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EBSD analysis of high strain rate application Al–Cu based alloys

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ABSTRACT

The high strain rate shear-compression behavior of two model Al–Cu and Al–Cu–Mn–Mg alloys was compared to a commercially available AA 2139 (Al–Cu–Mn–Mg–Ag) alloy. All three materials exhibited strain softening after the ultimate stress was reached, followed by a rapid degradation of mechanical properties after a critical strain level had been realized. Detailed microstructural characterization of the alloys at various interrupted strain levels revealed that the formation of shear bands led to the rapid degradation of properties. EBSD analysis subsequently revealed that shear bands typically formed within grains of various crystallographic orientations whose Schmid factor was maximized. Microstructural studies also correlated higher levels of local misorientation with higher concentrations of precipitates within the microstructures; samples with more alloying elements showed an increased amount of precipitation that provided more pinning points for dislocations. Evidence from this study suggests that improvements to the performance of aluminum alloys used for high strain rate applications could be made through careful control of the precipitation of secondary-phase particles as well as the orientation of the grain structure with respect to the loading direction.

1. Introduction

Highly alloyed aluminum alloys, such as AA 2139, are currently employed in the fabrication of vehicular armor due to their high strength to weight ratio and ballistic performance [1–3]. Recent advances in the fundamental understanding of these alloys [4–6] have contributed to the enhancement of mechanical performance and the development of constitutive models that are capable of accurately describing their characteristic response during deformation at high strain rates. Various studies have also shown that the mechanical properties of these alloys are closely influenced by their composition [6–8]. However, many of the existing constitutive models are phenomenological in nature and do not fully account for all of the microstructural changes that occur during deformation. In order to improve the predictive capability and develop a microstructure explicit model, a deeper understanding of the microstructural changes that occur during both processing and deformation is necessary.

The strength of 2139 aluminum can be attributed to the formation of precipitates within the microstructure, specifically the Ω and Θ' precipitates. Due to alloying with silver, 2139 can readily form Ω precipitates (Al_2Cu with an orthorhombic crystal structure) [1,7,9–14] that serve to enhance the mechanical properties of the alloy. The plate-like Ω precipitate has a tendency to nucleate intragranularly within the grain matrix itself along the {111} planes rather than at dislocations or grain boundaries. Since the interfaces along the precipitate are coherent to semi-coherent with the matrix during deformation [8–11,15], Ω precipitates tend to induce a residual tensile stress at their tips, which has the effect of uniformly distributing plastic deformation and inhibiting localized behavior, contributing to the added tensile strength of the alloy without reducing its ductility [10,16,17]. Θ' precipitates (also Al_2Cu but with a tetragonal crystal structure [18]), on the other hand, have a tendency to nucleate along {100} grain boundaries, low angle boundaries, and dislocation cores [11,12]. Due to the lack of interfacial coherency, Θ' precipitates do not contribute significantly in improving tensile strength but have been shown to enhance fracture toughness [19].

Aluminum alloys, like many other metals [20–22], are susceptible to forming adiabatic shear bands (ASB) during high strain rate deformation, as opposed to the typical deformation response during static or quasi-static deformation [23–25]. During high strain rate deformation, localized thermoplastic heating causes thermal softening and triggers the formation of ASBs that result in the accumulation of strain along narrow bands [26]. The formation of ASBs is typically considered to be detrimental to the ballistic performance of materials and is often associated with the presence of geometrical or microstructural defects. Recent studies have reported that dynamic recrystallization (DRX) may precede and retard the formation of adiabatic shear bands [27], but others indicate that DRX requires more time to occur than is available during high strain rate deformation [28,29] and could occur

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post-shear such as during the unloading process. To date, fundamental studies regarding the formation of adiabatic shear bands and how they may be controlled is still relatively limited, and further exploration may be required in order to fully understand the sequence of microstructural processes involved.

For this reason, this study was aimed at elucidating the effects of the grain orientations, secondary phases, and precipitates on the development and evolution of adiabatic shear bands in two model Al–Cu and Al–Cu–Mn–Mg alloys and 2139. EBSD mapping was used to reveal the Schmid Factor associated with different grain orientations [29,30], indicating which grains experienced higher or lower resolved shear stresses during compression [31]. Quantification of the dominant deformation mechanisms as a function of alloy composition and microstructure was performed to provide scientific insight on how to better engineer and optimize the properties of aluminum alloys for high strain rate applications.

2. Procedure

Shear-compression specimens (SCS) [32,33] were tested in a split Hopkinson pressure bar (SHPB) setup to quantitatively explore the susceptibility of different microstructures to adiabatic shear banding (ASB), which is the most commonly observed deformation mode in severe dynamic loading events such as ballistic impact or blast/shock loading. The shear-compression behaviors of three distinct alloy compositions were investigated: Al–Cu, Al–Cu–Mn–Mg, and AA 2139. The compositions of these alloys are shown in Table 1 by percent weight. Multiple shear-compression specimens were machined from each of the alloy compositions and microstructural changes in the material were quantified as a function of strain.

