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# The flow stress evolution and grain refinement mechanisms during hot deformation of Al-Mg alloy

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**Abstract**

In this study, the hot deformation behavior of an Al-1% Mg alloy with very coarse initial grain size was investigated in terms of flow stress evolution and grain refinement mechanism. The large grain size was employed to study the traditional continuous dynamic recrystallization (CDRX) behavior below the critical strain to reach classical geometric dynamic recrystallization (GDRX), i.e., the thickness between HAGBs decreases and impingement of serrated HAGBs finally leads to the formation of new grains. The microstructure evolution during hot deformation with different deformation temperatures and strain rates on Gleeble 3800 machine was examined systematically by EBSD. It is concluded that, among the investigated conditions, the stress increases with strain and could reach steady state at small strain of ~0.01 at higher deformation temperature, while it keeps increasing at lower deformation temperature. The grain refinement mechanism clearly depends on the hot deformation condition, (micro)shear band assisted grain refinement was observed at lower deformation temperature and CDRX dominated at high deformation temperatures. In the conditions favoring CDRX, the grain refinement process is strongly grain orientation dependent, there exist some stable orientations inside which the formation of subgrain boundaries are difficult. The connection between the stress-strain behavior and microstructural evolution during hot deformation is further discussed. As compared to the traditional CDRX mechanism, the results presented in this paper provide a clearer and more complete picture of the grain refinement mechanism during hot deformation of Al-Mg alloys.



*Keywords:* Aluminium alloy; Grain refinement; Hot deformation; Dynamic recrystallization

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## 1. Introduction

It is known that grain refinement, which usually leads to improved mechanical properties, can be achieved after strain-hardening through recrystallization for materials which do not exhibit phase transformations. The grain refinement during hot deformation of high stacking fault energy (SFE) materials, such as aluminium alloys, is usually explained by continuous dynamic recrystallization (CDRX) [1, 2] or geometric dynamic recrystallization (GDRX) [3, 4, 5, 6, 7, 8]. Different dynamic recrystallization phenomena during hot deformation of metallic materials have been recently reviewed by Huang and Logé [9]. The traditional CDRX mechanism, where low angle grain boundaries (LAGBs) formed during deformation progressively transform into high angle grain boundaries (HAGBs) when deformation continues, does not consider stable grain orientations inside which the misorientation of LAGBs saturates and never transform them into HAGBs even at larger strains [10, 11]. The formation of (micro)shear bands [9, 10], frequently observed during low deformation temperatures or high strain rates, is also not considered in this mechanism. Some authors even argued that CDRX does not exist, GDRX or dynamic recovery being enough to explain the observed grain refinement at large strains [12]. Later evidences, supported by the extensively studied on hot deformation of Al alloys [1,13,14], have proved that traditional CDRX does occur to a certain extent, especially when there is a change of strain path or when deformation is dominated by shear [9]. However, it was found that stable orientations persist even after severe plastic deformation, where the deformation route changes and strong shear deformation is usually involved [10]. It appears that there is currently no satisfactory theory accounting for all observed grain refinement mechanisms during hot deformation of high SFE materials.

In this study, samples of Al-1%Mg with coarse equiaxed initial grain size were deformed in different thermomechanical conditions by hot compression tests on a Gleeble 3800 machine, at strains far below the critical strain for GDRX [9, 13]. Microstructure evolutions were examined systematically by EBSD, and the connection to flow stress behavior was discussed.

