Preferred lattice misorientations in rolled aluminium: tracking experiments and crystal plasticity simulations

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Abstract. The microstructural heterogeneities developed in the grains of hot deformed Al have been quantified by applying the microtexture tracking technique to hot plane strain compressed Al-0.1%Mn. By successive EBSD measurements over the same (internal) surface at different strains, a large data set of lattice orientation and disorientation development has been obtained in over 150 grains up to strains of 1.2. Simultaneously, high resolution finite element simulations have been carried out to large strains with the same grain orientations; both experiments and simulations focus on the orientation distributions developed within individual grains. It is shown that 15-20% of the grains undergo orientation splitting, usually when the orientation is both symmetrical with respect to the loading and divergent in terms of potential lattice rotations. An analysis of the reorientation velocity field predicted for hot PSC provides a first order indication of the particular grain orientations expected to undergo orientation splitting together with their splitting modes. In the majority of grains which simply undergo orientation spreading, a detailed analysis of the disorientation axes within grains has been carried out. At low strains, there is a very high density of near TD disorientation axes which progressively evolve with plastic strain towards RD. A similar, although faster, evolution to both RD and ND is predicted by the FECP simulations. Original explanations for these low and high strain disorientation axes are proposed, based first on the influence of random local stress variations on lattice rotations and then on the reorientation velocity field.

1. Introduction
Plastic deformation is inherently inhomogeneous over many scales from the macroscopic to the micrometer level and, as is well known, this heterogeneity strongly affects microstructural evolution during subsequent annealing. Much work has been carried out in this area by cold deformation, but hot deformation can offer some advantages for fundamental studies. Although it is generally considered that hot deformation, by virtue of dynamic recovery processes, leads to a more homogeneous microstructure, many heterogeneities still occur, albeit on a coarser scale than at low temperatures. This turns out to have the advantage of facilitating their characterization by EBSD and also their modelling by finite element crystal plasticity (FECP) simulations. In particular, strain-induced orientation gradients across grains can be characterized and, in some cases, modelled. The present work is mostly focused on the hot deformation microstructures, microtextures and orientation distributions developed in hot, plane strain compressed, Al-Mn polycrystals.

Since FECP simulations can model the evolution of grain orientations during plastic straining it follows that experiments should be able to do the same. In the present case, we have chosen a tracking method by EBSD mapping of the inside surface of a split sample deformed in channel-die compression.
It was possible to follow over a hundred grains during successive plastic strains to a final strain of 1.2. Our previous publications have shown that the average behavior of the grains is reasonably consistent with the Taylor model of lattice rotations [1, 2], but that about 15% of the grains break up into two or more distinct orientations. This work looks in more detail at the local behavior within the grains, particularly grain orientation break-up and also the sub-grain orientation spreads.

2. Experimental and modelling procedures
The “microtexture tracking” technique consists of taking orientation maps of the same grains of a polycrystal on the internal surface of a split sample submitted to successive deformations in plane strain compression (PSC) [1]. A high-purity Al–0.1 wt%Mn alloy with an average grain size of about 300 µm and a weak initial crystallographic texture was used. The split sample was made of two identical parts of 8x3.5x10 mm each, assembled along the transverse direction (TD). Prior to deformation, an orientation map of the grain structure was carried out by EBSD on a 4x4 mm² region located at the centre of one of the two sample parts, using a step size of 5 µm. The sample was then deformed in plane strain compression at 400 °C (0.72 Tm) in a channel die rig [3] to successive strains of 0.19, 0.42, 0.77 and 1.20. After each deformation step, the sample was quenched to retain the “hot deformation” microstructure. Without any additional polishing, an orientation map was carried out on the same area of the sample, using the same spatial resolution. At ε = 0.41, 0.77 and 1.20, some finer orientation maps were also acquired on a region of the sample covering a few grains, using a step size of 0.5–0.6 µm. This enables one to directly analyze the evolution of the subgrains, which are 5–10 µm in size [4]. It was previously shown in Refs. [1, 4] that following the grain rotations on the internal surface of a split sample does not significantly affect the sample texture nor the local orientations. As a consequence, in the present work, the grains are considered to deform as grains in a real polycrystal.

