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Abstract. With the development of advanced electron and X-ray microscopy techniques, the local boundary migration during recrystallization and grain growth can be followed in 2D at a sample surface and/or in 3D inside bulk samples during in/ex situ annealing. The results show that locally boundaries migrate in a much more complex way than commonly imagined, for example by the development of local protrusions and retrusions and by migrating in a stop-go type of fashion. A quantitative analysis of the local boundary migration is essential for understanding this heterogeneous process. In this paper, methods for quantifying local boundary migration are summarized, including methods for quantification of local boundary migration velocities, misorientation (angle/axis pair) and plane normal of migrating/non-migrating boundaries, driving force (from both the energy stored in the deformation matrix and local boundary curvature), boundary mobility and activation energy. Through a quantitative analysis of both experimental and simulation data, local boundary migration is rationalized in several material systems during recrystallization and grain growth.

1. Introduction
In polycrystalline materials, grains with different crystallographic orientations are separated by grain boundaries. Boundary migration is involved when a grain grows into, or is consumed by, neighboring grains. The boundary migration rate, \( v \), is generally expressed as:
\[
v = MF,
\]
where \( M \) is the boundary mobility and \( F \) is the driving force for migration. \( F \) is typically considered as the driving force from the energy stored in the deformed matrix for recrystallization and as the boundary curvature induced driving force for grain growth [1]. New studies focusing on the local boundary migration during recrystallization have shown that the curvature-based driving force, \( F_c \), on recrystallizing boundaries from local protrusions/retrusions (developed when boundary segments migrate ahead/behind neighboring segments) [2,3] can be comparable in magnitude to that from the stored energy in the deformed matrix, \( F_s \). Therefore equation (1) is modified as [4,5]:
\[
v = M(F_s + F_c),
\]
for quantifying local recrystallization boundary migration.

Despite the simple equations, local boundary migration is not understood. This may be related to the fact that both boundary mobility and energy depend on temperature in complex ways, and are affected by material purity, and depend on the misorientation and plane normal of the boundary, which covers a 5D parameter space (3D for the misorientation plus 2D for the boundary plane normal) [1].
The migration of recrystallization boundaries is even more complex, where the local variation in the deformed microstructure has an additional impact. Experimentally, only equation (1) has been validated for boundary migration during the grain growth process using idealized samples, e.g. bicrystals [6–11]. However, the results cover only a very small proportion of the possible 5D parameter space, and the validity of such results on recrystallization boundaries and on boundaries in engineering materials have not yet been studied. Computer modeling using molecular dynamics has been conducted to determine both the boundary energy and mobility for a relatively large number of boundaries in different material systems [12–17]. Experimental validations on such results are, however, missing. As such new, sufficiently fast, experimental methods for quantification of boundary migration are required.

In the last 2-3 decades, several new experimental characterization techniques have been developed and matured, such as in-situ/ex-situ measurements using electron backscattered diffraction (EBSD) [18,19] and transmission electron microscopy (TEM) [20–22], 3D X-ray diffraction (3DXRD) [23–25], synchrotron and laboratory-based diffraction contract tomography (DCT) [26–30], and 3D X-ray Laue micro-diffraction [31–34]. With these techniques local migration of individual boundaries during recrystallization and grain growth can be followed in 2D or 3D in polycrystalline materials. By correlating the migrating boundary with the microstructure ‘swept’ by the boundary, the local boundary migration velocity, misorientation and plane normal (for 3D data), and driving force for individual boundary segments can be quantified. With such quantitative analyses, the physical mechanisms behind the local boundary migration behavior may be rationalized, which is essential for developing advanced models that can predict local migration behavior more precisely.

Nevertheless, quantitative analysis of local boundary migration based on crystallographic orientation data in polycrystalline is still relatively new. The analysis procedure is also rather complicated, involving sophisticated image processing steps, such as data registration and segmentation, morphological operations, and in some cases surface meshing. So far no commercial software is available to perform all these steps. The present paper aims to summarize the required analysis procedure. The results obtained using such quantitative analysis methods on several material systems are then be summarized for a better understanding of local boundary migration during recrystallization and grain growth.

