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The effects of aging treatment and strain rates on damage evolution in AA 6061 aluminum alloy in compression

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ABSTRACT

AA 6061 aluminum alloy in T4, T6 and T8 temper were subjected to quasi-static compressive loading at a strain-rate of $3.2 \times 10^{-3}\ s^{-1}$ and dynamic compressive loading at strain-rates between 7.0×10^3 and $8.5 \times 10^3\ s^{-1}$. The effects of strain rates and temper condition on the deformation behavior of the alloy are discussed. Under the quasi-static loading, deformation was relatively homogeneous and controlled by strain hardening, which is more pronounced in the naturally aged than the artificially aged alloys. Thermal softening played a dominant role under impact loading leading to strain localization along narrow bands called adiabatic shear bands (ASBs). Both deformed bands consisting of aligned second phase particles and transformed bands consisting of fine recrystallized grains were observed. The average size of the recrystallized grains in the transformed bands is about 600 nm and varies slightly depending on the temper condition. The fine grains are suggested to form by dynamic recrystallization. The T4 alloy showed the highest propensity for thermal softening, strain localization and cracking under impact loading while the T8 alloy showed the least tendency. The degree of recrystallization in the transformed band is influenced by temper condition with T8 alloy having the highest fraction of unrecrystallized grains inside the transformed bands. This is related to the temperature rise in the transformed bands that was estimated to be highest in the T4 alloy and lowest in the T8 alloy. The combined effects of high temperature and severe strain inside the transformed bands caused dissolution of second phase particles and induced microstructural changes that resulted in less silicon inside the transformed bands than in the adjacent region.

1. Introduction

Heat treatable aluminum alloys are strengthened by age hardening, a process by which fine precipitates are generated to impede dislocation motion and increase strength. Depending on the level of coherency with the continuous aluminum phase, dislocations can either cut through or loop around these precipitates during plastic deformation. In both cases, higher stress is required to keep the dislocations in motion. The strengthening effect is however higher for precipitates cutting through a coherent precipitate than for those looping around incoherent precipitates. The damage evolution in aluminum alloys under mechanical loading can be strongly dependent on aging condition and strain rates. For instance, Han et al. [1] reported that a subsequent aging of a peak-aged AA 7075 aluminum alloy at a lower temperature results in an 8% increase in its fracture toughness. On the other hand, a secondary aging of AA 8090-T8 alloy at a lower temperature resulted in a decrease in fracture toughness [2]. Pasang et al. [3] reported that a short time second step aging of AA 8090-T8771 aluminum alloy at a temperature that is slightly higher than the initial aging temperature resulted in substantial increase in short transverse fracture toughness, but only a slight decrease in strength. Aging condition has been shown to have a significant influence on the dynamic mechanical response of an aerospace Al–Li alloy [4]. Lithium segregation during aging of Al–Li alloys plays a prominent role in their fracture behavior during subsequent mechanical loading [3]. Both strain rates and aging condition affect the energy absorption capacity of aluminum alloy foams [5,6]. The solution heat treatment temperatures of artificially aged AA 7050 aluminum alloy affect the volume fraction of second phase particles and its fracture toughness [7]. The stability and the amount of various precipitates which strengthens an Al–Mg–Si alloys are not only influenced by the aging condition but also by the Si/Mg ratio [6,8]. The interrelationship between solute content, aging condition, strain rates and mechanical behavior of aluminum alloys can be very complex. It is therefore essential to investigate each commercial aluminum alloy under different temper and loading conditions.

Due to its moderate strength, good fracture toughness, good weldability and excellent corrosion resistance, AA 6061 aluminum

