

Influence of different wrought processes on the microstructure and texture evolution in AZ31 magnesium alloy

Hiromasa YOSHIKUMI^{1),2)*}, Motohiro YUASA^{1),2)},
Hiroyuki MIYAMOTO¹⁾, and Hidetoshi SOMEKAWA²⁾

¹⁾ Faculty of science and engineering, Doshisha University, 1-3 Miyakodani, Tatara, Kyotanabe, Kyoto, 610-0394, Japan

²⁾ Research Center for Structural Materials, National Institute for Materials Science, 1-2-1 Sengen, Tsukuba, Ibaraki, 305-0047, Japan

***Corresponding author:** H. Yoshizumi

E-mail: hyoshizumi3141@gmail.com

Tel: +81-774-65-6458

Abstract

AZ31 magnesium alloys were deformed by five wrought processes (forging, rolling, extrusion, equal-channel-angular-extrusion (ECAE), and caliber rolling) under the condition of an introduced equivalent plastic strain of approximately 1.5 at 573 K. The strain components induced by these wrought processes were computed using the finite element method (FEM). By combining the calculation results with the microstructure observations, the texture and microstructure evolution were discussed from the viewpoint of the strain state. The FEM showed that approximately the same equivalent strain was induced at the center of the alloys where observations were performed. Therefore, as the values of the equivalent plastic strain were fixed in all wrought processes, the effect of the strain component could be compared. The forged and rolled alloys had an intensive basal texture perpendicular to the compression and normal directions, respectively. In the others, a basal fiber texture parallel to the material flow direction was observed. These basal textures were oriented perpendicular to the compressive strain, and their intensities strengthened with an increase in the strain component. The microstructure became more refined and homogenized as the magnitude of the minimum principal strain decreased. This decrease corresponds to the transition of the deformation mode from the uniaxial compression mode to the multidirectional deformation mode as long as the same equivalent strain is induced. This result demonstrates that the microstructure obtained through dynamic recrystallization depends on the deformation mode, separately from the effect of the conventionally reported equivalent plastic strain.

Keywords

Magnesium alloy; Dynamic recrystallization; Finite element method; Basal texture; Strain

1. Introduction

Magnesium (Mg) is the lightest structural material. However, its poor mechanical properties are an obstacle to its widespread use. An effective method for improving this is microstructure refinement through dynamic recrystallization (DRX) during hot wrought processes [1-5]. In addition to grain size refinement, the texture changes as a result of DRX contributes to the promotion of formability [6, 7]. To optimize the process conditions for DRX, microstructural evolution has been investigated in various wrought processes (rolling [8-13], extrusion [14-16], equal-channel-angular-extrusion (ECAE) [17-20], high-pressure torsion (HPT) [21, 22], caliber rolling [23-25], and cyclic extrusion compression [26]). The deformed microstructure is known to differ depending on the wrought process. Compressed, rolled, and HPT-processed alloys have a larger number of twins in the initial stages of deformation [8-11, 27]. Conversely, twins were scarcely formed in the ECAE [20]. These results suggest that the deformed microstructure depends on the deformation mechanisms (i.e., basal, prismatic, and pyramidal slip systems and twins).

The deformation mechanisms operating in the wrought processes were investigated using crystal plasticity analysis. When slip systems and twins rotate the orientation of the crystal, a slip plane aligns perpendicular to the compression direction, and the texture is developed [28]. Because the critical resolved shear stress (CRSS) of basal slip is the lowest in the slip systems of Mg, the rotation of this slip is the most effective for basal texture development. As the rotational direction of the $\langle c+a \rangle$ pyramidal slip is inverse to that of the basal slip, the pyramidal slip activity plays a role in weakening the basal texture [29, 30]. The other deformation mechanisms (prismatic slip and twinning) hardly influence the final basal texture of the extruded alloy [31]. These deformation mechanisms are related to microstructure refinement together with the basal and pyramidal slips [20, 31].

Al-Maharbi et al. showed that the relative activities of deformation mechanisms are different in the processing path of the ECAE [20]. If the number of passes is the same as in ECAE, the introduced equivalent plastic strain is equal. The difference in the deformation mechanism is assumed to be caused by the induced strain state.

Because the internal strain cannot be measured directly during the wrought process, the strain state is quantified using calculations such as the finite element method (FEM) [25, 32-37] and the embedded-pin method [38] for various metal materials. The effects of the strain state on the microstructure evolution are discussed by combining the calculation results and the microstructure observations. There are two effects of the strain on the microstructure: The first effect is ascribed to the introduction of a higher equivalent plastic strain. According to Kamikawa et al. [38], accumulative roll bonding (ARB) with shear strain forms a more refined microstructure than ARB without shear strain because of the large equivalent plastic strain by shear strain. Inoue et al. [32] demonstrated that multidirectional compression could result in an ultrafine grain structure in low-carbon steel by introducing a homogeneous distribution of the equivalent plastic strain with a high value, compared

to uniaxial compression even at the same level of deformation. Because the driving force of DRX is the stored strain energy, DRX is activated according to the increase in the equivalent plastic strain [21]. However, this effect cannot explain the difference in the microstructure developed in each processing route of ECAE, as mentioned above, because the same value of equivalent plastic strain was induced.

Another effect is the deformation mode (e.g., uniaxial compression mode and plane strain compression, etc.). Inoue et al. [33]. showed that the bidirectional simple compressive mode and bidirectional plane strain mode deformed the cube texture and the α -fiber texture, respectively, in low-carbon steel. Cho et al. [34] compared the areas with and without shear strain at a fixed equivalent plastic strain in a compressed sample of Ni-30Fe and showed that shear strain accelerates the subdivision of austenite grains. These studies suggest that the microstructural evolution depends on the deformation mode, in addition to the effect of the equivalent plastic strain.

Sabat et al. [7] reported that different textures were developed in Mg alloys by three different wrought processes — rolling, multiaxial forging, and ECAE— even when the same true strain was induced. However, in this study, an analysis of the deformation mode was not conducted. Furthermore, there has been no systematic investigation of the relationship between the strain state and DRX in terms of the deformation mode. Thus, the effect of the induced strain state on the microstructural evolution has not been studied in different wrought processes.

In an individual wrought process, the effects of the processing conditions— temperature, processing speed, and number of processing passes, etc.— on microstructural evolution have been investigated. On the other hand, to the best of our knowledge, the effect of the difference in wrought processes has not been cleared. This study aims to investigate quantitatively the influence of difference in wrought processes on microstructural evolution from the viewpoint of the induced strain state. To impose different strain states at the same equivalent plastic strain, the AZ31 Mg alloy was deformed using five wrought processes (forging, rolling, extrusion, ECAE, and caliber rolling) introducing an equivalent plastic strain of 1.5 at 573 K. The strain components were computed using the FEM. By combining the microstructural observations with the calculation results, the effect of strain on the microstructure was discussed.

2. Experimental and numerical procedures

An extruded Mg alloy AZ31 (Mg-3.0 Al - 0.8 Zn-0.31 Mn, wt. %) was used. After annealing at 673 K for 1 h to weaken the deformed texture, the alloy was deformed using five wrought processes (forging, rolling, extrusion, ECAE, and caliber rolling) at 573 K. The processing conditions are listed in Table 1. These conditions were set to induce an equivalent plastic strain of 1.5. DRX is facilitated by an increase in induced strain energy. However, the microstructure is reported not to change significantly above the equivalent plastic strain of approximately 1.5 in various wrought processes [5, 10, 21, 25]. Thus, the induced equivalent strain was determined to be 1.5, because this strain value is

considered sufficient to investigate the effect of the wrought process on the microstructure. Additionally, although DRX is activated by an increase in the processing temperature [21], the DRXed grains grow rapidly, and the difference in the deformed microstructure is indistinct at processing temperatures above 623 K [39]. The processing temperature was set at 573 K to reduce the thermal effects on the microstructure.

The microstructures were observed at the center of the alloys by electron backscatter diffraction (EBSD) with scanning electron microscopy (SEM), and optical microscopy (OM). The crystallographic orientation was measured by X-ray diffraction (XRD) in two directions for each processed alloy. The forged alloy was observed in the direction parallel to the compression direction (CD) and in the transverse direction (TD) to the CD. The rolled alloy was observed from the normal direction (ND) to the rolling direction (RD) and the transverse direction (TD) to the RD. The extruded and ECAE-processed alloys were observed from the parallel to the extrusion direction (ED) and the normal direction (ND) to ED. The caliber rolled alloy was observed from ND and RD. To compare textures quantitatively, the orientation index was calculated from the XRD patterns [40]. The index corresponding to the (hkl) plane is defined by the following equation:

$$\text{Orientation index}(hkl) \equiv R(hkl)/R_0(hkl), \quad (1)$$

where $R(hkl)$ is the relative intensity of the (hkl) plane and $R_0(hkl)$ is that obtained from the peak pattern in powder form. In this study, $R(hkl)$ was calculated using the integrated intensity of the (hkl) peak, $I(hkl)$, by

$$R(hkl) = I(hkl)/\Sigma I(hkl). \quad (2)$$

The strain distributions after each wrought process were computed by FEM using the DEFORM-3D software. In the simulation, the alloy was defined as homogeneous, isotropic, and rigid-plastic body. The plastic response was obtained from the compression tests at 573 K and various strain rates from 0.01 to 1.0 s^{-1} . The sample was modeled using 100,000 tetrahedral elements, associated with our machine ability. The finite element meshes used in the simulations are shown in Fig. S1 in the Supplementary file. The meshes in the rolling appear coarse, compared with the others. Nevertheless, there was little difference in the calculation result of the rolling when the number of meshes was five times as many. Thus, 100,000 elements were enough for the calculation. The shear friction with a friction factor of 0.2 was used as the frictional condition [41]. All simulations adopted the same plastic response and frictional condition to focus on the difference in strain state caused by shape change, not thermal effects. For computing strain state under different processing conditions, the processing

temperature and conditions in the FEM simulation were identical to those used in the actual processes (Table 1). The numerically analyzed regions coincided with the observed microstructural regions.

3. Results

3.1 Microstructure observation

Figs. 1 and 2 show the EBSD results and the cumulative distributions of the grain size. The caliber rolled alloy has the most refined and homogeneous microstructure, with an average grain size of 5.4 μm (Fig.1 (i), Fig.2). In contrast, the microstructure of the forged alloy is composed of coarse and fine grains with a grain size of 70 μm and several micrometers, respectively (Fig.1 (a)). Fig. 3 shows the (0002) pole figures obtained from EBSD. In the forged and rolled alloys, an intensive basal texture perpendicular to the CD and ND appeared to develop. In the other alloys, a fiber basal texture parallel to the ED or RD was observed.

Because the observation of crystallographic orientation from EBSD is greatly influenced by grain size, the orientation data from XRD were used to quantify the basal texture intensity in this study. Fig. 4 shows the XRD patterns observed in the two directions. The orientation of the basal plane was examined by comparing these patterns. In the forged alloy, a significantly higher peak around 34° corresponding to the basal plane was observed for CD than TD (Fig. 4(a)). This indicates that the forged alloy had a basal texture perpendicular to CD. The other patterns show that the basal planes of the rolled alloy and caliber rolled alloy are perpendicular to the ND. In contrast, the extruded and ECAE-processed alloys were parallel to the ED (Fig. 4). The XRD results agree with the basal texture orientation examined using EBSD data. The orientation indices of the prismatic, basal, and pyramidal planes are summarized in Table 2. Comparing the orientation index of the basal plane in the direction perpendicular to the basal texture, as examined above, forging CD, rolling ND, extrusion ND, ECAE ND, and caliber rolling ND, the forged and rolled alloys have a more intensive basal texture than the other alloys. The basal textures examined in this study agree with those of several studies [8-11, 16, 23, 42, 43].