2. Quantification of procedure

The procedures described in this section are those required for analysing local boundary migration based on 2D/3D crystallographic orientation data collected from roughly the same sample area/volume over a certain time period during in situ or ex situ annealing.

2.1. Data registration

Data registration is typically the first task for the quantitative analysis. This is because data collected at different annealing time steps may not be aligned, especially for data collected during ex-situ annealing, where the sample has been demounted and remounted from the measurement stage for annealing treatment. Also in practice, different characterization techniques can be combined to optimize the spatial, temporal and angular resolutions for comprehensive characterization of the microstructural evolution [35]. In such a case, data registration between multiple data sources is also required. Moreover, if the time spent on collecting data at each individual time step is long, quite often the data can be spatially distorted due to thermal and/or mechanical drift of the sample and/or drift of the electron/X-ray beam [36]. In that case, data correction (registration) according to some distortion-free data has to be carried out for each time step.

For the registration, markers or control points (CPs) selected according to features in the microstructure or on the sample are typically required to determine the transformation function between data collected from neighboring time series. An example of how a spatially distorted EBSD map can be registered is given here to demonstrate the registration procedure (see figure 1). The SEM image collected after EBSD mapping is shown in figure 1a. The non-rectangular darker shaded region
shows clearly the drifting that has taken place during the mapping. In this case an electron channeling
counter (ECC) image (figure 1c) is used as a distortion-free reference image for the registration, as
such images can typically be collected within a relative short time (few minutes). A small horizontal
twin marked by the white arrow in figure 1c appears inclined in the EBSD map (figure 1b),
demonstrating the local spatial distortion in the EBSD map. To register the EBSD map, 60 CPs were
selected manually from the microstructure (figure 1d), based on which a transformation function was
computed. Many transformation functions can be used. For example, rotation, translation, similarity
and aff ine transformation are well-known functions, all of which transform an image rigidly, i.e. line
features in the image are reserved after transformation. More sophisticated non-rigid transformation
functions include polynomial and thin plate spline (TPS) functions. Registered EBSD maps using TPS
and polynomial functions are shown in figures 1e and 1f, respectively. The shape of the registered
EBSD map using a TPS function matches well the darker shaded region in figure 1a, and locally the
twin marked by the white arrow has also been correctly restored (see figure 1e), demonstrating the
accuracy of the data registration on both the global and local scales [36]. Registration with the more
commonly used polynomial function [37] is less satisfactory (see figure 1f), as polynomial functions
are more suitable for correcting global distortions while TPS is suitable also for correcting locally
varying distortions [36].

![Image](image_url)

**Figure 1.** Example showing the procedure for data registration of an EBSD map. (a) secondary
electron image showing carbon contaminated regions (darker shaded areas) after EBSD mapping. (b)
and (c) are EBSD map and ECC image, respectively, collected at the same surface region in a partially
recrystallized Ni sample. (d) Control point pairs collected manually based on the microstructure in (b)
and (c). (e) and (f) are registered EBSD maps using a TPS and a polynomial function, respectively.
The white arrows in (b), (c), (e) and (f) mark the same twin. Reproduced from [36].

The registration procedure for 3D time series data sets is similar to that described above for 2D
data, except that the transformation functions are 3D functions determined based on 3D CP
coordinates. For 3D data sets, the overall 3D sample shape and fiducial markers on the sample surface can be used as markers [38]. Crystallographic orientations of the same grain measured at different time steps can also be used to determine any rigid rotation of the sample. For systems containing many grains, the center of mass coordinates of the grains can be used as CPs [39]. However, the center of mass coordinate of a grain may change significantly during annealing if the grain grows or shrinks anisotropically. Transformation results based on this measure typically therefore need some crosscheck using other types of markers.

