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Abstract

Within the past decade Electron Backscatter Diffraction (EBSD) has experienced wide acceptance in metallurgical and geological labs. It is a Kikuchi diffraction device based on the SEM that enables individual grain orientations, local texture, point-to-point orientation correlations, and phases to be determined routinely at a sub-grain size level on the surfaces of bulk polycrystals while exploiting the excellent imaging capabilities of the SEM. After a brief retrospective some aspects of modern automated EBSD systems are discussed with particular emphasis on pattern processing, high speed and recent developments, but also on difficulties a newcomer may experience. Pole figure measurement by X-ray diffraction (XRD) is still the standard technique for studying crystallographic texture on a global statistical scale. Finally EBSD is compared with XRD pole figure measurement.

Keywords. EBSD, Kikuchi diffraction, Radon transform, BSE imaging, XRD.

Introduction

The great majority of natural as well as technologically used solids are not amorphous, but have a polycrystalline structure that is characterized by size, shape and arrangement of the constituting crystals and phases. The distribution of preferred grain orientations, or in short (crystallographic) texture [1], is a particularly important parameter of microstructure as being the link between a wide range of anisotropic properties of single crystals and those of polycrystalline material. If a material property depends on crystallographic orientation, an overall anisotropic property results also for the polycrystal, unless grain boundary effects are negligibly small and the grains are oriented at random. Texture can be beneficial, improving performance for instance in deep drawing, stiffness, and magnetization of sheet metal, or it can be detrimental, leading to brittleness, fracture, hysteresis losses and lower corrosion resistance. The main objectives of texture research are a better understanding and prediction of anisotropic materials properties (related to the bulk, grain or grain boundary) and texture-related processes (e.g. recrystallization, grain growth, and plastic deformation), the correlation of residual stress and texture as well as phase analysis and texture. Texture is also a fingerprint of the history of a work piece and gives valuable insight in rock formation in early days of the earth. The knowledge of texture is a prerequisite for all quantitative techniques of materials characterization, which are based on the interpretation of diffraction peak intensities such as the determination of the content of residual austenite in steel, the determination of residual stress by X-ray diffraction, and structure analysis from powder diffraction patterns. Many techniques have therefore been developed to measure crystallographic texture on a grain-specific as well as on a global scale (see review [2]). The most appropriate techniques to study these microstructural characteristics are electron and X-ray diffraction methods.

The long way from discovery of BKP to EBSD

The first Backscatter Kikuchi Patterns (BKP) have been reported already in 1928 [3], but more than 70 years passed by until Electron Backscatter Diffraction (EBSD), or more precisely Backscatter Kikuchi Diffraction (BKD) in the SEM, has emerged to be one of the most powerful techniques for orientation microscopy and crystal texture analysis in materials science and geology. Five years after the discovery wide-angle patterns from
cleavage faces have been reported at beam voltages from 10 kV to 40 kV [4]. The experiments showed that the bands are formed by backscattered electrons of virtually the same energy as the primary beam energy. The bandwidth is proportional to $\lambda/d_{hkl}$. Intensity profiles across two typical bands are sketched. The centerlines of the bands – now called Kikuchi bands – have been indexed as zone circles sectioning the cylindrical recording film.

A first review article in English language appeared in 1937 [5] with a collection of excellent backscatter Kikuchi patterns from a variety of single crystal cleavage faces. A comprehensive elaboration of the kinematic as well as dynamic theory of particle wave diffraction including application to Kikuchi diffraction in transmission and backscattering appeared in 1944 [6]. In 1954 the early experimental findings were confirmed [7]. In 1973 BKP were recorded with a video camera on an SEM [8]. Strangely enough the authors coined the new name Electron Backscatter Patterns (EBSP) for the long-time established term BKP.

