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Mapping of individual dislocations with dark field x-ray microscopy

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We present an x-ray microscopy approach for mapping deeply embedded dislocations in three dimensions using a monochromatic beam with a low divergence. Magnified images are acquired by inserting an x-ray objective lens in the diffracted beam. The strain fields close to the core of dislocations give rise to scattering at angles where weak beam conditions are obtained. We derive analytical expressions for the image contrast. While the use of the objective implies an integration over two directions in reciprocal space, scanning an aperture in the back focal plane of the microscope allows a reciprocal space resolution of $\Delta Q/Q < 5 \cdot 10^{-5}$ in all directions, ultimately enabling high precision mapping of lattice strain and tilt. We demonstrate the approach on three types of samples: a multi-scale study of a large diamond crystal in transmission, magnified section topography on a 140$\mu$m thick SrTiO$_3$ sample and a reflection study of misfit dislocations in a 120 nm thick BiFeO$_3$ film epitaxially grown on a thick substrate. With optimal contrast, the full width of half maximum of the dislocations lines are 200 nm, corresponding to the instrumental resolution of the microscope.

In conventional x-ray topography, a 2D detector or film is placed in the Bragg diffracted beam downstream of the sample (Tanner, 1976). The diffracted intensity is projected onto a two-dimensional image, a ‘topograph’. This technique allows one to visualize long-range strain fields induced by the dislocations. Three-dimensional mapping can be provided in several ways. First results were achieved by preparing ‘stereo pair’ diffraction topographs (Lang, 1959), (Haruta, 1965), which provide two views of the defects, followed by recording a number of closely spaced ‘section’ topographs (Medrano et al., 2006), (Ramar et al., 2010), (Liu et al., 2014). However, TEM is inherently limited to the study of thin foils. For non-destructive mapping of individual dislocations in the bulk X-ray imaging is prevalent.

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mapping dislocations. The former is a magnified version of classical topography. In the latter, an aperture is introduced in the back focal plane to define a certain range in reciprocal space. By scanning the aperture one can visualise the strain field around a dislocation, e.g. with the aim of identifying Burgers vectors.

We describe the optical principles and demonstrate the use of the methods by three examples. The first is a full field transmission study of dislocations within the interior of a 400 μm thick synthetic diamond crystal, the second a magnified section full field reflection study of a deformed SrTiO₃ sample and the third a topography study of a deformed BiFeO₃ thin film.

2. The dark field x-ray microscopy set-up

Dark-field x-ray microscopy (Simons et al., 2015) is conceptually similar to dark-field transmission electron microscopy. The experimental geometry and operational principle are shown in Fig. 1: monochromatic x-rays with wavelength λ illuminate the diffracting object. The sample goniometer comprises a base tilt, χ, and azimuthal angle η, the diffracting object. The sample goniometer comprises a base tilt, χ, and azimuthal angle η. An x-ray objective produces an inverted, an ω rotation stage and two orthogonal tilts, χ and φ. The sample is oriented such that the Bragg condition is fulfilled, as defined by scattering vector \( \vec{Q} \), scattering angle 2θ, and azimuthal angle η. An x-ray objective produces an inverted and magnified image in the detector/image plane. Furthermore, it acts as a band-pass filter in reciprocal space, which is crucial for polycrystalline specimens as spot overlap can be avoided in this way.

The method development has been motivated primarily by studies of polycrystalline samples. However, grains typically have to be aligned and studied one by one. For simplicity in this article we shall assume the sample to be a single crystal. Furthermore, following current practice the objective will be a compound refractive lens, CRL, (Snigirev et al., 1996) with N, identical parabolic shaped lenses with a radius-of-curvature R, and a distance between lenslet centres of T.

3. Methodology

3.1. Weak beam contrast mechanism

In this paper we shall assume that the scattering vector probed is in the proximity of a reciprocal lattice vector, \( \vec{Q}_0 \). We...
will neglect effects due to (partial) coherence and assume that dynamical effects only take place within a sphere in reciprocal space around the lattice point, \( \hat{Q}_0 \), with radius \( r_{\text{dyn}} \). By definition, when probing parts of reciprocal space with \( |\hat{Q} - \hat{Q}_0| > r_{\text{dyn}} \) kinematical scattering applies. We shall use the phrase ‘weak beam contrast’.