3.2 Strain calculated by FEM

Fig. 5 shows the contour maps of the strain computed using the FEM. The regions shown in Figs. 5 (b) and (c) are inside the red frame in Fig. 5 (a). Various studies have shown that the basal texture perpendicular to the direction of compression develops during uniaxial compression [28, 43, 44]. Similarly, during rolling, the basal texture formed is perpendicular to the compressive direction (i.e., ND) [8-11]. These results lead to the assumption that the basal plane aligns perpendicularly to the compression strain during wrought processes. To reveal the influence of the compression strain on the intensity of the basal texture, the strain was analyzed using the orthogonal coordinate system, in which the z-axis was along the direction, indicating a larger orientation index of the basal plane: forging CD,

rolling ND, extrusion ND, ECAE ND, and caliber rolling ND.

Fig. 5 (b) shows the distribution of the equivalent plastic strain $\bar{\varepsilon}$, which is defined as follows [33]:

$$\bar{\varepsilon} = \int_0^t \frac{d\bar{\varepsilon}}{dt} dt, \quad (3)$$

where $d\bar{\varepsilon}/dt$ is the increment of the equivalent plastic strain explained by

$$\frac{d\bar{\varepsilon}}{dt} = \sqrt{\frac{2}{3} \left[\left\{ \left(\frac{d\varepsilon_{xx}}{dt} \right)^2 + \left(\frac{d\varepsilon_{yy}}{dt} \right)^2 + \left(\frac{d\varepsilon_{zz}}{dt} \right)^2 \right\} + \frac{1}{2} \left\{ \left(\frac{d\gamma_{xy}}{dt} \right)^2 + \left(\frac{d\gamma_{yz}}{dt} \right)^2 + \left(\frac{d\gamma_{zx}}{dt} \right)^2 \right\} \right]}. \quad (4)$$

A close value of $\bar{\varepsilon}$ in the range of 1.65-1.9 is imposed in the center of the workpieces, where the microstructures are observed. This means that the processing conditions in Table 1 deformed the alloys at almost the same strain level.

Fig. 5 (c) shows the distribution of the strain component along the z-axis. All the wrought processes introduced compressive strain along the z-axis. Considering the crystallographic orientation from Fig. 4, this result indicates that the direction perpendicular to the basal texture is along the compression direction.

4. Discussion

4.1 Effect of compressive strain on basal texture development

The z-strain FEM result suggests that the basal texture developed perpendicular to the compressive strain during wrought processes. Fig. 6 shows the relationship between the intensity of the basal texture and the compressive strain along the z-axis. The basal texture intensifies with an increase in the compressive strain. This tendency is due to the crystal lattice rotation caused by the basal slip. When a slip system operates, the direction perpendicular to the slip plane rotates parallel to the compression. Owing to this orientation change, the basal slip introduces lattice rotation as the basal plane aligns perpendicular to compression [29]. Basal slip reorients the basal texture perpendicular to compression under various loading conditions [28, 44, 45]. Basal slip is activated to achieve a large compressive strain, enhancing the lattice rotation only to develop an intensive basal texture.

In addition to basal slip, non-basal slips (i.e., prismatic slip and $\langle c+a \rangle$ pyramidal slip) operate at high temperatures and change the crystallographic orientation. The prismatic slip hardly changes the orientation of the basal plane because the crystal lattice rotates around the c-axis owing to the slip system [15]. Conversely, the activation of the $\langle c+a \rangle$ slip changes the basal plane orientation oppositely to the rotation by the basal slip. Studies using crystal plasticity analysis have shown that the evolution of basal texture depends on the activity of the $\langle c+a \rangle$ slip [29, 30]. When the basal slip and $\langle c+a \rangle$ slip

operates simultaneously, the basal plane aligns the direction in which the rotations caused by them are balanced. Thus, the angle between the compressive direction and the c-axis widened with increased activity of the $\langle c+a \rangle$ slip. According to Agnew et al. [30], when the CRSS ratios for $\langle c+a \rangle$ slip to basal slip are 3 and 6, the tilt angles of the basal pole from the compression direction are simulated to be approximately 15° and 10° , respectively. From this result, considering the CRSS ratio for $\langle c+a \rangle$ slip to basal slip is approximately 6.2 in AZ31 at 573 K [46], the tilt angle of the basal pole from the compression direction is estimated as less than 10° . This estimation suggests that the effect of $\langle c+a \rangle$ slip on the evolution of the basal texture is significantly less than that of the basal slip.

The orientation index of the basal plane increased abruptly at a compressive strain of approximately 1.0 (Fig. 6). This is explained by the activity of the tensile twins. Fig. 7 shows the optical micrographs of the microstructures of the rolled and ECAE-processed alloys, which have an $\bar{\epsilon}$ of approximately 0.75. The rolled alloy with a large compressive strain had plenty of twins (Fig. 7 (a)), whereas the ECAE-processed alloy with a low compressive strain had no twins (Fig. 7 (b)). Several studies have reported that the $\{10\bar{1}2\}$ twins (tensile twins) are generated during the initial stage of rolling [8-11]. Thus, the twins observed in Fig. 7 (a) are considered tensile twins. The tensile twin is formed to achieve the strain along the c-axis, which cannot be achieved by the $\langle a \rangle$ slip. The c-axis of the tensile twinned region tilts 86.3° from the c-axis of the matrix [47]. This change in crystallographic orientation develops a basal texture perpendicular to the compression direction [43-45, 47]. To achieve a large compressive strain, the tensile twin operates as an additional deformation mechanism by rotating the grain orientation with the basal plane perpendicular to the compression. As a result, the basal texture becomes more intensive when the tensile twin acts in addition to basal slip.

At high temperatures, $\langle c+a \rangle$ slip serves as the deformation mechanism to achieve strain along the c-axis, suppressing the activity of the tensile twin [48, 49]. This is inconsistent with the activated generation of twins shown in Fig. 7 (a). The enhancement of the $\langle c+a \rangle$ slip is due to activated cross-slip from the basal plane to the non-basal plane with the aid of thermal vibration at high temperatures [50, 51]. However, at a high strain rate deformation, the stress reaches a high value before obtaining sufficient thermal energy to operate the $\langle c+a \rangle$ slip, even at a high temperature [52]. The deformation mechanisms with less influenced thermal activity (i.e., basal slip and tensile twin) operate before the $\langle c+a \rangle$ slip begins to act. Accordingly, the tensile twin plays a dominant role in the deformation along the c-axis instead of the $\langle c+a \rangle$ slip at a high strain rate, even in a high-temperature process. The FEM simulation computed the strain rate of the rolling to be approximately 100 times larger than that of the ECAE along the z-axis. This result supports the idea that tensile twins can be actively generated in a high degree of compression owing to the high strain rate despite the high temperature.

4.2 Relation between strain components and the orientation of deformed basal texture

All other strain components, except ϵ_{zz} , do not directly affect the intensity of the basal texture

(Figs. 8 (a)–(e)). However, the effects of normal strains can be discussed based on their correlations. The correlations $\varepsilon_{xx} = \varepsilon_{yy} = -\varepsilon_{zz}/2$, $\varepsilon_{zz} < 0$ holds for the forged alloy, $\varepsilon_{xx} = -\varepsilon_{zz}$, $\varepsilon_{yy} \approx 0$ hold for the rolled alloy, and $\varepsilon_{yy} \approx \varepsilon_{zz} \approx -\varepsilon_{xx}/2$, $\varepsilon_{xx} > 0$ hold for the extruded, ECAE-processed, and caliber rolled alloys (Table 3). These correlations indicate that the deformation mode occurs in a uniaxial compression mode in forging, plane strain compression mode in rolling, and biaxial compression mode in the others. Compared to the basal pole figure (Fig. 3), the alloys of the former two modes have a basal texture perpendicular to the z-axis, and the alloys of the last mode have a fiber basal texture parallel to the x-axis. This result is in agreement with previous studies on crystal plasticity analysis [15, 30]. The wrought processes of the uniaxial compression mode and the plane strain compression mode impose compressive strain along the z-axis, whereas that of the biaxial compression mode compresses around the x-axis. Because the basal plane is perpendicular to the compression, the basal texture depends on the deformation mode.

Although the deformation mode is discussed only in terms of the normal strain components, the ECAE-processed alloy has the same order of shear strain as the normal strain and cannot ignore the shear strain components. As the shear strain components rotate the principal axis [34], the principal axis rotates around the x-axis during two passes of ECAE owing to the shear strain and the rotation of route Bc. After the 1st pass of ECAE, considering the correlation of the normal strain ($\varepsilon_{xx} = -\varepsilon_{zz}$, $\varepsilon_{yy} \approx 0$) (Table 4), the deformation mode is the plane strain compression mode, and a basal texture perpendicular to the z-axis is formed, as is the rolled alloy (Fig. 9). The tilt angle of the principal axis from the orthogonal coordinate axis is small; the principal compressive axis is nearly parallel to the z-axis — the angle between them is 3.8°. After the 2nd pass of ECAE, the principal compressive axis rotated around the x-axis by 134°. Subtracting the 90° rotation from route Bc, the shear strain rotates the principal compressive axis by 44°, turning the plane strain compression mode into the biaxial compressive mode. This change develops a basal fiber texture parallel to the x-axis.

As mentioned above, basal texture development is due to the compressive and deformation modes. To determine the direction of the rotation caused by deformation behavior, three directions (i.e., the slip direction, the orientation of the slip plane, and the compressive direction) are necessary. The first and the second were obtained by the observation of crystallographic orientation, and the third was identified by the results of FEM. However, $\bar{\varepsilon}$ does not contain any information regarding the compressive direction. Because of the lack of orientation information, the texture development cannot be directly responsible for $\bar{\varepsilon}$ (Fig. 8(f)).

4.3 Quantification of the deformation mode with the principal strain

The basal texture development is mainly due to compressive strains. However, the effect of the wrought process on the microstructure cannot be related to one strain component in the orthogonal coordinate (the xyz-coordinate), because the driving force for the DRX is the stored strain energy. As

energy increases with an increase in the degree of the applied process, the relationship between DRX and the wrought process has been discussed in terms of $\bar{\epsilon}$ [32, 34, 38, 53]. Nevertheless, in this study, different microstructures were observed in wrought processed alloys with approximately the same $\bar{\epsilon}$ (Fig. 1). This result suggests that even if the same $\bar{\epsilon}$ is induced, the different microstructures evolve owing to the influence of the deformation mode. The deformation mode is discussed separately for the effects of the normal strain components and the shear components in Section 4.2. However, as all components contribute to the strain energy, the microstructure evolution should be discussed not by the individual strain component but by the deformation mode, considering the effect of all strain components.

The deformation mode is expressed by the correlation of the strain components; however, there are generally six strain components, and their correlation is complicated. In this study, the principal strains were calculated from the strain components listed in Table 3 and were used to assess the deformation mode. The principal strains are listed in Table 5. Using the principal strain components, every strain state was converted into a strain state without shear strain components [34]. The deformation mode can be expressed by the correlation of three normal strain components (i.e., the maximum principal strain ϵ_1 , the middle principal strain ϵ_2 , and the minimum principal strain ϵ_3). Furthermore, as introduced in the Appendix, these principal strains are related to each other; consequently, two of them are expressed as the functions of the other. Based on this dependence on the principal strain components, the deformation mode can be discussed using only one principal strain component. As the compressive component is assumed to be effective for the microstructure evolution, the deformation mode is assessed in terms of ϵ_3 .

