

Microstructure of individual grains in cold-rolled aluminium from orientation inhomogeneities resolved by electron backscattering diffraction

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Abstract

The deformation structure within individual grains of a deformed material is resolved by electron backscattering diffraction. The employed evaluation scheme for local orientation data is illustrated on cold-rolled aluminium. The orientation distribution of each grain is characterized by an averaged orientation spread and its anisotropy; the dependence of both parameters on grain size and grain orientation is discussed. The preferred rotation axis in each grain is determined and shows a dominant orientation spreading around the transversal direction. Characteristic features of the deformation structure (as alternating orientation differences or orientation gradients) are resolved from sign-carrying disorientation angles defined with respect to the preferred rotation axis in each grain. Five components of the dislocation density tensor – corresponding to part of the geometrically necessary dislocation content – are inferred from the local curvatures. The spatially inhomogeneous distribution of the dislocation density offers a new possibility for identifying dislocation boundaries.

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1. Introduction

Electron backscattering diffraction (EBSD) [1] has become a convenient tool for characterizing the microstructure of crystalline material as metals or rocks. Common for all conventional applications is that orientation data are gathered for numerous points in a regular grid on the surface of a specimen. Elaborate post-processing [2,3] must be used for extracting relevant information on the microstructure from the orientation data. The microstructure of a crystalline material is usually characterized by regions of homogeneous orientation. The slightest, discontinuous orientation difference indicates a border between two regions and the existence of a boundary. Individual regions separated by such boundaries are termed grains—at least in the as-grown, as-deposited or recrystallized state.

Plastic deformation introduces orientation differences in originally homogeneously oriented grains. Any local excess

of dislocations of one sign of the Burgers vector causes an orientation difference. As dislocations gather in deformation-induced boundaries, the grains become subdivided into smaller regions of different, but similar orientations. Within individual grains, different types of deformation-induced boundaries can be distinguished: randomly oriented and curved dislocation walls separating nearly dislocation free cells and straight dense dislocation walls running parallel to each other along preferred directions. After moderate cold-rolling, for instance, the second type of boundaries are inclined about 40° towards the rolling direction [4]. The disorientations across adjacent parallel boundaries are often of opposite sign [4].

The goal of the present investigation is a thorough characterization of the deformation-induced microstructure based on orientation differences within individual grains in several different manners. Firstly, the total orientation spread of a grain and its anisotropy in orientation space is considered. Secondly, the spatial distribution of the orientation variation is characterized to distinguish overall curvatures of the grains from local variations, in particular alternating orientation changes. Finally, from the local curvatures the geometrically necessary dislocation content

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is derived which gives a lower estimate for the local dislocation density. Such information is relevant, for instance, for developing microstructural theories for work-hardening based on dislocation accumulation [5].

2. Experimental example

The method of obtaining information on the microstructure within individual grains is illustrated for commercial pure aluminium AA1050 (99.5%) cold-rolled to 38% reduction in a single pass with an intermediate draught [6]. On a surface along rolling and normal direction, orientations are determined by EBSD at points on a regular square grid with a mutual spacing of 1 μm . In the orientation map shown in Fig. 1 grains are identified as contiguous areas surrounded by boundaries with disorientation angles larger than a selected threshold angle of 8°. Grains with less than 250 points or grains reaching the borders of the measurement grid are excluded from the analysis. From the orientation data on a 600 $\times$ 240 grid, 45 grains are identified in this manner [7].

3. Characterization of orientation spread

3.1. Description of orientations

The orientation of a crystalline lattice is described by the rotation required to achieve the orientation from a chosen reference orientation. This rotation (or more precisely the corresponding coordinate transformation) is characterized by a rotation angle ω and a rotation axis $\vec{r}$ which can be combined into a unit quaternion [8]:

$$\mathbf{q} = \begin{pmatrix} q_0 \\ \vec{q} \end{pmatrix} = \begin{pmatrix} \cos\left(\frac{\omega}{2}\right) \\ \sin\left(\frac{\omega}{2}\right) \vec{r} \end{pmatrix} \quad (1)$$

The components of the unit quaternion are linked to Bunge's orientation matrix $\mathbf{G}$ [10]:

$$q_0 = \frac{1}{2}\sqrt{G_{ii} + 1}, \quad q_i = -\frac{\epsilon_{ijk} G_{jk}}{4q_0} \quad (2)$$