We shall not be concerned with the symmetry of the unit cell, and reciprocal space and strain tensors both refer to a simple cubic system. Including crystallography is straightforward in principle, but the more elaborate equations make the treatment less transparent. Moreover, we will consider only the case of a synchrotron beam with an energy band \( \Delta E/E \) of order \( 10^{-4} \) or less. Unless focusing optics are used the incoming beam will have a divergence of \( \Delta \hat{z} \approx 0.1 \text{ mrad} \) or smaller. In comparison the numerical aperture of the objective is much larger, \( NA \approx 1 \text{ mrad} \).

In the following we estimate the width of the intensity profile from a single straight dislocation within this weak beam contrast model. This estimate will be used for a simple comparison from a single straight dislocation within this weak beam contrast. In a classical dislocation model the non-zero strain components in reciprocal space around the lattice point, \( \hat{Q}_0 \), will be spanned by a screw dislocation aligned with \( \hat{e}_{\text{rock}} \) parallel to the coordinate axes and with widths (FWHM) \( \Delta Q_{\text{rock}} \ll \Delta Q_{\text{roll}} \approx \Delta Q_{\text{2D}} \) and the resolution function is in fact an oblate spheroid.

Comparing Eq. 1 to Eqs. 3 and 4, it appears that for experimentally relevant values of \( r \), the intensities on the detector are the result of a 2D projection in reciprocal space: the objective’s NA effectively integrates over directions \( \hat{q}_{\text{2D}} \) and \( \hat{q}_{\text{roll}} \). In addition, the intensities are 1D projections in direct space, along the axis of the diffracted beam.

The resolution in the ‘rocking direction’ is in fact a convolution of the Darwin width of the sample and the divergence of the incoming beam. For simplicity, in Eq. 2 and throughout this manuscript we shall neglect the Darwin width.

Next, let us consider the model system of section 3.1. For \( q = 0 \) we integrate over \( e_{xy} \). The intensity distribution is then a function of only two variables \( I = I(y, e_{zy}) \). We can determine the path length along \( x \) for a given \( y \) and strain interval \( de_{zy} \) by inverting Eq. 1 and differentiating \( dx/de_{zy} \), see Appendix. As a result

\[
\Delta Q_{\text{rock}} = \left| \frac{f_{\text{rock}}}{2} \Delta \hat{z} \right|, \\
\Delta Q_{\text{roll}} = \frac{|\hat{Q}_0|}{2 \sin(\theta)} NA, \\
\Delta Q_{2D} = \frac{|\hat{Q}_0|}{2 \tan(\theta)} NA.
\]

This shows that \( \Delta Q_{\text{rock}} \ll \Delta Q_{\text{roll}} \approx \Delta Q_{2D} \).

\[
I(y, e_{zy}) \propto \int_{-\infty}^{\infty} f(y - y') \left| \int_{u_1}^{u_2} g(e_{zy} - \frac{u}{2}) \frac{du}{\sqrt{u^2 - \frac{g}{2\pi}}} - 1 \right| dy';
\]

with

\[
u_1 = -\frac{B}{2\pi y'}, \quad u_2 = -\frac{By'}{2\pi (y'^2 + (T/2)^2)}.
\]

Here \( f(y) \) is the point spread function and \( g(e_{zy}) \) is the resolution in \( e_{zy} \). In the following we shall assume both to be Gaussian distributions. \( T \) is the thickness of the crystal in the direction of the diffracted beam. \( \| \) symbolises the absolute value.
Simulations of the intensity profile across a screw dislocation are shown in Fig. 2 using parameters relevant to the experiments presented later, including a point spread function \( f(y) \) with a FWHM of 180 nm, a strain resolution function \( g(e_{12}) \) with a FWHM of 0.02 mrad and a sample thickness of 400 μm. With increasing offset in rocking angle the width of the curves asymptotically approaches the spatial resolution, while the peak position in direct space, \( r \), and strain (angular offset) approximately follows \( e = \frac{r}{2\pi\sigma} \).