Figure 1. (a) Experimental set-up and grain structures on the ND-RD plane before and after straining to 1.2, (b) 3-D Voronoi tessellation of same grain orientations used for FECP simulation. Rodrigues vector orientation colouring.

To model the local plastic strains and microtextural evolutions, a finite element crystal plasticity analysis was performed to simulate hot deformation of an equivalent Al polycrystal using an FECP model [5] and the initial grain orientations [6]. The microstructure was represented by a 1000-grain three-dimensional Voronoi tessellation of which 182 central grains were assigned the experimental orientations, as illustrated in Fig. 1b. The surrounding grains were assigned orientations taken randomly.
from a uniform distribution. This means that the initial polycrystal reproduces the experimental orientation distribution, but not the grain morphology and topology, due to the fact that the (3-D) experimental polycrystal morphology is not completely known. To follow these large plastic strains, remeshing was applied to correct the mesh for distorted elements, and the microstructural variables were transported from the old mesh to the new mesh, using the Neper polycrystal generator [7, 8]. The FECP simulation simulated the experimental deformation sequence, including the unloadings at ε = 0.19, 0.42 and 0.77. To enable the in-grain orientation spreads to develop properly, the grains were finely discretized into tetrahedral elements. The initial average mesh density was 600 elements per grain, and the final one (ε = 1.2) was 1250 elements per grain, so that element sizes are roughly 30-40 µm (compared to the grain size of about 300 µm).

Following previous studies on aluminium single crystals of similar compositions deformed at the same temperature [9, 10], we consider a mixed slip model, where different system families are potentially active: {111}<110>, {100}<110> and {110}<110> (24 systems). Slip is assumed to be viscoplastic with the slip rate $\dot{\gamma}^\alpha$ on a system $\alpha$ related to the resolved shear stress $\tau^\alpha$ through the power law:

$$\dot{\gamma}^\alpha = \dot{\gamma}_0 \left| \frac{\tau^\alpha}{g_\infty} \right|^{1/m} \text{sgn}(\tau^\alpha)$$  \hspace{1cm} (1)

$$g^\alpha = h_0 \left( \frac{g_s - g_\infty}{g_s - g_0} \right) \dot{\gamma}, \text{ where } \dot{\gamma} = \sum_\alpha |\dot{\gamma}^\alpha|$$  \hspace{1cm} (2)

By definition, sgn(x) = 1 if x ≥ 0 and -1 if x < 0. The values of the hardening parameters have been determined from experimental measurements, and are: $\dot{\gamma}_0 = 1$, m = 0.12, $h_0 = 4$ MPa, $g_0 = 8g_r$ MPa, and $g_s = 12g_r$ MPa, where $g_r = 1$ for {111}<110>, 1.4 for {100}<110> and 0.9 for {110}<110> (the $g_r$ values are from Ref. [10]). A complete description of the constitutive model and the FECP implementation can be found in Refs. [11, 12].

3. Results

3.1 Orientation distributions.

Orientation distributions in deformed grains are readily visualized by standard pole figures as shown in Fig. 2 where a) shows orientation spreading of a deformed grain and b) orientation splitting of the same grain at higher strain. However, their quantitative analysis is better performed using the quaternion and/or Rodrigues vector representations as originally described by Glez and Driver [13] and Barton and Dawson [14].

The fundamental Rodrigues region of a cubic crystal has the shape of a 3D polyhedron (a truncated cube) and provides a low distortion picture of orientations, or disorientations, without the degeneracy problems of many other representations. As an example, Fig. 2c plots out the Rodrigues vector distributions of orientations in the same, fragmented, grain at a strain of 1.2. The quaternion method provides the average orientations [15, 16] and also the size and shape of the disorientation spread around this average. Following Refs. [13, 14], the shape of this distribution is characterized by an eigen decomposition of the set of quaternion disorientations into eigenvectors $v_i$ (principal spread directions) and eigenvalues $\lambda_i$ (spread amplitudes). They are arranged so that $\lambda_1 > \lambda_2 > \lambda_3$ and $v_1$ is taken as the principal disorientation axis.