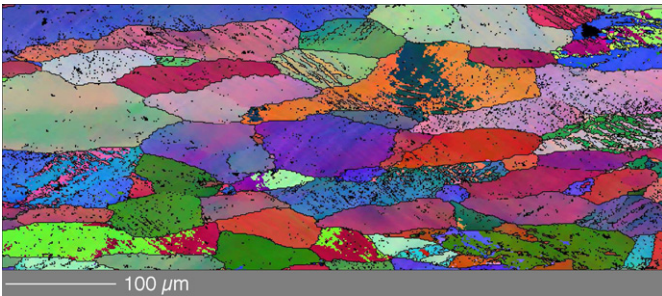


Fig. 1. Orientation map of commercial pure aluminium cold-rolled to 38%. Rolling direction is horizontal and normal direction vertical. Colors characterize the orientation difference between the local orientation at each measuring point and the reference orientation. Black lines indicate disorientations above 8° between neighboring measuring points.

or the commonly used Euler angles (ϕ_1, Φ, ϕ_2) :

$$q_0 = \left| \cos\left(\frac{\Phi}{2}\right) \right| \left| \cos\left(\frac{\phi_1 + \phi_2}{2}\right) \right| \quad (3)$$

$$q_1 = -q_0^{-1} \sin\left(\frac{\Phi}{2}\right) \cos\left(\frac{\Phi}{2}\right) \cos\left(\frac{\phi_1 + \phi_2}{2}\right) \times \cos\left(\frac{\phi_1 - \phi_2}{2}\right) \quad (4)$$

$$q_2 = -q_0^{-1} \sin\left(\frac{\Phi}{2}\right) \cos\left(\frac{\Phi}{2}\right) \cos\left(\frac{\phi_1 + \phi_2}{2}\right) \times \sin\left(\frac{\phi_1 - \phi_2}{2}\right) \quad (5)$$

$$q_3 = -q_0^{-1} \cos^2\left(\frac{\Phi}{2}\right) \sin\left(\frac{\phi_1 + \phi_2}{2}\right) \cos\left(\frac{\phi_1 + \phi_2}{2}\right) \quad (6)$$

An orientation difference between two orientations (in the coordinate system of the first orientation) is characterized by the rotation from one orientation (A) to the other (B) and given by the quaternion product [8]:

$$\Delta\mathbf{q} = \mathbf{q}_B \mathbf{q}_A^{-1} = \begin{pmatrix} q_0^B q_0^A + \vec{q}^B \cdot \vec{q}^A \\ -q_0^B \vec{q}^A + \vec{q}^B q_0^A - \vec{q}^B \times \vec{q}^A \end{pmatrix} \quad (7)$$

with a misorientation angle $\theta = 2 \arccos(\Delta q_0)$. Orientations can be described in several equivalent manners, because of the symmetry of the crystalline lattice. The symmetrical equivalent description with the lowest of all possible misorientation angles is selected and the corresponding orientation difference termed *disorientation*.

3.2. Ensemble of orientations

For an ensemble of m discrete orientations described by quaternions $\mathbf{q}_i$, its mean orientation:

$$\bar{\mathbf{q}} = \frac{1}{N} \sum_{i=1}^m \mathbf{q}_i \quad \text{with norm } N = \left| \sum_{i=1}^m \mathbf{q}_i \right| \quad (8)$$

can be defined [9]. For all individual orientations $\mathbf{q}_i$ their disorientations, i.e. the orientation differences from the mean $\bar{\mathbf{q}}$:

$$\delta\mathbf{q}_i = \mathbf{q}_i \bar{\mathbf{q}}^{-1} \quad (9)$$

are found under the condition of always selecting the symmetrical equivalent representation closest to the mean.

The orientation spread of an ensemble of orientations is characterized by a symmetric tensor [3]:

$$\mathbf{Q} = \frac{1}{m} \sum_{i=1}^m \delta\vec{q}_i \otimes \delta\vec{q}_i \quad (10)$$

of rank 2 spanned by the disorientation vectors $\delta\vec{q}_i$. As the average vector part of the disorientations vanishes – as a consequence of defining the mean orientation by Eq. (8) [3] – the tensor $\mathbf{Q}$ is

Table 1
Ideal texture components

Texture component	Ideal orientation	Symbol	Number of grains	Volume fraction
Cube	{001}<100>	□	–	–
TD rotated cube/shear	{001}<110>	◇	–	–
Goss	{011}<100>	△	1	2%
Copper	{112}<111>	▽	2	2%
Brass	{011}<211>	◁	1	1%
S	{123}<634>	◁	9	14%
Random		○	32	81%

the tensor of second-order *central* moments and independent of the average orientation $\bar{\mathbf{q}}$.