Fig. 10 shows that ϵ_1 and ϵ_2 are functions of ϵ_3 presented by Eqs. (A-12) and (A-13) in the Appendix under the condition of $\epsilon_{eq} = 1.5$. At the minimum ϵ_3 ($\epsilon_3 = -\epsilon_{eq}$), ϵ_1 and ϵ_2 are equal to $\epsilon_{eq}/2$, which means that the deformation mode is uniaxial compression. With an increase in ϵ_3 , the value of ϵ_1 increases, and that of ϵ_2 decreases. When the value of ϵ_3 is $-\sqrt{3}\epsilon_{eq}/2$, the correlations ($\epsilon_1 = -\epsilon_3$ and $\epsilon_2 = 0$) are established, and the deformation mode is the plane strain compression mode. At the maximum ϵ_3 ($\epsilon_3 = \epsilon_{eq}/2$), the other strain components lead to $\epsilon_1 = \epsilon_{eq}$ and $\epsilon_2 = -\epsilon_{eq}/2$, this deformation mode corresponds to the biaxial compression mode. Owing to Eqs. (A-12) and (A-13), the deformation mode changes from the uniaxial compression mode through the plane strain compression mode into the biaxial compression mode, as ϵ_3 increases.

However, the correlations between the principal strains varied according to the values of ϵ_{eq} obtained using Eqs. (A-12) and (A-13). Fig. 11 shows the correlation between the principal strains in each wrought process. The curves indicate the correlations of the principal strains in ϵ_{eq} for each wrought process, and the points indicated by the symbols correspond to the principal strains listed in Table 5. Although these wrought processes induce approximately the same $\bar{\epsilon}$, as ϵ_{eq} is obtained from the strain components using Eq. (A-8), ϵ_{eq} is influenced by the deformation mode. In the

uniaxial compression mode, which is forging, the principal axis direction does not rotate; consequently, strain is accumulated along the same direction. Conversely, during multidirectional deformation modes, such as caliber rolling, the principal axis direction rotates, and the reversed strain is imposed, resulting in a decrease in the magnitudes of the strain component. Accordingly, the decrease of ε_{eq} implies the transition from the uniaxial compression mode to the multiaxial deformation mode when approximately the same $\bar{\varepsilon}$ is induced. Simultaneously, with this decrease, the domain of ε_3 approaches the axis of $\varepsilon_3 = 0$ in Fig. 11 by the relationship presented in Eq. (A-14). Therefore, as long as a near value of $\bar{\varepsilon}$ is induced, the lower magnitude of ε_3 corresponds to the change from the uniaxial compression mode to the multidirectional deformation mode. Comparing the magnitudes of ε_3 , the multiaxiality of the deformation mode is more significant for caliber rolling, extrusion, ECAE, rolling, and forging.

Using only the correlation of the principal strain components, the deformation modes of the rolling and ECAE are classified as plane strain compression mode because the correlation of $\varepsilon_1 \approx -\varepsilon_3$ is established. However, this result is inconsistent with the fact that the different textures are obtained in these two processes owing to the difference in the deformation mode (4.1). Conversely, as assessed by the magnitude of ε_3 , the deformation mode of ECAE is closer to that of extrusion than that of rolling. This is consistent with the result that similar basal textures were developed in wrought processes with similar deformation modes. Therefore, the magnitude of ε_3 can quantify the deformation mode.

4.4 Effect of deformation mode on microstructure evolution

Fig.12 illustrates the relationship between the magnitude of ε_3 and grain size. Error bars indicate the standard deviation of the grain size. The average grain size and its deviation decreased with decreasing magnitude of ε_3 . This tendency indicates that the multidirectional deformation mode developed a more refined and homogeneous microstructure than the unidirectional compression mode. The achievement of microstructure refinement by multidirectional deformation has been attributed to the introduction of a larger $\bar{\varepsilon}$ [32]. However, since the applied $\bar{\varepsilon}$ is the same during all the wrought processes, the microstructural evolution cannot be related to $\bar{\varepsilon}$. Thus, the tendency in Fig.12 shows that the difference in the deformation mode influences the microstructural evolution, apart from the effect of $\bar{\varepsilon}$.

The forged alloy, which has the largest ε_3 , shows a large peak corresponding to the tensile twin boundaries ($86 \pm 5^\circ$) in the misorientation distribution observed by EBSD (Fig. 13). In the rolled alloy, even though a large peak is not shown at $86 \pm 5^\circ$ in Fig. 13, the twin is formed actively in the early stage of the process (Fig. 7(a)). This suggests that the tensile twins grow and consume the matrix with the number of passes in the rolling, resulting in the disappearance of the boundaries between the matrix and the twin. The observations mean that the tensile twin is activated by deformation modes with a large compressive strain. Conversely, there are seldom twins in the alloys of the multidirectional

deformation mode. Thus, the microstructure evolution in the processes with a large compressive strain is more influenced by the tensile twin than that in the other processes.

In the coarsest microstructure, the forged alloy, refined grains with a size of several micrometers are present around the coarse grains (Fig. 1(a)). The boundaries of the tensile twin were observed near the boundaries of a coarse grain with the basal plane perpendicular to the CD; furthermore, there are boundaries of $\{10\bar{1}1\}$ twins with a misorientation angle of $56 \pm 5^\circ$ (Fig. 14). Jin et al. [9] reported that the grains with c-axis parallel to the compression direction remain no recrystallized, while the surrounding microstructure is replaced by DRXed grains. The coarse grain in Fig. 14 is considered a grain formed during the initial deformation, and its surroundings are a twinned area. It should be noted that the $\{10\bar{1}2\}$ twin is formed under the loading condition of the c-axis compression in Fig. 14. The $\{10\bar{1}2\}$ twin is formed in tension along the c-axis or in compression with a direction parallel to the basal plane because this twin generates a tensile strain along the c-axis. Nevertheless, Koike et al. [54] demonstrated that this twin operates under compression along the c-axis to relax the strain concentration caused by strain incompatibility in the grain boundary. As the $\{10\bar{1}2\}$ twins are formed with the aid of this stress concentration, the twin boundaries are observed only near the grain boundary. The twinned region is known to be a preferential nucleation site for Mg [31, 56]. The refined grains in the forged alloy are considered to be DRXed grains induced the twins. Fig. 14 shows the operation of another twin— $\{10\bar{1}1\}$ twin. Several studies have shown that the inside of this twin has high accumulated strain and DRX occurs actively, while the mobility of this twin boundary is low and its effect on the microstructure evolution is ineffective [16, 55, 56]. Moreover, the peak indicating this twin was not large, as shown in Fig. 12. Therefore, the $\{10\bar{1}1\}$ twin was formed locally, and the effect of this twin was not significant.

Based on microstructure observations, DRX is limited in the vicinity of the grain boundary, the coarse grains remain, and the average grain size and its deviation are large in the forged alloy. The mixture of coarse un-DRXed grains and fine DRXed grains has been reported to be attributed to the inhibition of the grain growth by the pinning effect and the particle-stimulated nucleation (PSN) in various Mg alloys [57, 58, 59]. In Al-rich AZ series Mg alloy (AZ61, and AZ91), PSN is caused by the $Mg_{17}Al_{12}$ particles [60, 61, 62]. Conversely, to the best of our observations, AZ31 usually has very limited formation of the $Mg_{17}Al_{12}$ phase [60, 61, 63]. Actually, the $Mg_{17}Al_{12}$ phase particles were hardly observed by optical microscope and SEM. In the present work, the effect of the $Mg_{17}Al_{12}$ phase on the microstructure evolution is ignored. The difficulty of DRX in the core part of the coarse grain is due to the low activity of the prismatic slip system attributed to the orientation relation. In the coarse grain, as the c-axis is parallel to the compression, the $\langle c+a \rangle$ slip is favored, while the prismatic slip is not. The results of the crystal plasticity analysis showed that the relative activity of the prismatic slip was significantly smaller than that of the basal slip and the $\langle c+a \rangle$ slip in uniaxial compression perpendicular to the basal texture [64, 65]. According to Myshelyaev, et al. [66], the activity of

multiple slip systems forms substructures and recrystallization nuclei, leading to DRX grains. Sinivasarao et al. [67] reported that the optimum DRX condition for slip systems is that all three slip systems (basal, prismatic, and $\langle c+a \rangle$ pyramidal) are active. The coarse grains in the forged alloy are attributed to the lack of prismatic slip caused by the unidirectional compression mode.

Although the compression direction was also parallel to the c-axis in the late stage of rolling, the activity of the prismatic slip was higher than that of the uniaxial compression simulation [65, 68]. Consequently, DRX occurred more actively, and a more refined microstructure was obtained. Furthermore, the activities of the two non-basal slips are almost equal and have higher value during route Bc of the ECAE [20] and extrusion [15], which belong to the multidirectional deformation mode.

The above discussion suggests the following relations between the deformation mode and DRX phenomena: Under the uniaxial and plane strain compression, to achieve a large compressive strain in a direction, the tensile twin operates frequently in addition to the slip activities, leading to twin-induced DRX and evolution of heterogeneous grained microstructure with intensive basal texture. On the other hand, under the multidirectional deformation mode, the non-basal slips are preferred to the twins, resulting in the enhancement of DRX and the development of a more refined microstructure with weak basal texture. Generally, microstructure evolution during ECAE has been attributed to the shear deformation [17]. However, from the viewpoint of deformation mode proposed in this study, ECAE is categorized as multidirectional deformation as well as extrusion and caliber rolling, and their processed alloys have similar microstructure. This suggests that the DRX phenomena depend on deformation mode rather than a specific strain component such as a shear strain.

5. Conclusions

AZ31 Mg alloys were deformed by five wrought processes: forging, rolling, extrusion, ECAE, and caliber rolling, which introduced an equivalent strain of approximately 1.5 at 573 K. The strain components induced during them were computed using FEM. Based on the calculation results and microstructural observations, the following conclusions were obtained:

- (1) The caliber rolled alloy had the most refined and homogenized grain structure with an average size of 5.4 μm . Conversely, the forged alloy was composed of grains of non-homogeneous sizes. Each processed alloy exhibited a different basal texture. In the forged and rolled alloys, the intensive basal texture was developed perpendicular to the CD and RD, respectively. In the other cases, the basal fiber texture formed parallel to the material flow direction, which was weaker than the texture in the forged and rolled alloy.
- (2) The FEM results demonstrated that the induced close value of the equivalent plastic strain was in the range of 1.65–1.9 at the center of each alloy, where the microstructure was observed. By analyzing the strain component, it was found that the basal texture developed perpendicular to the compressive direction in all five wrought processes. The intensity of the texture increases as the compressive strain

increases.

(3) The magnitude of the minimum principal strain can assess the deformation mode (e.g., unidirectional compression mode, plane strain compression mode, etc.). The decrease in the value corresponds to the transition of the deformation mode from unidirectional compression to multidirectional deformation. The grain size tends to be refined as the value decreases. This indicates that the multidirectional deformation mode is more effective for microstructure refinement than the unidirectional compression mode.

Originality statement

I write on behalf of myself and all co-authors to confirm that the results reported in the manuscript are original and neither the entire work, nor any of its parts have been previously published. The authors confirm that the article has not been submitted to peer review, nor has been accepted for publishing in another journal. The authors confirm that the research in their work is original, and that all the data given in the article are real and authentic. If necessary, the article can be recalled, and errors corrected.

CRedit authorship contribution statement

Hiromasa Yoshizumi: Conceptualization, Investigation, Data curation, Validation, Writing – original draft. **Motohiro Yuasa:** Investigation, Validation, Writing – review & editing. **Hiroyuki Miyamoto:** Validation, Writing – review & editing. **Hidetoshi Somekawa:** Validation, Writing – review & editing.

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.

Data availability

Data will be made available on request.

