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DEFORMATION STRUCTURE VERSUS GRAIN STRUCTURE AFTER SEVERE PLASTIC DEFORMATION

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Dislocation boundaries form continuously during plastic deformation and subdivide grains of initially homogeneous orientation into smaller regions of deviating orientations The orientation differences increase with plastic strain and new high-angle grain boundaries form during severe plastic deformation, but a significant fraction of boundaries has lower disorientation angles reflecting their origin from an initial homogeneous grain, the orientations on both sides of deformation-induced boundaries are still related Orientation correlations can be resolved from spatial orientation data obtained e g by electron backscatter diffraction Deformation structures with inherent orientation correlations are distinguished from grain structures without orientation correlations between neighboring grains by the dependence of the disorientation angles on the distance between measuring points Quantitative analysis shows that the microstructure of severely deformed metals cannot be interpreted consistently as grain structure and must be characterized as deformation structure of fragmented grains with deformation-induced boundaries

1. Introduction

1.1 Grain structure. Most metallic specimens are built of many grains. Each individual grain in such a polycrystal is defined as contiguous region of homogeneous orientation separated from grains of different orientations by grain boundaries. Grain boundaries are associated with a finite orientation difference, but the orientation difference can be rather small, especially for strongly textured materials. Distinction of individual grains by their unique orientation separating them clearly from their neighbors is only possible for as-deposited, as-grown or fully recrystallized material. During plastic deformation grains of initially homogeneous orientations become subdivided by emerging deformation-induced boundaries. As evidenced by transmission electron microscopy (TEM), grains fragment into smaller regions [1] (subgrains, dislocation cells or cell blocks) of slightly deviating, but similar orientations. The orientation differences in the deformation structure increase strongly with plastic strain [2].

Electron backscatter diffraction (EBSD, e. g. [3]) can resolve the spatial arrangement of orientations. By means of a scanning electron microscope local orientations are determined from electron backscatter patterns at individual points on a specimen surface. Orientation data are usually gathered for many points on regular grids on the specimen surface. Any particular orientation of the crystalline lattice is described as rotation from a reference orientation [4]. The orientation difference between two orientations is characterized by a misorientation angle and axis. Taking into account the symmetry of the crystalline lattice (e. g. 24 symmetry operations for cubic symmetry), the lowest of all possible misorientation angles is selected and termed disorientation angle θ.

Identification of grain boundaries in an orientation map by a simple threshold for the disorientation angle might be misleading, as the disorientations within a single grain increase with proceeding deformation and some of the initial orientation differences between grains in the polycrystal might be rather small. In order to gain information on the grain structure, statistical methods must be employed. The most relevant parameter for characterization of grain structures is their grain size distribution. Because its evaluation from two-dimensional sections requires specific assumptions about the grain shape, only the chord length, i. e. the intercept length along a test line, and its distribution will be considered here, to avoid complications due to non-equiaxed grain shapes.

A strict mathematical scheme is presented for determining the chord length distribution from the dependence of the average disorientation angle on the step size between

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EBSD measuring points [5]. Such an indirect approach for determining the chord length from loosing any orientation correlation is applied to severely plastically deformed material where grain boundaries and deformation-induced boundaries cannot be distinguished by the amount of the disorientation angle any longer.

1.2 Severely plastically deformed metals. Grain sizes determined in materials after severe plastic deformation (e.g. after equal channel angular extrusion, ECAE [6]) depend strongly on the measuring method used. For instance, much smaller grain sizes are reported from direct inspection by transmission electron microscopy than from indirect determination from X-ray diffraction by finite size broadening or from evaluating orientation maps obtained by EBSD. These discrepancies pose not only a problem on the reliability of the reported grain size values, they also question the common characterization of severely plastically deformed material as nanostructures with grain sizes in the range of a few nanometer. It has been recognized that the differences between the reported values are due to fundamental differences in the measured quantity. With TEM, boundary spacings between all detectable boundaries are independent of their disorientation and, hence, the mean boundary spacing of all boundaries is determined directly, which is not necessarily the same as the grain boundary spacing.

Due to the relative ease of the technique, grain sizes are determined most commonly by electron backscatter diffraction. By EBSD orientations are determined on a regular grid and the disorientation angles between neighboring measuring points calculated. The density of point pairs with a disorientation angle above a certain threshold value along the entire measuring length can be interpreted as the density of boundary intersections, which is the inverse of the mean chord length. It is common practice to use a particular threshold value for the disorientation angle and to identify the presence of a boundary between two points on a regular grid by an orientation difference exceeding this threshold. The resulting measure for the chord length obviously depends on the step size of the measuring grid and the selected threshold value. In as-grown, as-deposited or recrystallized material ideally any small disorientation present is related to a grain boundary, but a meaningful threshold angle cannot be chosen below the noise level from the measuring accuracy of the EBSD system which for most commercially available systems is conservatively estimated about 2°.