## 2. Experimental

The as-received material was a direct chill cast Al-1% Mg, supplied by Novelis Switzerland. Due to the high solubility of Mg in Al, this material has limited second-phase particles, which makes the interpretation of the experimental results easier since it is well known that second-phase particle may interact with deformation [13] and recrystallization [15]. The as-received material has a very coarse initial grain size, more than 1500μm, as shown in Fig.1. The thermomechanical processing in this research was carried out by hot compression using a Gleeble 3800 machine. Cylindrical compression samples with dimensions 15mm×10mm (height × diameter) were machined from the as-received material. A J-type thermocouple was welded to the sample to monitor and control the temperature of the test specimens. The samples were lubricated with a thin layer of graphite foil. They were heated at 5°C/s to the target temperature and soaked for 3min at that temperature. Due to the excellent thermal conductivity of Al alloys, the temperature gradient within the sample is less than 2°C. The deformation was conducted under different temperatures (300°C and 400°C) and strain rates ($0.01s^{-1}$ and $0.1s^{-1}$). Deformation was interrupted at different strain levels, i.e., 0.2, 0.4, 0.6, 0.8 and 1.0, to follow the microstructure evolution. Deformed samples were quickly air quenched after deformation to avoid static recrystallization.

The deformed samples were cut parallel to the compression axis along the centerline. They were mechanically grinded and polished according to standard metallographic procedures, a final electro-polishing step (using electrolyte Struers A2 at room temperature at 30 V for 18 s) was applied to remove the deformed layer during mechanical polishing. Inverse pole figure (IPF) maps of the these samples, represented with respect to the compression direction (CD), were obtained by Electron backscatter diffraction (EBSD) in a FEI XLF 30 field emission gun scanning electron microscope (FEG-SEM) using the HKL software. For each condition, at least an area of size 500 μm × 370 μm (always in the center of the polished surface) was scanned using a fine step size of 0.4μm in order not to miss subgrain boundaries, even though only smaller maps are typically illustrated below to highlight some of the fine scale features. Grain boundaries with misorientation angles $\theta>15°$ were considered as high angle grain boundaries (HAGBs), and those with orientation angles $1°<\theta<15°$ were designated as low angle grain boundaries (LAGBs).

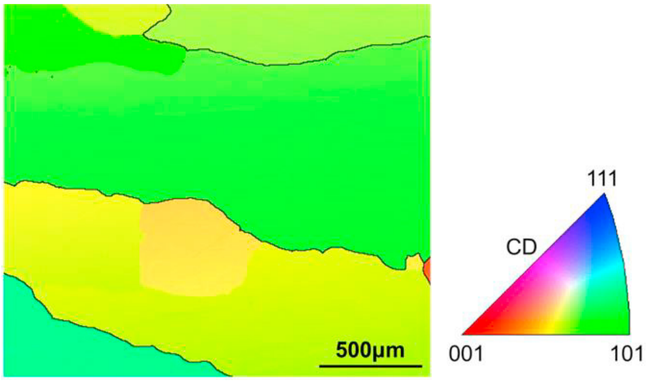


Fig. 1. EBSD map showing the microstructure of the as-received Al-1%Mg. The same IPF colour is used for the following maps

## 3. Results and discussion

### *3.1. True stress-True strain curves*

The flow stress curves of the samples deformed in three different conditions are shown in Fig. 2a. Let us take the one deformed at 400°C with strain rate of $0.01s^{-1}$ (red curve) as the reference case. It can be seen here that the stress increases very fast at the beginning of the test, and reaches a plateau of ~30MPa at a very small strain of ~0.01. This dynamic equilibrium state comes at a strain much smaller than that required for discontinuous dynamic recrystallization (DDRX) [9, 16]. When the strain rate is increased to $0.1s^{-1}$, the flow stress curve exhibited similar shape as the one deformed at smaller strain rate, but the stress level of the plateau increased to ~40MPa; there is only progressive slow increase of stress at large deformations. Even though the term "plateau" was used to describe the somewhat flat stress strain curve, it should be noted that further variations of the stress have been observed during hot torsion tests of Al alloys conducted at different temperatures and strain rates up to a strain >20 [7,17]. No test was conducted here at higher strain rates, in order to exclude the possible deformation heat. Another sample was deformed at a lower temperature of 300°C, and the stress-strain curve significantly differed from the former two cases. The continuous increase of stress with strain is obvious in this case. All the deformed samples kept a cylindrical shape, as shown in Fig.2b, which suggests that the lubrication condition employed in this study is satisfactory. Rough surface can be seen on these deformed samples which is an orange peel effect due to the large initial grain size.