The first commercial BKD systems for the SEM were marketed by Link-Oxford Ltd in 1986 with software developed by D. Dingley [9], and the second one in 1990 by HKL Technology (N.-H. Schmidt). The next steps ahead were automated indexing and scanning the sample with a mechanical stage in 1993 [10] and digital beam scan in 1995 [11]. The SEM of an advanced EBSD system with digital beam scan has to be operated under full control of the EBSD software [12] to enable dynamic focusing, automatic and dynamic system calibration, and usage of all working distances and magnifications of the SEM for analysis. The combination of stage with beam scans enables large areas to be studied with high accuracy and speed. At the end of last century EBSD has found rapid acceptance in materials science thanks to the wide availability of SEM, the ease of sample preparation from the bulk and the access to complementary information about the microstructure on a submicron scale. The combination of several easy-to-use techniques makes the SEM a powerful and almost universal analytical instrument in the hands of the physical metallurgist as well as the earth scientist. The reader is referred to the monographs [13, 14], a recent review [15] and the references therein.

Present state and future trends of EBSD

Backscatter Kikuchi patterns are generated on a phosphor screen by back-diffraction of a stationary beam of high-energy electrons from an almost perfect volume of crystal. The maximum of back-diffracted intensity is close to optical reflection, as a consequence of the strong forward scattering of high-energy electrons. In the SEM, the bulk specimen has to be steeply tilted to typically 20° to 30° from grazing incidence (Figure 1), in order to obtain patterns of sufficient intensity. For acquisition, a transmission phosphor screen is placed parallel to the incident beam, right in front of the tilted specimen, such that the intensity maximum falls almost on the center of the screen. A pattern spans over a wide angular range up to ±60°, and contains several principal zone axes. A high-sensitivity camera is used to record the patterns.

![Figure 1. Schematics of an EBSD system in the SEM.](image-url)

The geometry of a Kikuchi pattern can be interpreted as a gnomonic projection of the crystal lattice on the flat phosphor screen: The point of impingement of the primary beam on the specimen surface is the center of projection. The edges of the Kikuchi bands are hyperbola formed by the intersection of Kossel cones with the plane of the detector. The lattice planes can thus be imagined to be stretched out to intersect the screen in straight
centerlines of their related bands. Hence the angles between the centerlines correspond to the interplanar angles. The angular width of a Kikuchi band \{hk\} is twice the Bragg angle, \(\phi_{hk}\), according to Bragg’s law

\[
2d_{hk}\sin \phi_{hk} = n\lambda \tag{1}
\]

d_{hk} is the interplanar spacing, \(\lambda = 1.54\cdot U^{-1/2}\) kV·nm stands for the electron wavelength, and \(n\) is the order of reflection. Because of the short wavelength at usual accelerating voltages, \(U\), in the SEM the slightly curved band edges can be well approximated by straight lines. It is worth mentioning here that the angular distance between crystallographic zones does not depend on the acceleration voltage. The extinction rules apply for expected reflections (i.e. Kikuchi bands) according to the structure factor of the crystal lattice, and high order reflections may appear as a set of straight lines parallel with the band edges. So the geometry of a Kikuchi pattern (Figure 2) is unique for a particular crystal structure and crystal lattice orientation. Hence, it is sufficient for indexing a pattern to know the positions and widths of some bands in the pattern.

![Figure 2. Backscatter Kikuchi pattern of a Cadmium specimen recorded at 20 keV accelerating voltage.](image)

Band localization is performed by applying a Radon transform on the Kikuchi patterns and analyzing peak profiles in Radon space. The general Radon transformation is a functional to transform a manifold in an \(n\)-dimensional space [16]. With the normal parametrization of a straight line as the functional kernel, a line in a plane is thus transformed into a single point \(R(\rho, \phi)\) in 2-dimensional Radon space by

\[
R(\rho, \phi) = \int \int_{x,y} f(x, y) \cdot \delta(\rho - x \cdot \cos \phi - y \cdot \sin \phi) \ dx \ dy \tag{2}
\]

\(f(x,y)\) stands for the intensity distribution along the line, \(\delta\) denotes the Dirac function, \(\rho\) is the perpendicular distance of the line from the origin, \(\phi\) is the angle between the \(x\) axis and the normal from the origin to the line. Every point \(R(\rho, \phi)\) in Radon space thus corresponds to the sum of intensity values along a certain straight line in image space. Reconstruction of the pattern by applying the inverse Radon transform

\[
f(x, y) = \int_{0}^{2\pi} R(x \cdot \cos \phi + y \cdot \sin \phi, \phi) \ d\phi \tag{3}
\]

literally means that a peak in Radon space is projected back into image space along its incoming path. The intensity \(I(\rho, \phi)\) of every single point \(R(\rho, \phi)\) in Radon space is attributed equally to all pixels \((x_l, y_l)\) on the related straight line \(\rho = x \cdot \cos \phi + y \cdot \sin \phi\) in the reconstructed Kikuchi pattern.

