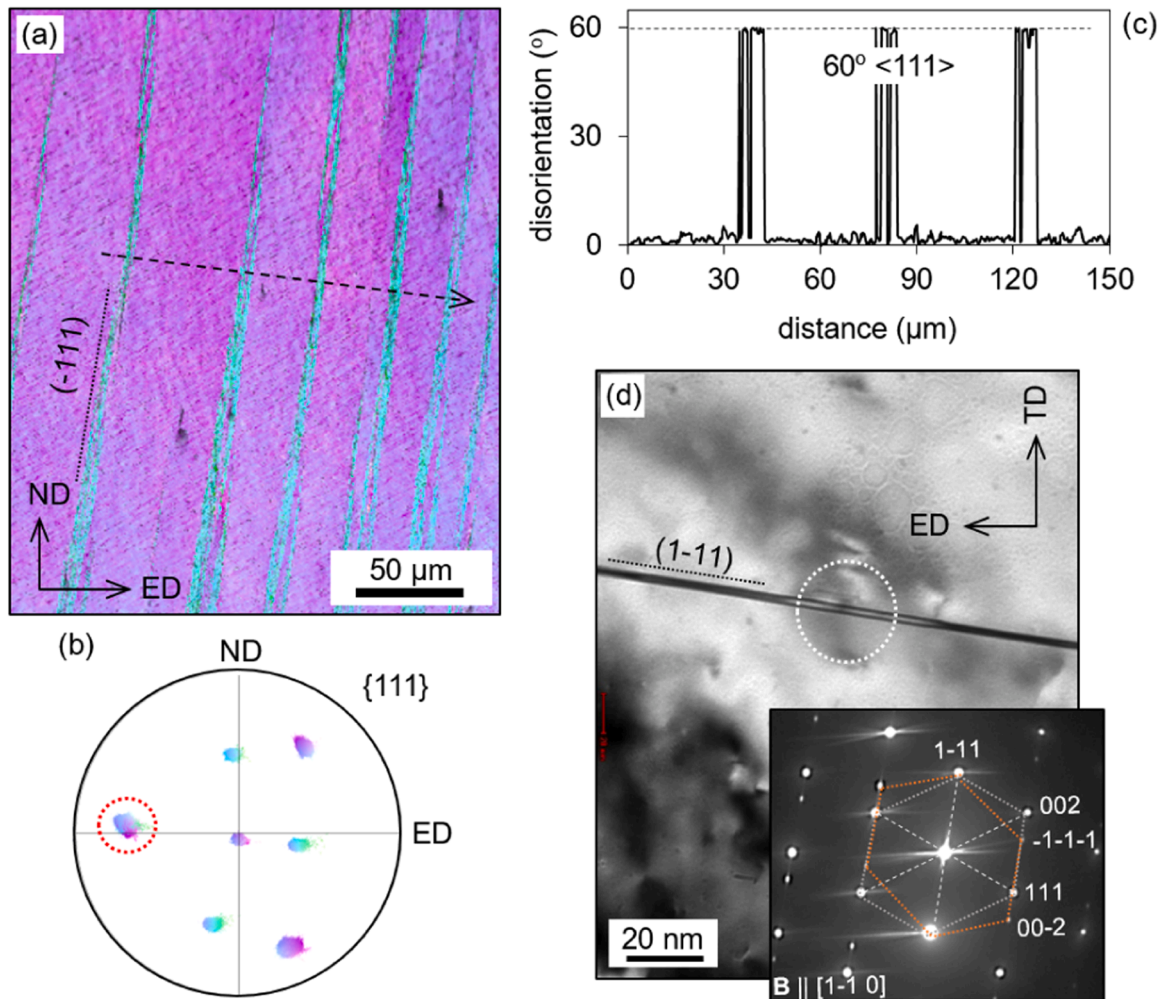


Fig. 6. Deformation twins in the Br-oriented single crystal observed on the $\{111\}$ planes less favourable for dislocation slip. (a) Orientation map showing the thin platelets of twins on the twinning plane with the normal situated near the ED and corresponding (b) the $\{111\}$ pole figure as well as (c) the disorientation line scan across the twin platelets. (d) TEM microstructure showing the thin needles of twins with a normal to the twinning plane near the TD. SEM/EBSD measurements in the ND/ED section with step size of 60 nm and TEM bright-field imaging.

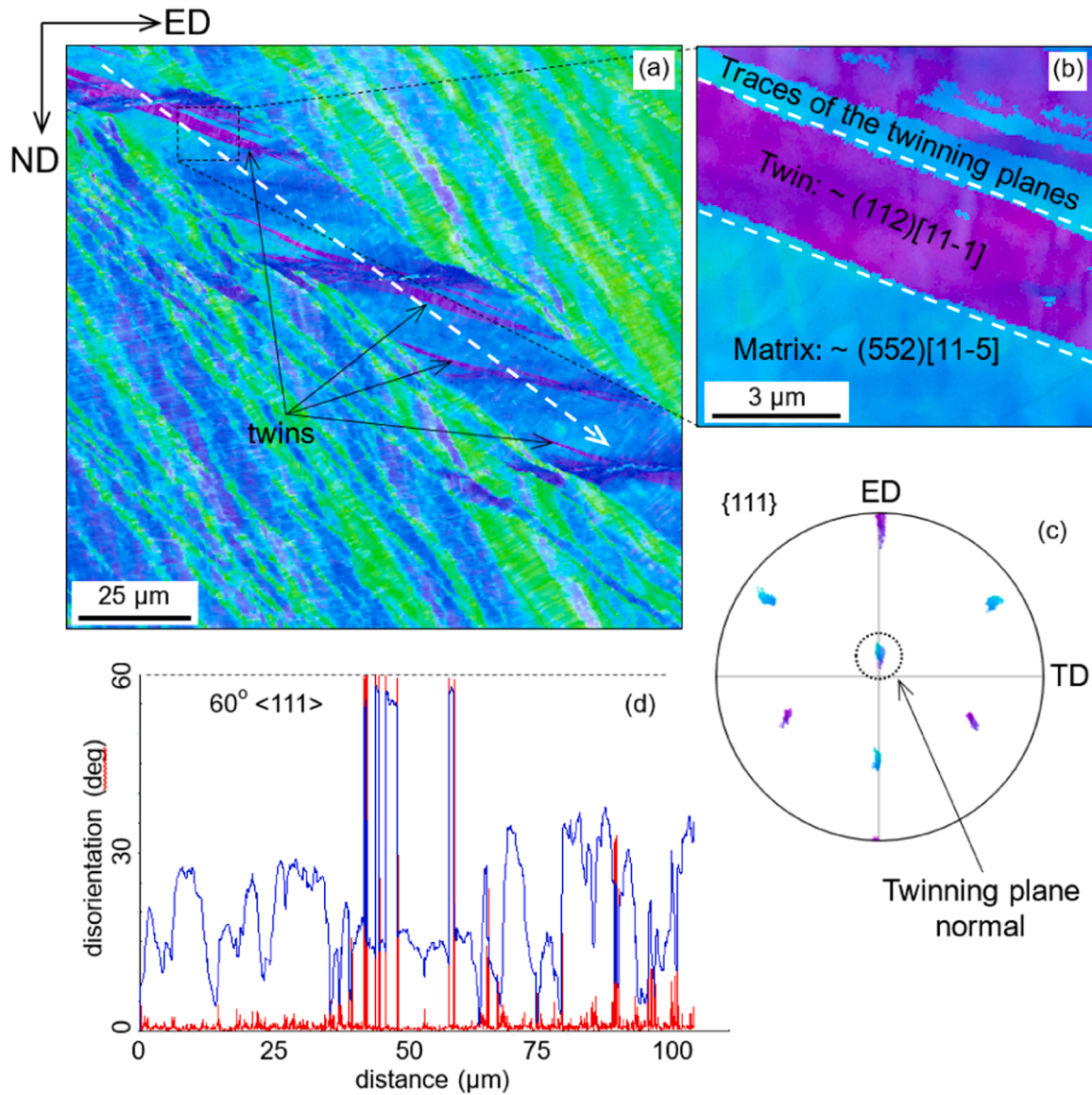


Fig. 7. Deformation twinning in G-oriented single crystal. (a) Orientation map showing the thin platelets of twins observed on one of the most favourable planes for dislocation slip. (b) Details from (a) and (c) corresponding $\{111\}$ pole figure showing the orientations of the primary matrix and twins. (d) The disorientation angle distribution along the white dashed line presented in (a). SEM/EBSD measurements in the ND/ED section. Step size of 50 nm.

$[-1-15]$, that is $\sim 16^\circ$ from the $(110)[001]$ orientation. This transformation leads to the formation of needles of twins with an orientation close to $(112)[11-1]$ (Fig. 7a,b).