1.3 Example. As example a high purity polycrystalline specimen of Al–0.13 wt.% Mg has been processed [7, 8] by equal channel angular extrusion (ECAE) through a die with an angle 120° between the channels without any bullet rotation (commonly referred to as route A) and deformed in 15 passes to an equivalent von Mises strain of 10. The processing details can be found in [9]. With high-resolution ESBD an orientation map has been gathered on a surface containing the extrusion direction (ED) and the normal direction (ND) with step sizes of 80 nm and 40 μm along ED and 16 μm along ND [8]. 69% of the measuring points have been indexed automatically, non-indexed points are ignored and no attempts are made to reconstruct the missing orientation data.

2. Chord length

Grain structures are characterized by their chord length distribution. The chord length l of a grain is the length of intercept of the grain with a test line. In traditional stereology, the number of intersections N_{is} of a (random) test line of length L with grain boundaries is counted. The number density $N_L = N_{is}/L$ of intersections is the inverse of the mean chord length $\bar{l} = 1/N_L$. Alternatively, individual chord lengths l_i of the intersected grains are recorded. From all distances l_i between two subsequent intersections the chord length distribution $F_{cl}(l)$ as well as the average chord length $\bar{l}$ is derived.

Both procedures are easily adapted to orientation data: If orientation data are gathered on grid lines with a constant step size Δx between subsequent points, disorientation

angles can be calculated between all N point pairs consisting of neighboring points along the grid lines. If a particular disorientation angle θ_i exceeds a certain threshold angle θ_{cr}, the existence of a grain boundary between the two points of a pair is assumed. From the number N_{cr} of such occurrences and the total line length $L = N\Delta x$ the intersection density $N_L = N_{cr}/\Delta x$ and the mean chord length $\bar{l} = N\Delta x / N_{cr}$ are inferred.

This procedure relies on a properly chosen sampling distance Δx and even more sensitive on the selection of a suitable threshold value θ_{cr}: As discussed in the introduction, neighboring grains in an ideal grain structure may have very small disorientations (especially for strong textures) and a rather small threshold value is advisable. Unfortunately, there are two restrictions for lowering the threshold to too small values: (i) With standard EBSD equipment and evaluation software, disorientation angles are determined reliably within 1° or 2°. (ii) Plastic deformation, on the other hand, gives rise to deformation-induced boundaries with significant disorientation angles. For instance, the average disorientation angle across a particular type of boundaries is 2° after 10% cold-rolling and increases with further deformation [10]. Both effects cause certain noise in the orientation determination. This makes the definition of a proper threshold value difficult (if not even impossible) and different evaluation schemes must be employed as the statistical approach described in subsection 4. A further effect is mentioned briefly: Occasionally, a particular electron backscatter pattern cannot be indexed properly and the orientation at a particular position cannot be determined. These non-indexed points have to be handled with care: the two suggested methods, ignoring the affected point pairs entirely or calculation of the disorientation with the next neighbor instead, bias the result in certain, but known and quantifiable manner. Other attempts presented in literature for establishing the missing information at a certain non-indexed point from weighted averaging over its neighbors or a filtering of the data are highly questionable, because their effect on statistics is difficult to access.

If the disorientation angles θ_i between all point pairs of (valid) neighboring points are determined and sorted in increasing order, the probability $P_\theta(\theta_{cr})$ for finding a disorientation angle θ less than an angle θ_{cr} is easily obtained from the serial number in the sorted list. The accumulated frequency distribution

$$P_\theta(\theta_{cr}) = 1 - \frac{N_{cr}}{N} \tag{1}$$

is intimately linked to the number N_{cr} of incidences of point pairs with angles above the critical angle.

Consequently, the resulting mean chord length depends both on the selection of a suitable threshold value θ_{cr} and the step size Δx

$$\bar{l}(\theta_{cr}) = \frac{N}{N_{cr}}\Delta x = \frac{\Delta x}{1 - P_\theta(\theta_{cr})}. \tag{2}$$

In case of the aluminium-magnesium alloy, the resulting value for the mean chord length determined after Eq. (2) from the orientation data depends strongly on the chosen threshold value as seen in Fig. 1. Along the extrusion direction, for instance, a threshold angle of 5° leads to a mean chord length of 4 μm, whereas a threshold angle of 15° results in a mean chord length of 6 μm, i.e. is enhanced by 50%. A threshold value close to the expected resolution limit of 2° decreases the mean chord length below 2 μm. This strong effect of the selected threshold value is unsatisfying and a scheme is proposed to allow an unbiased determination of the mean chord length without requiring the choice of a threshold angle.