Given that the disorientations are small, the principal angular spreads are obtained as:

$$\dot{\theta}^i = 2\arctan(\sqrt{\lambda_i})$$
Figure 2. Example of a grain undergoing fragmentation; a) {111} pole Fig. at strain 0.42 (spreading), b) pole Fig. at strain 0.77 (fragmentation) and c) Rodrigues vector representation of the disorientation distribution at strain 1.2. Original grain orientation given as points in a) and b).

3.2 Orientation Splitting.
From the set of 176 grain orientations that were followed experimentally during hot PSC (and simulated numerically using the FECP method), about 10% were observed to undergo clear orientation splitting at the final strain of 1.2. Their proportion increased monotonically, and almost linearly, with plastic strain showing that large strain experimental techniques are essential to characterize this phenomenon. Splitting is a phenomenon which can occur throughout a grain, i.e. within any of the RD-TD, TD-ND or RD-ND planes, but as a result of the 2-D EBSD measurements on the TD-ND plane, no splitting could be detected on this plane. The true fractions of orientation splitting should therefore be increased by 3/2 to about 15%. The 3-D FECP simulations, which do not have this limitation, predicted that about 20% of the grains should split at the final strain.

Fig. 3 gives an example of the orientation splitting of a grain initially near 20° ND rotated Cube which breaks up into 2 distinct orientations by opposite ND rotations – in both the experiments and the simulations. The upper {111} pole figures illustrate the orientation distributions determined by EBSD (a) and calculated by FECP (b), which are in good agreement.

Figure 3. Fragmentation of a grain of ND-rotated Cube orientation in (a) experiment and (b) simulation. The two modes are shown in red and blue on the pole Figs. and on the maps.
The lower orientation maps show the scale of splitting and their spatial distributions; perfect agreement between these EBSD and FECP maps is not really expected since the latter has an element size almost one of magnitude greater than the EBSD pixels. Nevertheless, similar tendencies can be observed.

As could be expected, the grains that split, generally into two orientations, possess symmetrical orientations with respect to the loading directions (Figs. 2 and 3). They were principally located near Cube {100}<001>, 45°/ND rotated Cube {110}<001>, U {110}<011> and Goss {110}<001> together with two grains near {233}<311>, an orientation on the TD <110> fiber. The reason for an increasing number of fragmented grains with strain is that fragmentation occurs more efficiently for orientations which are initially very close to the symmetry fibres. For orientations which are further away, the instability conditions discussed below are not so strong, and they develop significant spreading before they actually split. A more detailed analysis of orientation splitting requires a knowledge of the lattice stability with respect to possible rotations during plastic straining. This has been developed assuming a Taylor type model for hot, plane strain compression of Al and mixed slip as described above. The results of this stability analysis are illustrated on Figs. 4 and 5.

Fragmentation can be understood by examining the properties of the reorientation velocity field. This field is shown in Fig. 4 for hot plane strain compression of Al at the surface of the Rodrigues fundamental region, following previous work by Kumar and Dawson [17], but here with the 24 mixed slip systems. If an orientation is a stable equilibrium point, it constitutes a sink of the flow field. For an orientation to be asymptotically stable, all reorientation trajectories in the vicinity of an orientation must converge on the orientation. Using linearized stability analyses, the character of the equilibrium can be determined to be stable or unstable from the signs of the eigenvalues of the reorientation velocity gradient [6].

The regions surrounding the sinks are basins. A characteristic of most reorientation velocity fields associated with slip is that the orientations move quickly toward one of the fibres and then move more slowly along the fibre toward an equilibrium point. We can interpret the fragmentation results with respect to these characteristics of the reorientation velocity field, keeping in mind that neither the experimental observations nor the simulation adhere strictly to the Taylor assumption. The sink and basin features of the velocity field shown in Fig. 4 are evident, as is the close connection of these features to the symmetry fibres.