The model Al–Cu and Al–Cu–Mn–Mg alloys were processed in the laboratory as ~600 g heats. High purity, elemental additions were induction melted in a high purity alumina crucible under an inert cover gas and cast into a steel mold to produce rectangular ingots measuring 39 mm × 25 mm (cross section) and 80 mm long. After casting, the materials were subjected to a homogenization heat treatment at a temperature between 803 and 808 K (530–535 °C) for 3–5 h to minimize dendritic segregation in the cast structure. Following, the homogenized ingots were hot rolled just above their respective solvus temperatures (the same used for homogenization) until the plates were reduced to ~10 mm in thickness in order to break up the cast structure and refine the grain size. Recrystallization was then induced via annealing just above the solvus temperature. After quenching, the bars were immediately placed into an aging furnace and aged such that peak hardness was achieved. Afterwards, the shear-compression specimens were extracted from these aged plates. The 2139 shear-compression specimens were extracted directly from a ~50 mm thick cold rolled plate that was aged to peak hardness.

Samples were extracted such that the rolling direction was aligned along the length of the sample. Both sets of shear-compression specimens had a cylindrical shape with diameter, D, 6.35 mm and length, L, 12.7 mm. Angled slots of a width, W_0, of 1.27 mm were machined symmetrically on both sides of the sample at a 35.3° angle, alpha, relative to the base of the sample and such that they were centered relative to the sample's height, Fig. 1. The applied axial stress, P, to the sample would then be translated to shear stress upon compression and localized to the area within the machined slots. The remaining sample thickness within the slotted area, t_0, after machining was approximately 1.59 mm.

Deformation of the shear-compression specimens was conducted using a Split-Hopkinson Pressure Bar (SHPB) setup at room temperature, which is a classic technique for characterizing the high-strain-rate deformation response of materials [34]. In SHPB setup shown in Fig. 2, a striker bar accelerated using a gas gun generates a compressive stress wave in the incident bar upon impact. When the stress wave arrives at the specimen sandwiched between the two bars, part of this compressive wave is reflected back while the remaining part is transmitted through specimen to the transmission bar, during which the specimen is dynamically compressed. The strain rate associated with deformation as well as the dynamic stress–strain curve is established by using the incident, reflected and transmitted stress wave recordings via the strain gages on the bars after a time-shifting procedure. The total time window for experiments typically varies between 50 and 150 μs in SHPB compression experiments and, therefore, the dynamic deformation process can be considered to be adiabatic as shown by detailed heat conduction analysis [35].

By using shear-compression specimens, equivalent strain rates of approximately ~12,000/s were obtained in SHPB experiments. Once a complete stress–strain curve was established for each alloy system, controlled dynamic tests were conducted to predetermined strain levels using stop rings on the SHPB fixture such that the microstructure corresponding to these specific strain levels could be observed prior to shear failure. For the Al–Cu alloy, equivalent strains of 0.34 and 0.42 were investigated. For the Al–Cu–Mn–Mg alloy, equivalent strains of 0.30 and 0.47 were observed, and finally for the 2139 samples, equivalent strains of 0.11 and 0.22 were investigated.

After compression to the specified strain levels, samples were sectioned in half longitudinally and parallel to the previously machined slots such that the entire height of the sample could be metallographically prepared for EBSD characterization. Samples were prepared using standard metallographic techniques and a final polish of 0.06 μm colloidal silica suspension. EBSD characterization of the samples was completed using a JEOL-5900 LV SEM equipped with an Oxford Nordlys-HKL EBSD detector. Micro-Vickers hardness indents were used as fiducial markers such that large area mapping could be completed accurately. EBSD scans were conducted along the height of the sample, capturing the entire slot region and surrounding bulk material. Maps were stitched together and analyzed using the Oxford HKL Channel 5 software package.

Table 1
Major alloying elements of the systems investigated. Listed values reflect the weight percent of the respective element in the system.

	Al–Cu	Al–Cu–Mn–Mg	AA 2139
Cu	4.5	4.5	4.5–5.5
Mn	–	0.3	0.20–0.6
Mg	–	0.5	0.20–0.8
Ag	–	–	0.15–0.6

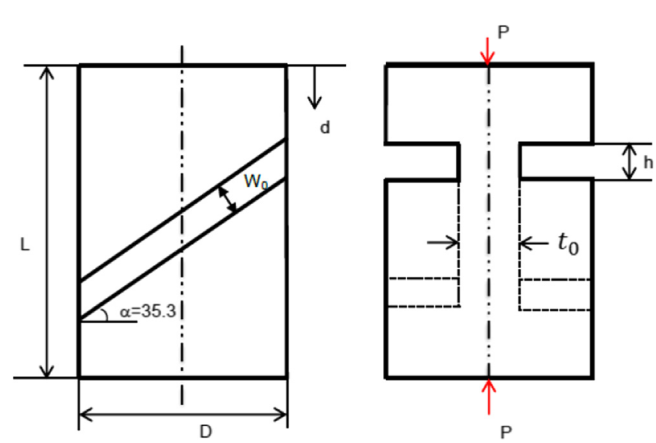

Fig. 1. Shear compression sample geometry as described in the text.