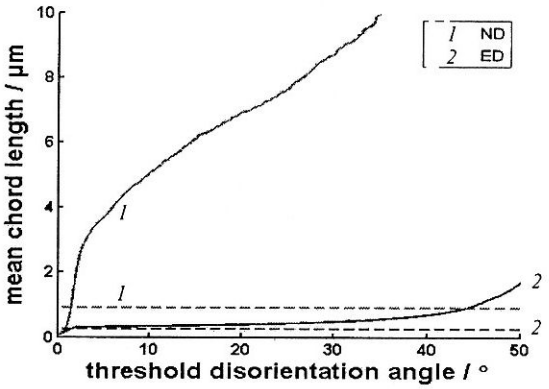
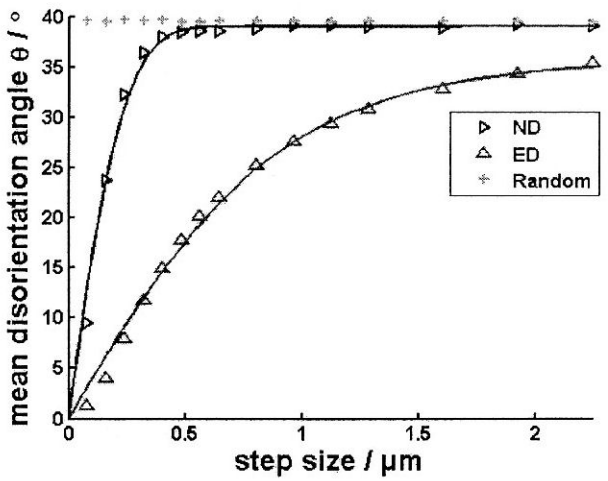


Fig 1 Mean chord length along normal and extrusion direction in Al–0 13 wt % Mg after ECAE processing to an equivalent strain of 15 in dependence on the selected threshold value θ_{cr} for the disorientation angle

3. Step size dependence

Disorientation distributions determined by EBSD are affected by the selected step size Δx in an analogous manner as the revealed mean chord length is affected by the selected threshold angle. Effects of an altered effective step size Δx are easily traced even from a single orientation map by calculating disorientation angles not only between neighboring points (corresponding to $\Delta x = \Delta x_0$), but determining them also between points in twice ($\Delta x = 2\Delta x_0$), trice ($\Delta x = 3\Delta x_0$) and multiple ($\Delta x = n\Delta x_0$) distances of the elementary step size Δx_0

From the orientation data obtained on Al–0.13 wt.% Mg all disorientation angles between points in different effective sampling distances $\Delta x = n\Delta x_0$ are calculated separately along both principal directions (extrusion and normal direction). The dependence of the average disorientation angle on the effective step size in Fig. 2 clearly exhibits an increase with increasing step size tending towards saturation for large step sizes.

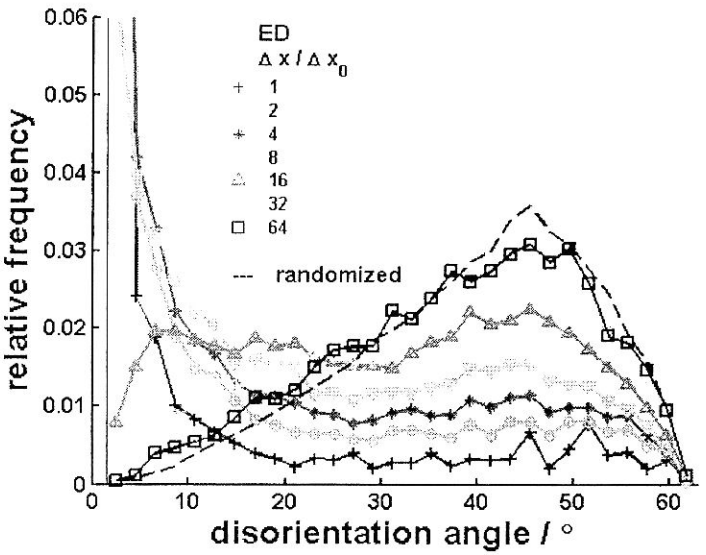


	ND	ED
$\langle\theta\rangle_{irr}$	0°	0°
$\overline{l}_{all}$ /μm	0 24	0 69
$\langle\theta\rangle_{DB}$	10 8°	16 6°
$\overline{l}_{GB}$ /μm	0 24	1 30
$\langle\theta\rangle_{gb}$	39 1°	35 6°

