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The limitations of Continuous Dynamic Recrystallization (CDRX) of Aluminium alloys

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Abstract: An approximate criterion for true CDRX is proposed for Al alloys bounded at low T by the limited low HAGB mobilities and at high T by high LAGB mobilities. It is shown that the shear mode is particularly important at high T and that conventional shaping processes (hot rolling, forging, extrusion) do not lend themselves to full CDRX.

Keywords; Continuous Dynamic Recrystallization, Aluminium, temperatures, strain paths

Table 1 Standard Terminology

SPD: severe plastic deformation	HAGB: high angle boundary
ECAE: equi-channel angular extrusion	GNB: geometrically necessary dislocations
HPT: high pressure torsion	DRX: dynamic recrystallization
MDF: multi-directional forging	DDRX: discontinuous dynamic recrystallization
ARB: accumulated roll bonding	CDRX: continuous dynamic recrystallization
LAGB: low angle boundary	DRV: dynamic recovery

1.Introduction

The present note addresses two major points:

i) The use of the term CDRX which often appears contrary to the standard definition of recrystallization, and

ii) The exact deformation modes by which CDRX can occur during hot deformation.

Huang and Logé [1] describe 3 types of CDRX in high SFE metals: i) CDRX by homogeneous misorientation increase ii) CDRX by progressive lattice rotation near grain boundaries and iii) microshear band assisted CDRX. The second type appears to be specific to a limited number of alloy systems so will not be treated here. The microshear band mechanism iii) is associated with the formation of large orientation gradients across shear bands, as in austenitic stainless steels, so that during repeated deformations (and strain path changes) a new, finer granular structure can develop, e.g. [4].The first, and original, type describes a progressive transformation of most LAGBs into HAGBs, by their accumulation of dislocations during straining, with a concomitant decrease of grain size [2, 3]; this type, whose mechanism is often controversial, will be analyzed here. The aim is to estimate the conditions of temperature and strain path for which true CDRX can occur in aluminium alloys.

Types i) and iii) of CDRX are considered to occur during SPD processes such as ECAE, Torsion, HPT, MDF and ARB. We shall come to the major issue of conventional shaping processes in the 2nd part.

2. Temperature effects

It has sometimes been stated that CDRX can occur by SPD at low temperatures, e.g. in the review paper by Toth and Gu [5] "*The DRX process during SPD at low temperatures is mainly a CDRX process in which the GNDs play an important role*."

The problem with the term CDRX to describe deformation microstructures developed by many SPD processes is that it does not conform to the basic definition of recrystallization as given by Doherty et al [6] "*the formation of a new grain structure in a deformed material by the formation and migration of high angle grain boundaries driven by the stored energy of deformation.*" Room temperature SPD of Al often leads to new grain structures whose formation is driven mainly by stored energy but during ambient deformation of most metals there is little chance of grain boundary migration occurring, so that fundamentally it is not a form of recrystallization. Is there any way of defining deformation conditions that would reconcile CDRX with the above definition of recrystallization?

Consider a dislocation cell in Al of typical size 1µm. One could define a minimum boundary migration rate for dynamic rex as given by 1/10 this distance divided by the typical straining time (say 1-2 secs) to give 50-100nm/sec. If the driving pressure P is taken as ½ $\rho\mu b^2$ then for Al (b=0.286nm, µ= 26GPa) P= 1.06 $10^{-9}\rho$ (N/m^2), taken here as $10^{-9}\rho$.

Assuming a typical dislocation density of 10^{14}-10^{15}, say 5 10^{14} m^{-2}, then this driving force would be 5 10^5 Jm^{-3}. And taking the above boundary velocities v gives a range of critical boundary mobility M = V/P = 4 10^{-13} – 8 10^{-13} m^{4}J^{-1}s^{-1} (to at least one order of magnitude since the pressures and velocities are obviously rather arbitrary). These critical mobility values would also be decreased by an order of magnitude since dynamic deformation tends to accelerate microstructure evolution [7]. Let us therefore take a conservative minimum mobility of 10^{-14} m^{4}J^{-1}s^{-1}.

Could this provide some guide to the boundary migration condition for CDRX to be a real type of recrystallization?

Some practical information on Al: very high purity Al (99.996) can recrystallize (slowly) at room temperature (0.32Tm) but most aluminium alloys recrystallize above about 250°C (0.56Tm). Consider some detailed experimental boundary mobilities measured by (static) recrystallization experiments on high purity Al [8] Al-17ppm Cu [78] and Al-0.1%Mn [9] and compare them with this critical mobility.