The EBSD image (i.e. the Kikuchi pattern) is made up by a discrete array \([x, y]\) of gray-tone image points (so-called pixels). So we consider a discrete Radon transform whereby the integrals in Eq. 2 and 3 are replaced by sums, and the Radon domain consists of an array of discrete cells on a Cartesian grid \((\rho, \phi)\). A Kikuchi band can be imagined to be composed of a bundle of straight lines that may be slightly twisted and shifted against each other. By a “forward” Radon transformation according to Eq. 2, the intensity profile of a Kikuchi band is mapped, pixel by pixel, to a ”butterfly-like” shape in Radon space with a bright center and dark cusps on the sides in \(\rho\) direction, rather than a single point as were the case of a single line.

A “backward” Radon transformation of the pattern according to Eq. 3 has some advantages for band localization. The Radon space is scanned point by point, and for each point \(R(\rho, \phi)\) the intensity along the corresponding straight line in the pattern is accumulated. However, the intensity profile can, as
contrasted to a “forward” Radon transformation, be in addition interrogated along this line. So short fragments and sections of the line in the background outside a band can be excluded from transformation by applying a run length threshold, as well as artifacts of “hot” spots in the pattern can be skipped by defining intensity thresholds. The result is a cleaner domain in Radon space. Noise can further be removed and contrast enhanced effectively in the Radon transformed pattern by applying a square intensity operator [17]. The task of locating lines or bands in the diffraction pattern is so reduced to the simpler task of searching for isolated peaks. The “butterfly peak” is localized by a peak search with constraints or by evaluating some coefficients of a 1D FFT of this Radon domain [18].

The Hough transform [19] was established for the detection of straight lines or contours in binary images in a complex hardware implementation almost half a century after the Radon transform had been published. Indeed the Hough transform is a simple “forward” variant of the more general Radon transform. It is a fast technique to localize sparse (line) contours since zero pixels can be skipped from transformation. A detailed introduction to theory and implementations of Radon and Hough transforms can be found in [20, 21]. For band localization a Hough transform has been “modified” to accommodate for gray tone patterns. The peaks in Hough space – which again correspond to bands in the pattern – are located by applying a normalized cross-correlation process with a large butterfly-shaped filter mask [22]. The main limitations of this approach are caused by the dependence of the band profiles and hence the size and shape of a suitable butterfly mask on the position of the particular band in the Kikuchi pattern, the type of pattern, the diffraction geometry and accelerating voltage, the type of crystal lattice and \( \{h \, k \, l\} \). The wider the bands are, the larger the masks should be. Large masks unduly increase computation time. In conclusion, cross-correlation with a filter mask tends to be ineffectual. One mask is in general not sufficient, but several masks are required to analyze one pattern. Determination of band positions and band widths is less accurate when using (inadequate) mask(s.)

When several bands in a pattern have been localized, indexing process is reduced to a comparison of pattern geometry with positions of lattice planes in the unit cell. Calculation is performed in computer programs that use look-up tables of the diffracting lattice families (“reflector tables”, their interplanar angles and diffracted intensities) for the lattice system under consideration. If several crystal lattices are possible in a multi-phase material the indexing routine checks for all the assumed lattice types and takes that one whose re-computed (theoretical) pattern fits best to the measured one (“Phase Discrimination”). From an indexed pattern, the orientation of the crystallite is obtained by elementary calculation of the Miller indices respectively the Euler angles of the reference directions (sample coordinate system, microscope coordinate system) related to the unit cell of the grain. The obtained orientation data, usually described by Euler angles, are analyzed with separate programs depending on the demands of the user (mappings, filters, GB, \( \Sigma \) GB, orientation stereology, quantitative texture analysis, ODF). The distribution of grain orientations is visualized in a quantitative way by assigning specific colors to the Miller indices, Euler angles or Rodrigues vector of the orientation in every pixel of the maps [23].