The microstructural changes occurring in the crystallites with $C(112)[11-1]$ and $S(346)[63-5]$ orientations are similar. The most characteristic feature is the occurrence of intense deformation twinning (Fig. 8a). After deformation up to a logarithmic strain of 0.9, the twinned regions occupy almost the entire sample volumes. The first dominant family of deformation twins appears as a result of a transformation on the (111) plane, and this variant of twins is marked as T^1 . In most of the analysed areas, the traces of the twinning planes are inclined at an angle of $\sim 20^\circ$ to the ED. This indicates that the compact clusters comprise platelets with a $C^{T1}(552)[-1-15]$ orientation divided by that with a $C(112)[11-1]$ (primary matrix) orientation. However, the areas near both free surfaces usually exhibit a stronger $(+)$ TD rotation of the crystal lattice, which determines the increase in the inclination of the (111) plane relative to the compression plane. This is illustrated in Figs. 8b and c, which show the microstructure of compact clusters of T-M layers with twinning planes inclined at angles of 27° – 30° relative to the ED. This

results in the appearance of orientations close to $(001)[110]^M$ (for the matrix) and $(110)[001]^{T1}$ (for the twins).

The microstructure and texture analyses were conducted after a given degree of strain, therefore, the beginning of the intense twin transformation on the (111) plane is unclear. This may be from the initial matrix, that is, the $C(112)[11-1]$ orientation or the orientation of the matrix already subjected to $(+)$ TD rotation in the 'direction' of the $(001)[110]$ orientation. Alternatively, the twinning on the (111) plane can be regarded as a 'continuous' process occurring for the entire range of $(112)[11-1]$ – $(001)[110]$ matrix orientations. The rotation of the crystal lattice from the initial $C(112)[11-1]$ orientation towards the $(001)[110]$ orientation increases the tendency to slip (and twinning) intensely on the $(-1-11)$ plane. Twinning can also occur inside twin lamellae (T^1) with a near $G(110)[001]$ orientation (that was, previously formed on the (111) plane), as shown on the $\{111\}$ pole figures in Fig. 8. This implies the appearance of second twin generation with orientations symmetrically equivalent to that related to the position of the original matrix, that is, $(114)[-2-1-1]^{T1,1}$ (Fig. 8c). Traces of the twinning planes in the microstructures observed in Figs. 8a and c are

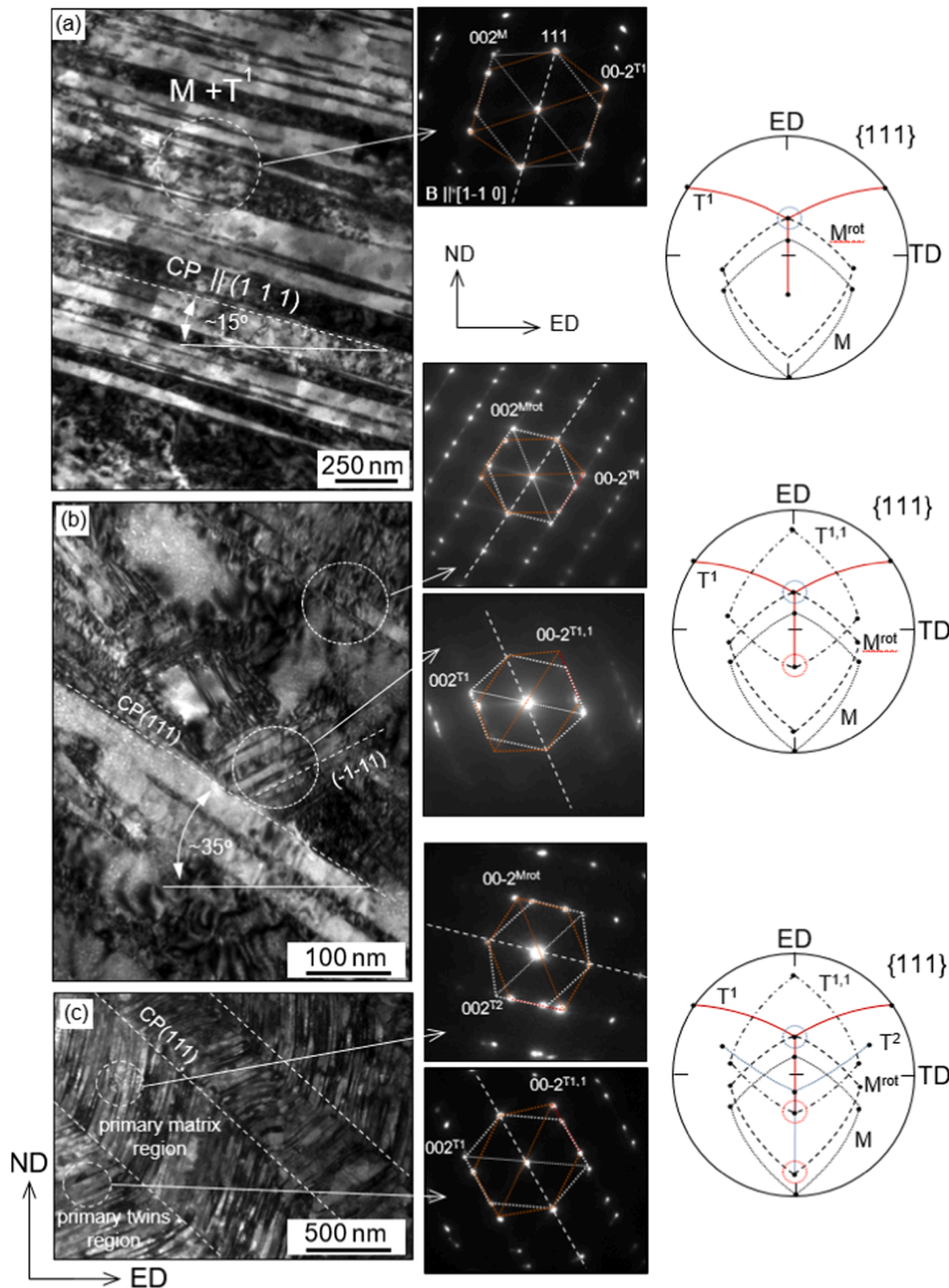


Fig. 8. TEM bright-field images documenting the twinning from the primary matrix in C-oriented single crystal on: (a) (111) plane, and (b) (111) and $(-1-11)$ planes. (c) Twinning on $(-1-11)$ plane in areas of re-oriented primary matrix and re-oriented primary twin regions (twinning from twins). The {111} pole figures show the orientations identified by selected area diffraction patterns from selected areas marked by dashed circles. Twinning planes are marked by red and blue dotted circles.

schematically presented in Fig. 9.

Twinning on different {111} planes in the matrix leads to interactions between the twin families. This is illustrated in the orientation map shown in Fig. 10. The twin lamellae of the T^1 orientation that appear as a result of the twinning of the initial matrix on the (111) plane are intersected by the twin platelets of T^2 orientation, resulting from the twin transformation of the initial matrix on the $(-1-11)$ plane. The intersection of these two sets of deformation twin families implies a rigid-body rotation of the first-formed twins (in the interaction region).

3.4. Deformation banding in Br- and G-oriented single crystals

During impact loading, both nominally stable orientations (in terms of quasi-static deformation) lose their stability relatively quickly and this process is controlled by the twins and deformation bands formation. In both the G- and Br-orientations, the deformation bands are bounded by regions of the structure with orientations close to the initial orientation.

Fig. 11a and b depict orientation maps showing the formation of narrow deformation bands against the background of a less deformed

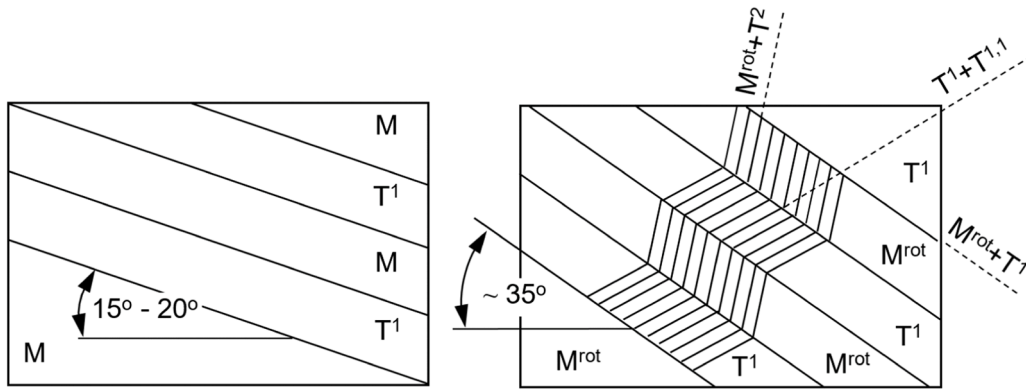


Fig. 9. Schematic presentation of the twinning plane traces situation in the ND/ED section corresponding to Fig. 7c. The traces of the twinning planes: (a) in areas with $\sim(112)[11\bar{1}]$ initial matrix orientation, and (b) in areas of a re-oriented matrix with $\sim(114)[22\bar{1}]$ orientation. $M + T^1$ primary twinning in initial matrix, $M^{\text{rot}} + T^1$ primary twinning in re-oriented matrix on (111) plane, $M^{\text{rot}} + T^2$ primary twinning in re-oriented matrix on $(\bar{1}1\bar{1})$ plane, $T^1 + T^{1,1}$ second-generation twinning in re-oriented twin platelet.