Figure 4. (a) Sample symmetry fibers for cubic symmetry represented on the surface of the Rodrigues fundamental region. (b) Reorientation velocity field represented on the surface of the Rodrigues fundamental region showing both divergence and convergence towards the α (blue) and β (red) fibres.
Thus, fragmentation is more likely with orientations that lie at the boundary of a basin, because they tend to be drawn equally strongly into two different basins and toward two different points of the predominant texture fibres. So, the boundaries of the basins can be considered as equivalent to ‘‘watershed’’ regions. An example is the case of the grains of U orientation. As can be seen in Fig. 4, U is an equilibrium orientation. Moreover, it is stable about RD, but unstable about ND. This favors fragmentation by opposite reorientations about ND. Such a fragmentation actually takes place for the experimental grain of orientation close to U as illustrated in Fig. 2.

![Diagram](image)

**Figure 5.** Fragmentation diagram for hot PSC Al showing potential reorientation velocity fields for sample symmetry orientations.

The second diagram, Fig. 5, depicts the reorientation velocity field for the principal grain orientations undergoing hot PSC. As above, instability is associated with potential divergence as seen for example for the 45°/ND Cube with respect to all rotations and also the Goss orientation with respect to TD and ND rotations. It is worth recalling that compared with other deformation conditions, hot PSC is not expected to generate large fractions of grains that undergo orientation splitting. Fragmentation is generally favoured by strain paths such as shear (e.g. torsion or ECAP) which do not develop stable grain orientations nor stable deformation textures. Cold work also strongly favours grain orientation splitting. The present study on hot, plane-strain compressed Al actually corresponds to a relatively low level of splitting, as occurs experimentally (~15%) and numerically (~20%) at strains of order unity. In this context, the agreement between FECP simulations and experimental results can be considered as satisfactory, at least in terms of lattice reorientations. In many cases the orientation splitting occurs in the predicted grain orientations about the predicted rotation axes. An analysis of the reorientation velocity field predicted for hot PSC by a Taylor type model provides a first order indication of the particular grain orientations expected to undergo orientation splitting together with their splitting modes. However, the agreement between experiments and simulations is not found for all orientations. In some cases, the discrepancies may be due to the influence of the neighbouring grains which are not identical in the present experiments and numerical simulations.
In conclusion, for this part of the work concerning grain orientation fragmentation, it appears possible to model grain orientation splitting (or its absence) in most, but not all grains undergoing hot PSC to high strains, i.e. the equivalent of industrial rolling.

3.3 Orientation spreading.
As pointed out above in the same set of experiments, the majority of the grains undergo orientation spreading, a general phenomenon which also merits further analysis. Of course there have been many studies of disorientation development in deformed grains, particularly during cold plastic straining and Risø has taken a major role in this field, e.g. [18, 19]. Contrary to cold deformation, hot plastic straining by conventional shaping methods has usually been found to lead to an average disorientation that stabilizes to a near constant value of a few degrees at strains of order 0.5 to 1 [20-23].

In the present study, we shall focus on an aspect of the problem that, until recently, has not received much attention: the disorientation axes of the intra granular orientation spread. The same methods as above are employed, but the set of grains whose behaviour is examined in detail is taken from the complementary set which did not fragment and which is sufficiently large to provide valid spread data, i.e. at least 1000 orientation measurements. A total of 92 grains met this requirement. The same set of 92 grains was considered in the simulation metrics, with a slight difference due to a limited number of grains that underwent orientation fragmentation in the experiment, but not in the simulation (or vice versa). In the following, we first analyze the evolution of the disorientation distributions over this set of grains. We then look for correlations between the disorientation distribution properties and the grain average orientations.