The tensor of second-order central moments of the disorientation distribution is advantageous for a concise characterization of an orientation spread [7]: not only the width of the orientation distribution is determined, but also the anisotropy in orientation space; for instance, an existing preferred rotation axis is identified. With this preferred rotation axis, a sign can be attributed to the disorientation angles. This allows differentiating between alternating orientation differences and long-range orientation gradients.

From its definition follows, that $\mathbf{Q}$ is a positive (semi-) definite 3×3 matrix. Its three non-negative eigenvalues λ_α are found by diagonalization. They are determined by the three standard deviations $\sigma_\alpha = \sqrt{\lambda_\alpha}$ of the disorientation distribution along the three eigenvectors $\delta\hat{\mathbf{q}}_\alpha$. The latter determine the three principal axes of the orientation spread in the crystallographic system of the mean orientation. The particular eigenvector $\delta\hat{\mathbf{q}}$ corresponding to the largest eigenvalue $\hat{\lambda} = \hat{\sigma}^2$ characterizes the dominant rotation axis $\tilde{\mathbf{r}}^* = \delta\hat{\mathbf{q}}$ of the disorientations in the ensemble. This definition of a preferred rotation axis is unambiguous and free of any arbitrary assumptions on preferred macroscopic axes. The direction $\tilde{\mathbf{r}}^*$ of the preferred rotation axis in the reference system is easily derived from

$$\begin{pmatrix} 0 \\ \tilde{\mathbf{r}}^* \end{pmatrix} = \bar{\mathbf{q}}^{-1} \begin{pmatrix} 0 \\ \tilde{\mathbf{r}}^* \end{pmatrix} \bar{\mathbf{q}} \quad (11)$$

3.3. Orientation spread within individual grains

For each of the 45 grains identified in the orientation map the average orientation $\bar{\mathbf{q}}$ is calculated from all m orientations $\mathbf{q}_i$ within the grain as well as all disorientations $\delta\mathbf{q}_i = \mathbf{q}_i\bar{\mathbf{q}}^{-1}$ with respect to this average. All grains are classified according to the disorientation angle between their mean orientation and certain ideal orientations defining specific texture components listed in Table 1. A grain belongs to the texture component with the smallest disorientation angle. If the latter exceeds 15° , the grain is classified as randomly oriented. As obvious from Table 1 most of the grains are of random orientation and the specified texture components constitute only smaller volume fractions.

For characterizing the orientation spread of each grain, the standard deviations σ_α along the three principal axes are calculated from the tensor $\mathbf{Q}$ of the second-order central moments for the individual grains. The magnitude of the orientation spread

is characterized as the equivalent isotropic spread by the geometrical mean $\bar{\sigma} = \sqrt[3]{\sigma_1\sigma_2\sigma_3}$ of the standard deviations of each grain. The anisotropy is characterized by the ratio between the largest standard deviation $\hat{\sigma}$ and $\bar{\sigma}$. In Fig. 2, both parameters are shown as a function of the grain area for all 45 grains. Small average orientation spreads are found for small grains only, larger grains show a larger equivalent isotropic spread, but some of the small grains have a large equivalent isotropic spread comparable with the largest grains. The anisotropy ratio has largest values for smaller grains. Similarly, the largest anisotropy ratio is found for grains with smallest average orientation spread. No dependence on the texture component is recognized.

The preferred rotation axis $\tilde{\mathbf{r}}^*$ in each grain is found from $\mathbf{Q}$ as the eigenvector corresponding to the largest eigenvalue. The directions of the preferred rotation axes (with respect to the reference system) from all 45 grains are displayed in Fig. 3a.