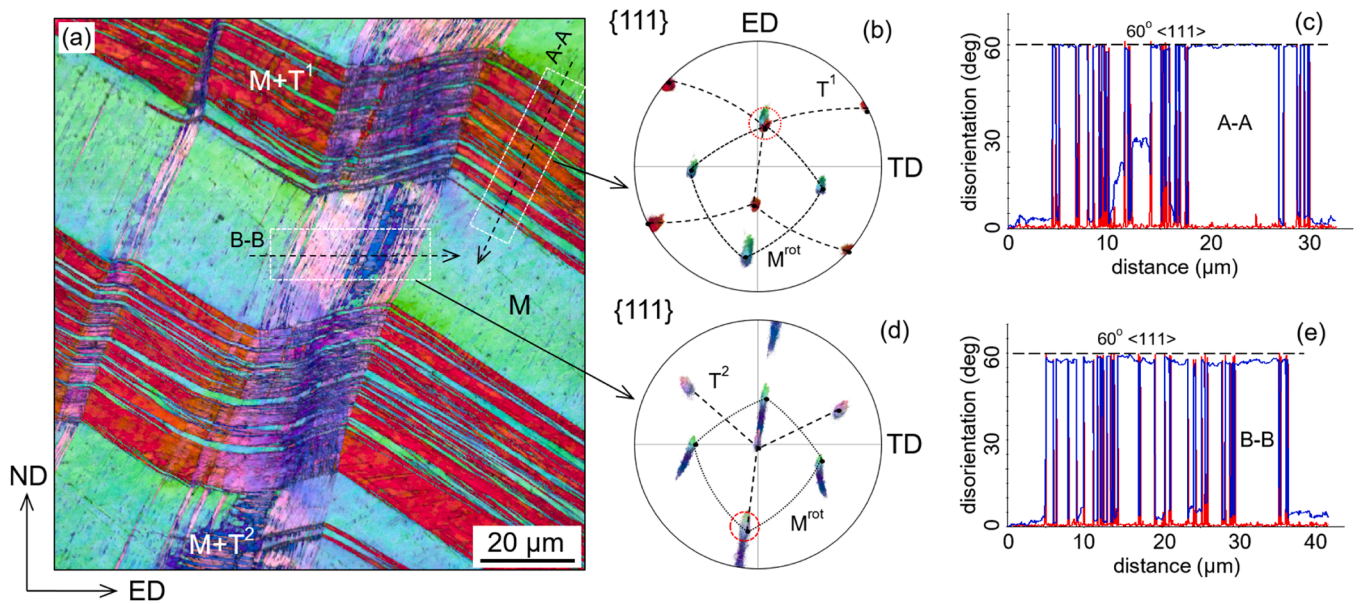


Fig. 10. (a) Orientation map showing interaction between two families of primary twins in re-oriented matrix of C-oriented single crystal. (b, d) The $\{111\}$ pole figures showing orientations in areas marked by dashed boxes. (c, e) Disorientation line scans across the compact clusters of deformation twins. SEM/EBSD measurements in ND/ED section. Step size of 50 nm.

matrix in a $Br(110)[1\bar{1}2]$ -oriented single crystal. The orientations from the band region and the neighbouring less deformed matrix are shown in the $\{111\}$ pole figures in Fig. 11c and d, which document the strong scattering of the initial orientation in the band region. The high-angle ($> 40^\circ$) boundaries in the microstructure are related to the interfaces between the band and neighbouring less-deformed matrix, whereas the low-angle ($< 15^\circ$) boundaries tend to separate the cells within the matrix and band. Fig. 11e shows that the disorientation angle across the band-matrix interface is $\sim 35^\circ$, whereas the disorientation axes in the disorientation relation between the band area and surrounding matrix are grouped near the TD ($\parallel \langle 112 \rangle$) (Fig. 11f).

Analyses conducted using the SEM/EBSD system in the $G(110)[001]$ -oriented crystal reveal elongated areas with orientations different from that of the neighbouring less deformed matrix, as presented in Fig. 12a and b. The disorientation between the points inside the band area and in the adjacent matrix does not exceed a few degrees, whereas the disorientation across the boundary between the matrix and band interior (Fig. 12c) reaches 50° with the disorientation axes grouped near the TD ($\parallel \langle 011 \rangle$) (Fig. 12f). This leads to a strong texture scattering, as

illustrated in the $\{111\}$ pole figure in Fig. 12d, e.

The stability of the Br and G orientations during semi-static deformation is determined by the activity of the two pairs of co-planar slips on the (111) and $(11\bar{1})$ planes [28,29]. In the case of the G orientation, all four slip systems are equally favourable, whereas in the case of the Br orientation, each pair comprises one more and one less favoured slip system. In the semi-static deformation of G- and Br-oriented single crystals, the intensification of slips in one of the favourable pairs of co-planar slip systems implies crystal lattice rotation, leading to a decrease in the Schmid factor of these systems, and therefore, the promotion of slips on the symmetrically situated plane.

Clearly, the deformation behaviours during impact loading are different. Detailed microstructure and texture analyses show that in the Br orientation, the disorientation axes correspond well to the $[-12\bar{1}]$ direction. Assuming that slip on the $\{111\}$ plane in the $\langle 110 \rangle$ direction implies the rotation of the crystal lattice around the $\langle 112 \rangle$ axis, the $[-12\bar{1}]$ rotation axis can be directly related to the dominance in a given area of a single slip system operating on the (111) plane along the $[10\bar{1}]$ direction and indicates that the band area is dominated by one of