Electron backscattering diffraction (EBSD) has become an almost mature standard technique for orientation microscopy and texture analysis on bulk polycrystals at a grain specific level [14]. Conventional EBSD systems are based on pattern acquisition and simultaneous on-line evaluation. Results depend on an accurate system calibration. All present phases must be known a priori. These systems discard the patterns once they have been indexed, making it impossible to repeat pattern solving with the same raw data set if the results, for some reason, turn out to be unsatisfactory.

A low “Hit Rate” (= fraction of correctly indexed data points), “orientation speckle” in orientation maps, and wrong peaks in the orientation density function (ODF) may have various reasons:

1. Noisy patterns are obtained if the beam current is too low and integration time too short or if the settings of the camera are incorrect.
2. Patterns are diffuse and fuzzy if sample preparation is insufficient, a foreign layer or contamination is covering the surface, if the specimen is highly deformed or residual stress is present, e.g. due to martensitic transformation, or simply because the beam spot is out of focus.
3. Overlaid patterns at grain boundaries or from
grains that are smaller than the beam spot respectively smaller than the inherent resolution of backscatter Kikuchi diffraction may result in misindexing.

4. Unexpected phases are not taken into account for indexing. The lattice constants of the prospective phases have to be known a priori for indexing.

5. System calibration is not correct or dynamic calibration has been disabled.

6. Background correction is insufficient.

7. Parameters of image processing or Hough parameters are not appropriately set so that band localization and indexing fail.

The user may feel some temptation to “beautify” the orientation maps by filtering the orientation data, or even worse, the software without notice performs filtering. Filtering is allowed only for removing single isolated pixels or for analyzing texture, for instance as a function of grain size, degree of deformation in combination with pattern quality or phase. A better strategy to improve unsatisfactory results is to repeat measurement, or if patterns are clear, to repeat pattern solving in cases 4 - 7 off-line with the original raw pattern sequence. Re-indexing of stored Hough peaks is fast, for instance in cases when not all phases were known initially [24]. However, image processing and Hough transformation may already have produced artifacts so that a reliable re-analysis is not always guaranteed.

It is therefore consequent to separate pattern acquisition from band localization and pattern solving. A high acquisition speed is achieved by streaming the sequence of raw unprocessed patterns, after binning on the sensor chip, on the hard disk [25]. Pattern solving is performed automatically off-line at any time whereby the parameters of the indexing program can be optimized conveniently so that misindexing is minimum. Re-examination of the results at any time by using the original raw data is mandatory for process and quality control. No compromise is made between speed and reliability [26]. In the following this technique is named “FastEBSD”. Speed of pattern acquisition and streaming is presently >1000 patterns per second on suitable specimens (Al, Ni; 20 kV, <= 20 nA), off-line indexing of pattern sequences >2000 orientations per second on single phase materials, and off-line re-indexing of stored Radon peaks about 10000 orientations per second, depending on the performance of a quad-core PC.

Off-line indexing is based on the same philosophy as EDS spectral imaging where complete X-ray spectra are acquired at scanned points of 2D arrays and evaluated off-line. FastEBSD is an economic strategy since the throughput of the SEM is increased considerably. Speed in the SEM session is independent of the crystal structure and number of phases contained in the sample.

Because of the high speed large sample areas with a dense setting of grid points are recommended as standard strategy. Regions of particular interest can then be selected later from the orientation maps. Large areas of similar size as in XRD can be examined with digital beam scan provided that the beam is dynamically focused when scanning across the steeply tilted specimen surface, dynamic calibration is enabled and sample preparation was successful. Earlier objections against EBSD concerning a non-representative low number of measured grain orientations are nowadays irrelevant.