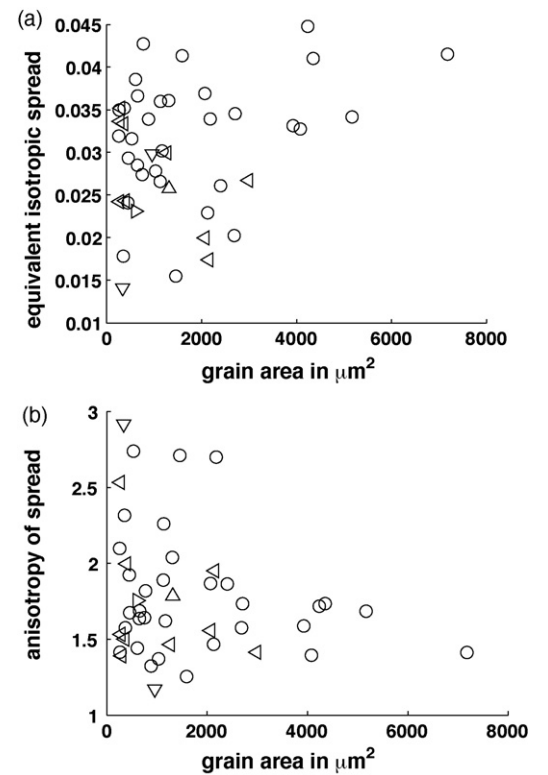


Fig. 2. Characterization of the orientation spread in 45 individual grains: (a) average orientation spread $\bar{\sigma}$ and (b) anisotropy ratio $\hat{\sigma}/\bar{\sigma}$ as a function of grain area in the orientation map. Symbols characterize the texture component defined in Table 1.

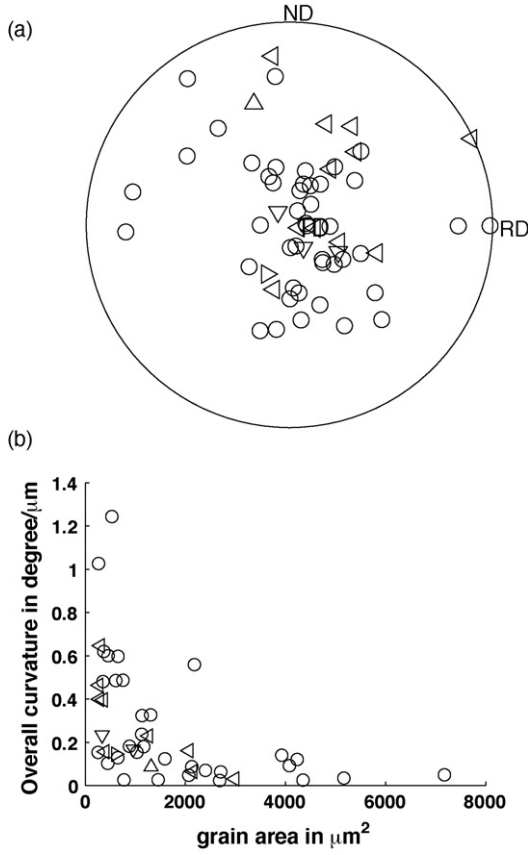


Fig. 3. (a) Direction of preferred rotation axis in 45 individual grains shown as pole figure along transversal direction. (b) Overall curvature in individual grains as a function of grain area in orientation map. Symbols characterize the texture component defined in Table 1.

Obviously, the distribution of the preferred rotation axes is not random. The dominant rotation rarely occurs around normal or rolling direction and the orientation spreads mainly around the transversal direction.

With respect to the preferred rotation axis $\vec{r}^*$, a proper sign can be assigned to the disorientation angles in a grain. The sign-carrying disorientation angle:

$$\theta_i^* = \frac{2\arccos\delta q_{i,0}}{\sqrt{1 - (\delta q_{i,0})^2}} (\delta \vec{q}_i \cdot \vec{r}^*) \quad (12)$$

highlights the orientation variations within two particular grains in Fig. 4. The grains of similar size but different orientation (a: S-component and b: random component) are selected to illustrate the spatial arrangement of orientations within the grains. Despite their comparable orientation spread (average as well as anisotropy), the appearance of both grains is rather different and resembles different characteristic features of deformation structures. For the grain in Fig. 4a, the disorientations alternate on short distances between positive and negative values and resemble the alternating disorientations across parallel dislocation boundaries observed by TEM [4]. The image also reveals a directionality of the alternating orientation changes and an existence of boundaries with an inclination angle of about 40° towards the rolling direction as experimentally observed. The