The distributions of the average disorientation angles are presented in Fig. 6 for the experiment (a) and simulation (b). Average values at successive strains are listed in Table 1. It should be recalled that these are point-to-average disorientations which usually tend to be significantly higher than point-to-point (or subgrain-to-subgrain) disorientations since they include orientation gradients across entire grains. As a consequence the axes associated with point-to-average disorientations are usually defined with greater accuracy than those of point-to point disorientations. This method was also adopted by Pantleon et al [24] in a similar study of cold rolled aluminum. For both experiment and simulation, the average disorientation angle increases rapidly early in the deformation ($\varepsilon < 0.5$) to comparable values between 5 and 6°. At strains above 0.5, the rate of increase drops off more rapidly in the experiment than in the simulation, such that the experimental average tends towards a saturation value of 7–8°. It is difficult to estimate a saturation value in the simulations. In both simulation and experiment, there is a high degree of variability in the average disorientation, ranging in the experiment from 3.6 to 19.7°.

![Figure 6. Average disorientation angles as a function of strain in (a) experiment and (b) simulation.](image-url)
Fig. 7 gives an example of the disorientation distribution of a deformed grain in Rodrigues space and which is clearly anisotropic. Using the eigenvalues of the covariant matrix of the Rodrigues orientation distribution as described above, this anisotropy can be conveniently quantified by the factor $\lambda_a = \lambda_1^{1/3} \sqrt[3]{\lambda_2 \lambda_3}$ which takes a value of 1 for an isotropic distribution. The average values of this anisotropy factor are given in Table 1 along with the average disorientation angles. The $\lambda_a$ values are typically of order 1.5-2, but there is an interesting difference between the experiments where $\lambda_a$ decreases with strain and the simulations where they tend to increase with strain.

![Figure 7](image.png)

**Figure 7** (a) Anisotropic disorientation distribution seen in the Rodrigues fundamental region and principal axes. (b) Angular distributions along the three principal axes.

**Table 1.** Average disorientation angles and anisotropy factors as a function of strain in the experiment and simulation.

| Strain | Average disorientation angle (°) | Anisotropy factor |
|--------|----------------------------------|-------------------|
|        | experiment | simulation | experiment | simulation |
| 0.19   | 3.5        | 3.3        | 1.91       | 1.62       |
| 0.42   | 5.2        | 5.7        | 1.75       | 1.64       |
| 0.77   | 6.5        | 8.0        | 1.55       | 1.84       |
| 1.20   | 7.2        | 9.9        | 1.49       | 2.13       |

The distributions of the preferential disorientation axes at successive strains are depicted in Fig. 8, as equal-area projections onto the sample RD-ND plane, for both experiment and simulation. The projections are reduced to one quarter using orthotropic sample symmetry. In addition, the simulated distribution at $\varepsilon = 0.02$ is also provided to analyze the early stage of plastic deformation. Fig. 8 shows that the experimental axes are strongly aligned with TD up to $\varepsilon = 0.42$. This initial TD disorientation axis is consistent with some earlier results on hot deformed Al single crystals by Glez and Driver [23] and work by Pantleon et al. on 38% cold rolled Al [24] and tensile tested Cu polycrystals [25]. However, the tendency for preferential TD axis is much stronger here than reported for cold deformation; typically the axis densities around TD are about twice those after cold rolling.
Figure 8. Preferential disorientation axis distributions as a function of strain, for (a) experiment and (b) simulation, shown as equal-area projections onto the RD-ND plane.
A second important result from Fig. 8 is the change in disorientation axis with strain. Looking at the experimental results, the initial strong preference toward the TD axis ($\varepsilon = 0.19$) weakens significantly at a strain of 0.77 and evolves to a preference toward mixed TD+RD at a strain of 1.2 (with some axes located halfway between RD and ND). In the simulation, the preferred axes are also close to TD at small strains. A rapid transition then occurs, between $\varepsilon = 0.19$ and 0.42, so that the preferred axes exhibited in the large-strain distribution is shared between RD and ND, without any TD. This change in disorientation axis with plastic strain does not appear to have been reported in previous works, which were carried out at smaller strains, but could have significant implications for the nature of the sub-grain boundaries. We shall examine below the possible mechanisms that lead to TD disorientation axes and then their evolution at high strains.