Fig. 2 Average disorientation angle $\langle\theta\rangle$ in dependence on step size Δx between points in a pair (along normal direction ND, and extrusion direction ED) for Al–0 13 wt % Mg deformed by ECAE to an equivalent strain of 15 Randomized average disorientation angles are obtained by applying the same analysis to an orientation map constructed by distributing exactly the same orientations randomly on the grid points The curves are fits (with optimal parameters summarized in the table) using a superposition of dislocation boundaries and grain boundaries with corresponding Rayleigh distributions as chord length distribution

For comparison purposes all orientations present in the orientation map have been randomly distributed on the map and evaluated again. This corresponds to calculating the average disorientation angle from the texture (or the uncorrelated MODF, respectively). Expectedly, the average disorientation angle (39.6°) obtained in this manner does not depend on the effective sampling distance and is about the same value as obtained (from the original orientation map) for large sampling distances along ND, where no orientation correlations are preserved any longer. The smaller value of the average disorientation angle at large distances along ED indicates a still preserved long range orientation correlation along the extrusion direction.

a



b

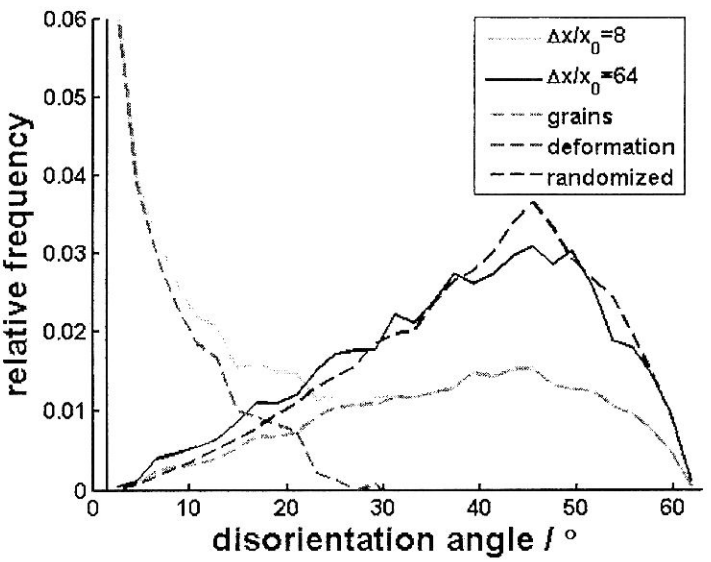


Fig 3 Disorientation distribution function for different step sizes Δx along the extrusion direction: (*a*) experimental distributions and (*b*) separation of distributions due to grain boundaries (magenta) and deformation-induced dislocation boundaries (red) Disorientation angles below 2° are not displayed The dashed black line represents the uncorrelated MODF

The uncorrelated MODF is derived from texture (by randomly picking orientations present in the map and calculating their disorientation) and ignores by definition any spatial correlation. It is close to a Mackenzie distribution [12], but not identical as the material is slightly textured. The deviations of the distributions for smaller step sizes at small disorientation angles are a consequence of orientation correlations in the deformed structure. During deformation, orientation differences in the original grains develop leading to subdivision of the original grains into fragments which develop mutual disorientations. These disorientations increase with plastic deformation and with proceeding (severe) plastic deformation some fragments start to behave independently instead of correlated. These two types of correlation behavior will be separated in section 5.

4. Grain structure [13]

Aiming towards an improved chord length evaluation, the distribution of disorientation angles is considered in detail for an ideal grain structure (without any orientation differences within the grains or any orientation correlation between the grains).The essential idea of the procedure is that when evaluating disorientation angles from orientation data additional to the relevant fraction of point pairs containing a boundary with a disorientation, there exists a fraction of irrelevant point pairs which do not contain a boundary at all, but for which a certain disorientation angle is recorded anyway.

4.1 Outline of problem. If individual orientations are measured along a grid line of length L with a step size Δx between subsequent points, the disorientation between all $N = \Delta x/L$ point pairs can be calculated. Only for N_{rel} point pairs, there will be a grain boundary between the two points, but no grain boundary will be found between the points of the other $N_{\mathrm{irr}} = N - N_{\mathrm{rel}}$ point pairs. The first mentioned point pairs form the relevant fraction and must possess non-vanishing disorientation angles θ_i, whereas the latter point pairs should not have any disorientation at all and are considered to be irrelevant. (This distinction between relevant and irrelevant point pairs has been introduced earlier in a similar approach of the author for analyzing deformation structures [14].) Due to the influences of the measuring technique mentioned in the previous subsection, there will be a certain disorientation originating from the noise, even if no boundary is present between the two points of a pair. Consequently, the total distribution of disorientation angles will be a superposition

$$f(\theta) = \xi_{rel} f_{rel}(\theta) + \xi_{irr} f_{irr}(\theta) \tag{3}$$

of a relevant distribution $f_{\mathrm{rel}}(\theta)$ from grain boundaries and an irrelevant distribution $f_{\mathrm{irr}}(\theta)$ from the accuracy of the measurement. An equivalent superposition holds for all moments of the distribution as the mean disorientation angle

$$\langle\theta\rangle = \xi_{irr} \langle\theta\rangle_{irr} + \xi_{rel} \langle\theta\rangle_{rel} . \tag{4}$$