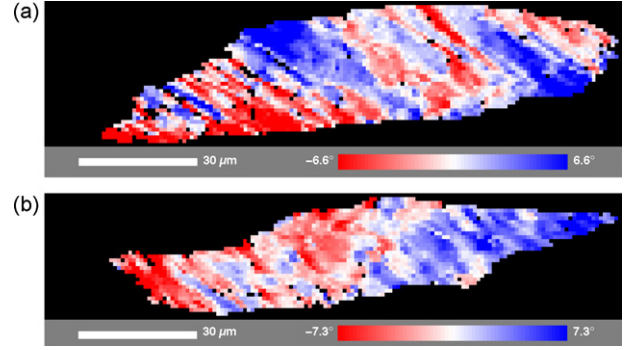


Fig. 4. Disorientations around preferred rotation axes for two grains (defined by a threshold disorientation angle of 8°) in a 38% cold-rolled aluminium polycrystal. Rolling direction is horizontal and normal direction vertical. Ranges are maximum values divided by 3.

grain in Fig. 4b shows corresponding features, but not as pronounced as for the first grain. This is due to continuous change of the sign-carrying disorientation angle from positive to negative values and an orientation gradient over the whole grain. The entire grain is bent and the curvature of $0.09^\circ/\mu\text{m}$ must be subtracted for resolving the underlying dislocation boundary structure in more detail [3]. Such an overall bending of grains is observed frequently: the 45 grains exhibit in average a curvature of $0.27^\circ/\mu\text{m}$ with a standard deviation of the same amount and smaller grains having an increased tendency for larger curvatures (see Fig. 3b).

4. Dislocation density

4.1. Dislocation density tensor

Dislocations are line defects causing a relative displacement of the crystalline lattice. They are described by a line vector $\mathbf{t}$ indicating their direction, their Burgers vector $\mathbf{b}$ characterizing the displacement and their position vector $\mathbf{r}_0$. This information is combined in Nye's dislocation density tensor [11]:

$$\alpha_{ij} = b_i t_j \delta(\mathbf{r} - \mathbf{r}_0) \quad (13)$$

Assuming compatibility, Kröner [12,13] has shown the direct relation between the dislocation density tensor and the geometry of the distorted lattice:

$$\alpha^T \equiv \text{curl} \beta^{\text{pl}} = -\text{curl} \beta^{\text{el}} \quad (14)$$

where β^{pl} and β^{el} are the plastic and elastic distortion tensors. The later is the sum of the elastic strain tensor ϵ^{el} and the tensor describing local lattice rotations ω . The components of the dislocation density tensor:

$$\alpha_{ij} = -\epsilon_{jkl} \beta_{il,k}^{\text{el}} = -\epsilon_{jkl} (\epsilon_{il,k}^{\text{el}} + \omega_{il,k}) \quad (15)$$

simplify in the absence of elastic stresses ($\epsilon^{\text{el}} \equiv 0$) and expressing the lattice rotations $\omega_{il} = -\epsilon_{ilm} \omega_m$ by the rotation vector $\omega_m = -\epsilon_{mil} \omega_{il}/2$:

$$\alpha_{ij} = -\epsilon_{jkl} \omega_{il,k} = \epsilon_{jkl} \epsilon_{ilm} \omega_m,k = \omega_{j,i} - \delta_{ij} \omega_{k,k} \quad (16)$$