For applications, a main challenge of any topography method is overlap of signal from dislocation lines. This effectively limits the approach in terms of dislocation density. It appears that in the weak beam contrast description the likelihood of overlap is determined by how far off the peak on the rocking curve one can go while still maintaining a contrast. The profiles shown in Fig. 2 are normalised. If not normalised, the amplitude of the profiles falls off rapidly with offset in rocking angle. Hence, signal-to-noise becomes critical.

Another concern is the nature of the tails of the distributions, \( f(y) \) and \( g(e_{12}) \). If these tails are intense, such as in Lorentzian distributions, the contrast deteriorates. Hence, being able to design and characterise the resolution functions is important. This can be achieved with an aperture in the BFP.

### 3.3. Mapping dislocations using an aperture in the back focal plane

Dark field imaging is one of the basic modalities of a TEM (Williams & Carter, 2009). By inserting an aperture in the back focal plane, one selects a certain region in reciprocal space and uses the diffracted signal within this region as contrast to image the sample. In Poulsen et al. (2018), we introduce the equivalent technique for hard x-ray microscopy. The relation between position \((y_B, z_B)\) in the back focal plane, the angular offset in rocking angle \(\phi - \phi_0\) and reciprocal space is

\[
q_{\text{rock}} = \frac{\Delta Q_{\text{rock}}}{Q_0} = (\phi - \phi_0) - \frac{\cos(N_\varphi)}{2\sin(\theta)} z_B \sin(\theta),
\]

\[
q_{\text{roll}} = \frac{\Delta Q_{\text{roll}}}{Q_0} = \frac{\cos(N_\varphi)}{2\sin(\theta)} y_B,
\]

\[
q_l = \frac{\Delta Q_l}{Q_0} = \frac{\cos(N_\varphi)}{2\sin(\theta)} z_B \cos(\theta),
\]

with \(\varphi = \sqrt{T/T}\) being a measure of the ‘refractive power’ of the lens, and \(f_N\) being the focal length. The last term in Eq. 7 and the \(\cos(\theta)\) factor in Eq. 9 originates in the fact that rocking the sample is a movement in a direction which is at an angle of \(\theta\) with the optical axis (the direction of the diffracted beam).

Unfortunately, if the aperture gap \(D\) is smaller than or comparable to the diffraction limit \(\lambda/\text{NA}\), the spatial resolution in the imaging plane will deteriorate. On the other hand, using wavefront propagation in Poulsen et al. (2018) we demonstrated that the aperture will not influence the spatial resolution if the gap is sufficiently large. For a specific application introduced below the minimum gap is 80 μm. In order to provide a high resolution both in reciprocal space and in direct space, we therefore propose to move a square aperture with a sufficiently large gap in a regular 2D grid within the BFP and to regain reciprocal space resolution by a deconvolution procedure as follows: let the positions of the center of the slit be \((y_B, z_B) = D/M \cdot (m, n)\), with \(m = -M, -M + 1, \ldots, M\) and \(n = -M, -M + 1, \ldots, M\). For fixed rocking angle \(\phi\) and for a given pixel on the detector, let the set of intensities measured in this detector pixel be \(S_{m,n}\).

Now, consider the intensities \(I_{m,n}\) for an aperture of size \(D/M\), in the hypothetical case that the diffraction limit can be neglected. Moreover, assume the diffracting object is bounded such that there is no diffracted intensity outside the grid. Then, in the first quadrant we have: for \(-M < m \leq 0\) and \(-M < n \leq 0\)

\[
I_{m,n} = S_{m,n} - S_{m,n-1} - S_{m-1,n} + S_{m-1,n-1}.
\]  
(10)

For the other quadrants similar expressions can be established. Hence, using this simple difference equation we can generate high resolution \(q\) maps.