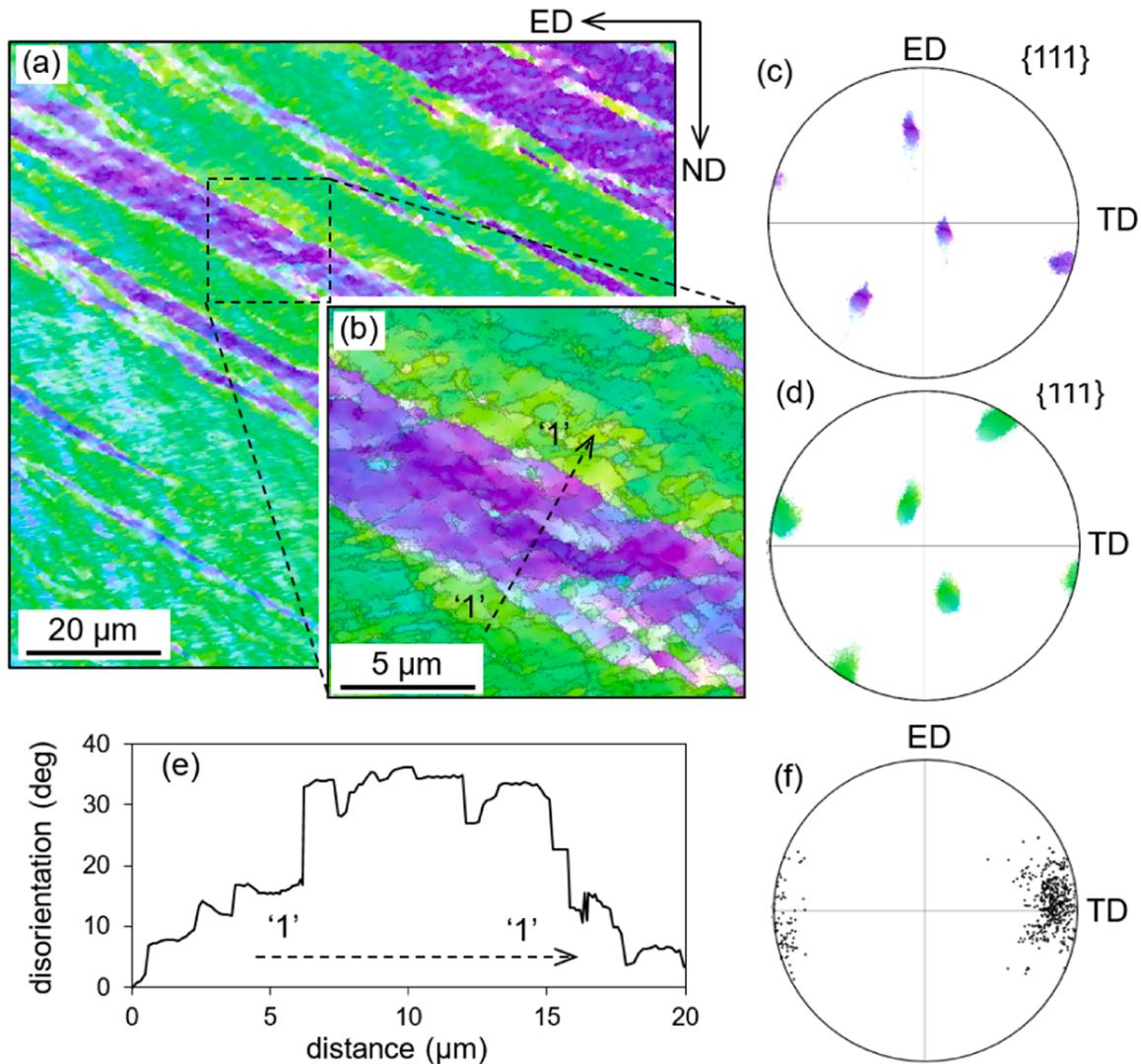


Fig. 11. (a) Orientation map showing the formation of deformation bands in Br-oriented single crystal. (b) Details of the orientation map presented in (a). The $\{111\}$ pole figures showing the difference in crystal lattice orientation between the (c) band interior and (d) matrix. (e) Disorientation line scan across the deformation band. (f) The disorientation axes distribution corresponding to the map presented in (b). SEM/EBSD measurements in ND/ED section. Step size of 50 nm.

the two favoured slip systems operating on the $(11\bar{1})$ plane (Fig. 13a). As for G orientation, the transformation of the matrix orientation into the orientation of the band interior can be described by a rotation around the $[-110]$ axis, which is parallel to the TD (Fig. 13b). Therefore, we deduce that the bands are formed because of the dominance of one pair of co-planar slip systems operating on the $(11\bar{1})$ plane. The above analysis leads to the conclusion that dynamic deformation causes the dominance of one or a pair of slip system(s) that are fixed and deepened owing to the lack of time for the symmetrically situated slip systems (or pairs of slip systems) to react.

3.5. Shear banding in C and S orientations

As expected, SBs intensively form in the $(112)[11\bar{1}]$ and $(346)[63\bar{5}]$ -oriented single crystals. The mechanism of SBs formation during impact loading is analogous to that observed in *fcc* metals with low-SFE after quasi-static deformation at RT [13,14,34] or with a medium SFE, but deformed at 77 K [13,14,35–37] or lower [38]. The beginning stages of SB formation are related to the local deflection of the twin-matrix layers within narrow areas, which increases the inclination of the twinning plane traces relative to the ED (Fig. 14a). This leads to the formation of kink-type bands, which are precursors to SBs [13,36,37].

This process is inherently associated with (opposite) the global tendency of rigid rotation of the T-M layers to locate the twinning planes outside the band parallel to the compression plane [35–37]. This increases the intensity of the $(111)[11\bar{2}]^{\text{Matrix}}$ and $(111)[\bar{1}\bar{1}2]^{\text{Twin}}$ texture components [34,38].

In the shear band area the crystal lattice of the twin platelets rotate towards the $G(110)[001]^{\text{Twin}}$ orientations, whereas the crystal lattice of the preserved platelets of the original matrix inside the SB rotates towards the $(001)[110]^{\text{Matrix}}$ orientation. This rotation tendency increases the Schmid factor of the slip systems operating on the (111) plane and increases the propensity to twin transformation of the remaining untwinned matrix (Fig. 14b). This leads to the complete disappearance of the orientations related to original matrix and an increase in the intensity near the $G(110)[001]$ orientations in the band interior (Fig. 14c). Simultaneously, the rotation of the crystal lattice of the twin platelets to the G orientation opens the possibility for twinning on the $(11\bar{1})$ plane of the previously twinned areas and the appearance of orientations close to the position symmetrically equivalent to the original matrix.

The mechanisms leading to the texture formation in the band interior are presented very clearly in the SEM/EBSD orientation maps in Figs. 15 and 16. Twinning on the (111) plane transforms the $C(112)[11\bar{1}]$ orientation into the $C^T(552)[\bar{1}\bar{1}5]$ orientation, according to

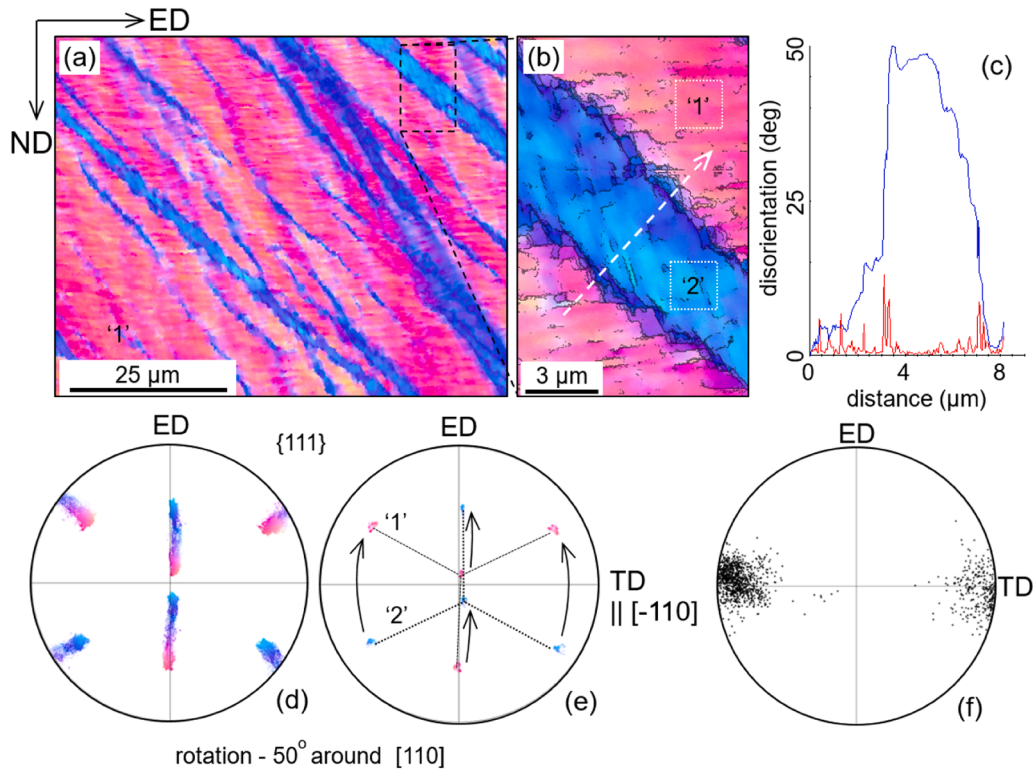


Fig. 12. (a) Orientation map showing deformation bands in G-oriented single crystal. (b) Details from (a). (c) The disorientation line scan across the deformation band. (d) The $\{111\}$ pole figure from the orientation map presented in (a), and (e) the $\{111\}$ pole figure showing the orientations in the band interior and neighbouring matrix. (f) The disorientation axes distribution corresponding the map presented in (b). SEM/EBSD measurements in ND/ED section. Step size of 50 nm.

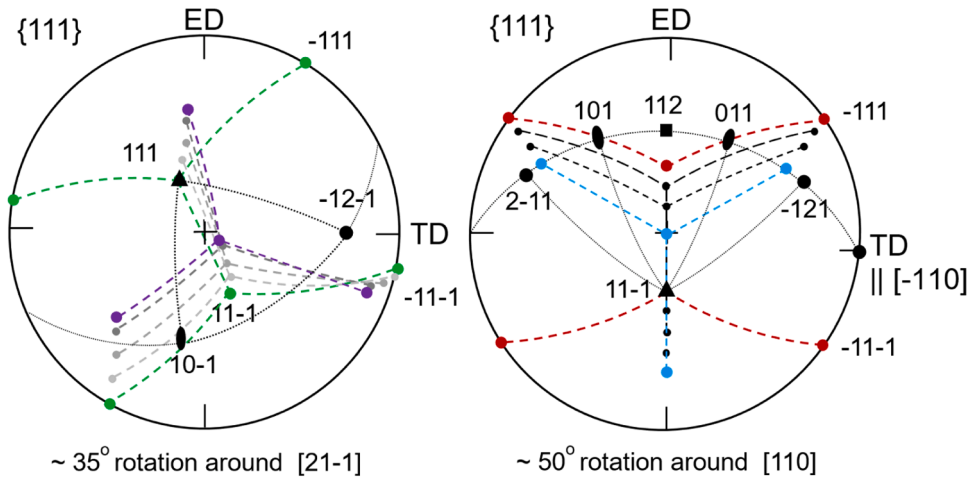


Fig. 13. Schematic presentation of the orientation changes due to deformation band formation in single crystals of (a) Br and (b) G orientation.