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Appendix A

In this appendix, the correlations between the principal strain components are introduced. The principal strains are the solutions λ of the following equation:

$$\det(\boldsymbol{\varepsilon} - \lambda \mathbf{I}) = 0, \quad (\text{A-1})$$

where $\boldsymbol{\varepsilon}$ is the strain tensor:

$$\boldsymbol{\varepsilon} = \begin{pmatrix} \varepsilon_{xx} & \varepsilon_{xy} & \varepsilon_{xz} \\ \varepsilon_{yx} & \varepsilon_{yy} & \varepsilon_{yz} \\ \varepsilon_{zx} & \varepsilon_{zy} & \varepsilon_{zz} \end{pmatrix}, \quad (\text{A-2})$$

and $\mathbf{I}$ is a unit matrix. Calculated from Eq. (A-1), the following relation is derived:

$$\lambda^3 - (\varepsilon_{xx} + \varepsilon_{yy} + \varepsilon_{zz})\lambda^2 + (\varepsilon_{xx}\varepsilon_{yy} + \varepsilon_{yy}\varepsilon_{zz} + \varepsilon_{zz}\varepsilon_{xx} - \varepsilon_{xy}^2 - \varepsilon_{yz}^2 - \varepsilon_{zx}^2)\lambda - \det(\boldsymbol{\varepsilon}) = 0. \quad (\text{A-3})$$

From the law of volume conservation ($\varepsilon_{xx} + \varepsilon_{yy} + \varepsilon_{zz} = 0$),

$$(\varepsilon_{xx} + \varepsilon_{yy} + \varepsilon_{zz})^2 = 0, \quad (\text{A-4})$$

$$\varepsilon_{xx}\varepsilon_{yy} + \varepsilon_{yy}\varepsilon_{zz} + \varepsilon_{zz}\varepsilon_{xx} = -\frac{1}{2}(\varepsilon_{xx}^2 + \varepsilon_{yy}^2 + \varepsilon_{zz}^2). \quad (\text{A-5})$$

The shear components of the strain tensor are those of the nominal strain γ_{ij} :

$$\gamma_{ij} = \frac{1}{2}\varepsilon_{ij}. \quad (\text{A-6})$$

By Eqs. (A-5) and (A-6), the coefficient of the primary term of λ in Eq. (A-3) is

$$\begin{aligned} & \varepsilon_{xx}\varepsilon_{yy} + \varepsilon_{yy}\varepsilon_{zz} + \varepsilon_{zz}\varepsilon_{xx} - \varepsilon_{xy}^2 - \varepsilon_{yz}^2 - \varepsilon_{zx}^2 \\ &= -\frac{3}{4} \cdot \frac{2}{3} \left[(\varepsilon_{xx}^2 + \varepsilon_{yy}^2 + \varepsilon_{zz}^2) + \frac{1}{2} (\gamma_{xy}^2 + \gamma_{yz}^2 + \gamma_{zx}^2) \right] \\ &= -\frac{3}{4} \varepsilon_{eq}^2, \end{aligned} \quad (\text{A-7})$$

where the ε_{eq} is the von-Mises equivalent plastic strain defined by

$$\varepsilon_{eq} = \sqrt{\frac{2}{3} \left[(\varepsilon_{xx}^2 + \varepsilon_{yy}^2 + \varepsilon_{zz}^2) + \frac{1}{2} (\gamma_{xy}^2 + \gamma_{yz}^2 + \gamma_{zx}^2) \right]}. \quad (\text{A-8})$$

Therefore, considering the law of volume conservation and Eq. (A-4), Eq. (A-3) is rearranged as

$$\lambda^3 - \frac{3}{4} \varepsilon_{eq}^2 \lambda - \det(\boldsymbol{\varepsilon}) = 0. \quad (\text{A-9})$$

As the solutions of Eq. (A-9) are the principal strains (i.e., $\varepsilon_1, \varepsilon_2, \varepsilon_3$), the left-hand side of Eq. (A-9) is converted as follows:

$$\begin{aligned} \lambda^3 - \frac{3}{4} \varepsilon_{eq}^2 \lambda - \det(\boldsymbol{\varepsilon}) &= (\lambda - \varepsilon_1)(\lambda - \varepsilon_2)(\lambda - \varepsilon_3) \\ &= \lambda^3 - (\varepsilon_1 + \varepsilon_2 + \varepsilon_3)\lambda^2 + (\varepsilon_1\varepsilon_2 + \varepsilon_2\varepsilon_3 + \varepsilon_3\varepsilon_1)\lambda - \varepsilon_1\varepsilon_2\varepsilon_3. \end{aligned} \quad (\text{A-10})$$

Comparing the coefficient of the primary term of λ , the following equation is obtained:

$$\varepsilon_1\varepsilon_2 + \varepsilon_2\varepsilon_3 + \varepsilon_3\varepsilon_1 = -\frac{3}{4} \varepsilon_{eq}^2, \quad (\text{A-11})$$

By using the law of volume conservation ($\varepsilon_1 + \varepsilon_2 + \varepsilon_3 = 0$) in Eq. (A-11), ε_1 and ε_2 can be expressed as the following functions of ε_3 :

$$\varepsilon_1 = \frac{1}{2} \left(-\varepsilon_3 + \sqrt{3(\varepsilon_{eq}^2 - \varepsilon_3^2)} \right), \quad (\text{A-12})$$

$$\varepsilon_2 = \frac{1}{2} \left(-\varepsilon_3 - \sqrt{3(\varepsilon_{eq}^2 - \varepsilon_3^2)} \right). \quad (\text{A-13})$$

To satisfy the relation such as $\varepsilon_1 > \varepsilon_2 > \varepsilon_3$, the domain of ε_3 is below:

$$-\varepsilon_{eq} < \varepsilon_3 < -\varepsilon_{eq}/2. \quad (\text{A-14})$$

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Captions

Figures

Fig. 1: Grain boundary maps and inverse pole figure maps of (a) and (b) the forged alloy observed from CD, (c) and (d) the rolled alloy observed from ND, (e) and (f) the extruded alloy observed from ND, (g) and (h) the ECAE-processed alloy observed from ND, and (i) and (j) the caliber rolled alloy observed from ND. The red and blue lines correspond to high-angle grain boundaries with misorientations larger than 15° and low-angle grain boundaries with misorientations between 4° and 15° .

Fig. 2: Cumulative fraction of grain size in the forged, rolled, and caliber rolled alloy.

Fig. 3: (0002) pole figures of (a) the forged alloy, (b) the rolled alloy, (c) the extruded alloy, (d) the ECAE-processed alloy, and (e) the caliber rolled alloy.

Fig. 4: XRD patterns of (a) the forged alloy, (b) the rolled alloy, (c) the extruded alloy, (d) the ECAE-

processed alloy, and (e) the caliber rolled alloy.

Fig. 5: Schematics of (a) the regions where FEM results are analyzed, (b) contour maps of equivalent plastic strain, and (c) component of strain in the z-direction.

Fig. 6: Relationship between compressive strain along the z-axis and orientation index of the basal plane.

Fig. 7: Microstructures of (a) the rolled alloy after the 7th pass observed from ND and (b) the ECAE-processed alloy after one pass observed from ED.

Fig. 8: Orientation index of the basal plane is plotted against (a) the component of strain in the x-direction, (b) the component of strain in the y-direction, (c) shear strain in the xy plane, (d) shear strain in the yz plane, (e) shear strain in the zx plane, and (f) equivalent plastic strain.

Fig. 9: (a) Grain boundary map, (b) inverse pole figure map, (c) (0002) basal pole figure, and (d) XRD patterns of the ECAE-processed alloy after one pass. (a), (b), and (c) were observed from ND by EBSD.

Fig. 10: The maximum principal strain and the middle principal strain are plotted against the minimum principal strain at the von-Mises equivalent plastic strain of 1.5.

Fig. 11: The maximum principal strain and the middle principal strain are plotted against the minimum principal strain at the von-Mises equivalent plastic strain of each wrought process. The place indicated by the symbol corresponds to the strain state listed in table 5.

Fig. 12: Relationship between the magnitude of the minimum principal strain and grain size. Error bar indicates the standard deviation of grain size.

Fig. 13: Misorientation distribution of the forged, rolled, and caliber rolled alloy.

Fig. 14: Grain boundary map and inverse pole figure of the forged alloy observed by higher magnification than Fig. 1 (a) and (b).

Tables

Table 1: Processing condition and induced equivalent plastic strain in wrought processes

	condition	number of passes	equivalent strain
forging	80 % reduction compression speed 11 mm/s	-	1.61
rolling	10% reduction /pass outer speed of roller 0.2 m/s	14	1.70
extrusion	extrusion ratio 5 extrusion speed 0.5 mm/s	-	1.61
ECAE	$\phi = 110^\circ$ and $\psi = 20^\circ$ [35] extrusion speed 0.1 mm/s	2 passes in route Bc	1.52
caliber rolling	C = 40.0 – 25.9 mm [25] outer speed of roller 0.5 m/s	6	1.27

Table 2: Orientation index of the prismatic plane, basal plane, and pyramidal plane obtained from two directions

	prismatic plane	basal plane	pyramidal plane
forging CD	0.3	9.3	0.8
forging TD	2.4	0.6	1.9
rolling ND	0.1	10.5	0.4
rolling TD	4.0	0.3	1.1
extrusion ND	0.9	2.8	2.1
extrusion ED	4.9	0.3	1.6
ECAE ND	1.2	2.5	2.2
ECAE ED	2.9	0.3	1.7
caliber rolling ND	0.3	4.0	1.1
caliber rolling RD	3.2	< 0.1	1.4

Table 3: Equivalent strain and strain components at the center of each processed alloy computed by FEM

	$\bar{\varepsilon}$	ε_{xx}	ε_{yy}	ε_{zz}	γ_{xy}	γ_{yz}	γ_{zx}
forging	1.87	0.94	0.93	-1.87	< 0.01	0.01	< 0.01
rolling	1.90	1.36	0.09	-1.44	0.02	< 0.01	0.03
extrusion	1.67	1.45	-0.72	-0.73	-0.04	< 0.01	0.06
ECAE	1.68	0.50	-0.24	-0.26	0.94	-0.56	0.83
caliber rolling	1.65	0.84	-0.31	-0.47	0.09	0.01	-0.08

Table 4: FEM result of the ECAE-processed alloy after one pass

$\bar{\varepsilon}$	ε_{xx}	ε_{yy}	ε_{zz}	γ_{xy}	γ_{yz}	γ_{zx}
0.68	0.57	< 0.01	-0.58	-0.01	-0.01	0.50

Table 5: Principal strains and the von-Mises equivalent plastic strain calculated from strain components in Table 3

	ε_1	ε_2	ε_3	ε_{eq}
forging	0.94	0.93	-1.87	1.87
rolling	1.36	0.09	-1.44	1.62
extrusion	1.45	-0.72	-0.73	1.45
ECAE	0.80	0.03	-0.83	0.94
caliber rolling	0.84	-0.31	-0.47	0.82