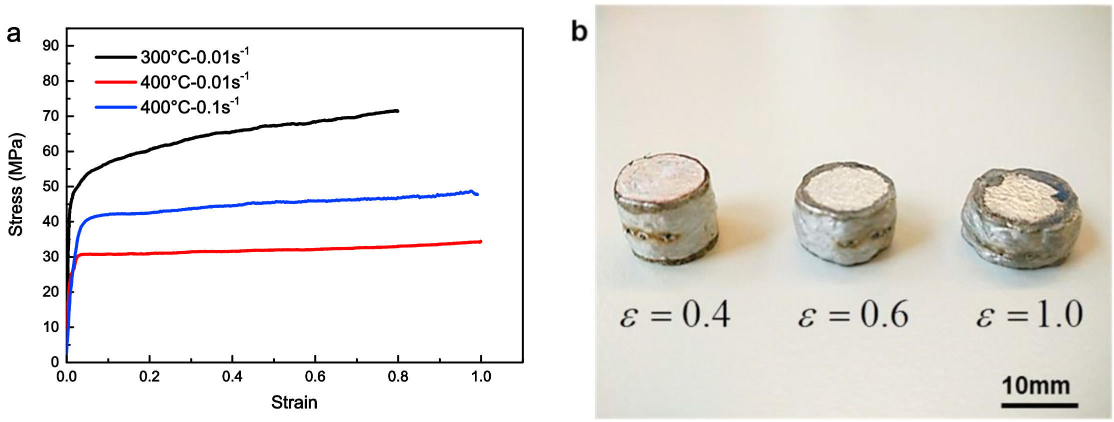


Fig.2 a) Flow stress curves; b) Typical sample geometries after deformation to different strains

### *3.2. Hot deformation at small strains*

The microstructures corresponding to different strain levels are now examined. First, it is of interest to check the microstructure at small strain levels corresponding to the formation of plateau on the stress-strain curves. The microstructure of the sample deformed to 0.2 (much larger than the strain needed to reach stress plateau) at 400°C with a strain rate of $0.01s^{-1}$ is shown in Fig.3. It is evident that the microstructure is still far away from the steady state, where substructure should be fully developed, with a size uniquely dependent on the thermomechanical condition. In fact, the formation of subgrain structure was mainly apparent near original grain boundaries, and the misorientation angles of these subgrain boundaries are mostly <4°. The majority of the sample remains free from subgrain structures.