Wassermann's approach [39]. The orientation identified in the SB region indicates the lattice rotation of the remaining non-twinned matrix from its original $C(112)[11-1]$ orientation to an orientation close to $-(114)[22-1]$ (M^{rot}). This increases the tendency of twinning on the (111) plane and is responsible for the fading areas related to the original matrix in the band interior. This in turn, leads to the dominance of orientations in the vicinity of the $G(110)[001]$ orientation, as observed in C-oriented Cu single crystals deformed at 77 K during quasi-static loading in the channel-die [13,35–37]. The texture of the SB interior is always dominated by components related to the position of the dominant family of twins, that is, the orientations from the nearest neighbourhood of the G orientation (Fig. 16). However, the areas

adhering to the band core, forming the transition region [32,40] in which the primary matrix still exists, are described by a set of positions from $(114)[22-1]$ to $(111)[11-2]$ for the matrix lamellae and from $(110)[001]$ to $(111)[-1-12]$ for the twins (Fig. 16).

Analogous changes due to the formation of twins and SBs are observed in crystals with the S orientation (Fig. 17). However, regarding to S-oriented crystals, the textures of the SB interior and areas outside the band are rotated by several degrees around the axes close to the ND owing to the initial crystal orientation.

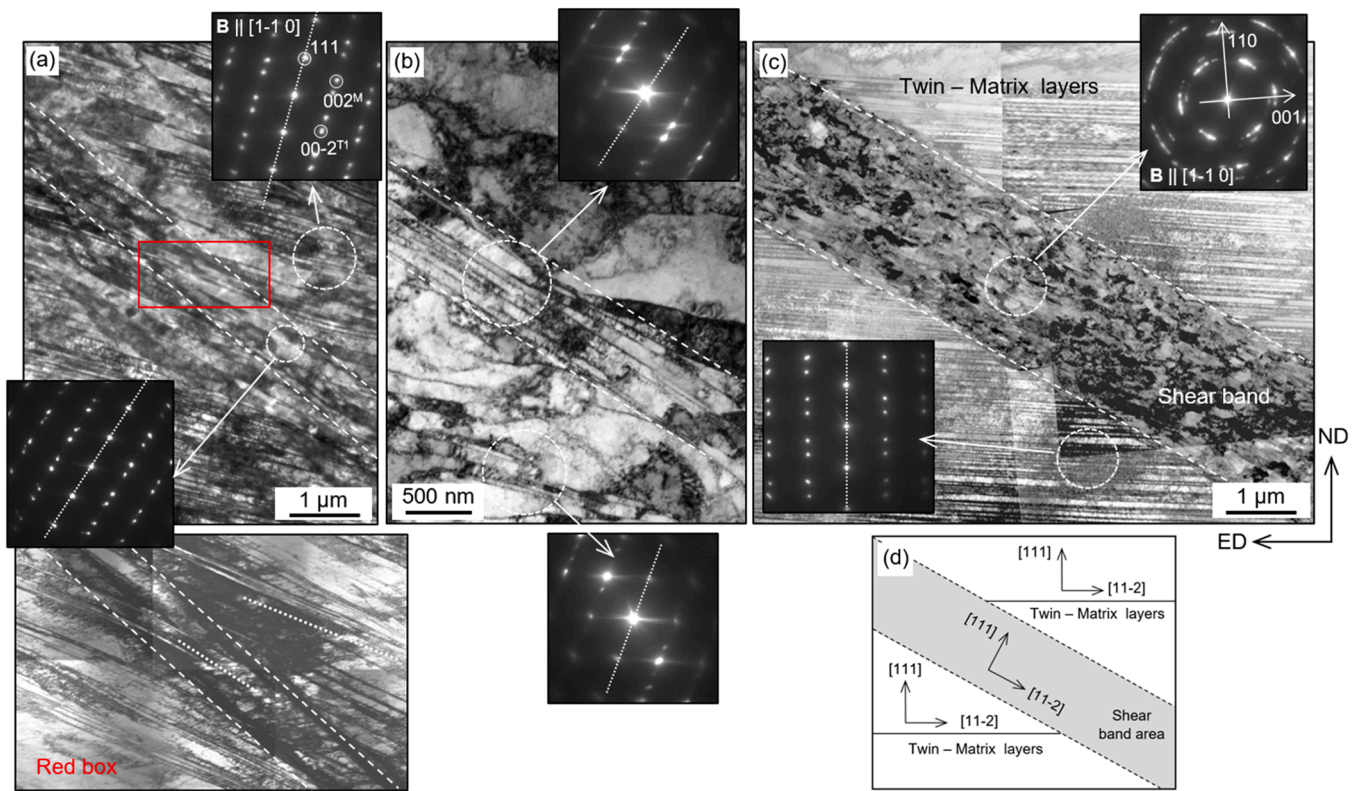


Fig. 14. TEM bright-field images showing important stages of SBs formation in C-oriented single crystal. (a) Formation of kink-type bands inside compact clusters of T-M layers, (b) twinning in re-oriented matrix inside the band region, (c) well-developed SB penetrating the background structure of compact clusters of twins situated parallel to the compression plane. Selected area diffraction patterns show the orientations in areas marked by dashed circles.

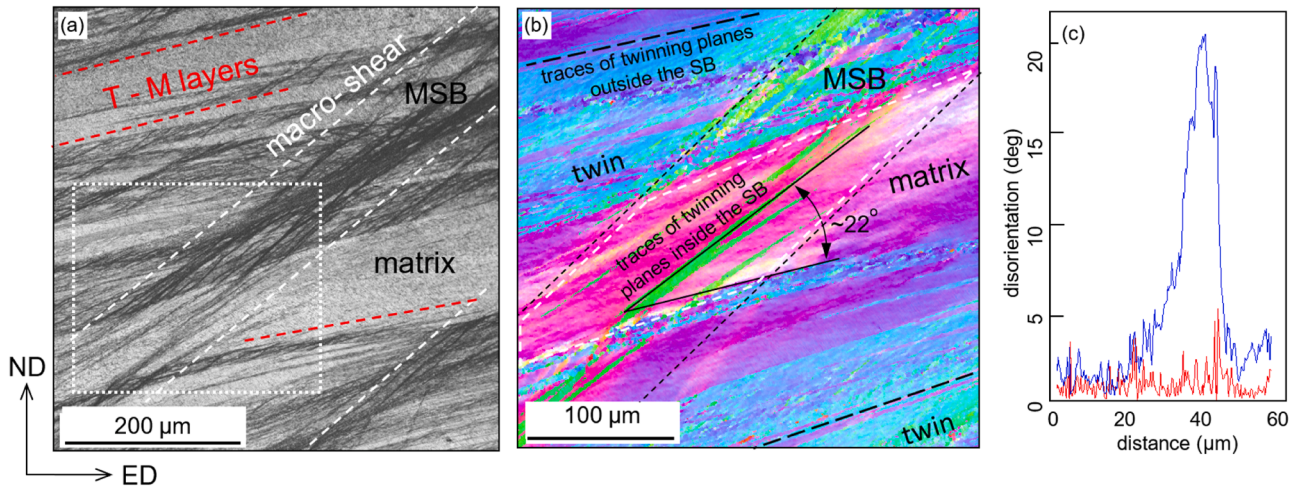


Fig. 15. Twinning in re-oriented matrix platelets inside the SB. Orientation map presented as a function of: (a) the image-quality factor and (b) the inverse pole figure colouring. (c) Disorientation line scan across the SB and along the line inside the matrix platelet free of twins. SEM/EBSD measurements in ND/ED section. Step size of 70 nm.