The weighting factors are the relevant fraction $\xi_{\mathrm{rel}} = N_{\mathrm{rel}}/N$ and the irrelevant fraction $\xi_{\mathrm{irr}} = 1 - \xi_{\mathrm{rel}}$ of point pairs, respectively. For an ideal grain structure, the relevant distribution is a direct consequence of all orientations present in the material, i. e. the texture, and can be calculated from the texture as part of the uncorrelated misorientation distribution function MODF (at least, if no orientation correlations between grains are present). The irrelevant distribution is an instrumental distribution and depends on the EBSD equipment, the pattern recognition software as well as the particular patterns or the particular orientations investigated.

4.2 Relevant fraction. For a grain structure of known chord length distribution $f_{\mathrm{cl}}(l)$ the relevant fraction of point pairs can be derived theoretically in dependence on the step size Δx in the following manner: A point pair cannot contain any grain boundary, if the

first point of the pair is located further than Δx apart from the next grain boundary. The irrelevant fraction

$$\xi_{irr} = \frac{\int_{\Delta x}^{\infty} (1 - \Delta x) f_{cl}(l)\, dl}{\int_{0}^{\infty} f_{cl}(l)\, dl} \tag{5}$$

is determined by the probability of finding a point located in a distance larger than the step size Δx from the next grain boundary along the grid line. Consequently, the relevant fraction

$$\xi_{rel} = 1 - \xi_{irr} = \frac{1}{\bar{l}} \int_{0}^{\Delta x} \left(1 - \int_{0}^{l} f_{cl}(l')\, dl' \right) dl \tag{6}$$

(derived by partial integration from Eq. (5)) depends strongly on the sampling distance Δx. Independent of the particular chord length distribution, two limiting cases can be identified: For large step sizes, at least a single grain boundary is located between nearly all point pairs and all of them contribute to the relevant fraction ($\xi_{rel} - 1$). For small Δx, the linear relationship $\xi_{rel} = \Delta x / \bar{l}$ (which has been obtained already in an earlier attempt [14] on deformation structures) is confirmed.

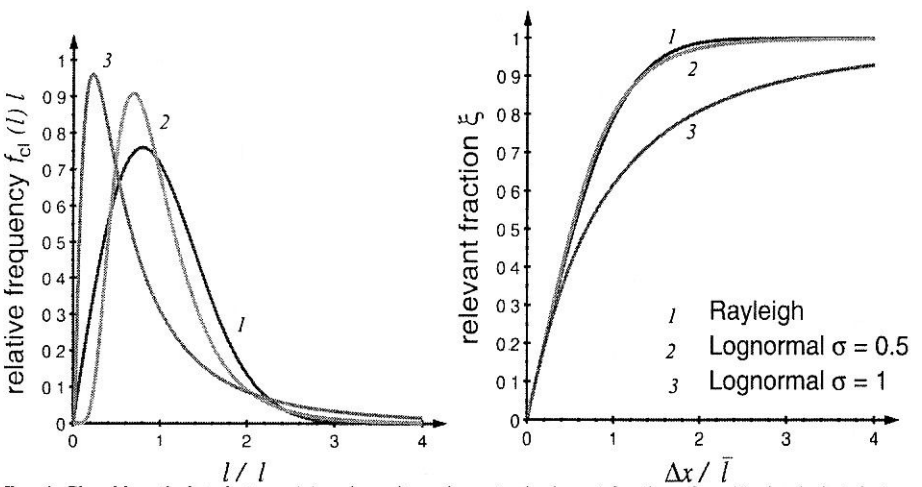


Fig 4 Chord length distributions (*a*) and resulting theoretical relevant fractions for a Rayleigh distribution and two logarithmic-normal distributions with different coefficients of variance (*b*)

The dependence of the relevant fraction on $\Delta x/\bar{l}$ is illustrated in Fig. 4 for two chord length distributions commonly used for grain structures. If the chord length distribution can be described by a Rayleigh distribution

$$f_{cl\,R}(l) = \frac{\pi}{2} \frac{1}{\bar{l}^2} \exp\left(-\frac{\pi}{4} \frac{l^2}{\bar{l}^2} \right), \tag{7}$$

the relevant fraction becomes

$$\xi_{rel}(\Delta x) = \operatorname{erf}\left(\frac{\pi}{2} \frac{\Delta x}{\bar{l}} \right). \tag{8}$$

