1 Introduction

1.1 Aluminum alloys

Automotive industry demands an increasing use of aluminum alloys to reduce weight so as to improve performance and fuel consumption. However, in comparison to steel aluminum alloy sheet materials have lower formability in cold stamping processes. The 5000 series aluminum alloy (AA 5xxx) offers an alternative approach which can be deformed to quite a high percentage at elevated temperature by the so-called superplastic forming (SPF) process[1, 2]. Superplasticity is the ability of a material to undergo very large uniform neckless tensile deformation normally over 500% elongation prior to failure at a temperature well below its melting point ($T_m$). Because the deformation mechanisms fall into the grain boundary sliding (GBS) regime[3], fine grain size of 10 µm, high operation temperature of $0.9T_m$ and slow strain rate are required as for the typical AA5083 material.

Aluminum based materials that rely solely on grain boundary sliding for their superplastic properties have relatively high purity; in order to acquire their fine grain size, they require significant thermomechanical processing and thus, are quite costly. Furthermore, the low deformation strain rate, $\dot{\varepsilon}$, results in long forming times, and consequently, its application in the automotive industry is fairly limited. By promoting grain boundary sliding, a dramatic decrease of the forming time may be obtained, that is associated with $\dot{\varepsilon}$ approximately equal to $10^{-1}$ s$^{-1}$ and a maximum elongation to failure above 550% [4, 5]. This requires the reduction of the grain size in the sub-micrometer or nanometer scale, by applying severe plastic deformation (i.e. Equal-Channel Angular
Pressing, or ECAP) [5, 6]. These processes, however, are currently not capable of producing low cost material in the industrial scale.

The automotive industry requires alloys with a uniform maximum elongation between 200 and 300%. Since rolled products are preferred for volume component production, “engineering superplasticity” is concerned mostly with materials that do not follow the definition of superplasticity in the strictest sense and for that reason their deformation is often characterized as “enhanced ductility” or “quasi-superplasticity”[7, 8]. Since this thesis is concerned with this type of materials, the term “superplasticity”, or “coarse-grained superplasticity”, even though it is not exactly applicable in all circumstances, will be maintained.

For volume component production, the most widely used alloy is the high purity fine grained AA5083 (with grain size equal, or less than 10 µm). The forming time is related to $\dot{\varepsilon}$ of approximately $10^{-3}$ s$^{-1}$ and the operating $T$ is lower than 500°C. At these conditions, the material deforms not only by grain boundary sliding but also by a second process, which is based on viscous glide or solute drag creep (SDC) of dislocations. This mechanism is grain size insensitive and demonstrates strain rate sensitivity value of $m \approx 0.3$. Indeed, a recently developed constitutive model that incorporates both mechanisms seems to predict accurately the results obtained by gas-pressure bulge tests[9] A further decrease in the cost of the primary material was found to be possible by using coarse-grained AA5083, i.e. it does not require extensive thermomechanical treatment for microstructure modification. These alloys with an initial grain size of 70 µm, exhibited a maximum elongation to failure in excess of 300% at 440°C and a strain rate of $10^{-2}$ s$^{-1}$. The operation of solute drag creep alone, in coarse grained AA5083 was found to
promote dynamic recovery leading to a significant grain refinement of the microstructure (i.e. the average grain size decreased to 43 $\mu$m), and thus, an enhancement of plasticity[10]. Even though viscous glide and climb of dislocations occurred simultaneously, the former was considered rate controlling. Consequently, coarse-grained materials may well fulfill the industrial requirements and within this scope, the use of the low purity coarse-grained AA5182 would constitute the next step for further cost reduction.

The presence of additional solutes, however, may arrest the viscous glide of dislocations rendering their climb (which depends on the thermal vacancy concentration) as the rate controlling mechanism of deformation. The presence of more precipitates at the GB interfaces may hinder the intragranular dislocation motion and result in significant stress concentrations. Even though, the presence of more precipitates at the GB may prevent large grain growth, most of the deformation may be restricted intragranularly. The subdivision of the grains into sub-grains and their continuous refinement may reach the point where discontinuous dynamic recrystallization (DDRX), cannot be prevented. Regions of the microstructure with a large distribution of softer grains (i.e. having orientations where more slip systems can be activated to carry out the deformation) will most likely deform preferentially resulting in non-uniform plasticity.