In Poulsen et al. (2018) it is also found that the FWHM of the resolution function in the BFP can be \(\Delta Q_{\parallel}/Q_0 = 4 \times 10^{-5}\) or better in all directions, which is substantially smaller than the angular range of the diffracted beam. We conclude that by placing an aperture in the back focal plane we can generate a 5D data set. Hence, we can associate each detector point with a reciprocal space map. Then the only remaining integration is in the thickness direction in real space. We anticipate this enhanced contrast to be useful for identifying Burgers vectors and for improved forward models. In particular this may enable studies of samples with higher dislocation densities as one can separate dislocations that are overlapping in the greyscale images.

A significant simplification arises if we use the formalism of elasticity theory. Then each point \((x_s, y_s, z_s)\) in the sample...
is associated with one point in reciprocal space corresponding

to the three strain components: \((e_{xx}, e_{yy}, e_{zz})\). Let the recorded

intensities be \(I(\tilde{q}, y_d, z_d)\) with \((y_d, z_d)\) being the detector coor-
dinates, \(\tilde{q} = (q_{\text{rock}}, q_{\text{roll}}, q_{\text{tilt}})\) and strain vector \(\tilde{e} = (e_{xx}, e_{yy}, e_{zz})\).

Then for \(\omega = 0\) we have

\[
I(\tilde{q}, y_d, z_d) \propto \int \int dx, du \ dv \ f(y_d - u, z_d - v) \ \int d^3 \tilde{q}' \ g(\tilde{e}(x, u; M, v; \tilde{M}) - \tilde{q}').
\]  

Here \(M\) is the magnification in the x-ray lens, \(f\) is the detector
point-spread-function and \(g\) is the reciprocal space resolution
function. With the square aperture in the BFP, the function \(g\) is \(\delta_{200}\)
box function in two directions.

With respect to implementation, it may also be possible to
transfer additional TEM modalities. In particular, annular dark
field imaging is a candidate for fast 3D mapping of dislocations.

Blocking the central beam by an elegant way to remove spu-
raneous effects due to dynamical diffraction.

4. Experimental demonstrations

To illustrate the potential and challenges of our approach, we report
on the results from three different type of use. Three samples were
studied at beamline ID06 at the ESRF over two beamtimes and under slightly different configurations (as the
beamline instrumentation evolved during this period).

In all cases, a Si (111) double monochromator was used to
generate a beam with an energy bandwidth of \(\sigma_\gamma = 0.6 \cdot 10^{-4}\n\) (rms). The goniometer with all relevant degrees of freedom, cf.
Fig 1, is placed 58 m from the source. Pre-condensing is per-
formed with a transfocator (Vaughan et al., 2011) positioned at
a distance of 38.7 m from the source. For section topography, a
1D condenser was used to define a horizontal line beam. Other-
wise, a slit defined the dimensions of the beam impinging on
the sample. Two detectors were in use, firstly a nearfield cam-
era, placed close to the sample, which may provide classical
topographs and topo-tomograms without the magnification by
the x-ray objective. Secondly, a farfield camera placed at a dis-
tance of \(\approx 5.9\) m for imaging the magnified beam in the image
plane of the microscope. Both detectors were FRELON 2k \(x 300\)
2k CCD cameras, which are coupled by microscope optics to
a LAG scintillator screen. The objective comprised \(N\) identi-
cal parabolically shaped Be lenses with a radius of curvature
\(R = 50 \mu\) m and thickness \(T\). A square slit with adjustable gaps
and offsets was placed in the BFP. The surface normals of all
detectors and slits were aligned to be parallel to the optical axis.
The nearfield camera and the aperture in the BFP could be trans-
lated in and out of the diffracted beam.

4.1. Transmission experiment

The sample was an artificially grown diamond plate, type
IIa, with a thickness of 400\(\mu\)m, see Burns et al. (2009). It was
mounted in a transmission Laue geometry. The 17 kV inci-
dent beam had a divergence (FWHM) of 0.04 mrad, and dimen-
sions of 0.3 mm \(\times 0.3\) mm. With \(N = 72\) and \(T = 2\) mm, the
focal length of the objective was \(f_0 = 0.245\) m. The effective
pixel size of the near and far-field detector was 0.62 \(\mu\)m and 1.4 \(\mu\)m, respectively. The magnification by the x-ray objective
was measured to be \(M = 16.2\), implying a numerical aperture of \(NA = 0.643\) mrad and an effective pixel size of
93 nm. The detector was then binned 2 \(\times\) 2. Using Eqs. 2 –
4 the FWHMs of the reciprocal space resolution function in
the three principal directions become \((\Delta q_{\text{rock}}, \Delta q_{\text{roll}}, \Delta q_{\text{tilt}}) =
(0.00062\AA^{-1}, 0.0055\AA^{-1}, 0.0055\AA^{-1})\).