Table 2; HAGB mobilities in some Al alloys

	HAGB Mobilities (m^4 J^{-1}s^{-1})		
	RT (0.32Tm)	0.5Tm	0.7Tm
High purity Al	7 10^{-13}	2 10^{-8}	

Al-17ppm Cu	10^{-18}	10^{-10}	
Al-0.1wt%Mn		$2\ 10^{-16}$	$7\ 10^{-13}$

Taking the above estimated critical mobility (10^{-14} m^4 $J^{-1}s^{-1}$) as a criterion then high purity Al deformed by say ECAE at room temperature could undergo CDRX but probably not Al-Cu. Al-0.1%Mn at 0.5Tm is a borderline case but since most industrial Al alloys contain more solute than 0.1wt % one can consider that 0.5Tm is very much a lower limit of Al alloys for real CDRX.

However, several SPD studies of Al and its alloys at temperatures below 0.5Tm describe the development of very fine grain structures. Pure Al is a particular case; the mechanisms by which this occurs in Al of various purities are diversely reported as DDRX [10], DRV and grain growth [11] and triple point motion in rolled CP Al [12]. Apart from this special case, the obvious mechanism in the absence of boundary migration is dynamic recovery, and a particularly pronounced form of dynamic recovery at very large strains. Dynamic recovery, of say Al deformed 10-30%, is usually associated with the formation of a dislocation cell structure composed of LAGBs of typical misorientation a few degrees. But the process can continue during large strains to transform LAGBs into HAGBs (and without grain boundary migration). In fact the Riso group and others [13-15] have shown that ambient large strain deformation leads to an increase of average misorientation $\Delta\theta$ according to a power law relation $\Delta\theta \propto \varepsilon^n$ with $n \simeq 0.5$, so that at large strains a new HAGB structure is formed (still with a significant proportion of LAGBs). This has sometimes been termed “extended dynamic recovery”, but this term does not really convey the idea of a transformation to a new fine, grained structure. A better term would indicate that this is about as far as dynamic recovery can go, e.g. “complete, ultimate or asymptotic dynamic recovery”. I shall use ultimate dynamically recovered state or UDRV for these low temperature deformation-induced microstructures consisting of a fine new granular structure developed without significant boundary migration. We can recall here the standard definition of recovery “*recovery can be defined as all annealing processes occurring in deformed materials that occur without the migration of a high-angle grain boundary*" [6].

In this scheme an Al alloy deformed by SPD at room temperature could develop a UDRV structure, but at moderate temperatures > 0.5Tm or 200°C would undergo CDRX. What happens at much higher temperatures (>0.7Tm)?

CDRX, by progressive transformation of LAGBs into HAGBs, basically requires that the LAGBs do not move much – if they migrate rapidly then they can annihilate. The activation energies for LAGB migration are often considered about twice those of HAGBs so at low temperatures they do not move but can do so at much higher temperatures of order 0.7Tm, and therefore limit the upper range of CDRX. Some experimental data from sub grain growth rates of Al-0.05Si [16] and Al-0.1Mn [17] at 0.7Tm give LAGB mobilities in the range of 10^{-12} to 10^{-14} $m^4J^{-1}s^{-1}$, i.e. sometimes higher than the grain boundaries. Note that, as for boundary migration, dynamic effects would tend to substantially increase these LAGB mobilities so in practice 0.7Tm is very much an upper limit. In this context it has also been conclusively shown by Quey and Driver [18] that an Al-0.1wt%Mn alloy when deformed at 400°C

(0.72Tm) by plane strain compression does not undergo CDRX, but instead the subgrain structure is continuously renovated by LAGB dissolution and reformation to a near steady state with constant average misorientation. This is standard high temperature dynamic recovery by repolygonisation (DRV) as argued by McQueen and Kassner [19].

So for large strain experiments on Al-0.1%Mn on can reasonably postulate UDRV at room temperature (0.32Tm), true CDRX above 0.5-0.55Tm and then standard DRV above 0.65-7Tm. These are obvious estimates but are consistent with the accepted definitions of recovery and recrystallization. It can be recalled here that one of the first experimental studies to identify CDRX by ECAE (on an Al-Mg-Li-Sc alloy) used a deformation temperature of 300°C or 0.61Tm [20].