First, we have checked that these disorientation axes are essentially related to the sample axes. We have plotted the disorientation axes in the crystal coordinates and found them to be nearly uniformly distributed in the crystal stereographic triangle. At small strains, this was expected from the TD distribution of the preferential disorientation axes and the uniform distribution of the average grain orientations. Second, it is known that the grain rotation axes during PSC also tend to align with TD, so one could ask whether there is a stronger correlation with sample TD or with grain rotation axis? The frequencies of these two distributions have been determined and it was found that the sample TD frequencies (both experiments and simulations) are about twice those of the grain rotation axis [26].

So what favours these disorientation axes? Clearly, they are related to local variations in slip rates, which generate local variations of lattice rotations, but which must possess specific properties. An elementary analysis of spatial variations in slip rates under the conditions of PSC where the predominant slip systems have planes inclined near 45° between RD and ND and orthogonal directions in this plane would suggest local rotations about the direction perpendicular to the slip plane normals and directions, i.e. TD. A more general analysis can be made by looking at the effect of variations in stress state. The latter would occur as a consequence of microstructure heterogeneities principally between dislocation cells or sub-grains and their interior. One expects them to vary in a near random way about the average stress state of the grain which would be near the Taylor, Bishop and Hill stress. By virtue of Equation 1 stress variations give rise to variations of slip rates. Using the viscoplastic relation (1) between stress and slip rates, it can be shown [26] that the rotation velocity field varies with the stress vector according to

$$\frac{\partial \mathbf{r}}{\partial \sigma_v} = -\sum_{\alpha} \frac{\partial \gamma^\alpha}{\partial \tau^\alpha} (\mathbf{t}^\alpha \otimes \mathbf{p}^\alpha)$$

(3)

where $\mathbf{p}^\alpha$ stands for the vector form of the symmetrical Schmid tensor, and $\mathbf{t}^\alpha$ the spin vector. This relation has been applied to the initial grain orientations to determine the rotations expected of small random stress variations.

The distribution of the associated preferential directions is provided in Fig. 9a as a projection on the RD-ND plane. The distribution exhibits an intense focus about TD that is in remarkable quantitative agreement with observations of Fig. 8 (at $\varepsilon = 0.02–0.19$). The distribution can also be represented over the Rodrigues fundamental region (Fig. 9b) by batons aligned with their associated principal axes and coloured according to their components along sample directions: green for TD, red RD or blue ND. This figure confirms that a TD preferential disorientation axis is expected independently of grain orientation.
Figure 9. Preferential disorientation axes simulated by random stress variations shown (a) as an equal-area projection onto the RD-ND plane and (b) at the initial average grain orientations in the Rodrigues fundamental region. On (b), the blue and red lines stand for the α and β fibers, respectively.

The general trend to develop TD preferential disorientation axes can therefore be explained from the expression of $\partial \dot{\gamma} / \partial \nu$ (Equation 3), which involves the slip geometry and the slip behaviour. Under the effect of a random, isotropic stress distribution about the nominal value, all slip systems are subject to equivalent resolved shear stress distributions. On a given slip system, the resolved shear stress perturbations lead to slip rate perturbations proportional to $\partial \dot{\gamma} / \partial \nu$. These values are significantly higher for the active systems, which have high $\tau_\alpha$ values and spin vectors close to TD. This ultimately leads to a reorientation velocity perturbation which has a higher component along TD.

This analysis applies well to small strains, where the orientation spreads are relatively small. When an orientation distribution is wide enough, its orientations are subject to different reorientations even under the same stress or strain. This influences the orientation distribution anisotropy and can be analyzed from the reorientation velocity field, basically as done for orientation fragmentation above. The Eigen decomposition of the velocity gradient indicates how an orientation distribution narrows or broadens about specific directions during deformation. The directions (and the associated intensities) change widely over orientation space, but as shown in Fig. 10, the values of greatest interest are those in the vicinity of stable fibres. In this figure, the same colouring scheme is applied for the components of the reorientation velocities. Most are reddish (RD) with some blueish (ND) and the corresponding stereographic projections indicate near RD (+ some ND) axes.