An in-plane \{111\} reflection was used for the study. The
length of the diffraction vector and Burgers vector are \(|\mathbf{Q}_0| =
3.051\AA^{-1}\) and \(|\mathbf{B}| = 2.522\AA\), respectively. Using the formal-
ism of Als-Nielsen & McMorrow (2011), the corresponding Pendellösung length, and Darwin width are \(\Lambda_\gamma = 35 \mu\)m and
\(w_\gamma^0 = 0.0119\) mrad (FWHM), respectively. Hence, the in-
coming beam divergence dominates the Darwin width. The data set
involved 36 \(\omega\) projections over a range of 360 degrees. For each
projection images were acquired in a 31 \(\times\) 31 grid in rocking angle \(\mu\) (with steps of 0.0016 deg) and \(2\theta\) (steps of 0.0032 deg).
Exposure times were 1 second.

![projection images](image)

**Figure 3**
Projection images of a large single crystal diamond in the transmission exper-
iment. Nearfield detector image with no x-ray objective and corresponding
dark field image acquired with the diffraction microscope, both for \(\mu - \mu_0 =
0.002\) deg. The magnification of the microscope is \(M = 16.2\). The direction of
the rotation axis is marked by an arrow.
θ 'rocking' and 'longitudinal' contrast are validated. As expected, the signal is corrupted by dynamical diffraction effects until the radial direction (obtained by a simultaneous transverse strain of δµ = ±0.002, as marked by the 5 pixel thick black lines. The lineplots are normalized to max intensity. The red lines indicate the interpolated position of the dislocation line. The latter is inverted for ease of comparison. The difference in the field-of-view, FOV, is evident, as is the fact that the objective magnifies the image without visible distortions. The dominant cause of discrepancy is instead considered to be alignment of the microscope, that was problematic at the time due to the ad hoc character of the set-up.

4.2. Magnified section topography experiment

Within the weak beam regime one may reduce the likelihood of overlap of dislocations in the images by narrowing the incident beam in the vertical direction (see Fig. 2). By introducing a condenser we can furthermore improve the S/N ratio, at the expense of an increased divergence. In principle, one can adjust the height of the incoming beam to match the spatial resolution. 3D mapping can then be performed layer-by-layer. However, identifying points is more difficult than identifying lines, and 1D condensers providing a micrometer-sized beam tend to be more efficient than those producing a nanometer-sized beam. Hence, it may be optimal to operate with an incoming box beam having a large aspect ratio. We shall use the term ‘magnified section topography’ for this setting.

In this experiment, the sample was a wedged shaped piece of SrTiO₃, where surfaces had been polished mechanically. It was mounted in a transmission Laue geometry, using an in-plane aperture of the objective.
The {110} reflection for the study. The 15.6 keV beam was condensed by a CRL with 55 1D Be lenslets to generate a beam (FWHM) of size 4.2 × 300 μm². The objective configuration was in this case N = 45, T = 1.6 mm, leading to a focal length of fN = 0.406 m. The measured x-ray magnification was 12.32 and consequently the numerical aperture had an rms width of σθ = 0.24 mrad. The far-field detector had an effective pixel size of 122 nm. A rocking scan was made over a range of 0.5 deg, with 70 steps and exposure times of 1 second.

Fig. 6 shows a raw image. The top point of the wedge is far to the left of this image. Generally speaking the weak beam scattering signal is confined to two regions, adjacent to the two external boundaries (top and bottom in the figure). We speculate that these have formed during polishing. As shown in the figure, at a certain distance to the top of the wedge, point dislocations are created that bridge the gap between the two surface layers. The intensity profile across one of these vertical lines is shown in Fig. 7. It exhibits a FWHM of 210 nm. In Fig. 6 in the vicinity of the prominent vertical dislocations a network of other dislocations pointing in near random directions are seen. Their linewidths are in some cases below 200 nm, but the statistics is poor. 200 nm is comparable to the spatial resolution of the instrument.