Figure 1

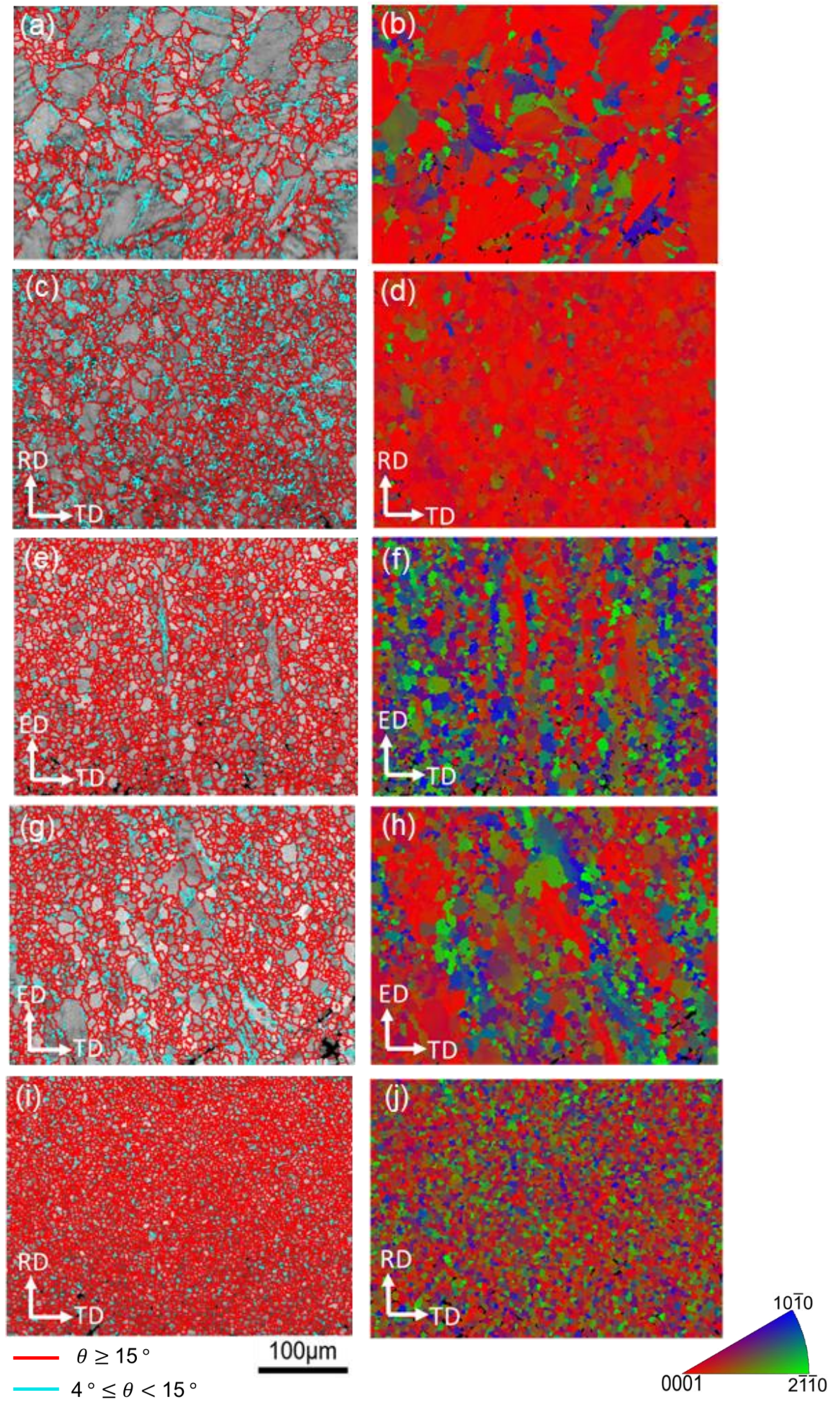


Figure 2

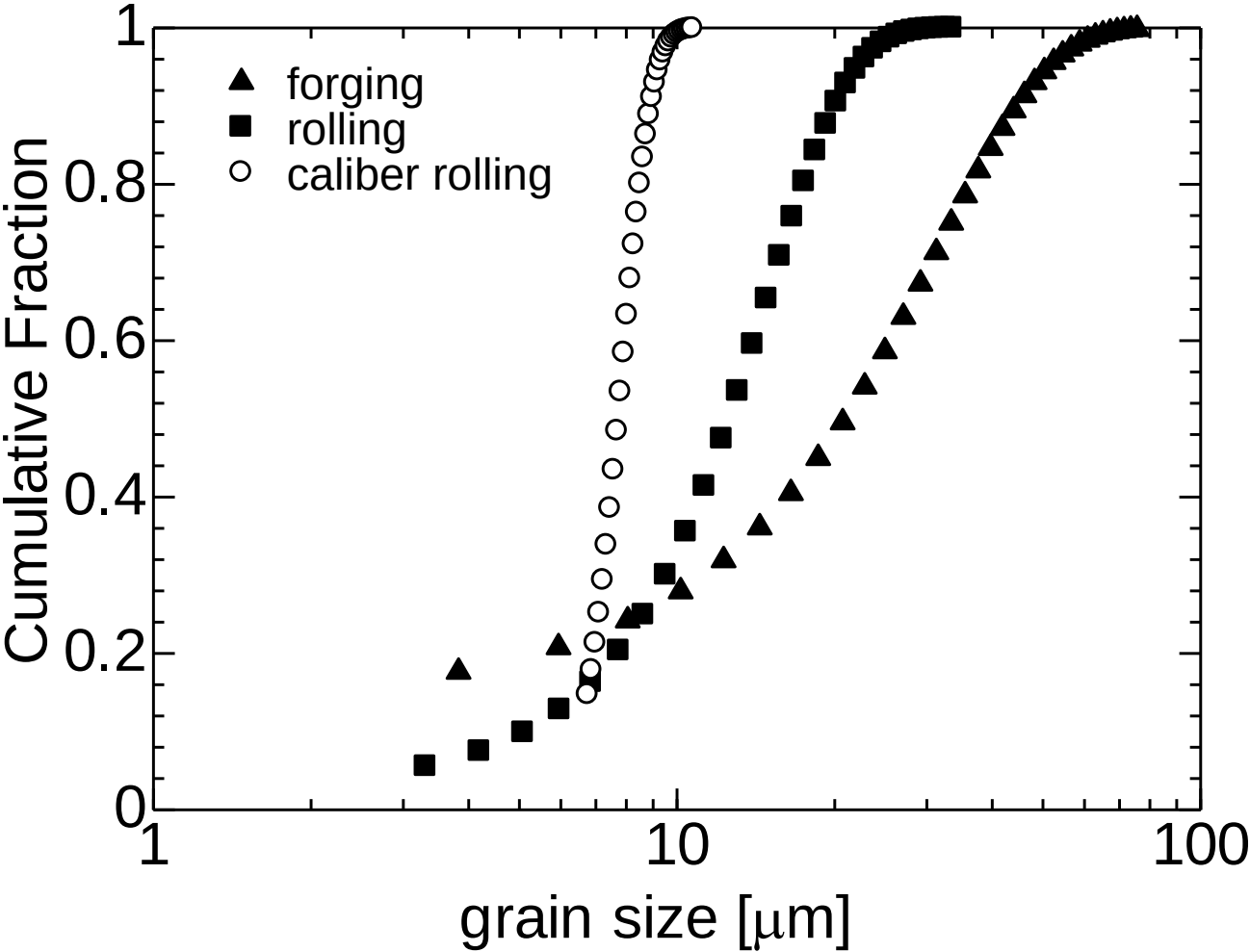


Figure 3

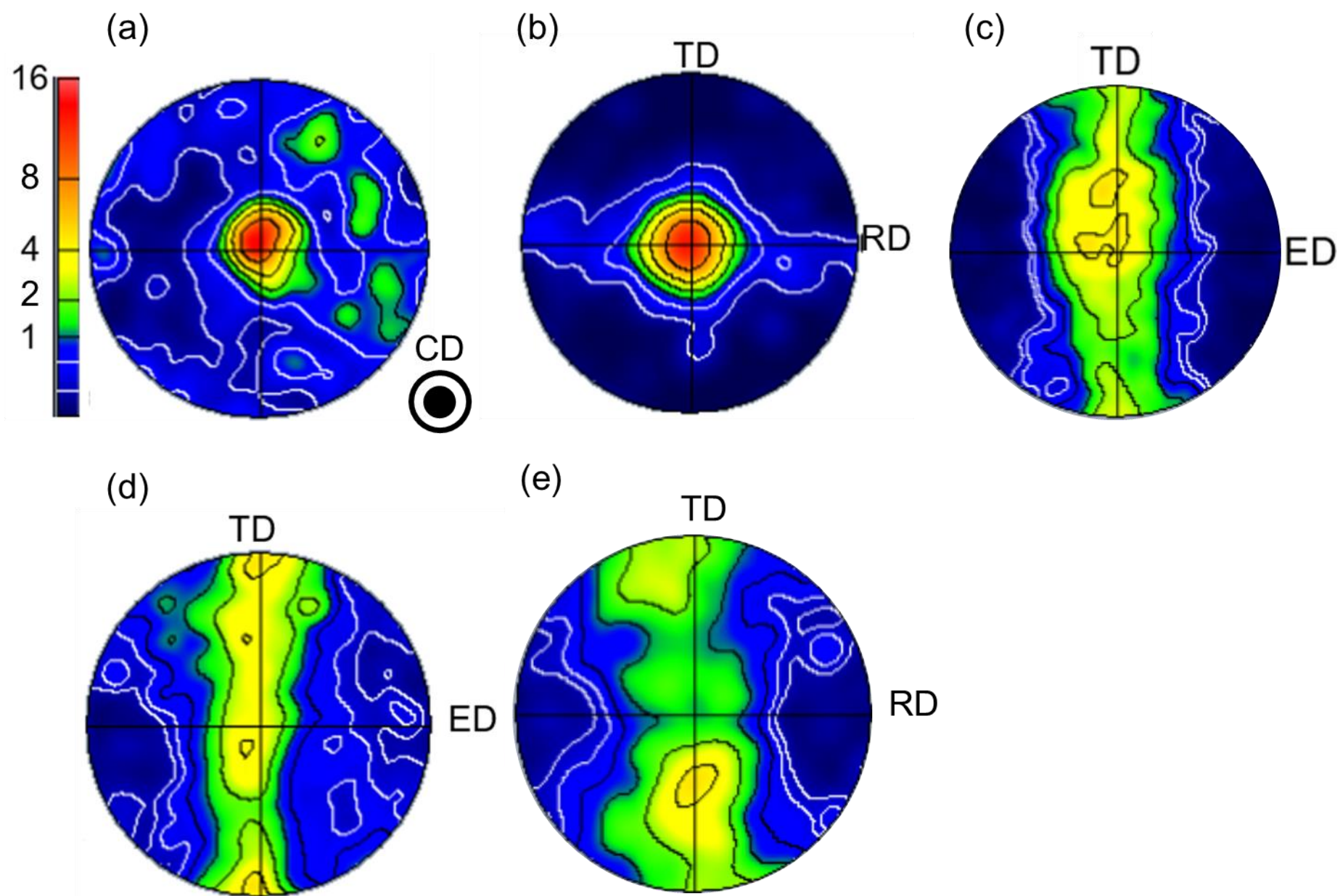


Figure 4