4. Discussion

This study on Cu single crystals with (346)[63–5], (112)[11–1], (110)[1–1–2], and (110)[001] initial orientations unveils the microstructure and texture evolutions during deformation under plane state of strains at very high strain rates. The experimental data presented here show that the microstructure of the Cu single crystals during impact loading up to a logarithmic strain of 0.9 differs significantly from that observed during static or semi-static deformation. This differentiation is mainly related to the appearance of deformation twins, which are not

observed in statically deformed Cu at RT. The presented microstructures are similar to those observed after semi-static deformation at 77 K [35–37] or lower temperatures [38], where deformation twins can appear in Cu crystallites of specific orientations. However, some issues remain open. Two of the most important relate to:

- (i) the selection of active twinning systems and their intensity in terms of initial grain orientation,

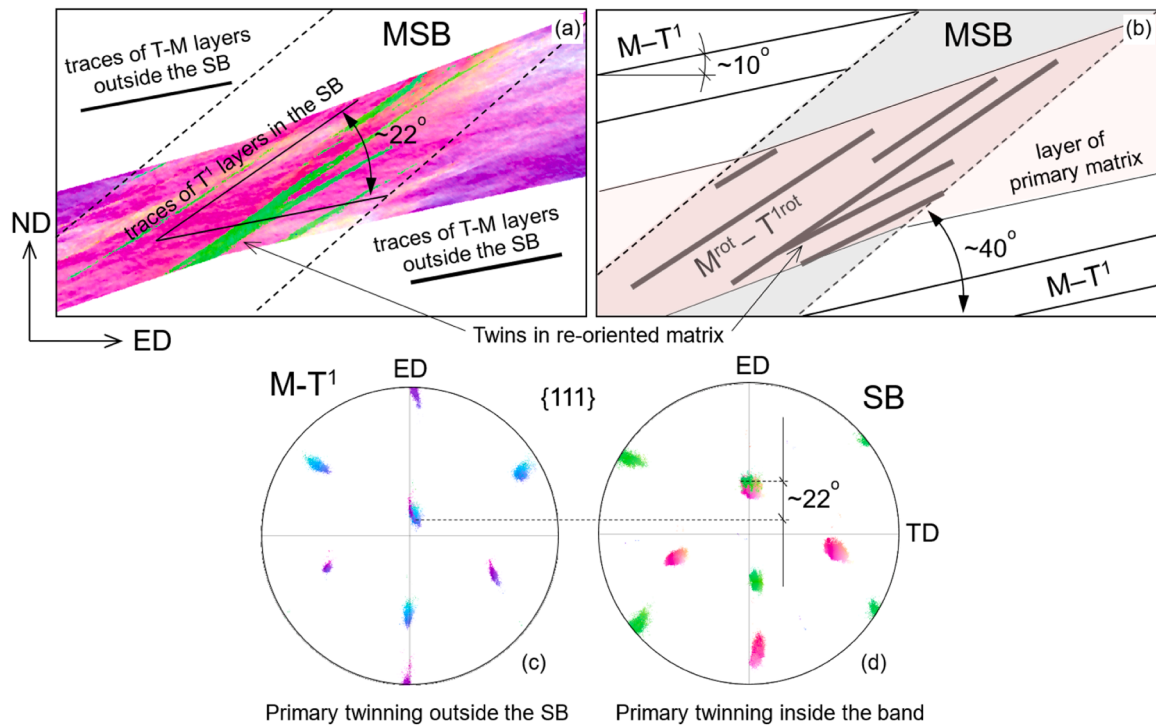


Fig. 16. Orientation changes due to twinning in re-oriented part of the matrix platelet inside the SB. (a) Part of the orientation map presented in Fig. 13 showing orientation changes in the primary matrix platelet and (b) corresponding $\{111\}$ pole figures showing the orientations of the T-M layers in the areas (c) outside and (d) inside the SB. SEM/EBSD measurements in ND/ED section. Step size of 70 nm.

- (ii) the mechanism of SB formation in impact-loaded copper and the formation of specific texture components within the band interior.

In addition to primary twinning, much attention has been paid to describing the crystallographic determinants of second generation twinning in dynamically deformed crystals. The orientations of these secondary twins are discussed in terms of the orientation of the initial matrix and primarily twinned platelets. Regarding the SB, the discussion focuses on the characterization of the mechanism responsible for the destabilisation of the compact clusters of T-M layers. Fundamental issues include the disclosure of the role of SBs in the global texture formation of copper during high-speed deformation and mechanisms controlling the background lamellar-structure transformation into an ultra-fine-grained structure inside the SBs.

4.1. Twinning systems and the grain-orientation-dependent intensity of twins formation

The microstructure of copper after RT deformation strongly depends on the strain rate. In the range of low strain rates, greater strengthening during channel-die deformation was ascribed to a greater density of dislocations, whereas strengthening during impact loading was ascribed to a greater density of dislocations and deformation twins [41–46]. According to our best knowledge, no existing works describe the evolutions of the microstructure and texture in oriented crystallites after deformation under the plane state of strains in the range of extremely high strain rates. All previous studies were performed based on uniaxial compression tests of cylindrical or hat-shaped axisymmetric specimens. Hence, any data from the literature should be compared with the results obtained in the present study with a certain degree of caution. However, several common features can be observed during high-strain-rate uniaxial compression and deformation under the plane state of strain.

The results obtained by Hong et al. [32] based on the experiments of the dynamic deformation of polycrystalline Cu strongly suggest that

grain orientation plays a deterministic role in the activation of deformation twins. A high density of twins was observed in uniaxially compressed Cu single crystals with compression axes located in the area of the standard triangle situated between the $\langle 100 \rangle$ - $\langle 112 \rangle$ - $\langle 123 \rangle$ directions. However, they were not observed in single crystals with compression axes situated between the $\langle 101 \rangle$ and $\langle 111 \rangle$ directions [33, 47–49]. These results correlate well with the one obtained here or in copper that was rolled or channel-die compressed at 77 K under quasi-static loading [13,14,49], where the extremely strong twinning in crystallites with $(112)[\parallel \text{ND}][11-1](\parallel \text{RD})$ orientation was observed. The compression axis along the $\langle 100 \rangle$ direction also resulted in intense twinning owing to the extremely strong instability of the crystallites. For the $(100)[011]$ - or $(100)[001]$ -oriented crystallites channel-die deformed at low strain rates, the formation of two symmetrically equivalent C $\langle 112 \rangle$ - $\langle 111 \rangle$ orientations (in the first case) and four symmetrically equivalent S $\langle 123 \rangle$ - $\langle 634 \rangle$ orientations (in the second case) was observed, already in the range of intermediate deformations ($> 50\%$) [12]. This, in turn, strongly favours the formation of deformation twins in C- or S-oriented areas. Therefore, to enhance structural refinement via deformation twinning, the use of materials with a strong $\langle 100 \rangle$ -type fibre texture is beneficial.

The results obtained during impact uniaxial compression of the $\langle 110 \rangle$ -oriented crystallites show that they do not form deformation twins [32]. The compression direction of $\langle 110 \rangle$ occurs in the two orientations analysed during the channel-die compression tests, that is, Br $(110)[1-2]$ and G $(110)[001]$. Both orientations are stable during the semi-static channel-die deformation at RT and 77 K and did not form deformation twins. However, during impact loading under a plane state of strains, the twins were formed. This confirms the hypothesis that increasing the strain rate to $4.7 \times 10^5 \text{ s}^{-1}$ has a greater influence than lowering the temperature to 77 K or strong decreasing the SFE ($< 10 \text{ mJ/m}^2$), e.g., in Cu-Al alloys.

Another issue is the extent to which the results for single crystals are representative for polycrystalline metals. This issue has been analysed recently by Cao et al. [33] during the dynamic uniaxial compression of

