Deformation band formation, characteristics, history

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Abstract
Deformation bands (DMB) are crystal slabs that develop slip on systems different from neighbours; they rotate in different directions developing intervening transition boundaries (TB) that rise rapidly in misorientation. In polycrystals, grains divide into DMB, slipping on 2 or 3 systems to provide the 5 required components with minimum energy, as confirmed through OIM boundary misorientations and band rotation poles. TEM exposed the subgrains but seldom TB, except in single crystals as layers of cells. As the TB extend and align into layer bands, the microtextures are similar in cold and hot working. In hot working, TB are quite narrow and able to migrate; at large strains, lengthening and rotating TB (like grain boundaries) are responsible for rapid accumulation of high angle facets of many subgrains. Nevertheless sub-boundaries persist defining steady-state cellular dimensions and flow stress.

1. Introduction
When considering the plastic deformation of single-phase crystalline metals, there are three categories of constraints although the third one is effective only at high temperatures (T>0.5T_M melting K). The dislocations move in glide, cross-slip or climb, according to the lattice slip systems to develop a substructure (cell walls (CW) below 0.5T_M and subgrain boundaries (SGB) above that) that is fairly uniform throughout the grain but with increased density near grain boundaries (GB) [1,2]. If there are no cracks, the deformation must conform to plastic stability requiring 5 independent components of the strain tensor. In polycrystals, a grain can lower its strain energy by dividing into volumetric regions called deformation bands (DMB) with intervening transition boundaries (TB). As the DMB slip on separate sets of only 2 slip systems, they rotate differently (net slip vector into elongation axis); the dislocations from the 4 systems interact in the TB that rises more rapidly in misorientation \( \psi \) than walls within a DMB. At high T, where boundaries are mobile to varying degrees, GB, TB and SGB migrate and rearrange as required to minimize surface energy [1,2].

The presentation of cold working proceeds from single crystals to polycrystals as observed by x-rays (XRD) optical and transmission electron microscopy (TEM) as understanding grew of strain hardening [3,4]. Examination of TB during annealing has thrown much light on static recovery (SRV) and nucleation of recrystallization (SRX). The role of DMB in texture formation from x-rays was confirmed through micro textures from TEM selected area diffraction (SAD) and by electron back scattering SEM-EBS (diffraction D or imaging I). Finally, the characteristics of TB in hot working are explained with the previous techniques aided by orientation imaging microscopy (OIM).

Dislocation mechanisms in single- and poly-crystals
On initially polished surfaces with varied orientations in single slip crystals, the slip bands at strains \( \varepsilon \approx 0.05 \) developed kinks transverse to the bands (XRD asterism), arising from bending constraints or internal perturbations (particles) [3-7]. As \( \varepsilon \) rose to 0.1-2, the kinks broadened into DMB, exhibiting
secondary slip bands; this could also be related to rotational overshoot in orientation for slip on a system. In crystals oriented for poly-slip, bands start almost immediately developing large rotations; in channel compression forming textural regions expected \[89\]. In rolled single crystals of Fe Si and of Al alloys, the DMB were clearly defined with alternate rotations by etching \[10\]. In TEM of single crystals, the DMB were easily located and confirmed by SAD an alternating rotation direction \[10\], as could be predicted by slip patterns \[10,11\]. The TB, often referred to as transition bands, generally consisted of layers of very fine cells (sharper and smaller ~0.05 \(\mu m\) than the cell ~0.5 \(\mu m\)), with walls of ~5° rising towards 35° as more layers formed with rising \(\varepsilon\) \[10\].

In polycrystals, the grains broke up into DMB from very low strains and rotated through large angles that were characteristic of the initial orientation of the grain relative to the imposed stress axes (consistent with single crystal rotations \[10\]) \[3\]. In Al-0.5Cu and Al-0.5Cu-1Si, compression to 0.69 \[6,7\] produced neat alternating DMB in the single-phase alloy at 20°C \[5\] but at -196°C, the DMB were finer and more irregular as in Cu-Si particle alloy at 20°C \[68B\]. None of the TEM micrographs illustrated TB but the substructure in the DMB was described as low energy dislocation structures \[6,7\], thus evidence of the first constraint acting alone. The TB are clearly observed in orientation imaging microscopy (OIM) and are permanent even on reversing the strain \[12-14\].