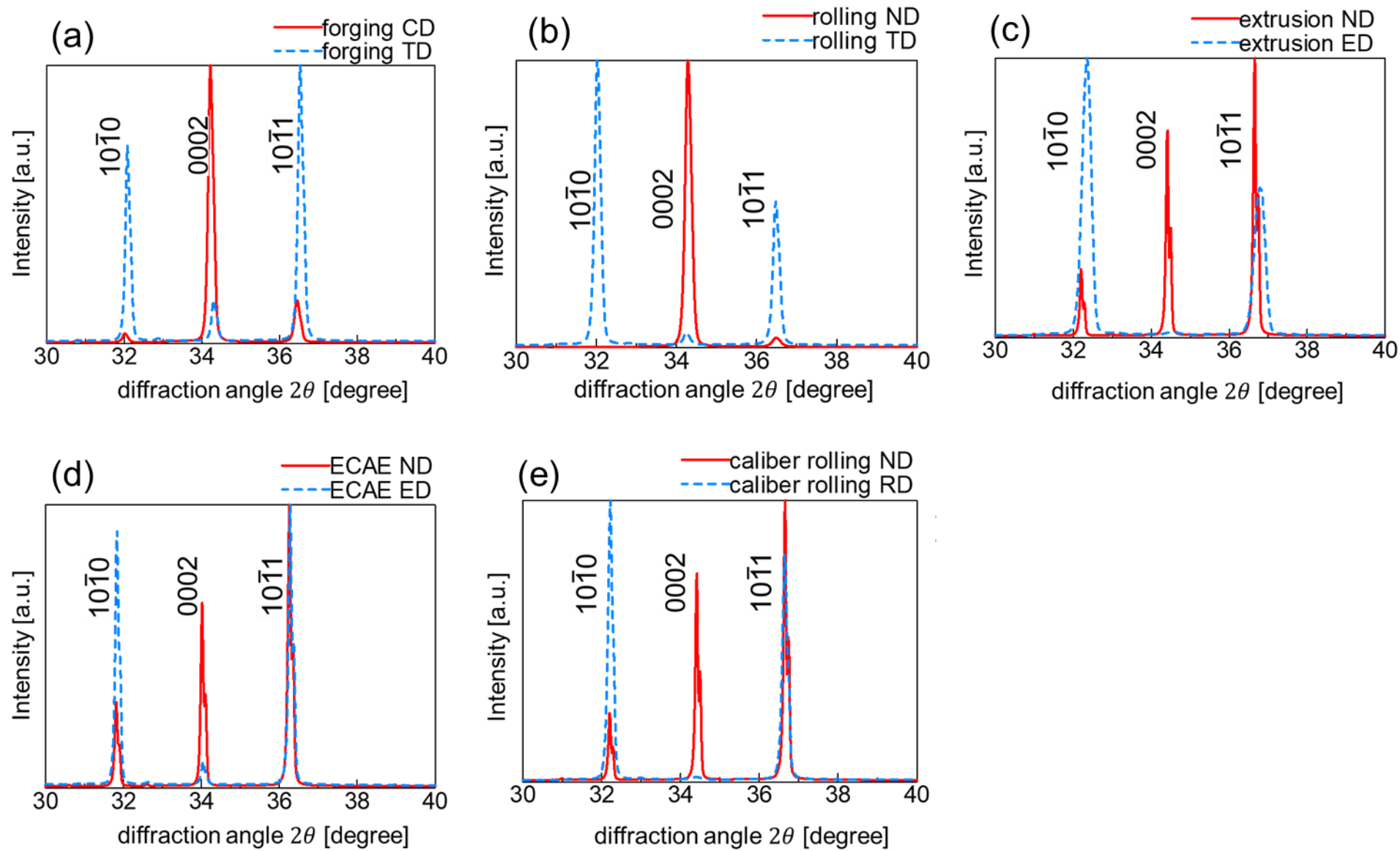


Figure 5

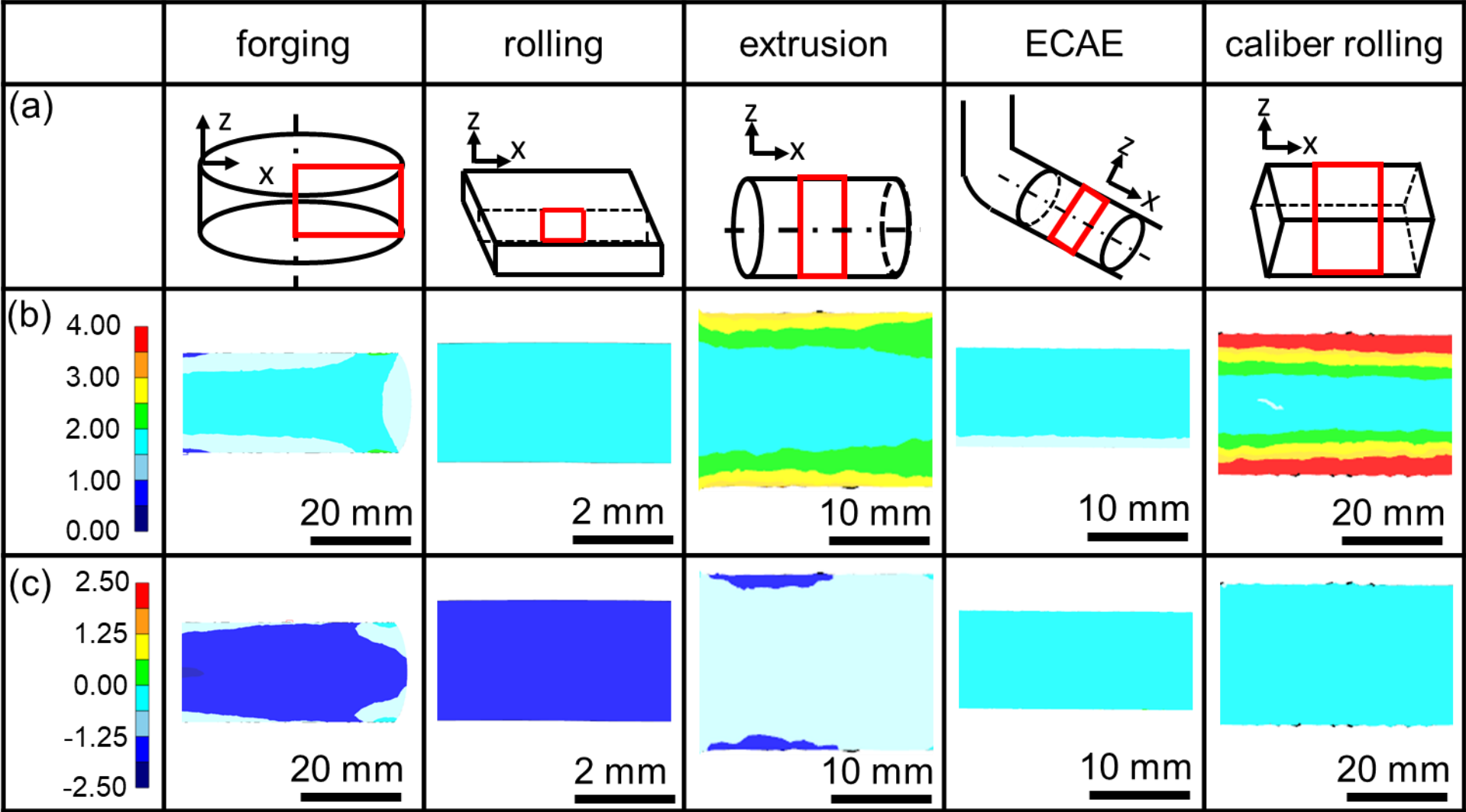


Figure 6

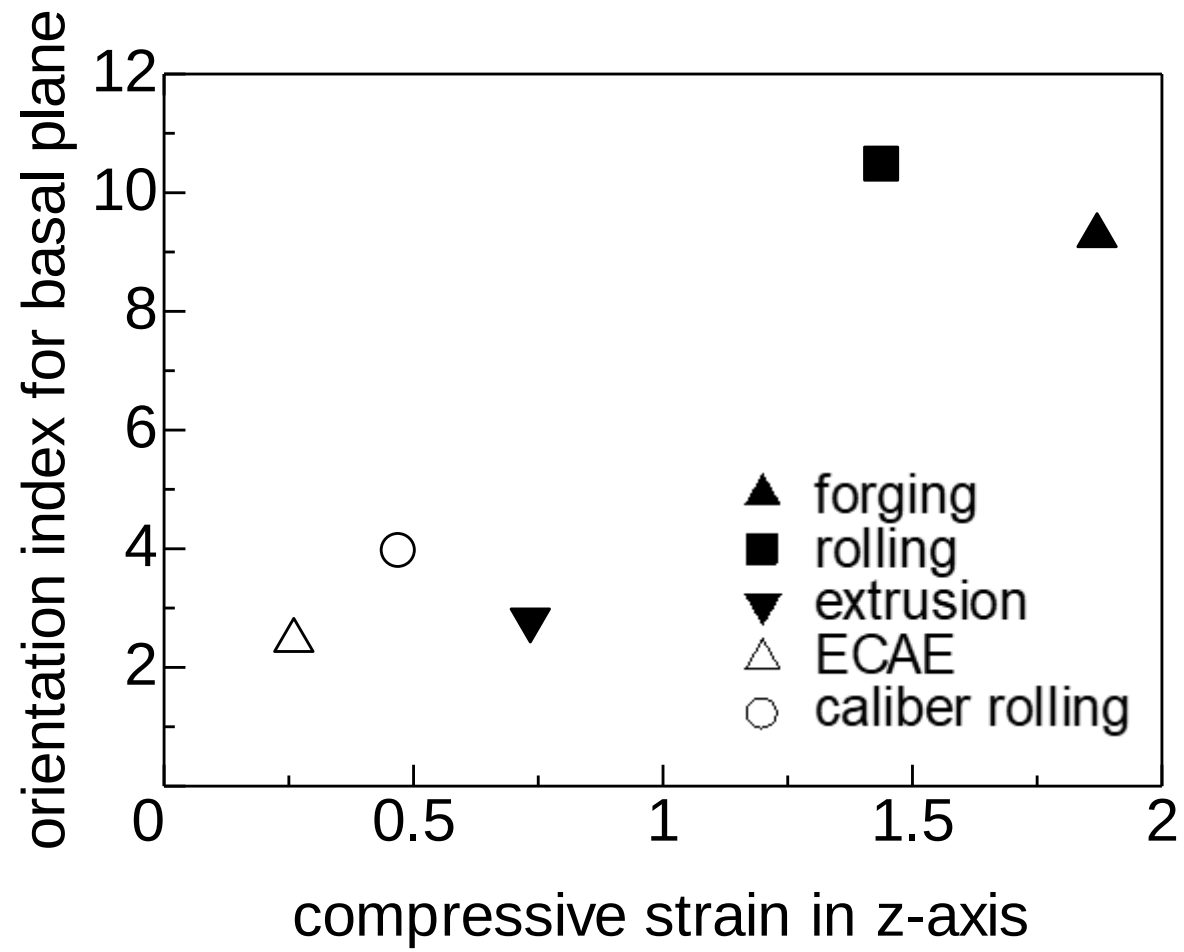


Figure 7

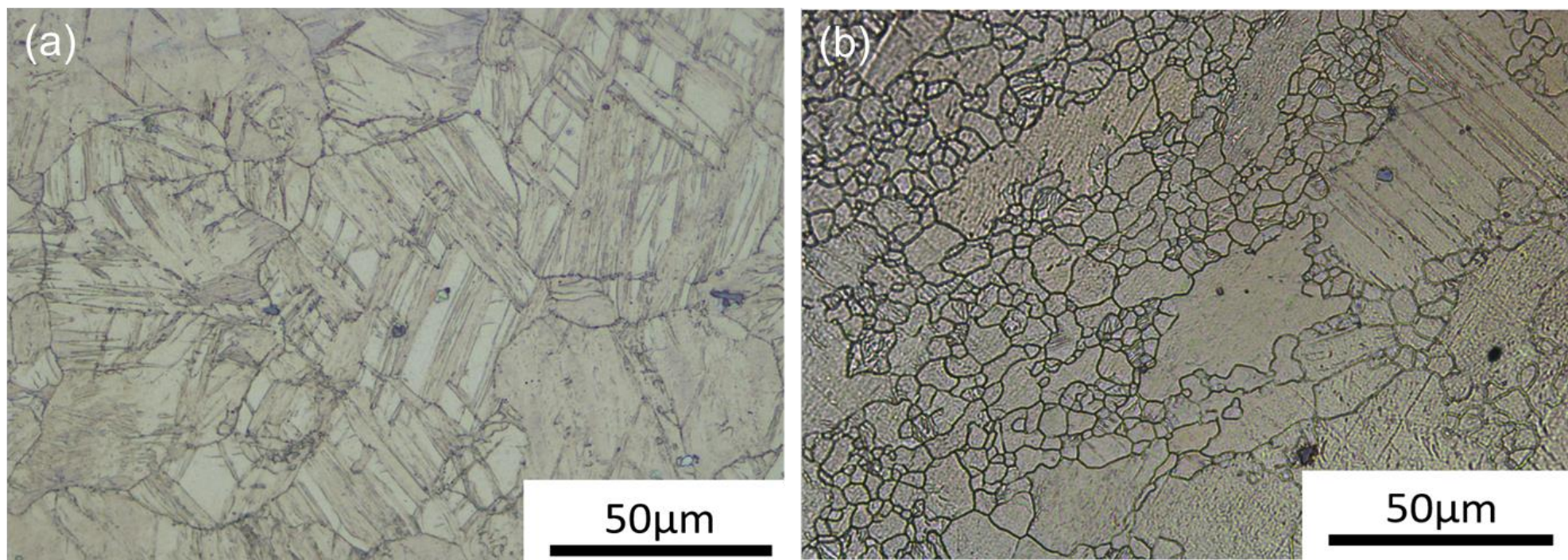