**Figure 2.** Examples showing data segmentation and methods for calculation of migration velocity based on aligned ex-situ data. (a) and (c) are aligned ECC images showing the microstructure of partially recrystallized pure Al at the starting state (annealed at 250 °C for 10+15 min) and after 6 steps of annealing at 250 °C for 15 min each. (b) EBSD map showing the same microstructure as that in (a). The colors in (b) represent the crystallographic direction parallel to sample ND (see the inset pole figure). (d) Traces of the recrystallization boundary after each annealing step; 7 steps at 250 °C and the last 2 steps at 260 °C for 15 min each. Areas of different color/shading show the regions consumed during different annealing steps. Vertical lines are used for determining local migration velocity using a line intercept method. (e) Segmented EBSD data using the area masks in (d) for the first annealing step. (f) Segmentation to local regions marked by the five numbered boxes. The dashed lines in (d) and (f) represent the boundary grooves seen in (a) and (c), developed during electropolishing before the ex-situ annealing experiments. Reproduced from [33].

2.1.1. Data segmentation. After data registration, time series of data can then be segmented into subsets for further quantification. An example of a registered 2D data set is shown in figure 2, where the migration of a recrystallizing boundary into a deformed matrix in pure Al (50% cold rolled) has been followed using ECC after ex-situ annealing at 250 °C and 260 °C for various time intervals [40]. An EBSD map (figure 2b) has also been collected before the ex-situ study and registered according to the corresponding ECC image. By tracking the boundary position at each annealing step, the relative
position of the boundary of interest can be determined as a function of annealing time (see the coloured boundary traces in figures 2d and f). From the aligned boundary traces, the area ‘swept’ by the recrystallizing boundary during each annealing step can be segmented (see the areas of different colour in figure 2d). The segmented areas can then be used as masks to segment the EBSD data for local quantification of temporal changes (see e.g. figure 2e). In addition, the boundary can also be segmented spatially along the boundary into local segments for quantification of spatial variations (see e.g. the five boxes in figure 2f).

2.2. Quantification of local boundary migration velocity

Three commonly used methods [35,41] for quantification of local boundary migration velocity are summarized in the following:

i) The migration velocity, \( v \), of the boundary can be determined using equation (3)

\[
v = \frac{A_{\text{Cons}}}{lt},
\]

where \( A_{\text{Cons}} \) is the area through which a migrating boundary of a length, \( l \), migrates over a given annealing time interval, \( t \). Both \( A_{\text{Cons}} \) and \( l \) can be quantified from the aligned data (see for example the colored areas in figure 2d). For 3D boundaries, \( A_{\text{Cons}} \) and \( l \) are replaced with volume, \( V_{\text{Cons}} \), and area, \( A \), of a migrating boundary, respectively. The measured velocity is an average migration velocity for local boundary segments, with length/area of \( l/A \) in 2D/3D.

ii) The velocity can also be measured by a line-intersect method, i.e. by drawing a line across the migrating boundary and measuring the migration distance during each time interval (see the red intersect lines in figure 2d). Using this method, the measured velocity values are valid for specific points on the boundary along a specific direction. The values depend therefore on the line direction, which sometimes can be difficult to define when the migration pattern is complex and the migration direction is difficult to define.

iii) To overcome the line direction problem, a Euclidean distance transformation method can be used to determine the migration distance for specific points on the boundary. For each pixel/voxel on the initial boundary, the Euclidean distance transformation determines the migration distance as the minimum distance from the pixel/voxel of interest to all the pixels/voxels on the new boundary. An example is shown in figure 3, where the migration distance for a 3D recrystallizing boundary in pure Al is colored according to the migration distance in figure 3e [41]. The migration direction determined with this method is different for each pixel/voxel. This method is well suited for boundary migration along multiple directions for convex grains.

2.3. Boundary characteristics for migrating and non-migrating boundary segments

A boundary can be separated into migrating and non-migrating parts, based on the quantified local migration velocities. The non-migrating parts are defined as those parts that migrate less than a certain amount (e.g. the step size of the measurement), while the remaining parts are defined as the migrating parts. In practice all pixels/voxels within consumed regions (see e.g. figure 2e) are considered for calculation of misorientations (angle/axis pairs) for migrating boundary segments, as at a certain time during annealing the migrating boundary has reached this position.