FastEBSD warrants sample contamination to be kept at a minimum. Adverse drift effects are reduced, either caused by charging or during in situ tensile tests or heating experiments. It is self-evident that the EBSD scan has to be faster than the changes induced in the microstructure as a result of an in situ experiment. Cold field emitters with typically low long-term stability are accommodated. FastEBSD is in particular favorable for 3D orientation microscopy. It enables rapid acquisition of backscatter Kikuchi patterns, and these are evaluated off-line while the next slice is removed by FIB serial sectioning. Because speed of on-line analysis with conventional EBSD systems is relatively slow, EBSD data collection adds significant overhead to the time required to lay open and analyze a section slice [27, 28].

EBSD is less accurate as compared to XRD. The intrinsic reasons are a lower dispersion Δλ/λ, Δθ/θ of electron diffraction which makes determination of lattice constants via the Bragg angle inaccurate. Backscatter Kikuchi patterns furthermore suffer from a high uneven background, noise and diffuse band contours. The calibration parameters of the system setup, in particular the pattern center and the sample-to-screen distance, can be determined from pattern geometry with some errors only. Conventional EBSD programs are optimized more
for speed rather than for phase discrimination. Phases can be discriminated if their lattice constants differ by roughly some percent. Since one lattice constant (a) is skipped in the program as a scaling factor, different phases of the same cubic Laue group cannot be distinguished from each other at all. Fortunately the phases present in a sample are known beforehand in most cases, but for some multiphase samples additional knowledge is required to achieve a reliable EBSD analysis. Sources of additional information can be for example the morphology of the material (grain size, grain shape), surface relief after phase-specific etch, integral electron backscattering coefficient (material contrast in the BSE image), Pattern Quality (i.e. the crispness of the BKP) in particular in case of a martensitic transformation or partial recrystallization, and element concentrations. Since a thorough evaluation is time-consuming and usually needs interaction by the operator, FastEBSD with pattern streaming followed by off-line interpretation is the procedure of choice.

An obvious approach in an analytical SEM is the combination of EBSD and EDS measurement [28-30]. Acquisition of EDS spectra, however, is considerably slower than acquisition of backscatter Kikuchi patterns. Spatial resolutions of EDS in the plane and in-depth beneath the sample surface are, with typically 0.5 µm, more than one order of magnitude less than spatial resolutions of EBSD, and errors of EDS analysis are typically at about 5 weight-% at medium concentrations. In addition phases are defined by the crystal lattice rather than the element concentrations since a material need not be in thermodynamic equilibrium. Simultaneous EDS and EBSD suffer from an unfavorable compromise in acquisition geometries. So EDS analysis can give nothing more than an indication of possible phases to be checked for in the indexing procedure at a sampled spot of EBSD acquisition. In conclusion, phase differentiation and phase identification can be quite complex tasks that lend themselves to separate off-line acquisition and analysis of EDS spectra and backscatter Kikuchi patterns.

A Kikuchi pattern represents the directional distribution of backscatter electrons (BSE) that are projected on the two-dimensional phosphor screen. The intensity of BSE, emitted in a solid angle element, can so be measured by integrating the intensity that falls in the selected segment on the phosphor screen. If this is done for all patterns of a sequence a BSE map of the microstructure can be constructed from scan point to scan point without having to index the patterns [31]. Depending on the selected element of solid angle of BSE topographical contrast maps or/and material contrast maps are obtained. They reflect the relief state and composition of the sampled surface. Topographical contrast can be used for checking the quality of sample preparation. Material Z contrast helps to discriminate between phases which are hard to distinguish by indexing the patterns in EBSD. As contrasted to EDS, spatial resolution in BSE images is the same in magnitude as resolution of EBSD analysis.

It is common to mount BSE and FSE solid state detectors close to the margin of the EBSD screen (Figure 3). The EBSD detector in our FastEBSD system [25] functions in the same way with the advantage that the free space in the specimen chamber is not further constrained. No additional time is spent for acquisition on the SEM. The selected solid acceptance angles are not fixed, but can be optimized by freely setting acquisition boxes on the patterns (Figure 4). The signal intensities from different boxes in each pixel can be mixed and weighted by arithmetic functions (add, subtract, multiply, divide). Gray values can be transformed to pseudo-colors (Figure 5).