The superplastic deformation of coarse-grained aluminum alloys at high temperatures is a complex phenomenon and therefore, cannot be considered solely on the basis of solute drag creep. Grain size dependent and grain size independent mechanisms take place simultaneously and modify the microstructure to such extent that it exhibits little resemblance with the original [11]. The grain “dynamic reconstruction” which is
demonstrated by extensive refinement is often explained in terms of a process referred to as “continuous recrystallization” or “continuous dynamic recrystallization” (CDRX)[12, 13]. Theoretical modeling, as well as in-situ direct observations have shown that this is most likely a “recovery dominated transformation of the microstructure that occurs homogeneously” [14, 15]. Contrary to the ordinary dynamic or discontinuous dynamic recrystallization, this process does not produce large changes in the texture and is, presumably, a balance between the progressive increase of the sub-grain boundary (SGB) misorientation by dislocation accumulation, which transforms them into low-angle grain boundaries (LAGBs) and then to high-angle grain boundaries (HAGBs), and the migration of the adjoining high-angle grain boundaries which annihilate the highly deformed sub-grain structure. As a result, the microstructure is refined up to the point where both mechanisms (i.e. the progressive increase of the sub-grain misorientation and their absorption by high-angle grain boundary migration) reach steady-state equilibrium. According to this mechanism, recovery of the dislocation substructure can occur either by their condensation into new low-angle grain boundaries or by their absorption by migrating pre-existing high-angle grain boundaries. Even though this mechanism in not well understood, it is of central importance in the superplastically deformed alloys because it does not only soften and restore the ductility of the material, it also produces a completely new grain structure with modified grain size, and shape.

1.2 Scope of the thesis

In Chapter 2, the material composition and experimental methods used throughout this thesis are introduced. The main techniques employed are Scanning Electron
Microscopy (SEM) attached with Orientation Imaging Microscopy (OIM also known as EBSD for Electron Backscattered Diffraction) facility and Transmission Electron Microscopy (TEM), and their aspects relevant to the observation and characterization of the microstructures like the grain sizes and grain boundaries are reviewed.

Chapter 3 presents the mechanical performance studied by uniaxial tensile tests on dog-bone aluminum sheet specimens. The experiments include systematic measurements of the superplastic behavior under tension over a wide range of deformation temperature and strain rate as well as under different actuation modes of the extension bar, \textit{i.e.} at constant cross-head speed or true strain rate deformation mode. In conjunction with the mechanical properties, the anisotropy and the dislocation microstructures were investigated. At the secondary necking instabilities, the average dislocation velocity increases, most dislocations break away from their solute atmospheres and thermally activated deformation occurs with high activation energies. Grain boundary sliding is prohibited due to the presence of coherent precipitates that pin effectively the grain boundary motion.

In Chapter 4, the results of EBSD measurements are presented and discussed. The experiments were performed on specimens of coarse-grained AA5182 after uniaxial tension along and perpendicular to the rolling direction. The novelty of the approach relies on the fact that the EBSD map was partitioned into deformed, recovered and recrystallized microstructure. The evolution of the texture, grain size and volume fraction, sub-grain boundary, low-angle grain boundary (LAGB) and the high-angle grain boundary (HAGB) density were investigated with increasing local strain. A stable grain size and a balanced Cube texture and Goss texture with similar proportion of each
produced large elongations. Continuous dynamic recrystallization led to homogeneous grain refinement. During deformation a dislocation controlled process produced many sub-grain boundaries (SGBs) and low-angle grain boundaries pinned by precipitates. During recovery the low-angle grain boundaries are converted into high-angle grain boundaries. Recrystallization resulted in a large volume fraction of small grains separated by high-angle grain boundaries. Continuation of the continuous dynamic recrystallization refines further the microstructure until the precipitates can no longer prevent grain boundary long range motion. Then discontinuous dynamic recrystallization leads to necking and failure.

Chapter 5 deals with the deformation mechanisms of mechanical behavior of coarse-grained Aluminum alloy. Strain rate change mechanical tests were conducted on four specimens from two AA5182 alloy materials. The specimen with grain size of 21 µm demonstrated stress exponents and activation energies characteristic of grain boundary sliding, solute drag creep and dislocation glide creep (DGC) at low, intermediate and high strain rates, respectively. Solute drag creep and dislocation glide creep are dominant mechanisms governing the deformation and contributing to the high strain rate superplasticity. Based on this specimen, deformation mechanism map is constructed presenting the dominant mechanism over regions with different temperatures and stress level. It is also demonstrated that dislocation glide creep can be the only dominant responsible mechanism of superplasticity for some specimens under certain conditions. At the end of this thesis a summary and outlook is presented in Chapter 6.
References