For determining the boundary plane normal based on 3D data set, segmentation is carried out only for the starting boundary for each annealing step. An example of the segmentation for the 3D boundary in figure 3 is shown in figure 4a and 4b. The voxelized data is then meshed into triangles for calculation of the boundary plane normal distribution (see figure 4c and 4d).
Figure 3. An example showing the calculation of the local migration velocity for a 3D migrating boundary. (a) 3D view of the gauge volume in partially recrystallized Al mapped by micro-diffraction. (b) and (c) are subsets of the volume in (a) visualizing the recrystallized grain and the deformed matrix, respectively. (d) the deformed part of the gauge volume after two annealing steps. The colors in (a)-(d) represent crystallographic orientations along the sample rolling direction (RD) (see the color code below (a)). (e) the initial recrystallizing boundary, colored by the migration distance. The black voxels in (a) represent non-indexed voxels. The box size along x, y, z, is 28, 32, 95µm, respectively. The gray and black arrows in (c) mark faceted and curved parts of the boundary, respectively. [41]

Figure 4. Partitioning of the 3D recrystallization boundary in figure 3a, based on the calculated migration distance at each position on the boundary, into (a) migrating parts and (b) non-migrating parts (defined here as parts migrating by less than 1 voxel) [41]. (c) and (d) show the plane normal distributions in the crystal system for the boundary segments in (a) and (b), respectively.
2.4. Quantification of driving force

The driving force for recrystallization is considered as both the driving force from the energy stored in the deformed grains and that induced by boundary curvature (see equation (2)), while only the latter is typically considered for grain growth. For both cases, a knowledge of grain boundary energies is required, which is typically determined using the Read-Shockley equation to describe the energy as a function of the misorientation of each boundary. A recent 3D study using LabDCT has shown that this equation describes in general well the boundary energies in real engineering materials [42].

For recrystallization, the stored energy can be calculated using the methods described in [43]. In some cases, it may be required to perform orientation filtering to reduce orientation noise [40,44]. Similar to the case for quantification of boundary misorientations, the analysis can also be carried out for local areas/volumes consumed by the growing grains, which represents the ‘true’ stored energy that leads to the migration, and for areas/volumes of a certain thickness in front of non-migrating boundary segments. This type of quantification is more accurate for understanding the local boundary migration than the averaged stored energy determined from whole samples [19,45,46].

For both recrystallization and grain growth, the curvature-based driving force, \( F_\sigma \), is calculated numerically based on the boundary shape. The driving force, \( F_{\sigma,i} \), for a position, \( i \), at a boundary is determined using equation \( F_{\sigma,i} = 2\sigma / r_i \), where \( \sigma \) is the boundary energy and \( r_i \) is the radius of the curvature at that position. \( F_\sigma \) is then calculated by integrating \( F_{\sigma,i} \) over the length, \( \Delta l \), of boundary segments within the area investigated, i.e. \( F_\sigma = \int_{\Delta l} F_{\sigma,i} \). So far, this kind of quantification has been carried out mainly for 2D data sets. For 3D data sets, the integration has to be performed over the whole area of a boundary.

\[ F_\sigma = \int_{\Delta l} F_{\sigma,i} \]

Figure 5. Migration velocity \( v \) for the boundary segments in the five boxes marked in figure 2f as a function of total driving force, \( F = F_S + F_\sigma \) for recrystallization at 250 °C. (a) – (e) are for boundary segments in the boxes I–V, respectively. The error bars represent standard deviations. Numbers are used to identify annealing intervals. [40]

2.5. Quantification of local mobilities

With both \( v \) and \( F \) quantified for local boundaries segments, the boundary mobility can be determined for cases when a linear function fits the \( v \) and \( F \) data well. For example, the \( v-F \) plots for the boundary segments in the five boxes in figure 2f for recrystallization at 250 °C are shown in figure 5, based on which the apparent mobility for the boundary segments within boxes III and IV is determined to be
6.2 ± 0.4 × 10^{−14} \text{m}^4\text{s}^{−1}\text{J}^{−1}. It is not possible to determine the apparent mobilities for the boundary segments in other boxes due to the large scatter in the \(\nu\)-\(F\) plot (i.e. poor linear fitting), which matches a large orientation spread in the deformed matrix in front of the those local boundary segments [40].