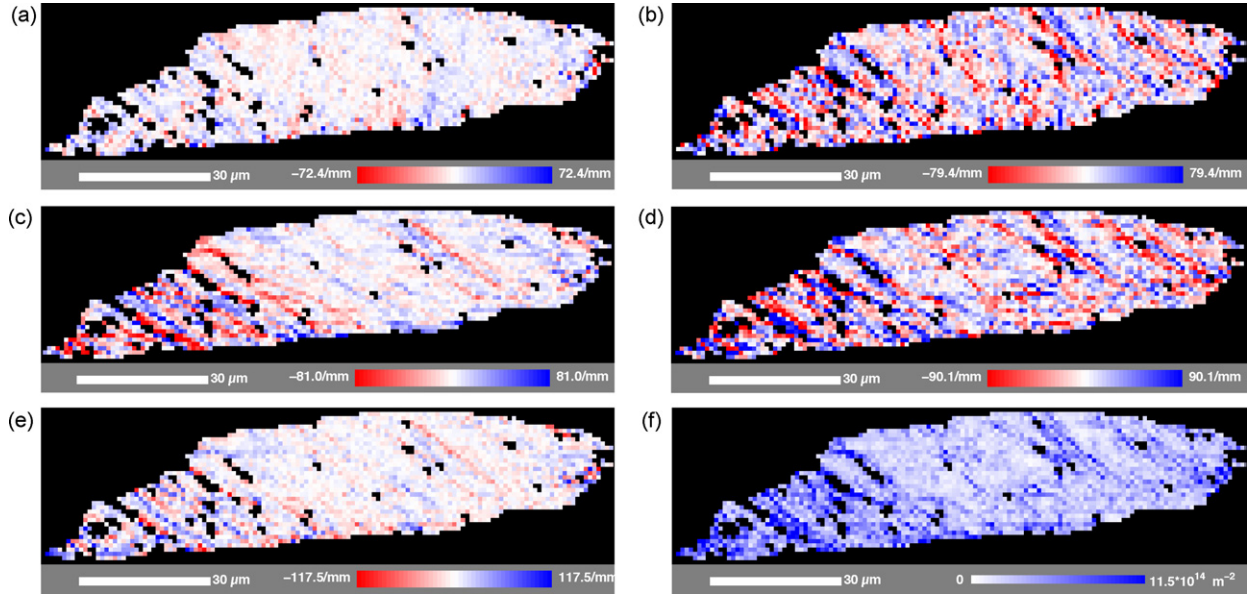


Fig. 5. Obtainable components of the dislocation density tensor from a grain in a 38% cold-rolled aluminium polycrystal: (a) α_{12} , (b) α_{13} , (c) α_{21} , (d) α_{23} and (e) α_{33} . (f) Total dislocation density estimated from the available components of the dislocation density tensor. Ranges are chosen as maximum values divided by 10.

Introducing the lattice curvature tensor $\kappa_{ji} = \omega_{j,i}$, Nye's original relation between the dislocation density tensor:

$$\alpha_{ij} = \kappa_{ji} - \delta_{ij}\kappa_{kk} \quad (17)$$

and the curvature tensor κ is finally obtained (with a different sign convention as in [14,15]).

4.2. Dislocation density tensor and EBSD

As lattice curvatures can be derived from local lattice orientations, the dislocation density tensor can be determined from spatially resolved orientation measurements. With conventional EBSD, orientations on a planar surface along two directions (e.g. x_1 and x_2) can be obtained. The difference in the local lattice rotations between neighboring points:

$$\Delta\vec{\omega} = \theta\vec{r} = \frac{2 \arccos \Delta q_0}{\sqrt{1 - \Delta q_0^2}} \Delta\vec{q} \approx 2\delta\vec{q} \quad \text{for } \theta \ll 1 \quad (18)$$

is given by the disorientation $\Delta\mathbf{q}$ between them and the components of the curvature tensor:

$$\kappa_{kl} = \frac{\partial \omega_k}{\partial x_l} \approx 2 \frac{\Delta q_k}{\Delta x_l} \quad (19)$$

can be found from the disorientation vectors $\Delta\vec{q}$. As spatial information is available only along two directions x_1 and x_2 , solely the six components κ_{i1} and κ_{i2} of the curvature tensor can be obtained, but not the components κ_{i3} . From the six accessible curvature components, five components of the dislocation density tensor can be found [16]:

$$\alpha_{12} = \kappa_{21}; \quad \alpha_{13} = \kappa_{31} \quad (20)$$

$$\alpha_{21} = \kappa_{12}; \quad \alpha_{23} = \kappa_{32} \quad (21)$$

$$\alpha_{33} = -\kappa_{11} - \kappa_{22} \quad (22)$$

Evidently, with conventional EBSD it is possible to obtain five components of the dislocation density tensor and not only three components (α_{13} , α_{23} and α_{33}) as hitherto stated [14,15]. (The reason for the (only apparent) problem for components α_{12} and α_{21} in other approaches is their use of the orientation matrix instead of the rotation vectors or quaternions, cf. [16]). The obtainable components of the dislocation density tensor are illustrated in Fig. 5a–e for the particular grain of S-orientation.

