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A practical method to determine the complete orientation of intragranular boundaries

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Mineral boundaries play crucial although often poorly understood roles in many grain scale processes (e.g. diffusion, dislocation creep, dynamic recrystallization). Intragranular boundaries occur within individual grains and accommodate differences in crystallographic orientations of $<10^{\circ}$. As misorientations are typically defined by rotations that bring adjacent lattices into parallelism, intragranular boundaries are regarded as being crystallographically constrained. However, whilst misorientation is conveniently and efficiently measured via EBSD, it represents only three of the five degrees of freedom required to define complete orientations of intragranular boundaries. The other two degrees of freedom relate to the physical orientation of the boundary plane (i.e. plunge/azimuth of the plane normal), which has historically proved difficult to measure. Here, a practical method suggested by SEM electron channelling patterns (ECPs) is presented for determining complete orientations of intragranular boundaries.

ECPs are similar to EBSD patterns but cover much smaller angular (i.e. crystallographic) distances, defined by the 'rocking angle' (RA) of the incident beam about a point on the sample surface. Only EC bands 'dipping' $\geq$(90 - RA)$^{\circ}$ are visible in ECPs. However, unlike in EBSD patterns, a one-to-one relationship exists between each point in an ECP and its equivalent spatial position in the sample. Thus, 'partial-ECPs' can be obtained from each side of intragranular boundaries by rocking the incident beam about a point on the boundary. Spherical geometry considerations impose the following constraints: 1. the boundary plane is parallel to an EC band that is parallel to the boundary trace; and 2. the misorientation (rotation) axis is normal to an EC band that is not displaced across the boundary trace. The following parameters can be readily measured from partial ECPs to fully define intragranular boundary orientation: 1. The boundary plane dip (θ) = 90 - (d_1 + d_2)/2, where d_1, d_2 are perpendicular distances between the ECP centre and the boundary parallel EC band in both partial patterns; 2. for pure tilt boundaries, the rotation angle (ω) = d_1 - d_2; or 3. for pure twist boundaries, ω = d_{t1} - d_{t2}, where d_{t1}, d_{t2} are distances in both partial patterns between the ECP centre and any EC bands normal to but displaced across the boundary trace. Furthermore, if partial ECPs are restored to their original (i.e. pre-boundary formation) configurations, they define different 'diffraction areas' relative to the partial patterns; the difference indicates 'excess volume' due to boundary formation.

Although based on ECPs, the method is adapted to EBSD via Euler Angles and Spherical Kikuchi Maps. The former constrain orientations each side of a boundary and are located centrally on the latter. ECPs are simulated on the maps, with boundary trace drawn centrally. If the simulated patterns are juxtaposed along the trace, diffraction bands are displaced according to the *complete* boundary orientation, determined via θ, ω. We illustrate our method by considering quartz subgrains dynamically recrystallised at different temperatures. We show that subgrains are separated by combinations of tilt, twist and general (i.e. tilt + twist) boundaries with different specifications to conventional angle/axis pair misorientations.