2.6. Quantification of activation energy

Boundary mobility is a thermally activated parameter, and can be expressed by:

\[
M = M_0 \exp(-Q/RT),
\]

where \(Q\) is activation energy, \(M_0\) a pre-exponential factor, \(R\) the gas constant, and \(T\) temperature. Much work has been carried out on grain growth using bicrystal experiments (e.g. [7]). In contrast very little effort has been given to local analysis of individual recrystallizing boundaries, due to the complexity of the boundaries. Here a new method is proposed, which is based on the local analysis of in-situ/ex-situ data sets.

By measuring the boundary apparent mobilities at different temperatures using the method described in section 2.5, the activation energy can be obtained. For example, if the boundary apparent mobilities measured at temperatures \(T_1\) and \(T_2\) are \(M_1\) and \(M_2\), respectively, the activation energy can be determined as:

\[
Q = \ln(M_2/M_1)RT_1T_2/(T_2-T_1). 
\]

For example, for the boundary segments within box III in figure 2, the activation energy determined using equation (5) is \(\sim 280\ \text{kJ/mol}\). This value is well within the range determined based on grain growth experiments for pure Al [7,9].

Note that one has to be careful when determining activation energy for local recrystallizing boundaries. The characteristics (misorientation and plane normal) of the boundary should not change during the investigated annealing period. For example, the boundary segments within box IV in figure 2 migrates into a region with new misorientation during the last annealing steps at 260 °C with a relatively low mobility [40]. If the activation energy for boundary segments is determined using equation (5) assuming that the boundary characteristics have not changed, an activation energy value of \(\sim 100\ \text{kJ/mol}\) is obtained. This value is only about 1/3 of that for the boundary segments with the same misorientation in box III, and this is clearly not correct.

3. Local boundary migration

The local boundary migration has been followed and quantified using the methods described in section 2 in several material systems, including recrystallization boundary migration in heavily deformed as well as medium strained metals, and boundary migration during grain growth. Several examples are presented in the following. Quantitative studies of local boundary migration based on phase field simulations are also exemplified.

3.1. Boundary migration in heavily deformed metals

3.1.1. Triple junction motion. After rolling deformation to high strains (\(\varepsilon_{\text{vM}}>3.5\)), the microstructure contains typically lamellar boundaries almost parallel to the sample rolling direction (RD), connected by incidental dislocation boundaries (IDBs) within the lamellae (see e.g., figure 6a). During annealing, boundary migration in connection to triple junction (TJ) motion prior recrystallization nucleation has been seen in Al and Ni [21,47–49]. Figure 6 shows an example of TJ motion in heavily deformed pure Ni followed during in situ annealing using TEM. The migrating TJs and the moving direction is marked by the red arrows in figure 6c. It is seen that TJ motion leads to microstructural coarsening without much changing of the morphology of the microstructure (figures 6b and 6d). By combining in situ TEM with transmission Kikuchi diffraction (TKD), the boundary characteristics of the migrating TJs can be studied. It has been shown that both low angle boundaries (LABs) and high angle boundaries (HABs) can migrate in connection with TJ motion; see the thin and
bold lines, respectively, in figures 6c and 6d. The misorientation distribution for the moving boundaries have been quantified based on ex-situ EBSD studies (see figure 7). The results show that the misorientations for the migrating boundaries cover a large range, 3-63°. The relative frequency of moving LABs is much lower than that of HABs, considering the fact that the fraction of LABs is about 50% in the deformed matrix in this material [50,51]. Among all the migrating boundaries, boundaries with rotation axes close to \(\langle 111\rangle\) have a relatively high frequency (see figure 7).