This methodology could be applied to other metallic systems but obviously requires more data on grain boundary mobilities.

3. Strain path effects.

It is also important to examine the deformation modes (strain paths) that promote or inhibit CDRX, particularly in relation to conventional shaping processes. The first publication claiming CDRX was by Perdrix and Montheillet [2] using hot torsion tests (i.e. shear) on CP Al. Virtually all the SPD processes that have been developed since to promote "CDRX" are based on macroscopic shear deformation (e.g. ECAE, HPT). The other two methods denoted ARB and MFG, although not macroscopically shear, can be considered to rely extensively on microshear banding (as in the above definition by Huang and Logé). However, in 2000 Gourdet and Montheillet [3] published some apparent indication that CDRX could occur during standard, uniaxial, hot compression of Al crystals. 3 single crystal orientation of Al were compressed at 380°C (0.7Tm) along the <100>, <111> and <110> directions. They showed that the <100> and <111> crystals developed large numbers of HAGBs at strains of order unity, in contrast to the <110> crystal which to quote *"..for the <011> orientation, all the measured misorientations are lower than 15°, even at ε=1.5. Furthermore, no evolution is noticeable between ε=0.9 and 1.5, and thus, the formation of grain boundaries is unlikely, even at larger strains*."
How would the majority of randomly oriented grains in a polycrystal behave during hot compression? From elementary crystal plasticity it is known that, during compression, any (fcc) crystal orientation within the stereological triangle (apart from those on the <100>-<111> line) rotates towards the stable <110> orientation. As a consequence, after strains of order 0.5, they would behave like the <110> orientation, i.e. without the formation of HAGBs. It follows that the creation of HAGBs in the unstable <100> and <111> orientations does not represent typical grain behaviour. This means that standard hot compression of fcc metals is inappropriate for CDRX.
The same is true of hot PSC. A detailed analysis [21] of grain orientation evolution on the inner surface of a split sample of Al-0.1%Mn during hot PSC up to strains of 1.2 showed that a large majority developed unimodal orientations (without HAGBs) while rotating towards standard PSC texture components (i.e. stable orientations). It turns out that only about 15% of the grains on the inner surface underwent orientation fragmentation by crystal deformation instability. Note this means that large EBSD maps are required to clearly distinguish unstable and stable grains. The results of this

academic study are consistent with the hot rolling behavior of industrial Al alloys that are known not to recrystallize during processing. High strength aeronautic alloys (e.g. 7000 series) are often alloyed with about 0.1% Zr which pins the grain boundaries and so inhibits recrystallization throughout the entire hot rolling process. After large strains of order 3-5 they develop strongly aligned (fibrous) grain structures but within each grain there is no sign of CDRX, just a dynamically recovered dislocation microstructure.

Extrusion processes often include a mixture of elongation and surface shear but so far there have not been clear reports of CDRX during hot extrusion.

One concludes therefore that standard shaping operations such as hot forging, hot rolling and extrusion do not lend themselves to full CDRX of Al.

In general shear deformation, as in torsion, does not provide stable, convergent, grain orientations, typical of PSC, so that the natural tendency is to favour orientation instabilities, and therefore grain orientation fragmentation (by true CDRX or towards UDRV). The current emphasis on macroscopic shear modes (ECEA, HPT) or microscopic shear assisted CDRX is understandable from an academic point of view but does not, however, appear to be really transposable to current industrial practice. Nevertheless, one should not exclude the possibility that processes like ARB and MDF can be developed to promote extensive micro shearbanding in specific alloys, and hence fine grain structures.

4.Conclusions

i) An approximate criterion for true CDRX is proposed based on the elementary requirement that a certain amount of grain boundary migration should take place to be consistent with the standard definition of recrystallization. It leads to a deformation temperature window bounded at low T by the limited HAGB mobilities and at high T by high LAGB mobilities.

ii) The criterion has been quantified by data on boundary mobilities in some binary Al alloys: for most Al alloys it leads to a maximum temperature window: $0.5<CDRX<0.7T_m$.

iii) For high strain, low temperature, deformation processes that generate new fine grained structures the predominant mechanism is often dynamic recovery; the result of this transformation is termed here an ultimate dynamically recovered state, UDRV.

iv) It is shown that the shear mode is particularly important at high T and that conventional shaping processes (hot rolling, forging, extrusion) do not in general lend themselves to full CDRX.

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