A logarithmic-normal distribution (containing an additional parameter σ describing the coefficient of variance ($\sqrt{\exp(\sigma^2)}-1$)

$$f_{cl,ln}(l) = \frac{1}{\sqrt{2\pi}\,\sigma\, l}\exp\left(-\frac{\left(2\ln\left(l/\overline{l}\right)+\sigma^2\right)^2}{8\sigma^2}\right) \tag{9}$$

leads in an analogous manner to a relevant fraction

$$\xi_{rel,ln}(\Delta x) = \frac{1}{2}\left[1+\mathrm{erf}\left(\frac{\left(2\ln\left(\Delta x/\overline{l}\right)+\sigma^2\right)}{\sqrt{8}\sigma}\right)\right] + \frac{1}{2}\frac{\Delta x}{\overline{l}}\left[1-\mathrm{erf}\left(\frac{\left(2\ln\left(\Delta x/\overline{l}\right)+\sigma^2\right)}{\sqrt{8}\sigma}\right)\right]. \tag{10}$$

4.3 Determination of chord length distribution from experimental orientation data. The dependence of the relevant fraction Δx $\xi_{rel}(\Delta x)$ on step size can be derived from the dependence of the average disorientation angle $\langle\theta\rangle$ on step size in Eq. (4)

$$\langle\theta\rangle = \xi_{rel}(\Delta x)\left(\langle\theta\rangle_{rel} - \langle\theta\rangle_{irr}\right) + \langle\theta\rangle_{irr}. \tag{10}$$

In principle, the chord length distribution can be obtained from the second derivative of the relevant fraction with respect to the step size (cf. Eq. (6))

$$f_{cl}(l) = -\overline{l}\left[\frac{d^2\xi_{rel}}{d(\Delta x)^2}\right]_{\Delta x = l} \tag{11}$$

Due to numerical errors caused by differentiation, such a procedure is however not advisable. Alternatively, a reversed procedure can be applied: a proper distribution function (with free parameters) is selected as chord length distribution, the resulting relevant fraction calculated from Eq. (6) and fitted to experimental data of e.g. the average disorientation angle $\langle\theta\rangle$ $n\,\Delta x_0$ dependence on the step size according to Eq. (10). The latter can be gained from experimental data gathered with an elementary step size Δx_0 by calculating the disorientation angles and their average not only between neighboring points in the elementary distance Δx_0, but also for all possible pairs of points in larger distances $\Delta x = n\,\Delta x_0$. In this manner, average disorientation angles $\langle\theta\rangle$ $n\,\Delta x_0$ can be determined for all possible integers n.

4.4. Superposition of distributions [14]. The superposition of different fractions of disorientation angles leading to a dependence of the mean disorientation angle on the step size is also reflected in the distribution function for the disorientation angles. The shape of the distribution depends on the effective step size Δx, as seen in Fig. 3, *a*. Distributions obtained with different step sizes Δx_i differ in the corresponding relevant fractions. From a comparison of two different distributions obtained with different step sizes Δx_1 and Δx_2 the irrelevant distribution

$$f_{irr}(\theta) = \frac{\xi_{rel}(\Delta x_2) f_{\Delta x_1}(\theta) - \xi_{rel}(\Delta x_1) f_{\Delta x_2}(\theta)}{\xi_{rel}(\Delta x_2) - \xi_{rel}(\Delta x_1)} \tag{12}$$

and the relevant distribution

$$f_{rel}(\theta) = \frac{f_{\Delta x_2}(\theta) - f_{\Delta x_1}(\theta)}{\xi_{rel}(\Delta x_2) - \xi_{rel}(\Delta x_1)} + f_{irr}(\theta) \tag{13}$$

is obtained easily. Based on the linear dependence of the relevant fraction on the effective step size, this procedure has been utilized before [14].

5. Deformation structure

For severely plastically deformed material (as illustrated in Fig. 2), the determined mean boundary spacing depends strongly on the chosen threshold angle and step size; no appropriate choice for θ_{cr} becomes obvious. Therefore, the evaluation procedure developed for grain structures (which has been applied successfully to recrystallized material [13]) is modified for heavily deformed structures; in particular, for clarifying the debated issue of resolving the proper grain size of severely plastically deformed material.