Figure 6. An example showing TJ motion in heavily deformed pure Ni prepared by accumulative rolling bonding to a von Mises strain of 6.4. (a) and (b) are TEM images showing the microstructure before and after in situ annealing. Letter A marks the same position. (c) and (d) are drawings showing boundary networks and TJs. Bold and thin lines indicate HABs and LABs, respectively. In (c) arrows indicate the migration direction of Y-junctions. Changes caused by TJ motion are shown in grey in (d). The LABs and HABs are defined as boundaries with misorientation <15° and \(\geq 15°\), respectively, determined based on TKD results. [48]

Figure 7. Misorientation (angle/axis pairs) distribution for migrating boundaries in connection with TJ motion during annealing at 200 °C for 1h of heavily deformed Ni prepared by ARB to a von Mises strain of 4.8. The result is obtained from an ex-situ EBSD study.

3.1.2. Recrystallization boundary migration. Boundary migration during recrystallization has been followed by ex-situ EBSD in heavily deformed Ni, Cu and steel. It is found that the recrystallizing boundaries migrate relatively faster along RD than along ND [52,53]. A large misorientation range is seen for the migrating and non-migrating boundaries. A relatively high frequency of HABs with rotation axes close to \(\langle 111\rangle\) is seen for both FCC and BCC metals (see figure 8) [53–55]. A comparison between the misorientations for migrating and non-migrating boundaries has shown no significant differences, except that there is a relatively high fraction of LABs and boundaries with misorientation close to 60°\(\langle 111\rangle\) for the non-migrating boundaries (see figure 8a,b) [52]. The local stored energies quantified based on EBSD maps within the consumed regions and the regions in front of non-migrating boundaries are also similar [53,54]. The results have also shown that the boundary segments along RD are relatively flat and typically migrate slowly [52,53].

The results have further shown that some parts of the non-migrating boundary segments at any given annealing step can migrate upon further annealing, i.e. the boundary segments migrate in a stop-go fashion. Quantification of local stored energy in front of the non-migrating and migrating boundary segments after different annealing steps shows that concurrent recovery of the deformation matrix is not the reason for the stop-go motion [53].
3.2. Boundary migration during recrystallization of pure Al deformed to medium strain

3.2.1. 3D boundary migration. Local boundary migration during recrystallization in 50% cold rolled pure Al has been studied using 3D synchrotron Laue micro-diffraction (see figure 3). The 3D boundary has a staircase shape, consisting of both curved and facet parts (see figure 3c and d). Quantitative analysis shows that the curved parts migrate faster than the facets (see figure 3e). The boundary migrates by movement of the curved parts laterally along the facets, thereby maintaining the staircase shape of the boundary. As a result, the facet edges are either extended or shortened during boundary migration [41].

![Migration distribution for migrating and non-migrating boundaries during recrystallization.](image)

**Figure 8.** Migration distribution for migrating and non-migrating boundaries during recrystallization. (a) and (b) are for migrating and non-migrating boundaries, respectively, for cold rolled pure Ni [52]. (c) and (d) are for migrating boundaries in cold rolled pure Cu [53] and steel processed by dynamic plastic deformation [55].

Misorientation analysis shows that there is no significant difference between the misorientation distributions for curved parts and facets, i.e. migrating and non-migrating boundary segments, respectively. The stored energies quantified based on the misorientations for the regions consumed by the curved migrating boundary are similar to those for the regions in front of non-migrating facets (see figure 9). The curvature-based driving force is relatively small and is not the main reason for the observed difference between the migrating and non-migrating boundary segments [41].