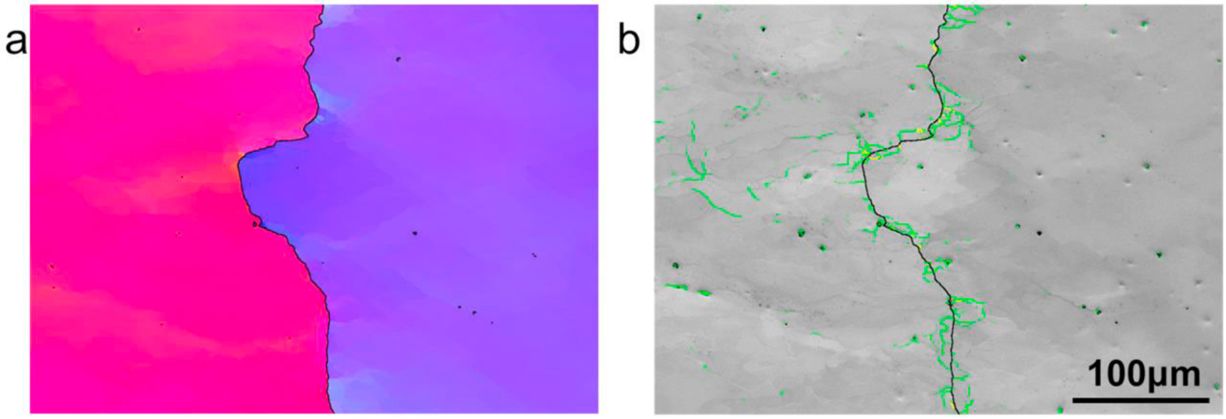


Fig.3 Microstructure of the sample deformed to 0.2 at 400°C with strain rate of $0.01s^{-1}$. a) IPF EBSD map; b) Band contrast EBSD map, where LAGBs with 1°<θ<4° are plot in green, 4°<θ<8° are plot in yellow , 8°<θ<10° are plot in fuchsia, 10°<θ<15° in red, grain boundaries with θ>15° are considered as HAGBs and plot in black.

### *3.3. Hot deformation at low temperature*

The flow stress curve of the sample deformed at a lower temperature of 300°C differs from the others, the microstructure after a deformation of 0.8 at $0.01s^{-1}$ was also examined, as illustrated in Fig.4. While it is clear that the observed area belongs to the single grain with pink color, elongated grains with blue color are formed within the grain, separated by HAGBs, as shown in Fig.4a. In general, these elongated blue grains are parallel to each other, which implies that their formation mechanism is crystallographically dependent, which is, however, not the focus of this paper. The focus of this study is the formation of HAGBs within the single grain. It is clear that these HAGBs are not formed in a progressive manner, since the majority of the grain exhibit subgrains with misorientation angles<4°. This formation mechanism is therefore clearly different from the traditional CDRX. These HAGBs are likely to be formed through the assistance of micro-shear bands. Even though tests with larger deformation were not conducted in this condition, it is expected that more HAGBs can be formed in this manner.

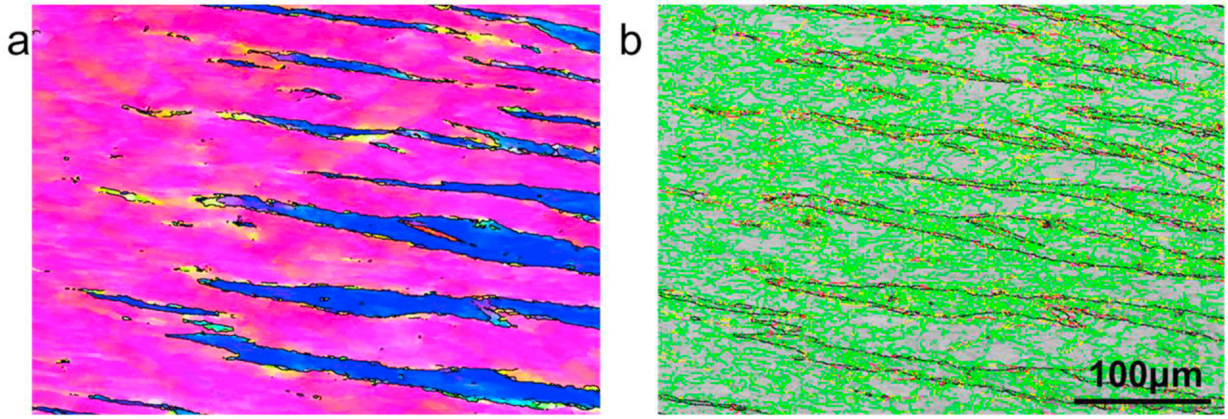


Fig. 4 Microstructure of the sample deformed to 0.8 at 300°C with strain rate of $0.01s^{-1}$. a) IPF EBSD map; b) Band contrast EBSD map with LAGBs, the color code is the same as those used in Fig.3b.