Figure 3. Conventional solid state backscatter and forward-scatter electron detectors mounted on an EBSD device in the SEM.
Figure 4. An example of four acquisition boxes on a Cadmium backscatter Kikuchi pattern. Z marks a box for material contrast, T and T’ are boxes to construct topographic contrast maps and O indicates a box for acquisition of an orientation map. Grains with bands or zone axes close to this window appear bright.

Figure 5. Topographic contrast (a.) and material Z contrast (b.) maps of a silver solder sample after excessive ion etch by FIB milling.

EBSD measurement can also provide an indication of local elastic and plastic lattice strain. The reader is referred to a recent review [32] and the references therein. Patterns of excellent quality are obtained from diffracting volumes of perfect crystal only, but Pattern Quality (PQ), i.e. the crispness of the patterns (sometimes also named Image Quality, IQ), degrades with increasing dislocation density and subgrains because the 3-D lattice is locally distorted and rotated so that a range of near Bragg conditions is operative. A range of similar Kikuchi patterns is generated, overlaid and the resulting pattern is blurred.
Pattern Quality can be determined from the sharpness of Kikuchi lines bordering the bands in the pattern as well as from the height and slope of the Radon peaks. An elegant method of PQ measurement is the determination of first coefficients of a 1-D Fast Fourier Transform in $\rho$-direction of Radon peaks [18]. PQ, however, is nothing more than a semi-quantitative measure of plastic lattice strain and dislocation density since its magnitude depends on many other factors, for instance the size of probed crystal volume respectively beam size, accelerating voltage, contrast settings of the camera, and very sensitively on the sample surface condition (foreign contamination or reaction layer, unwanted deformation from preparation). PQ is therefore not used as a means of quantification of residual stress but for the construction of PQ maps to visualize the local surface state.

Elastic strain produces a “stretch” of the lattice. The related shift of bandwidths in the pattern is in the range of no more than a few pixels even with a high-resolution detector, too small for reliable evaluation. However, the lattice stretch may clearly be noticed in a displacement of zone axis in the pattern [32]. By comparison with a pattern from an unstrained crystal volume of the same orientation low lattice strain can be quantified down to $10^{-4}$ via a cross-correlation analysis. The same approach can be made for evaluating faint lattice rotations. A high-resolution EBSD detector is required, acquisition time of a pattern is about 1 sec. Evaluation is carried out interactively and off-line.

**Comparison of EBSD with XRD**

Traditionally (crystallographic) texture refers to the statistical distribution of grain orientations in polycrystals, as obtained from X-ray or neutron pole figure measurement and mathematically described by the Orientation Density Function (ODF) [1]. So the spatial arrangements of grains and grain boundaries are not explicitly considered. But with the availability of spatially resolving techniques such as Kikuchi diffraction in the SEM the field of texture research is being increasingly extended to other microstructural features such as grain boundaries, interfaces in general and the 3D characterization of microstructure. This widened experimental view has now again great impact on theory of material science as well as day-to-day applications.

Fully automated EBSD has developed into a mature alternative to X-ray pole figure measurements in quantitative texture analyses of selected small areas of any shape. Outstanding advantages are high spatial resolution down to typically 50 nm, access to orientation correlations and orientation stereology, high speed, and representation of texture and grain boundary character distributions visually and quantitatively via orientation maps. Angular instrument resolution is usually higher (about 0.5° with EBSD, whereas X-ray pole figures are measured with typically 3° to 5° angular step width). By means of consecutive sectioning, either by mechanical abrasion or ion milling, the microstructure can be reconstructed in three dimensions.

EBSD, though powerful, is not a universal technique to meet all requirements of texture analysis. It is limited to materials in which the grains are larger in diameter than several tens of nanometer. The surface should be reasonably flat and free from foreign layers. Surface strains must not be excessive, and the specimens must be compatible with the general requirements of electron microscopy. In particular, the specimens should be conductive and should not decompose in vacuum or under the electron beam.