An analysis of boundary plane normals from the 3D studies reveals a significant difference for the curved migrating and facet non-migrating boundary segments (see figure 4). However, MD simulations reveal that the mobility difference for the curved and facets parts is small, suggesting that the mobility difference between the migrating and non-migrating parts can be small, i.e. the boundary plane normal is also not the main reason for the observed difference for this particular 3D boundary.
By correlating the deformed microstructure and the migration pattern of the recrystallizing boundary, it is found that the migration of the curved parts follows the (111) set of dislocation boundaries (see figure 9e). This result suggests that the curved part is likely to be promoted by these favorably aligned dislocation boundaries, which supports a previous suggestion based on destructive 3D [56] and in-situ 2D investigations [35,57].

3.2.2. Boundary mobility. The local boundary migration during recrystallization in the same pure Al sample discussed above in section 3.2.1 has been followed during ex-situ annealing at 250 °C using ECC and EBSD (see figure 2). Based on the local analysis in the regions marked by the five boxes in figure 2, the local migration behaviour has been analysed in detail, from which the local mobilities have been quantified (see section 2.5).

By correlating the deformed microstructure and the boundary traces (see figure 10), some further information can be obtained. It is found that most of the pinning points (e.g. A–D in figure 10) coincide with ⟨110⟩ rotation axes (green in figure 10b), while the misorientation angles are quite scattered. The local stored energies at these points are relatively high (figure 10c). It is therefore very likely that the boundary segments with misorientation axes close to ⟨110⟩ have comparatively lower apparent mobilities than those with rotation axes close to ⟨111⟩.

3.3. Boundary migration during grain growth

3.3.1. Grain growth of cube textured Ni. Grain growth of a cube textured pure Ni has been followed during ex-situ annealing at 600 °C using EBSD. It is found that the misorientation angles for the migrating boundaries cover a wide range (see figure 11). Compared to the misorientation distribution...
for all the boundaries in the sample (figure 11a, twin boundaries are not included), the results show that except for boundaries with misorientations lower than 5° the two distributions are similar [58].

Figure 10. Boundary traces overlaid on an EBSD map for the corresponding area. In the EBSD map, each pixel in the deformed matrix is colored according to the misorientation angle (a), the misorientation rotation axis (b) between the pixel and the recrystallized grain, and the local stored energy (c). A–D mark pinning points. The local stored energy at each pixel is calculated from a subset of 11 × 11 pixels centered at each selected pixel using the method described in section 2.4. [40]

Figure 11. (a) Evolution of boundary misorientation distributions during annealing for annealing times of 3.5, 43.5 and 243.5 min in pure Ni. (b) Misorientation distribution for all boundaries observed to migrate, summarized over all annealing steps. Reproduced from [58].

A close-up examination of the individual migrating boundaries in this sample shows that some boundaries with misorientation close to Σ3 boundaries (following the Brandon criteria) are highly mobile. These boundaries typically have a large deviation angle (>4°) to the ideal twin relationship, i.e. 60°(111), and therefore are not twin boundaries (see figure 12a-d). A coherent or incoherent twin boundary can be elongated by the migration of their connected boundaries (see figure 12e and f); similar results are seen in pure Al [59]. More surprisingly, it is also observed that a boundary can migrate even though it has a negative curvature (see the black boundary around grain C in figure 12c and d). This underlines the importance of the grain boundary network connectivity.
3.3.2. 3D grain growth of pure iron.

The grain growth of pure iron has been followed in 3D during ex-situ annealing using LabDCT and synchrotron DCT (see e.g., figure 13). Similar to pure Ni, a large misorientation spread is seen for the migrating boundaries, and the misorientation distribution for these boundaries is similar to that for all the boundaries in the characterized volume [60]. The 3D result shows that in this sample locally some boundaries having negative curvature also migrate. Quantification of the growth rate for individual grains as a function of mean integrated curvature of the grains shows that grain growth slows down and the overall boundary mobility decreases during grain growth [39].