The sample was a 120 nm thick film of ⟨001⟩-oriented BiFeO₃, grown via pulsed laser deposition on a SrRuO₃ electrode layer and ⟨110⟩-oriented DyScO₃ single crystalline substrate. This was mounted for a reflection study on the (002) reflection — at 2θ = 22.6 deg. In this case the 15.6 keV beam from the transfocator was only moderated by a slit close to the sample. The objective and detector configuration were identical to those of section 4.2. The aperture in the BFP had a square opening of 80 μm. Within the approach of section 3.3 this aperture was translated in a 2D grid with a step size of 30 μm. At each position a rocking scan was made with a step size of 0.001 deg and with exposure times of 2 seconds.

Deconvoluting the signal according to Eq. 10 each point in the sample plane was associated with a reciprocal space map. The voxel size of this map is \( \Delta Q|Q| = (1.7 \cdot 10^{-5}, 1.6 \cdot \)
505 Zooming in on one dislocation, we illustrate in Fig. 8 the richness of the results obtained. To the left is shown the result with no aperture in the BFP for two offsets in rocking angle. The remainder of the subplots are corresponding results based on the aperture horizontally. Using Eq. 8 this is converted into a relative shift $q_{\text{roll}}$. The fitted center position and width (FWHM) are shown in column 2 and 3, respectively. In columns 4 and 5 are shown the result of an analogous fit to the intensity profile arising from scanning the aperture vertically. Using Eq. 9 this is converted into a relative shift $q_{\text{roll}}$. All shifts in turn can be directly related to strain components $e_{yx}$ and $e_{zx}$, while the rocking profile gives access to $e_{zx}$.

The rocking profiles (not shown) exhibits a clear asymmetry, analogue to that shown in Fig. 4. The second column of Fig. 8 reveals that the rolling profiles have a similar left-right asymmetry. Near the dislocation core the profile has a dip in the center, evident as a large increase in the FWHM of the one-peak fit (third column). In contrast there is no noticeable variation in the longitudinal direction (columns 4 and 5). These findings are consistent with the response from the strain field from a single dislocation with the Burgers vector pointing in the direction of the surface normal, as anticipated for misfit dislocations.

![Figure 8](image_url)

**Figure 8** Images of a dislocation in a BiFeO$_3$ film acquired at an offset in rocking angle from the main peak of $\phi = 0.01$ deg (row above) and $\phi = 0.015$ deg (row below). The contrast is set differently in the two rows. First column: no aperture in the back focal plane; red is maximum intensity, blue is background. Other four columns: results from scanning an aperture of fixed size in the back focal plane. For each pixel on the detector, Gaussian type fits were made to the profile in the rolling and longitudinal directions, respectively. Shown are the center-of-mass positions and the FWHM in units of $\lambda Q/|Q|$, as determined by Eqs. 8 and 9. The unit on the axes is $\mu$m and refers to the detector plane.

5. Discussion

Dark field microscopy is fundamentally different from classical x-ray topography, as rays emerging in various directions from one point in the sample plane are focused onto a spot in the image plane, rather than leading to a divergent diffraction beam. This implies that the detector can be placed many meters away and that the space around the sample is limited by the objective, not the detector. Moreover, the high spatial resolution allows to visualise the core of the strain field. This simultaneously enables the dislocations to appear as thin lines and scattering to be sufficiently offset from the Bragg peak that weak beam conditions apply. Below we first present the perceived main limitations of the technique and discuss options to overcome these. Next we briefly outline the scientific perspective.