Pole figure measurement by X-ray diffraction is still the standard technique for studying crystal texture as referred traditionally to the statistical Orientation Density Function (ODF) of grain orientations in polycrystals. In special cases it offers several advantages over electron diffraction that are in particular: A high accuracy is an intrinsic feature of XRD due to the accurate knowledge of characteristic wavelengths and large diffraction angles. So a high dispersion enables the separation of textures of materials with similar lattice constants. Small grain sizes down to the nanometer range (“X-ray amorphous” materials) are not a real limitation. XRD is a gentle method. The specimen may stay on air and is not exposed to damaging radiation nor excessive heat transfer. Requirements on specimen preparation are moderate, and preparation is often non-destructive. XRD can be combined with
residual stress analysis. Ghost problems in pole figure inversion are no longer a problem due to the harmonic positivity method of ODF analysis [33]. Noticeable advancements in experimental diffraction methods have been made recently. Highlights are the development of automated X-ray goniometers including position sensitive linear and area detectors, CCD and energy dispersive detectors, new X-ray optics (mirror and capillary optics), and the use of synchrotron radiation.

However, spatial resolution of X-ray pole figure measurement is often insufficient, and local microstructure is not visualized. Martensite, for instance, cannot be distinguished from ferrite by XRD, whereas EBSD enables separate texture analysis of both phases by interrogating the pattern quality in the measured points. Further constraints result from stepwise specimen tilts through large angles. The beam spot is defocused and information depth varies so that inconsistent pole figure data are acquired unless the probed volumes are homogeneous in three dimensions. XRD is a slow technique unless an area detector is employed.

**Conclusion**

Spatially resolved diffraction analysis from Kikuchi patterns in the SEM has now reached a high standard. It enables individual grain orientations, local texture and point-to-point orientation correlations to be determined routinely from small volumes of material on bulk surfaces of polycrystals. It is about to become a tool for process and quality control due to the wide availability of SEM, the ease of sample preparation and of operation of commercial EBSD systems, the high speed of data acquisition – at present significantly more than 1000 orientations per second –, the ability to discriminate phases and to represent texture and grain boundary character distributions quantitatively and visually via orientation maps. A high speed and measures to enable re-examination of the results by solving stored raw pattern sequences off-line at any time are mandatory requirements for process and quality control, for dynamic experiments, for the study of sensitive materials that degrade under the electron beam or in vacuum, and for three-dimensional orientation microscopy using focused ion beam milling for serial sectioning. The quantification of plastic strain measurement is still a major challenge. EBSD is not limited to investigate local texture of small areas only, but can compete well with X-ray diffraction pole figure measurement to reveal global texture.

EBSD, though powerful, is not a universal technique to meet all requirements of texture analysis. It is limited to materials in which the grains are larger in diameter than several tens of nanometer. The surface should be reasonably flat and free from foreign layers. Surface strains must not be excessive, and the specimens must be compatible with the general requirements of electron microscopy.

Pole figure measurement by X-ray diffraction (XRD) is still the standard technique for studying global crystal texture. In special cases XRD offers several advantages over EBSD. The most outstanding benefit is an intrinsic feature of XRD: a high accuracy due to the accurate knowledge of characteristic wavelengths and large diffraction angles. So a high dispersion enables the separation of textures of materials with similar lattice constants. XRD is a gentle method. The specimen may stay on air and is not exposed to damaging radiation nor excessive heat transfer. Requirements on specimen preparation are moderate, and preparation is often non-destructive. Large specimens from the bulk are accommodated. XRD can be combined with residual stress analysis.

However, spatial resolution of X-ray pole figure measurement is often insufficient such as for studying texture gradients, and local microstructure is not revealed. Martensite, for instance, cannot be distinguished from ferrite by XRD, whereas EBSD enables separate texture analysis by evaluating the pattern quality in the measured points. Further constraints are defocusing effects or inconsistent pole figure data from inhomogeneous materials as a consequence of specimen tilts through large angles. XRD is a slow technique unless an area detector is employed.

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References


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