Figure 12. Examples to show the evolution of twin boundaries during grain growth: numbers in each figure indicate deviation angle from ideal 60°〈111〉. [58]

Figure 13. Experimental results from a 3D grain growth experiment in pure iron annealed at 800 °C. Above: 3D grain maps for time-step 1, displaying (a) all grains and (b) only interior grains. Below: one section of the 3D grain map for time steps 1 (c), 8 (d), and 15 (e). The colours represent the grain orientation along the sample RD (see the insert triangle), while black and white lines in (c)-(e) represent boundaries with misorientations above and below 15°, respectively. [39]
3.4. Modelling of local boundary migration during recrystallization

A phase-field model and a numerical model have been developed to simulate local boundary migration into spatially varying deformation energy fields during recrystallization [61–64]. An example of phase-field modelling is shown in figure 14, where the movement of an initially planar boundary into energy fields with 2D variations representing two sets of dislocation boundaries lying at equal, but opposite angles to the boundary are considered. When the boundary migrates from bottom to top of the energy field, parts of the boundary, indicated with the red oval, show a stop-and-go type of movement with pinch-off: boundary parts near minima in the energy density field remain for a long period at the same position, while the rest of the boundary continues to advance. At a certain point, the stagnated parts are pinched off, leaving behind a small part of deformed material (see figure 14b). As the boundary migrates from left to right, asymmetrical protrusions and retrusions develop along the migrating recrystallization front with a considerably higher curvature at the retrusions than at the protrusions, and no boundary segment stagnates (see figure 14c). The average migration velocity is therefore different for the two cases. The results suggest that the geometrical alignment of the dislocation boundaries has a strong impact on the formation of local protrusions and retrusions as well as on the local boundary migration. More work has to be done to quantify the effect of protrusions/retrusions on such non-uniform migration behaviour.

Figure 14. Simulation results obtained for a recrystallization boundary migrating in a deformation energy field with variations. (a) Representation of the considered energy field using a color scaling with black representing the highest energy density and white an energy density equal to 0. (b), (c) Grain boundary positions as a function of time obtained in the simulations, plotted every 2000 time steps starting from time step 60000. (b) Boundary moving from bottom to top. (c) Boundary moving from left to right. [61]

4. Conclusions and outlook

Based on 3D(x, y, time) and 4D(x, y, z, time) data, several methods have been developed for quantifying local boundary migration, with which the local boundary migration in several material systems during both recrystallization and grain growth has been quantified in detail. Based on such analyses, the physical mechanisms underlying the migration process have been quantified for the selected materials. The conclusions are:

i) The misorientations of migrating boundaries cover a wide range for both recrystallization and grain growth in both FCC and BCC metals. Boundaries with misorientation angle < 5° are in all cases rarely mobile, which agrees with the common knowledge that LABs have lower mobilities and with the orientation pinning mechanism [65]. High angle boundaries with misorientation rotation axis close to \{111\} have relatively high mobilities compared to other high angle boundaries for both FCC and BCC metals. The results further show that the misorientation...
distributions for the migrating and non-migrating boundaries are quite similar for recrystallization boundaries. This is partly because that the boundaries are connected in a network: the non-migrating parts can eventually be driven by the connected migrating parts through the development of local retrusions, which provide an additional curvature based driving force.

ii) For the migration of recrystallizing boundaries, the geometrical alignment of the dislocation and lamellar boundaries in the deformed microstructure is another important parameter defining the boundary migration. It affects the overall recrystallization boundary shape, e.g. staircase shaped consisting of curved parts and facets, and the formation of local protrusions and retrusions, which can in turn accelerate the local boundary migration speed. Boundaries migrate differently when they meet the dislocation/lamellar boundaries at different angles, even if the misorientations of the boundaries are very similar.

iii) The local apparent mobility and the activation energy for a recrystallization boundary with misorientation around $50^\circ \langle 221 \rangle$ has been quantified to be $6.2 \pm 0.4 \times 10^{-14} \text{m}^4 \text{s}^{-1} \text{J}^{-1}$ and $280$ kJ/mol, respectively, for pure Al annealed at $250$ °C. The results show also that the boundary segments with misorientation axes close to $\langle 111 \rangle$ have comparatively higher apparent mobilities than those with rotation axes close to $\langle 110 \rangle$.

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