For cold worked polycrystals deformed by rolling through strains of 1 to 4, TEM does not definitely pinpoint TB in the small field of view. Cell structure (~0.5 \(\mu m\)) undergoes modifications due to micro bands cutting through it, forming blocks (~5 \(\mu m\)) with denser walls (BW). Some of these, also called geometrically necessary boundaries (GNB), rise fairly rapidly in misorientation, if neighboring blocks slip on systems different, whereas the cell walls (called incidental dislocation boundaries, IDB) rise slowly \[15-17\]. The cells maintain an aspect ratio of about 1.2 indicating that CW (IDB) rearrange, thus being transitory (non-permanent) \[1,2\]. Because of misalignment of slip systems in neighboring grains, dislocations form at GB smaller cellular structures, with denser walls than in grain centers. Additional dislocations are also induced by stress concentrations from the neighboring grain. While \(\psi\) may reach high values, the pattern of distribution at GB is quite different from the layers of fine cells in TB. The TB (and some BW) attain permanency and lengthen like GB, as grains deform and rotate into the elongation axis, separating bands a few cells thick with the same orientation; each is one component of the deformation texture. Micro-texture measurements showed that such regions with narrow strong textures created the X-ray texture designated as a spread of an ideal orientation averaged across the specimen \[24\]. From low \(\varepsilon\) and high \(\varepsilon\) layer bands; DMB are initially 20-100 \(\mu m\) wide, depending on grain size and impurity effects \[1,18,19\].

**Annealing after cold work**

In annealing after cold straining up to about 0.5, nucleation is observed in optical microscopy to occur primarily along GB. As \(\varepsilon\) rises above that, nucleation occurs more frequently along internal interfaces that are likely TB\[17,19\]. In single crystals with only TB nucleation, TEM clarified that dislocations from cell interiors and some weaker CW move to some surrounding walls that rise in \(\psi\) and appear sharper \[10,11\]. This process, called coalescence by Hsun Hu, continues with a cell enlarging and increasing in \(\psi\), until the energy differential induces rapid migration \[10\]. Variations on the mechanism in proximity to GB include bulging of the original GB related to differences in intersecting walls \[19\]. In studies of SRX texture formation, nuclei originate in the layer bands, depending on their cell size and perfection and on boundary structure that is related to their micro-texture. The variations in rate of SRX of a deformation micro-texture band and not its initial volume fraction, controls its strength in the SRX texture \[17,19\].

**Hot working change in constraints**

Above 0.5\(T_M\), the mobility of the boundaries SGB, TB and GB become significant principally at high strains \[1,2,20\]. From the start in hot working at constant \(T\) and strain rate \(\dot{\varepsilon}\), dislocation behavior is changed through dynamic recovery (DRV) that, above a strain \(\varepsilon_S\), results in steady state deformation
with constancy of flow stress $\sigma_S$ and equiaxed subgrain size $d_S$ (= 1.27 $W_S$ wall spacing), as well as of dislocation spacing in SGB, $W_S = b_\psi / \psi$, (Burgers vector) and in subgrains $\rho = 0.5$ ($\rho$, Frank network density) [20,21]. All substructure parameters are dependent on each other and on $\sigma_S$/$G$ ($G$, shear modulus). The subgrains were observed initially by XRD, etching in ferritic alloys and by polarized optical microscopy (POM) in Al. Through POM, DMB were reported in hot worked Ni, uranium and Al [22]; however, the contrast alternation in a few Al grains slightly above $\varepsilon_S$ arises from rows of subgrains that reorganize at higher $\varepsilon$, where DMB are much wider in both POM and OIM [12-14].

Through TEM (imaging and SAD), the characteristics of SGB ($d_S$, $W_S$, $\psi$, $\rho = 0.5$) were confirmed but TB as layers of fine cells were not seen; not more than one boundary of high $\psi > 15^\circ$ was noticeable in each view field, due to SAD or to foil tilting for $\varepsilon < 3$ [1,12-14,20,21]. In hot rolling to $\varepsilon = 2.3$ (90% reduction in one pass (20 s$^{-1}$)) of Al, Cu, Ni and austenitic stainless steels ($\gamma$-Fe), elongated grains exhibited deformation textures similar to cold working of Cu (change from brass type at low T in $\gamma$-Fe) [23-25]. In Cu and 7030 brass, there were internal grain markings, possibly TB that gave rise to nucleation along GB with upon slower cooling [23].