The predicted directions evolve smoothly along the β and α fibres from RD about the Brass orientation, to RD+ND about the S and Copper orientations, and to ND about the Goss orientation.

In summary, these two mechanisms favor different disorientation axes and compete to provide an evolution of the preferential disorientation axis distributions with strain that is dominated at low strain by one mechanism and at large strain by the other. Assuming a constant stress variability in the grains, the stress-variability mechanism predicts a linear growth of the average disorientation angle with stable, intense anisotropy (anisotropy factor of 3 in average) and preferential disorientation axis distribution about TD. In contrast, the reorientation-velocity mechanism influences the anisotropy properties only when the orientation distributions are large enough and most grains have reached the stable texture fibres. By continuously broadening the orientation distributions along a direction between RD and ND (the first eigenvector of the reorientation velocity gradient, which depends on the orientation) rather than along the original TD, this second mechanism leads to a transition to RD or RD+ND.

It should be noted that orientation distributions would necessarily align with the first eigenvector of the reorientation velocity gradient only at very large strains. Because a grain orientation distribution
which reaches the stable orientation fibre is not uniformly distributed about an average orientation, but rather with a strong TD component, it needs a significant amount of additional strain for the disorientation axes to migrate to RD or RD+ND. In the experiment, this movement occurred for only half of the grains at 1.2, but is expected to be complete at higher strain.

![Figure 10](image)

**Figure 10.** Preferential disorientation axes resulting from the reorientation velocity conditions shown (a, b) as an equal-area projection onto the RD-ND plane and (c, d) at the final average grain orientations in the Rodrigues fundamental region. (a, c) refers to the experiment and (b, d) to the simulation. On (c, d), the blue and red lines stand for the $\alpha$ and $\beta$ fibers, respectively.

Interestingly, the transition of the preferential disorientation axes from TD to RD or RD+ND also indicates a change of the dislocation structures, i.e. the character of the dislocation subgrains. Specifically, knowing that the slip system spin vectors are preferably distributed about TD at all strains, a TD preferential disorientation axis at small strain indicates a higher content of edge dislocations (Burgers vector perpendicular to the rotation axis) while an RD or RD+ND preferential disorientation axis at large strain indicates a higher content of screw dislocations (Burgers vector parallel to the rotation axis).

4. Conclusions

The orientation distributions developed within individual grains from small to large strains at 0.72 Tm have been quantified and compared to high resolution FECP simulations. By means of the microtexture tracking technique applied to hot, plane strain compression of an Al-0.1 wt.% Mn alloy, a large data set of orientation and disorientation development has been obtained in over 150 grains up to strains of 1.2. Simultaneously, FECP simulations have been carried out at the scale of about $10^3$ elements per grain to
compare with the same experimental grain orientations to large strains. A mixed slip, viscoplastic model is used for these high temperature deformation conditions.

Apart from the fact that most grain reorientations are close to those of the Taylor model, it is shown that a significant proportion undergo orientation splitting (15-20%), usually when the orientation is both symmetrical with respect to the loading and divergent in terms of potential lattice rotations (termed watershed orientations). An analysis of the reorientation velocity field predicted for hot PSC provides a first order indication of the particular grain orientations expected to undergo orientation splitting together with their splitting modes. In the majority of grains, which simply undergo orientation spreading, a detailed analysis of the disorientation axes within grains has been carried out.

At low strains, there is a very high density of near TD disorientation axes which progressively evolve with plastic strain towards RD. A similar, although faster, evolution to RD+ND is predicted by the FECP simulations. Original models for these low and high strain disorientation axes are proposed: i) at low strains random local stress variations create slip rate perturbations which are significantly higher for the active systems characterized by high $\tau^\alpha$ values and spin vectors close to TD; ii) at high strains most grain orientations are close to the rolling texture components whose reorientation velocity fields favour RD+ND axes.

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