### 3.4. Hot deformation at high temperature (large strain)

The space limitation does not allow to include the microstructures of all interrupted tests, even though all of them have been examined. The microstructure of the sample tested at $0.1s^{-1}$ to 1.0, at 400°C, is shown in Fig.5. It is clear from Fig.5a that HAGBs have formed at different locations without creating yet a continuous network, which strongly suggests they are not sections of original HAGBs, and they do not originate from micro-shear bands. A plot of all boundaries in Fig.5b reveals that they must have been formed through traditional CDRX since LAGBs with misorientation angles $10° < \theta < 15°$ (red lines) are found close to the discontinuous HAGB segments. It is further noticed that no HAGB is found in the grain on the top of Fig.5b, where the majority of LAGBs have orientations<4°. The upper grain is therefore in a stable orientation with respect to the compression strain path.

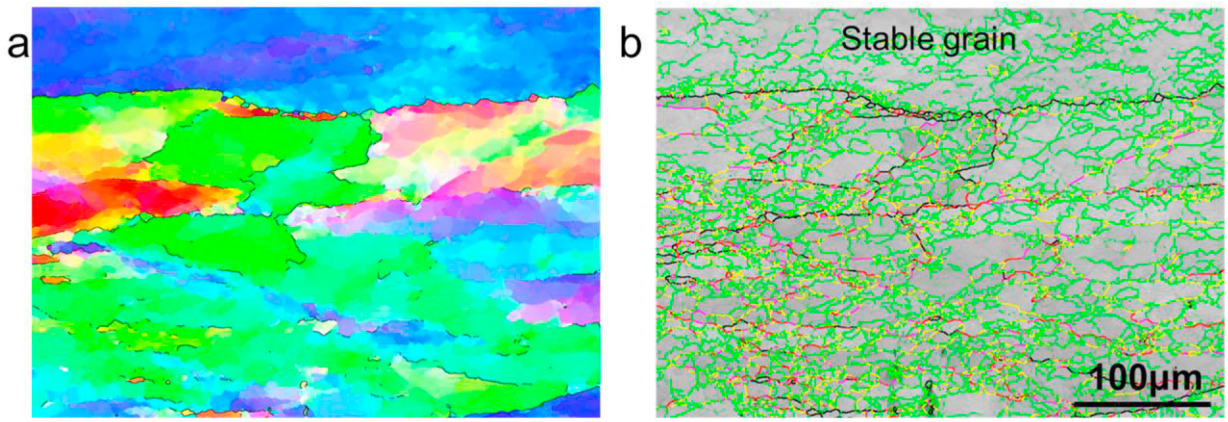


Fig. 5 Microstructure of the sample deformed to 1.0 at 400°C with strain rate of $0.1s^{-1}$. a) IPF EBSD map; b) Band contrast EBSD map with LAGBs, the color code is the same as those used in Fig.3b.

Hot deformations at even higher temperature remains to be tested, as the strong dynamic recovery could balance the formation of dislocations during deformation, leading to no net increase of dislocations which then stops the increase of LAGBs misorientation angles. Also of interest are the microstructures of the samples deformed to even higher strains. In this regard, hot torsion tests to larger strain levels are being performed, the results of which will be reported in a coming paper.

## 4. Conclusions

Various Al-1%Mg specimens with coarse initial grain size were submitted to uniaxial hot compression testing in different thermomechanical conditions. The main results are:

1) The plateau in the stress-strain curves does not correspond to a steady state microstructure, and could appear at small strains of ~0.01 where the subgrain structure is still not well developed.

2) The grain refinement mechanisms during hot deformation of high SFE materials vary with deformation conditions. Microshear band assisted grain refinement dominates at low deformation temperatures while traditional CDRX is the main mechanism operating at high temperatures.

3) In case of CDRX, the formation of HAGBs is grain orientation dependent. There exist stable crystallographic orientations within which the LAGBs are reluctant to increase their misorientation angles even though HAGBs are formed in other orientations within the same deformed sample.

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