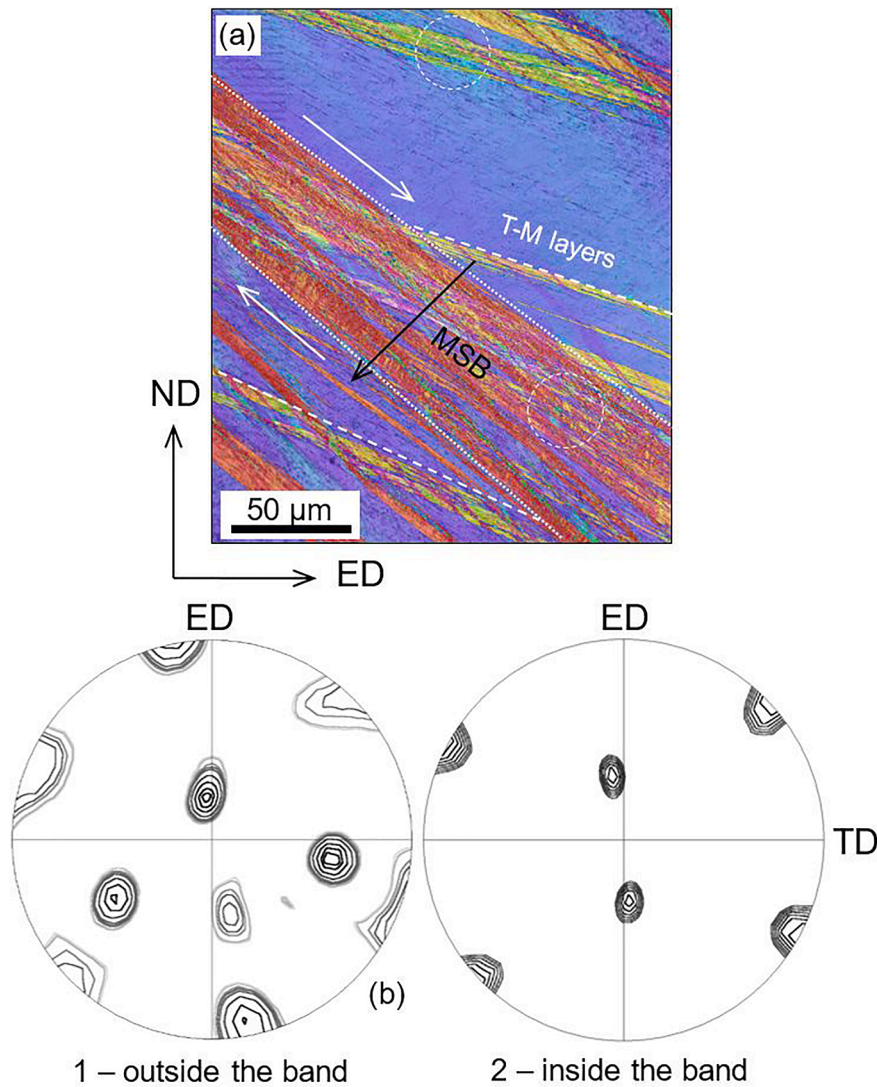


Fig. 17. Internal texture of the well-developed SB. (a) The orientation map shows a well-developed SB in an S-oriented single crystal. The $\{111\}$ pole figures show (b) the texture of the T-M platelets outside the band and (c) the texture of the band interior. SEM/EBSD measurements in ND/ED section. Step size of 50 nm.

polycrystalline Cu. The results showed a strong correlation between the susceptibility to twinning and the density of twin formation with crystallite orientation, which is similar to that observed in single crystalline samples. This is consistent with earlier studies that showed that the behaviour of individual crystallites in a polycrystalline aggregate is analogous to that of single crystals with the same orientation [47, 50–52], beyond thin layers near grain boundaries [53,54]. Intense deformation twinning has also been observed in C- and S-oriented grains of Cu sheets explosively welded to steel or copper plates, where the states of strain and stress are similar to those observed during rolling or channel-die compression [26,27].

The contribution of slip and deformation twinning to strain accommodation is difficult to determine without analysing intermediate states (the issue is not analysed in this work). In crystallites with G and Br orientations, the dominant strain accommodation mechanism is dislocation slip. Deformation twinning (in the form of thin needles) appears accidentally and only in areas near the free ends of the sample. Since the twins occupy only a small area, it can be concluded that the twins do not play a significant role in strain accommodation. In the C and S orientations, the contribution of deformation twinning is significant giving rise to a large accommodation of strain with the dominant role of twinning on the (111) plane. However, because of other twinning systems, as well as second-generation twins, it is difficult to assess their

contribution to global deformation.

4.2. Shear bands formation vs. orientation changes

The evolution of the microstructure of many *fcc* metals compressed statically in a channel-die at RT indicated that shear band nucleation always occurs after creating obstacles for dislocation gliding [32,35–37, 40]. In metals with a low SFE, compact clusters of T-M platelets were developed, and the twin boundaries act as barriers for dislocation motion, similar to the boundaries between flattened (sub)grains in high and medium SFE *fcc* metals.

The mechanism of SBs formation during impact loading is similar to that observed during static or semi-static straining and comprises three main steps. The initial stage (i) is connected to T-M lamellae bending in narrow volumes, leading to kink-type band formation. This process gradually increases the inclination of the twinning planes relative to the compression plane and makes them progressively parallel to the shear band plane. This changes the crystal-lattice orientation in the matrix and twin platelets in the SB region and is the most important source of significant texture scattering. The second stage is determined by (ii) deformation twinning in the re-oriented matrix (dominant), which leads to the disappearance of the matrix orientation and secondary twinning in earlier twinned platelets (less noticeable). Finally, after the complete

disappearance of the original matrix in the SB region (iii) pure shear is observed along one of the $\{111\}$ -type planes situated along the SB plane (Fig. 13d). The latter process transforms the lamellar structure of the twin and matrix platelets into a micro-/nano- grained structure of a well-developed SBs.

The intensities of the individual processes determine the texture of the SBs interior. The origin of localised bending has been analysed in detail in earlier studies on twinned Ag [36], Cu [13,49] and single-phase Cu-Al alloys [13,37]. The model that was proposed is based on observations by Kamijo and Sekine [55] and Leffers and Bilde-Sørensen [56], which indicate little dislocation slip on slip planes intersecting the twin boundaries. Because the plane of maximum shear stress (SB plane) does not coincide with any of the $\{111\}$ -type planes (in the twin and matrix platelets), twin boundaries are an effective barrier for dislocation motion on slip planes other than the twinning plane. This causes a dislocation re-grouping and contributes to stress concentration at the potential SB boundaries (Fig. 18a), i.e. along the plane of maximum shear stress. In this model, an additional assumption was made that the current state of anisotropy determines the inclination of the SB plane [37]. The dislocation concentration along the SB plane induces a rigid-body rotation of the twin and matrix platelets towards the plane of maximum shear stress in the region of the 'potential' SB. This leads to the double kinking of the twin and matrix lamellae, which can be regarded as a precursor of a well-developed SBs (Fig. 18b). The dislocations stored on both sides of the SB accommodate the bending of the

twin and matrix lamellae until they create a localised slip along the shear plane (Fig. 18c) and strong microstructure refinement inside the band (Fig. 18d). The orientation of the crystal lattice in the twin and matrix lamellae in the SB region tends to change progressively owing to the rigid-body rotation, as shown in Figs. 13–16.

The crystal-lattice rotation within the sheared region clearly explains why the SBs plane is significantly different from the $\{111\}$ plane in the T-M layers outside the band. Previously, the observed non-parallelism of the $\{111\}$ planes in the SB area and the neighbouring twinned matrix led some authors to the conclusion that slip within SBs is non-crystallographic [57,58]. Nevertheless, as shown above, the lattice rotation in the SB area progressively brings the (111) plane into a position parallel to the SB plane, which opens up the possibility of shearing along the favourable (111) slip/twinning plane (Fig. 14d).

Another consequence of this crystal-lattice rotation in the SB region is the increased activity, extinction, or activation of new slip and twinning systems. However, the description of certain phenomena during SB formation is the subject of strong controversy. The most obvious example of these controversies is the explanation of the mechanisms responsible for 'disappearance' in the SB region of twin relationship and orientations derived from the original matrix. Several mechanisms have been proposed to explain this phenomenon. *The first* involves the reverse twinning (or de-twinning) of existing twins induced by twin-slip interactions [40,46,59–61]. *The second plausible mechanism* is based on the consideration that different tendencies of the crystal-lattice rotations