4.3. "Total" dislocation density

The "total" dislocation density is given by the sum of the absolute values of all components of the dislocation density tensor:

$$\rho_{\text{tot}} = \frac{1}{b} \sum_{i=1}^3 \sum_{j=1}^3 |\alpha_{ij}| \quad (23)$$

As some of the components cannot be assessed by conventional EBSD, the sum cannot be calculated. Alternatively, the apparent dislocation density, i.e. the sum off all available components:

$$\rho^* = \frac{1}{b} (|\alpha_{12}| + |\alpha_{13}| + |\alpha_{21}| + |\alpha_{23}| + |\alpha_{33}|) \quad (24)$$

may serve as approximative measure for the total dislocation density. An inhomogeneous distribution of the apparent dislocation density becomes obvious for the particular grain illustrated in Fig. 5f. Regions of lower dislocation density are separated by regions of higher dislocation density. In this manner, dislocation boundaries can be resolved by their enhanced dislocation density. Remarkably, boundaries can be recognized in both figures highlighting either the sign-carrying disorientation angle around the preferred rotation axis (Fig. 4a) or the apparent dislocation densities (Fig. 5f) by the marked alternating orientation change or by the enhanced dislocation density, respectively. The

most pronounced boundaries run parallel from the upper left corner to the lower right corner with an inclination of about -40° towards the rolling direction—in accordance with experimental observations by TEM [4].

4.4. Comments on the dislocation density

The calculated local curvatures and, hence, the derived dislocation densities depend strongly on the step size of the measurement grid, because variations in the lattice rotations on smaller distances than the step size become ignored. Consequently, dislocations causing orientation differences on a length scale less than the step size cannot be traced by the method. Only the dislocation content geometrically required by the orientation measurements is obtained and a large amount of dislocations may remain unresolved. Reducing the steps size resolves curvatures on a smaller length scale and allows determination of a larger fraction of the dislocations. As all dislocations are defined by the local geometry of the lattice [12,13], a concise determination of all dislocations would require determination of the local orientations on a scale below the dislocation distances. Even with the best available spatial resolution of 10 nm, this is not achievable and, for instance, narrow dipoles cannot be resolved.

More importantly, the registered dislocation content is strongly impaired by the limited angular resolution of the technique. Orientation differences can be accurately resolved only for disorientation angles above a critical value θ_{cr} which is typically between 0.5° and 1.5° . Disorientation angles below that value cannot be resolved accurately due to a lack in accuracy of the band detection, etc. The limited accuracy causes noise in the orientation data, spurious local curvatures of the order of $\kappa_{cr} = \theta_{cr}/\Delta x$ and, hence, an artificial dislocation density:

$$\rho_{cr} = \frac{\theta_{cr}}{\Delta x b} \quad (25)$$

If a minimum angle of 1° and the Burgers vector for Al (0.286 nm) is assumed, a typical step size of $1 \mu\text{m}$ (as used in the present investigation) will result in an acceptable noise in the dislocation density of $6.1 \times 10^{13} \text{ m}^{-2}$ corresponding to slightly deformed material (e.g., aluminium cold-rolled to about 5% [17]). On the other hand, a step size of 20 nm would result in an enormous artificial dislocation density of $3 \times 10^{15} \text{ m}^{-2}$ corresponding to heavily deformed material, e.g. aluminium cold-rolled to 98.2% [18]. The step size can therefore not be reduced below a certain length without creating artifacts and a certain amount of dislocations will always remain unresolved.

4.5. Full dislocation density tensor

Determination of the four missing components of the dislocation density tensor:

$$\alpha_{11} = -\kappa_{22} - \kappa_{33}; \quad \alpha_{31} = \kappa_{13} \quad (26)$$

$$\alpha_{22} = -\kappa_{11} - \kappa_{33}; \quad \alpha_{32} = \kappa_{23} \quad (27)$$

would require determination of the lacking curvature components κ_{i3} and hence orientation measurements along the third