Single crystals deformed hot did develop high-$\psi$ TB that did not lead to nucleation in Al on continued straining or holding [26], although such TB did give rise to dynamic recrystallization (DRX) in Cu or Ni at similar $T/T_M$ [27]. In hot torsion of polycrystals of Cu, Ni and $\gamma$-Fe (C, HSLA and stainless), nucleation of discontinuous dDRX took place at relatively low critical $\varepsilon_C$ (1 - 0.2, as T rose, $\varepsilon_f$ fell). The new grains reached final size as a continuous necklace along GB and then in repeated necklaces until initial grains consumed [28,29]. Compared to single crystals, dDRX occurs before TD’s reach a suitable structure for nucleation. In steady state after work softening, nucleation appears to be random at GB of grains produced by dDRX, maintaining the texture unchanged [25].

Under high strains, the third constraint alters the microstructures compared to cold working. The GB and the TB become serrated as they migrate down SGB in order to balance the surface tensions (minimize surface energy) [1,2,12-14,18,20-23,28,29]. However, the SGB constantly rearrange to maintain subgrains equiaxed. There is continual slow rearrangement of the GB and TB in the course of which there is a net migration into grains or DMB that have smaller subgrains, i.e., higher Taylor factor [1,2,25]. This causes a change in torsion texture and a gradual softening across $\Delta\varepsilon \approx 10$ [18]. The micro textures have been measured to $\varepsilon = 60$ by scanning TEM Kikuchi patterns and agrees with the macro texture from XRD and is quite different from the dDRX texture in Cu and Ni [18,25]. At strains where layer banding was attained, the pattern of $\psi$ values indicated that between each original GB, there were several TB, the number rising with strain [18].

Since the serrations have a wavelength and amplitude near $d_S/2$, those on neighboring grains start to meet as the average grain thickness approaches $2d_S$ [1,2,17,18,20,21,30]. This shortens the grains affected and thickens the neighboring ones; this is accelerated at the thin ends of pancaked grains [2,20,21]. In result, the layer bands always remain thicker than $d_S$ and the microstructure in POM or in OIM takes on a spheroidal appearance with size near $d_S$ [21,36], with $\varepsilon_S$ being maintained; both being similar to those of specimens of 20 X larger initial grain size [18]. A good name for this is grain-defining gDRV [2], although grain shortening with an internal substructure was called geometric $^*$gDRX [2], although grain shortening with an internal substructure was called geometric $^*$gDRX (no nucleation or growth) [18].

Since through constant rearrangement of the transitory SGB maintaining the subgrains constant in size and equiaxed despite rising strain, more subgrains must come into contact with the permanent GB and TB that are lengthening as the grains pancake [12,20,25,30] (like marbles filling a balloon that is being flattened), a count of all subgrain facets show that the fraction of boundaries >15\degree rises as the fraction of SGB decline [1,12-14]. TEM shows that $\psi$ or $W_S$ of SGB remain constant to $\varepsilon = 16$ [31] and that at $\varepsilon = 60$, there are many such SGB. OIM statistics for $\varepsilon = 0.2$ to 6 exhibit such a change while fraction with $\psi = 10 - 15$ remains unchanged [12,14]. Moreover, when the $\varepsilon_f$ is reversed, 0.2 to 0.5, then to 0, the fractions transfer from high to low with $d_S$ and $\varepsilon_S$ having constant magnitude [13]. This contradicts the theory of continuous cDRX that the SGB continuously increase in $\psi$ to account for the rising fraction with $\psi>15$ [2,12,14,20].
Conclusion
In 20˚C straining, the plastic equilibrium constraint gives rise to multi-layered transition boundaries between deformation bands quite different from walls due to dislocation behavior alone. Such TB have much greater density of dislocations in multi layers, rising rapidly in misorientations and on annealing, provide nucleation sites. The deformation bands rotate to give rise to the various components of the deformation texture. When above 0.5 T\textsubscript{M}, DRV alters dislocation behavior, deformation bands still create a similar texture. The transition boundaries are much narrower (like GB), but still have high misorientation and permanence and are easily defined by OIM. Balancing surface tensions causes serrations of the non-transitory boundaries and a spheroidal microstructure no smaller than the steady state subgrain size best named grain defining dynamic recovery (gDRV).