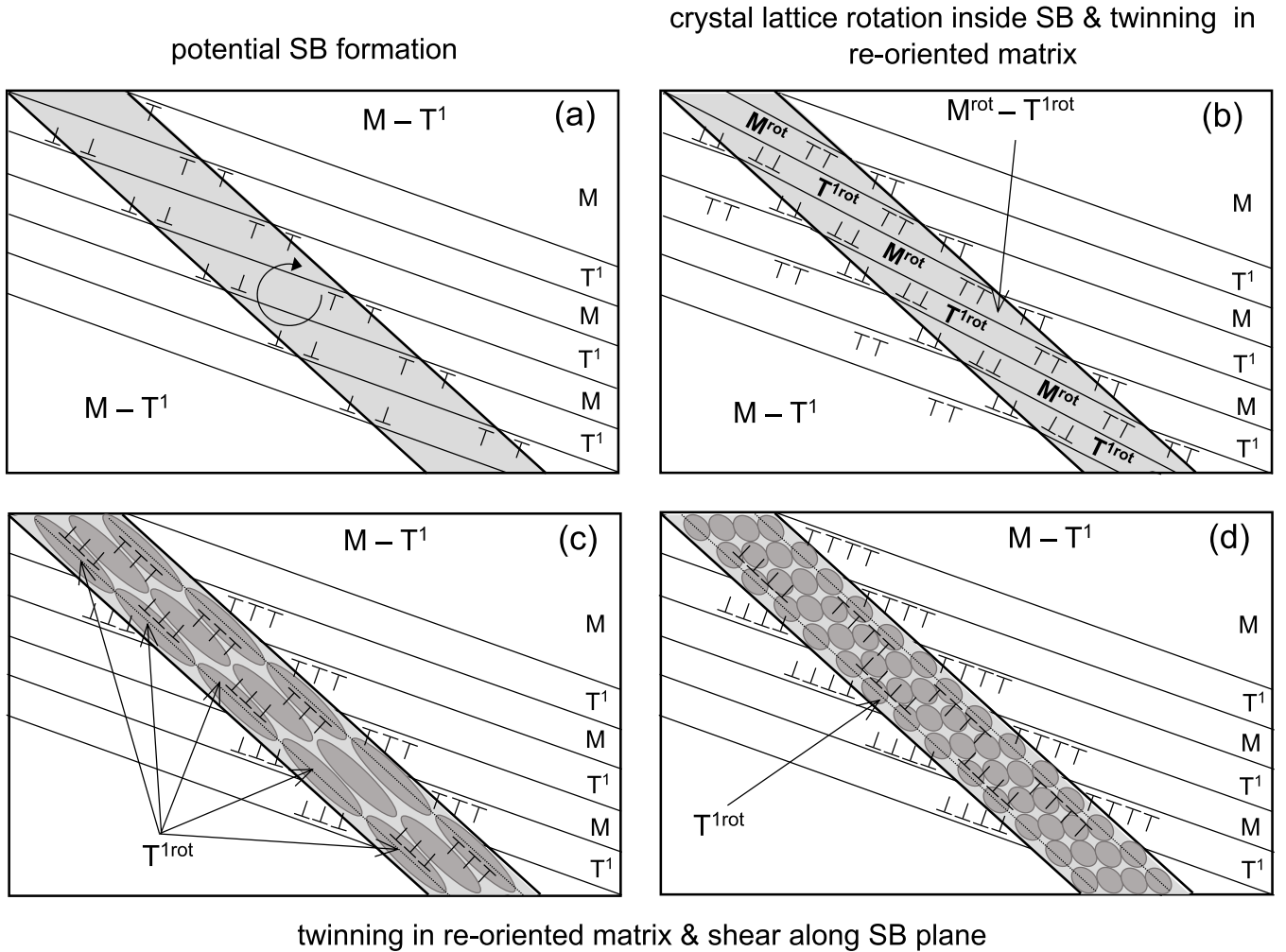


Fig. 18. Schematic presentation of microstructural changes due to SB formation in twinned structures of fcc metals. (a) Dislocation re-grouping along SB boundaries leading to (b) local kinking of T-M layers and short-distance dislocation slip along re-oriented twinning planes in the SB area. (c) Long-distance dislocation slip along SB plane and (d) intense dislocation slip along the shear plane leading to structure refinement in the SB interior.

(due to the operation of various slip systems) in the T-M platelets inside the sheared region may lead to a loss of coherency of the twin boundaries [40,49,62,63]. This mechanism was identified in our previous study [49] and was successfully applied to explain the deformation texture transition from $\{110\}<001>$ to $\{110\}<112>$ observed in C (112)[11–1]-oriented single crystal of pure Ag in the range of large strains [36]. The third mechanism involves primary twinning in re-oriented matrix platelets and secondary twinning in earlier-twined areas [37].

At this stage, unequivocally stating the validity of the given mechanisms for the formation of a specific texture in the SB interior is difficult. The most controversial issue relates to the importance of secondary twinning (twinning from twins) and reverse twinning (de-twinning). Hong et al. [40] argued that “... the secondary-twinning mechanism is energetically unfavourable compared with the de-twinning mechanism since the energy barrier to create an intrinsic stable stacking fault from a perfect lattice is larger than that to create a twin fault from a pre-existing twin plane”. This study does not confirm this hypothesis. Secondary twins were observed in the SB interior. However, their intensities were very low; therefore, they are weakly visible in the texture of the SB interior. The changes that occur in the area of the matrix platelets during the process of incorporation into the SB area are critical to the formation of the texture image. The present work demonstrates that crystal-lattice rotation in the sheared region and twinning in re-oriented matrix lamellae (that have not yet been transformed) play a key role in the texture formation of the SB interior. Moreover, the disappearance of components connected with the primary matrix leads to dominance near the G(110)[001] orientations, which are inherent in the orientation of the twinned platelets [64–66]. This mechanism has also been suggested in our earlier studies [37,49] and its validity has been documented by the orientation maps presented in Figs. 15–17. This indicates that reverse twinning (if it occurs at all) cannot play any role in the texture formation of brass-type SBs because it should transform the orientation of the primary twinned areas into that of the original matrix. Based on the SEM/EBSD measurements, the orientations connected to the primary matrix were not represented inside the SB region; therefore, such a transformation could not occur. The orientation of secondary twins that appeared accidentally inside the primary twinned platelets was symmetrical to the primary matrix; thus, they cannot be related to the primary matrix.

The analyses confirmed the validity of the description of the structural changes based on the value of the Schmid factor of slip and twinning systems operating in the SB region. This applies to both the twin transformation in the re-oriented platelets of the original matrix with an orientation close to $(100)[011]^{\text{Matrix}}$ in the SB region and the possibility of a second generation of twins appearing in twin platelets with an orientation close to $(110)[001]^{\text{Twin}}$.

In these twin and matrix orientations, the (111) and (11–1) planes, are highly favourable for dislocation motion and twinning. They are situated symmetrically with respect to the plane perpendicular to the ED and therefore, the slip and the twinning systems have the same Schmid-factor values. For the particular case of twin platelets, twinning on the (11–1) plane introduces an orientation symmetrically equivalent to the (+)TD-rotated orientation of the initial matrix. In orientation measurements based on the selected-area-diffraction technique in TEM, this differentiation can be difficult to identify and is often interpreted as a process of reverse twinning (because the measured orientations are characterised by symmetrically equivalent Miller indices). Hence, extra consideration should be given in analyses that assume the possibility of reverse twinning in the SB region.

5. Conclusions

The effect of impact loading on the microstructural and textural evolutions of pure copper was studied through controlled channel-die compression experiments on single crystals with (346)[63–5], (112)[11–1], (110)[1–1–2], and (110)[001] orientations deformed up to 60

%. SEM and TEM examinations of the post-impact samples revealed well-defined heterogeneous microstructures in all single crystals, which were different from those observed in statically channel-die deformed copper single crystals with the same orientations. The dominant mechanisms governing the structural and textural evolutions include deformation twinning, deformation and shear banding. Twins were observed in all four orientations. In the G- and Br-orientations, twins occurred rather rarely and always in the form of single isolated needles, whereas in the C- and S-oriented crystals, the twins formed compact clusters that occupied a significant volume of the sample. A detailed analysis of the single-crystalline samples led to the following conclusions:

- (i) Single crystals of nominally ‘stable’ Br and G orientations exhibited unstable textural and structural behaviour. In both orientations, deformation bands with a disorientation relative to the surrounding matrix of up to 40° were developed. This resulted in the strong scattering of the initial orientation.
- (ii) In the Br-oriented crystallites, twin transformations were demonstrated on all four $\{111\}$ planes, whereas in the G-oriented crystallites, twinning occurs on two planes strongly favoured for dislocation motion.
- (iii) In S- and C-oriented copper single crystals, the evolutions of the microstructure and texture were similar. Deformation twins formed intensively, with the dominant family resulting from twin transformation on the (111) plane. The presence of three families of deformation twins was demonstrated. Moreover, the formation of second-generation (secondary) twins was often observed in areas where the primary twins reached the G (110)[001] orientation.
- (iv) The intense deformation twinning in the C- and S-oriented crystals favoured the formation of SBs. In the nucleation stage, shear banding resulted from the loss of stability of the compact clusters of deformation twins within narrow areas, leading to the formation of kink-type bands. In more advanced stages of the SBs formation, twinning in the re-oriented platelets of the matrix in the interior of the SBs led to the appearance of a texture dominated by the (110)[001] orientation.
- (v) Finally, we conclude that the influence of an impact loading at RT on the deformation substructures of copper single crystals is similar to that observed following quasi-static deformation at low temperatures.

CRedit authorship contribution statement

H. Paul: Conceptualization, Data curation, Formal analysis, Funding acquisition, Investigation, Methodology, Project administration, Writing – original draft, Writing – review & editing. **I. Mania:** Data curation, Investigation. **R. Chulist:** Investigation. **S. Puchlerska:** Formal analysis, Investigation. **M. Prazmowski:** Investigation.

Declaration of competing interest

The authors declare that they have no known competing financial interests or personal relationships that could have appeared to influence the work reported in this paper.

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