In plastically deformed materials each grain of the original grain structure is subdivided by deformation-induced boundaries. Grain boundaries and deformation-induced dislocation boundaries coexist. During the early stages of deformation, the orientation differences associated with deformation-induced boundaries will be small and the orientation on both sides of the boundaries still correlated. As the orientation difference between fragments increases, the orientation correlation between some fragments is lost and some of the deformation-induced boundaries will transform to grain boundaries. Two different types of contributions to the disorientation distribution must hence be distinguished: one from deformation-induced boundaries which have not yet matured to grain boundaries and still preserve an orientation correlation; they are hence addressed as dislocation boundaries (DB); and a second from boundaries which have lost any orientation correlation, i.e. the grain boundaries (GB). The first contribution is characteristic for the deformation structure within grains ($f_{corr} = f_{GB}$)., the other for the grain structure and should be given by the uncorrelated MODF ($f_{uncorr} = f_{GB}$). The total distribution is consequently described as superposition of three different contributions

$$f(\theta) = \xi_{irr} f_{irr}(\theta) + \xi_{corr} f_{corr}(\theta) + \xi_{uncorr} f_{uncorr}(\theta). \qquad (14)$$

All weighting factor are related to the fraction ξ_{all} of all boundaries and the fraction $\xi_{gb} = \xi_{uncorr}$ of grain boundaries with lost orientation correlation, as the former is the complement of the irrelevant fraction $\xi_{all} = 1 - \xi_{irr}$ and the fraction of dislocation boundaries preserving an orientation correlation in the deformation structure $\xi_{corr} = \xi_{all} - \xi_{uncorr}$ is the difference between the fractions of all boundaries and the one of the grain boundaries.

$$f(\theta) = (1 - \xi_{all}) f_{irr}(\theta) + (\xi_{all} - \xi_{GB}) f_{corr}(\theta) + \xi_{GB} f_{uncorr}(\theta). \qquad (15)$$

Each fraction of boundaries depends on step size and is entirely determined by the chord length distribution or distance distribution of the respective boundaries, e.g. ξ_{GB} from the chord length distribution of the (uncorrelated) grain boundaries.

For a description of the experimental data on the aluminium-magnesium alloy, Rayleigh distributions for the chord length distributions of both boundary fractions, of all boundaries and of grain boundaries only with independent mean chord length $\bar{l}_{all}$ and $\bar{l}_{GB}$, respectively, are assumed and the corresponding fractions ξ_{all} and ξ_{GB} derived from Eq. (8). As seen in Fig. 2, a rather good correspondence between experimental observation and theoretical description is obtained by fitting the dependence of the mean disorientation angle on step size according to

$$\langle\theta\rangle = (1 - \xi_{all})\langle\theta\rangle_{irr} + (\xi_{all} - \xi_{corr})\langle\theta\rangle_{corr} + \xi_{uncorr}\langle\theta\rangle_{uncorr} \qquad (16)$$

Taking into account all boundaries a shorter mean chord length becomes obvious along the normal direction (0.24 μm) than along the extrusion direction (0.68 μm). An even more pronounced anisotropy is seen for grain boundaries only with 0.24 μm along ND and 1.30 μm along ED. Considering the macroscopic deformation such an elongation of the grain structure is expected. The results show that along the extrusion direction in average each grain is subdivided by a deformation-induced boundary into two subgrains or fragments, respectively. Whereas along the normal direction only a few dislocation boundaries are present and the mean of the chord lengths of all boundaries and that of grain boundaries only are nearly the same.

When describing the disorientation distributions as superposition of different contributions according to Eq. (15), the irrelevant part can be ignored, if only disorientation angles above 2° are considered as the mean disorientation angle of the irrelevant part is almost vanishing. The disorientation distributions in Fig. 3a are therefore described as superposition of only two contributions corresponding to the correlated $f_{\text{uncorr}}(\theta) = f_{\text{GB}}(\theta)$.and the uncorrelated distribution $f_{\text{uncorr}}(\theta) = f_{\text{GB}}(\theta)$ with weighting factors depending on step size. As discussed in section 4.4, two individual contributions can be separated, if the superposed distribution is known for two different step sizes. This is illustrated in Fig. 3b, where two distributions with $\Delta x = 8\Delta x_0$ and $\Delta x = 64\Delta x_0$ respectively, are used to calculate $f_{\text{GB}}(\theta)$ and $f_{\text{DB}}(\theta)$ analogous to Eqs. (12) and (13). Both resolved disorientation distributions are displayed in Fig. 3b to show their relative contribution to the distribution obtained with $\Delta x = 8\Delta x_0$. The resolved distribution for grain boundaries $f_{\text{GB}}(\theta)$ resembles the shape of the uncorrelated MODF. The distribution function originating from the dislocation boundaries becomes a smoothly decaying function ranging up to disorientation angles of 30°.