Figure 8

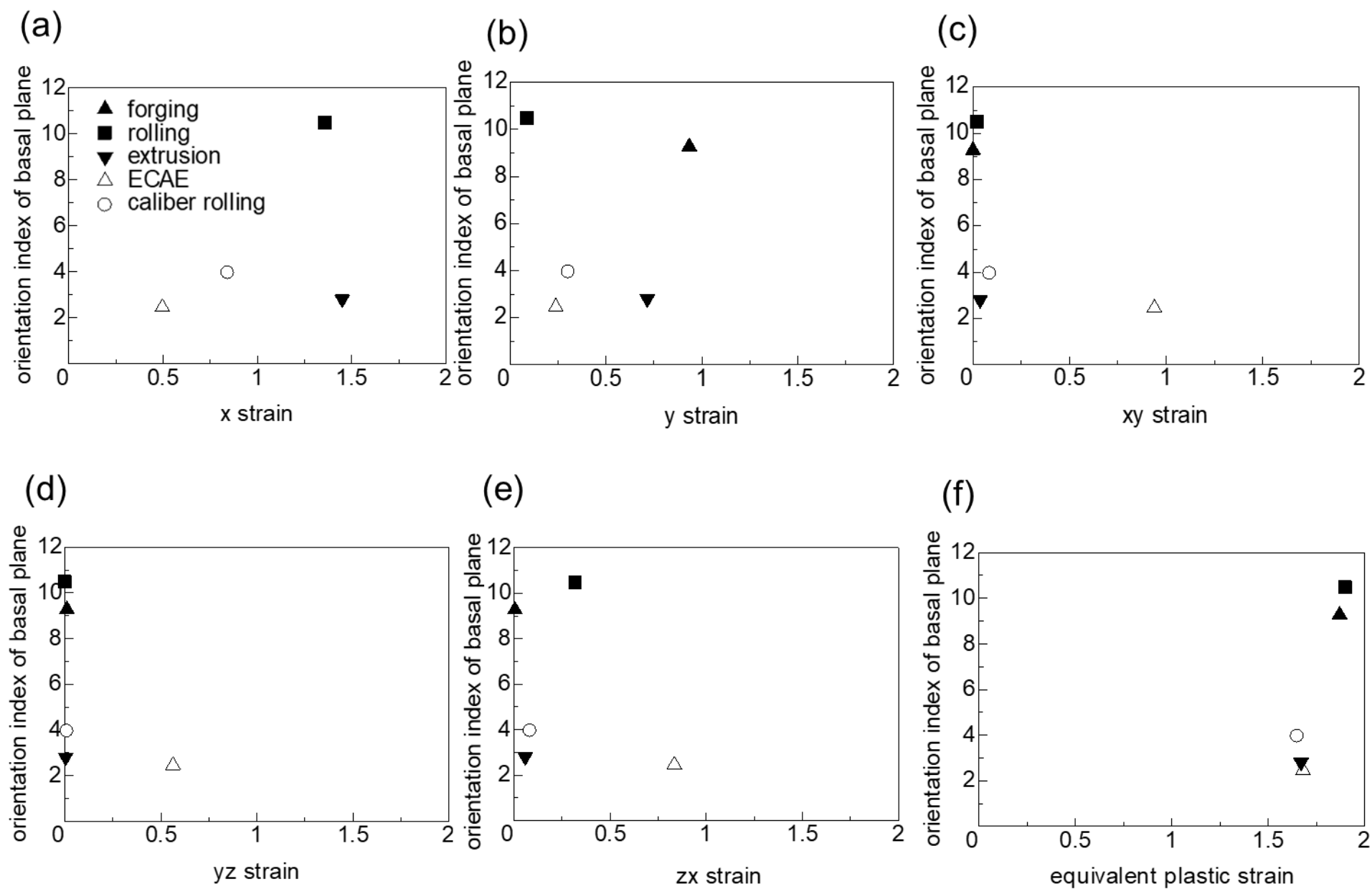


Figure 9

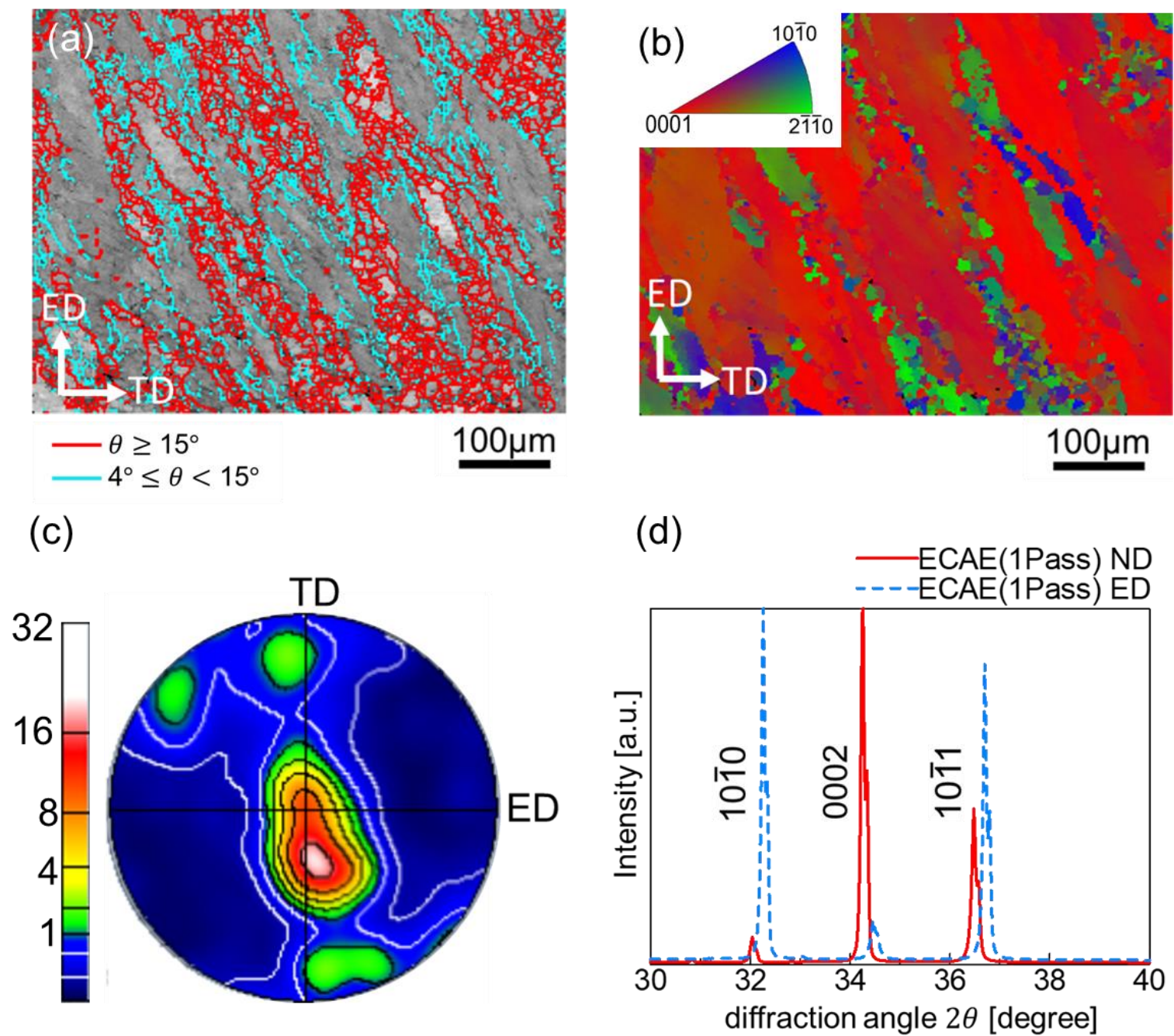


Figure 10

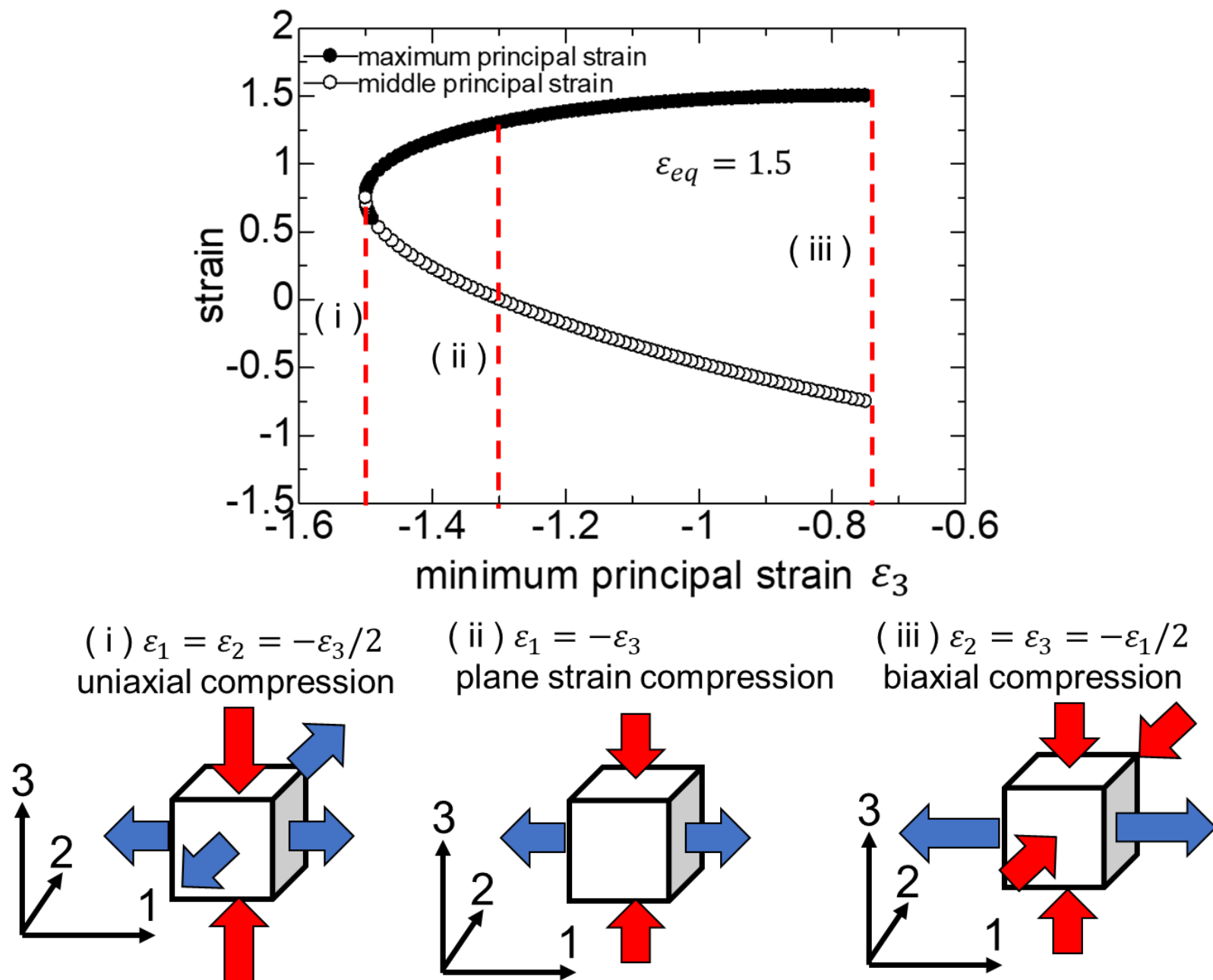