Dynamical diffraction effects. The ‘weak beam’ condition presented strongly simplifies the data analysis and interpretation. In practice, it is likely that dynamical or coherent effects needs to be considered in some cases. A treatment of dynamical scattering in the context of x-ray topography can be found in e.g. Gronkowski & Harasimowicz (1989) and Gronkowski (1991). However, as mentioned previously, the geometry of data acquisition is fundamentally differently in a microscope. A dynamical treatment of the scattering of a dislocation line in the context of a microscope exists for TEM (Hirsch et al., 1960), but has to the knowledge of the authors yet to be generalized to x-ray microscopy. In a heuristic manner with dark field microscopy we attempt to overcome the issue with dynamical effects in two ways:

- By improving both the spatial and angular resolution it becomes possible to probe parts of reciprocal space which are further from $Q_{\text{dyn}}$.
- By combining projection data from a number of viewing angles we anticipate that ‘dynamical effects can be integrated out’. Similar strategies have led the electron microscopy community to apply annular dark-field imaging for providing accurate crystallographic data.

Spatial resolution. The spatial resolution sets an upper limit on the density of dislocations that can be resolved. With increasing spatial resolution, one can monitor the strain and orientation fields closer to the core. At the same time, dynamical diffraction effects becomes smaller as one is probing parts of reciprocal space that are further away from the Bragg peak. In practice, the limitation of the technique is currently set by aberrations caused by the lens manufacture and by signal-to-noise considerations. With the possibility of providing a reciprocal space map for each voxel in the sample, cf. section 3.3, overlap of the diffraction signals from dislocation lines can be handled.

To our understanding there is no fundamental physics prohibiting a substantial increase in the spatial resolution of dark field microscope. With ideal CRL optics hard x-ray beams may be focused to spot sizes below 10 nm (Schroer & Lengeler, 2005). Using zone plates as objectives, at x-ray energies below 15 keV, bright field microscopes are in operation with resolutions at 20 nm. For work at higher x-ray energies, there has recently been much progress with multilayer Laue lenses, which seem to promise imaging with superior numerical apertures and much reduced aberrations (Morgan et al., 2015). Finally, the next generation of synchrotron sources will be 10 – 100 times more brilliant than the current sources (Eriksson et al., 2014).
This will benefit both spatial resolution (via improved signal-to-noise) and time resolution.

**Probing only one diffraction vector.** As for any other diffraction technique, the contrast in visualizing the dislocations is proportional to \( \mathbf{q} \cdot \mathbf{B} \). Dislocations with a Burgers vector nearly perpendicular to the \( \omega \) rotation axis are therefore invisible. In order to map all dislocations and/or to determine all components of the strain tensor one has to combine 3D maps acquired on several reflections.

**Scientific outlook.** The higher resolution in 3D offers new perspectives on dislocation geometry, including measurements of distances and dislocation curvatures (and the balance of line tension by local stresses). This may be relevant for models of dislocation dynamics, and the visualisation of dislocations under e.g. indentations. With respect to dynamical diffraction effects, we remind that extinction lengths for 30 keV x-rays are about 100 times larger than the corresponding extinction lengths for 200 keV electrons. This points to high resolution studies of dislocation dynamics in foils at least 10 \( \mu \text{m} \) thick.

Studies of dislocation structures within grains or domains are facilitated by the fact that dark field microscopy is easy to integrate with coarse scale grain mapping techniques such as 3D X-ray Diffraction Topography, DCT (King et al., 2008) (Ludwig et al., 2009).

6. Conclusion

We have demonstrated an x-ray microscopy approach to characterizing individual dislocations in bulk specimens. The method combines high penetration power, a data acquisition time for 3D maps of minutes, and the possibility to study local internal regions by magnifying the images. The spatial resolution is in this proof-of-concept work 200 nm. The limitation is the quality of the focusing optics and the signal-to-noise ratio. With improved x-ray sources and optics this opens the door to studies of dislocation dynamics, and the visualisation of dislocations under e.g. indentations. With respect to dynamical diffraction effects, we remind that extinction lengths for 30 keV x-rays are about 100 times larger than the corresponding extinction lengths for 200 keV electrons. This points to high resolution studies of dislocation dynamics in foils at least 10 \( \mu \text{m} \) thick.

The method can be extended to mapping of the \( e_{1x} \), \( e_{2y} \) and \( e_{3z} \) fields by scanning a fixed gap aperture in the back focal plane of the objective and by rocking the sample.

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