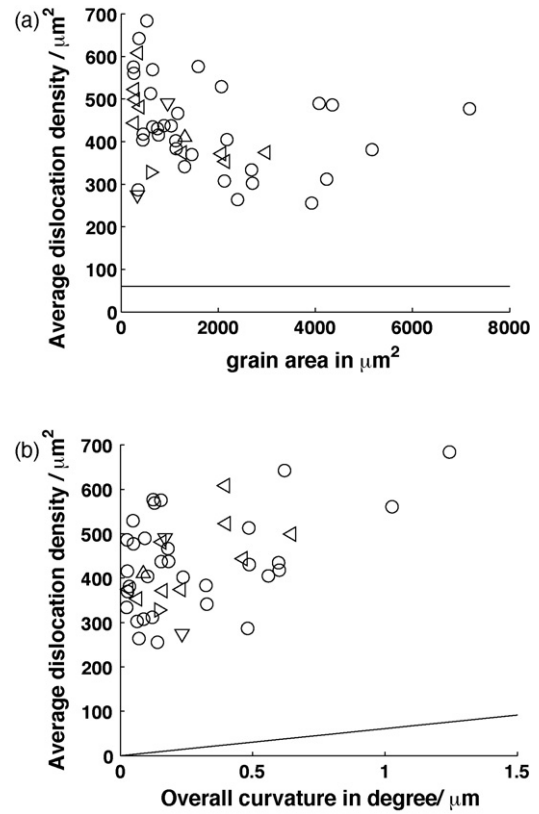


Fig. 6. Average apparent dislocation density in 45 individual grains as a function of (a) grain area in the orientation map and (b) overall curvature of the grain. Lines indicate (a) the accuracy limit and (b) the minimum dislocation density required from the overall bending. Symbols characterize the texture component defined in Table 1.

direction. This can be accomplished by serial sectioning and conventional EBSD on the new surfaces. The removal of individual layers can either be achieved by mechanical and electrochemical polishing outside the microscope or *in situ* in the SEM by a focussed ion beam. Layer thicknesses of the order of $2 \mu\text{m}$ are attainable by polishing [19], whereas the *in situ* technique allows removal of layers down to 50 nm [20]. If one aims for calculating the dislocation density tensor, the chosen step size in the 2D orientation determination must in each case correspond to the thickness of the removed layers, as the two components α_{11} and α_{22} depend on derivatives along different directions and the accuracy of the curvature determination ($\theta_{cr}/\Delta x$) is strongly affected by the step size. If the layer thickness is chosen differently, the accuracy of the different curvature components becomes different and their addition meaningless.

4.6. Average apparent dislocation density within individual grains

The average apparent dislocation density as defined in Eq. (24) has been analyzed for each of the 45 grains individually. It varies between 2.6×10^{14} and $6.8 \times 10^{14} \text{ m}^{-2}$, in all cases well above the mentioned accuracy limit of $6.1 \times 10^{13} \text{ m}^{-2}$. No correlation with the grain area is observed in Fig. 6a: some of the smallest grains have the highest average dislocation con-

tent, other the smallest average dislocation content. In larger grains locally enhanced dislocation densities average out. In Fig. 6b the average apparent dislocation density of each grain is shown as a function of the overall curvature (around the preferred rotation axis and along the direction of largest orientation spread) derived in the previous section. An increase of the average dislocation density with the overall curvature is indicated. For the overall curvatures, the minimum dislocation density $\rho_{\min} = \kappa^*/b$ required for creating the overall bending is calculated and included in Fig. 6b. Obviously, the measured average apparent dislocation density is well above the dislocation density required from the overall bending—indicating that the bending is of minor relevance for the total dislocation content.

5. Conclusion

An elaborate evaluation scheme for orientation maps obtained from EBSD is utilized for the investigation of the microstructure within individual grains of cold-rolled aluminium. Based on local orientation measurements the orientation distributions are analyzed in terms of the orientation spread in orientation space and in terms of local inhomogeneities within the grain. The orientation spread is characterized by means of the second-order central moments of the disorientation distribution in orientation space. It is found that only some of the smaller grains in the investigated sample show a small average spread and a large anisotropy of the orientation distributions. The preferred rotation axes of the individual grains are less likely oriented along the normal or rolling direction than along the transversal direction. Two distinct spatial patterns of orientations (long-range orientation gradients and alternating disorientations) can be distinguished by means of a sign-carrying disorientation angle with respect to the preferred rotation axis. From the local inhomogeneities, five components of the dislocation density tensor can be resolved. Determination of the apparent dislocation density allows resolving deformation-induced boundaries as regions of high dislocation density.

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