References
[1] H.J. McQueen and W. Blum, *Al Alloys Physical and Mechanical Properties*, (ICAA6), T. Sato, ed., Japan Inst. Metals (1998), pp. 99-112.
[2] H.J. McQueen, E. Evangelista, M. Cabbibo and G. Avramovic-Cingara, *Can. Metal Quart.*, 47 (2008), 71-82.
[3] C.S. Barrett and L.H. Levenson, *Trans. AIME*, 137 (1990), 112-127.
[4] B. Jaoul, *Étude de la Plasticité et Application aux Métaux*, Dunod, Paris, 1965.
[5] R.E. Smallman and C.S. Lee, *Mater. Sci. Eng.*, A184 (1994), 97.
[6] S.S. Kulkarni, E.A. Starke Jr. and D. Kuhlman Wilsdorf, *Acta Mater.*, 46 (1998), 5283-5301.
[7] D. Kuhlman Wilsdorf, *Acta Mater.*, 47 (1999), 1697-1712.
[8] G.Y. Chin, *Inhomogeneity Plastic Deformation*, ASM, Metals Park, OH (1973), pp. 83-112.
[9] M.C. Theyssier, B. Chenal, J.H. Driver, N. Hansen, *Phys. Stat. Sol., (a)* 149 (1995), 367-378.
[10] H. Hu, *Recrystallization of Metals*, Gordon and Breach, NY (1963), pp. 311-364.
[11] A. Berger, P.J. Wilbrandt, P. Haasen, *Acta Metall.*, 31 (1983), 1433-1443.
[12] G. Avramovic-Cingara, H.J. McQueen and D.D. Perovic, *Light Metals/Metaux Legers 2004*, D. Gallienne, R. Ghomaschi eds., Met. Soc., CIM, Montreal (2004), pp. 141-152.
[13] G. Avramovic-Cingara, H.J. McQueen, *Mater. Res. Forum*, 519-523 (2006), pp. 1659-1664.
[14] G. Avramovic-Cingara and H.J. McQueen, *Aerospace Materials Manufacturing, Emerging Techniques, Met. Soc.*, CIM Montreal, M. Jahazi, et al., eds. (2006), pp. 173-186.
[15] D.A. Hughes, Q. Liu, D.C. Chrzan and N. Hansen, *Acta Mater.*, 45 (1997), 105-112.
[16] D.A. Hughes and A. Godfrey, *Hot Deformation of Al Alloys II*, T.R. Bieler, et al., eds. TMS, AIME, Warrendale, PA. (1998), pp. 23-36.
[17] R.D. Doherty, D.A. Hughes, F.J. Humphreys, J.J. Jonas, D. Juul-Jansen, M.E. Kassner, W.E. King, T.R. McNelley, H.J. McQueen and A.D Rollett, *Mater. Sci. Eng.*, 238 (1998), 219-274.
[18] J.K. Solberg, H.J. McQueen, N. Ryum and E. Nes, *Phil. Mag.*, 60 (1989), 447-471.
[19] F.J. Humphreys, M. Hatherly, *Recrystallization, Related Phenomenon*, Pergamon, Oxford 1995.
[20] H.J. McQueen, W. Blum, *Mater. Sci. Eng.*, A290 (2000), 95-107.
[21] W. Blum, Q. Zhu, R. Merkel and H.J. McQueen, *Z. Metallkd.,* 87 (1996), 14-23.
[22] F.N. Rhines, *Inhomogeneity of Plastic Deformation*, ASM Metals Park, OH. (1973), pp. 251-284.
[23] H.J. McQueen, *Trans. Japan Inst. Met.*, 9 Suppl. (1968), 170-77.
[24] H.J. McQueen and H. Mecking, *Z. Metallkd.,* 1987, 78, 387-395.
[25] H.J. McQueen, *Proc. ICOTOM 12, J.A. Szpunar,ed., NRC Res. Pub., Ottawa* (1999), pp. 836-841.
[26] M.E. Kassner, *Metal. Trans.*, 20A (1989), 2182-2185.
[27] P. Karduck, G. Gottstein and H. Mecking, *Acta Metal.*, 31 (1983), 1525-1536.
[28] H.J. McQueen, E. Evangelista and N.D. Ryan, *Recrystallization ('90) in Metals and Materials*, T. Chandra, ed., TMS-AIME, Warrendale, PA. (1990), pp. 89-100.
[29] C.M. Sellars, *Phil. Trans. Roy. Soc.*, A288 (1978), 147-158.
[30] T. Petterson and E. Nes, *Met. Trans.* 34A (2003), 2727-2736; 2737-2744.
[31] M.E. Kassner and M.E. McMahon, *Met. Trans.*, 18A (1987), 835-846.