CONCLUSION

An evaluation scheme for orientation maps from EBSD is presented. Based on a classification of relevant and irrelevant point pairs and a distinction between correlated and uncorrelated disorientations, the contributions from different types of boundaries are separated by a statistical approach. In this manner, grain boundaries are distinguished from deformation-induced dislocation boundaries within the grains. The mean chord length and the disorientation angle distribution are evaluated separately for dislocation boundaries (still preserving orientation correlations) and grain boundaries (without any orientation correlation). Applying the approach to an ECAE processed aluminium-magnesium alloy reveals that the structure obtained after severe plastic deformation cannot be described properly as grain structure, but must be described as deformation structure with grains subdivided by deformation-induced boundaries in individual fragments.

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ЦЕНТРАЛЬНЫЙ НАУЧНО-ИССЛЕДОВАТЕЛЬСКИЙ ИНСТИТУТ
КОНСТРУКЦИОННЫХ МАТЕРИАЛОВ «ПРОМЕТЕЙ»

САНКТ-ПЕТЕРБУРГСКАЯ ОБЩЕСТВЕННАЯ ОРГАНИЗАЦИЯ
«ОБЩЕСТВО МАТЕРИАЛОВЕДОВ»

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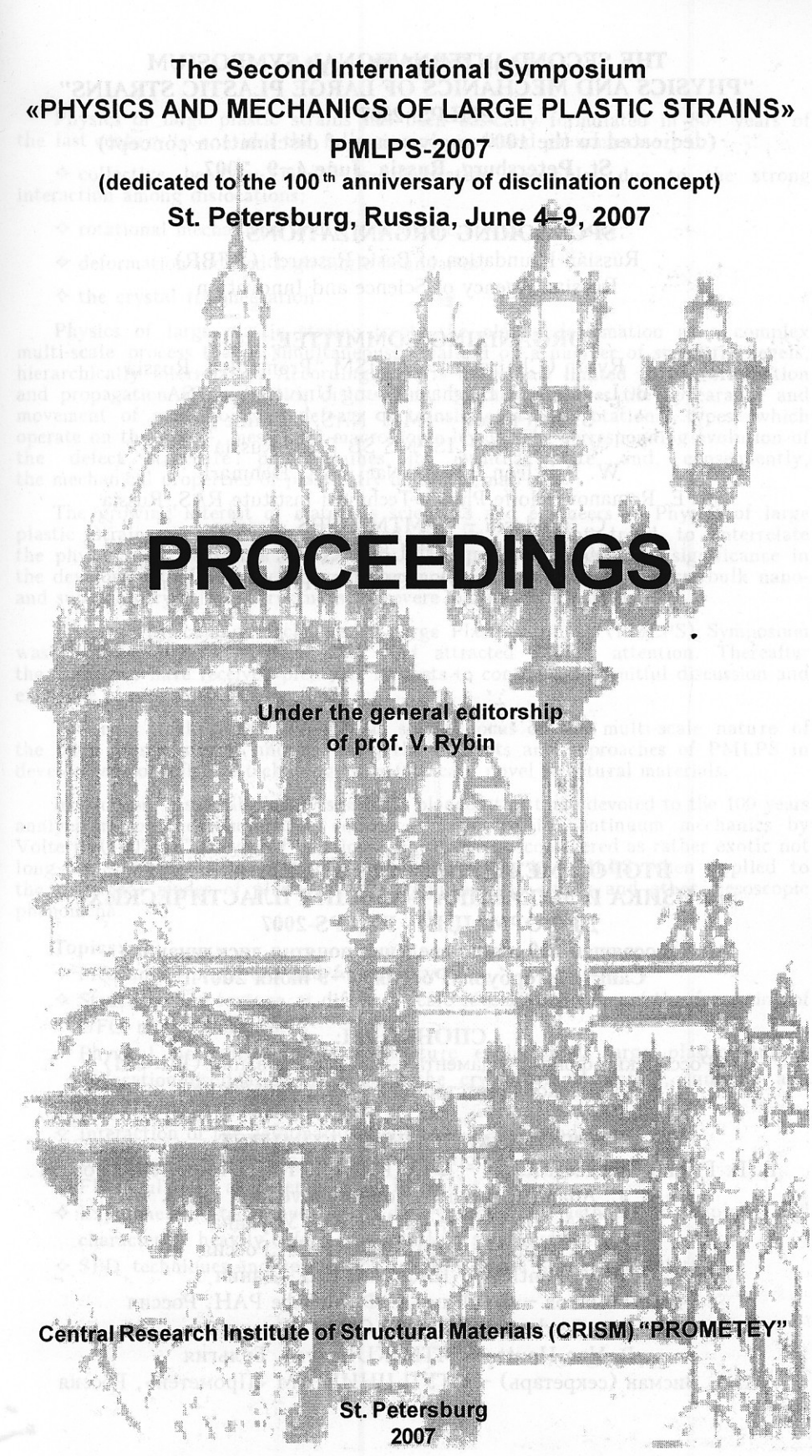

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