Figure 11

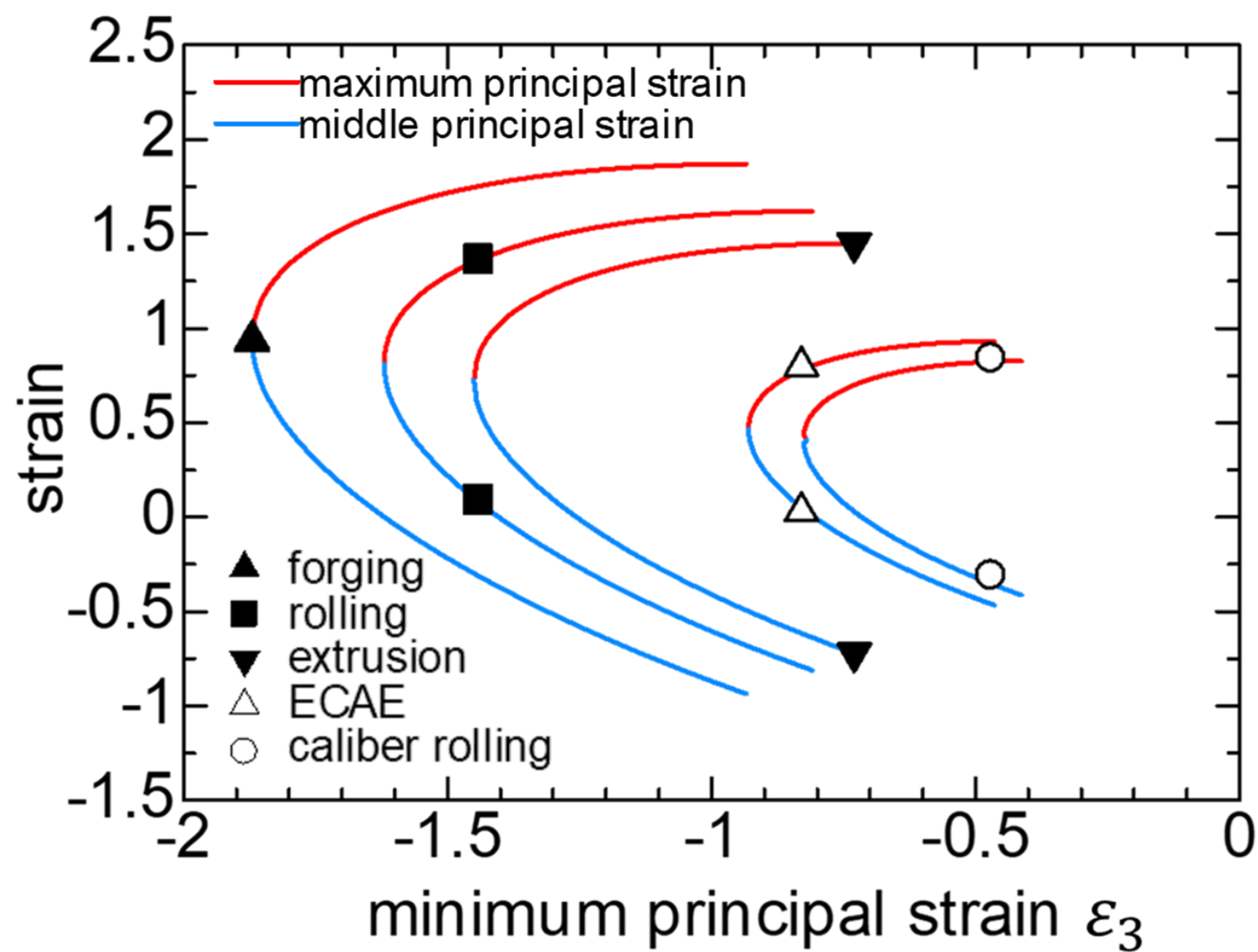


Figure 12

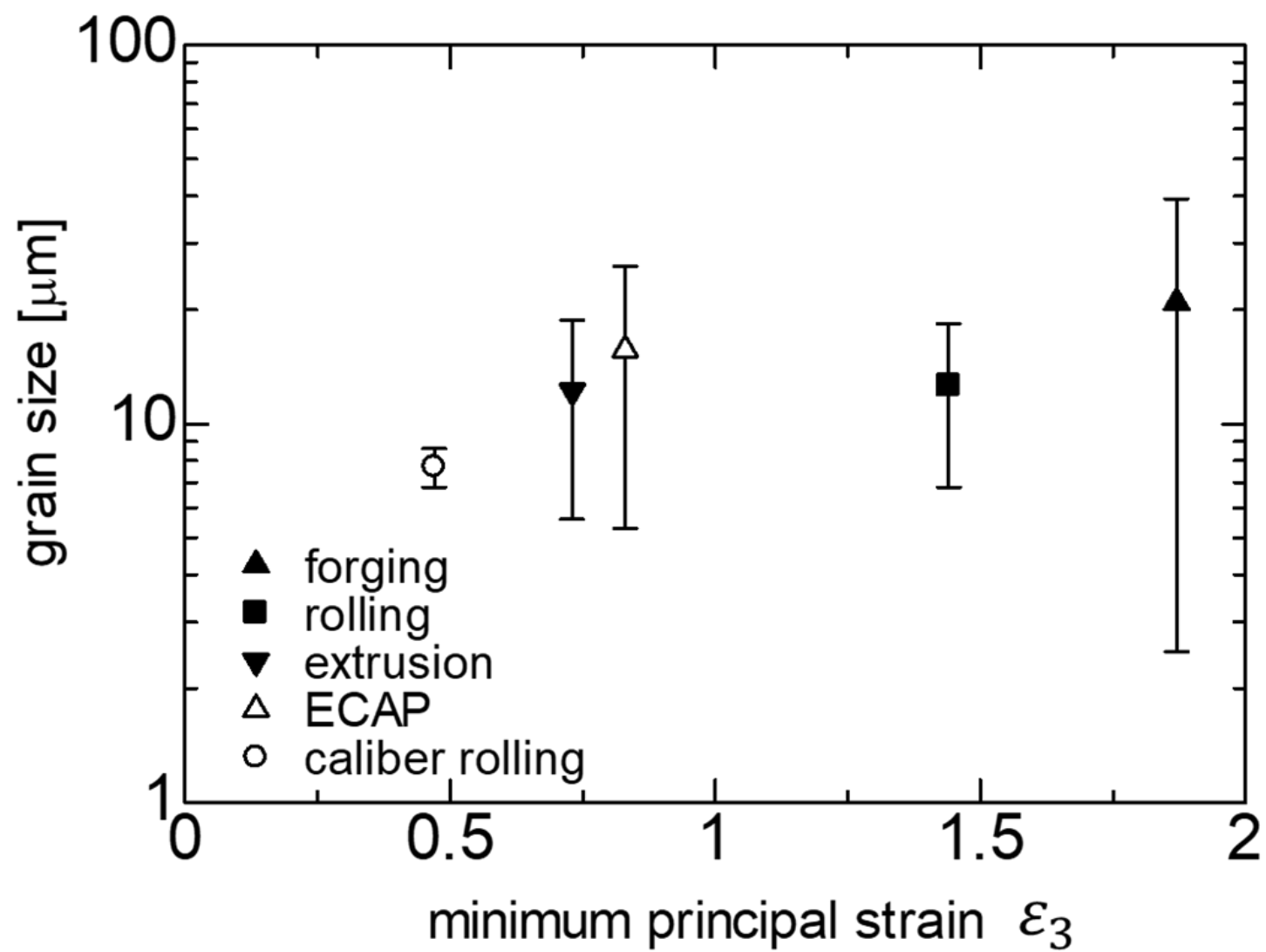


Figure 13

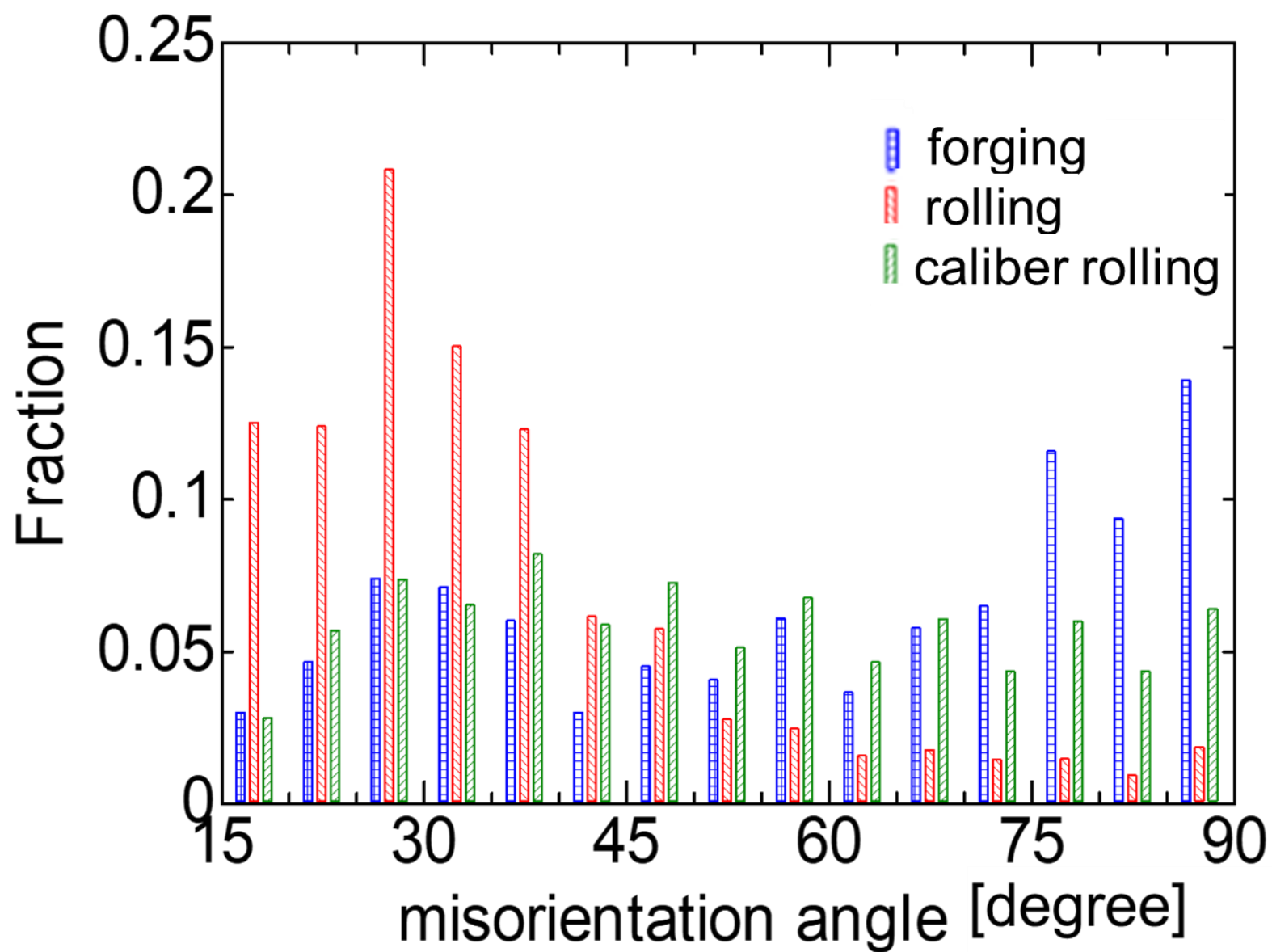


Figure 14