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alloy is widely used for structural applications in T6 temper [9–11]. It is an Al–Mg–Si alloy which can contain between 0.14% and 0.4% Cu. The precipitation hardening of Supersaturated Solid Solution (SSSS) of Al–Mg–Si alloys is suggested by Dutta and Allen [12] to proceed in the following sequence: Supersaturated Solid Solution (SSSS) → solute clusters → GP-1 zone → GP-2 zone or $\beta'' \rightarrow \beta' \rightarrow \beta$ (Mg_2Si). While the shape of GP-1 zone could not be resolved, TEM investigations showed the β'', β' and β precipitates to be needle-like, rod-like and plate-like, respectively. Edwards et al. [13,14] suggested that solute clustering begin with independent clusters of Mg and Si atoms followed by dissolution of Mg clusters and finally formation of Mg/Si co-clusters. The solute clusters and co-clusters offer the initiation sites for the nucleation of the subsequent intermediate fine precipitates called GP-1 zone, which subsequently transforms to β'' precipitates [12,13]. The optimum strength of the peak-aged AA 6061 alloy is due to the high strengthening effect of the β'' needles which are coherent with the aluminum matrix along the needle axis. The metastable β' phase that forms beyond the peak-aged condition is semi-coherent and will lead to small decrease in hardness in a slightly over-aged condition [9,15,16]. A Si phase with varying morphologies can form alongside the stable β phase during prolonged post-peak aging of Al–Mg–Si alloy with excess silicon [17]. An intermediate natural aging before artificial aging can have a negative influence on precipitation kinetics of an Al–Mg–Si alloy depending on the natural aging time [14,16]. Copper addition, as in the case of the AA 6061 alloy, can reduce the detrimental effect of natural aging on the peak-age hardness of Al–Mg–Si alloys [18].

The good mechanical and corrosion behavior of AA 6061 aluminum alloy, coupled with its low density, make it very attractive for structural application in automotive panels. The dynamic impact response of this alloy in T6 temper condition has been investigated both in compression and in torsion [19,20]. In the current study, the effects of temper condition on its deformation and failure under both quasi-static and high strain-rate loading are compared. Under dynamic impact loading, intense strain localization along narrow bands called adiabatic shear bands initiate failure in structural materials including aluminum alloys [4]. This occurs as a result of a local rise in temperature along narrow paths as a substantial fraction of mechanical work of impact is converted to thermal energy. Mechanical instability and loss of load carrying capacity occur along these paths leading to formation of adiabatic shear bands. The effects of aging condition on the microstructural evolution accompanying the occurrence of adiabatic shear bands in AA 6061 aluminum alloy under impact loading are investigated using Scanning Electron Microscopy (SEM), X-ray Photoemission Electron Microscopy (X-PEEM) and Near Edge X-ray Fine Absorption Structure (NEXAFS) spectroscopy. The quasi-static and dynamic stress–strain responses and damage evolution in the alloy at microscale are discussed.

2. Materials and experimental procedure

Commercial AA 6061-T6 aluminum alloy rod was used in this study. The composition range in wt.% is >95% Al, 0.8–1.2% Mg, 0.4–0.8% Si, 0.15–0.4% Cu and 0.04–0.35% Cr. Some specimens of the alloy were heat treated to obtain T4 and T8 temper while some were investigated in the as-received T6 temper condition. The T4 temper was developed by solutionizing at 540 °C for 2 h, quenching in water and natural aging for 14 days at room temperature. The T8 temper was developed by solution heat treatment at 540 °C for 2 h followed by water-quenching, 17% cold work, and artificial aging at 160 °C for 18 h. There was a time lapse of about 48 h between the solution and precipitation heat treatments. The presence of Cu in the alloy will mitigate whatever adverse effect this time lapse might have on precipitation kinetics during the artificial aging process [18]. The test specimens were cylindrical in shape with diameter and length of 9.5 and 10.5 mm, respectively. Microhardness, quasi-static compression and dynamic impact tests were conducted on these specimens.

The microhardness test was carried out using the Mitutoyo Vickers machine MVK-H1 under an applied load of 200 gf according to ASTM E 384-11 standard [21]. Quasi-static compression test was conducted using Instron R5500 machine following ASTM E9-09 standard [22]. The test specimens were loaded at a strain rate of $3.2 \times 10^{-3}\ s^{-1}$ up to a maximum applied load of 100 kN. Unlike mechanical testing under quasi-static loading, there is no standard test method for high strain-rate testing. Split Hopkinson bars have been developed for the characterization of high strain-rate response of structural materials, especially for strain rates in excess of $10^3\ s^{-1}$. Design, testing and applications of split Hopkinson bars have been extensively discussed by Chen and Song [23] and Ramesh [24]. The instrumented direct impact Hopkinson pressure bar used in the current investigation is fully described in another previous publication [25]. The specimens were impacted by a 1.905-kg hardened steel projectile at momentums of 33, 39 and 44 kg m/s, which generated strain rates that ranged between 7.0×10^3 and $8.5 \times 10^3\ s^{-1}$. The polished surface of the specimens for microstructural examination were etched using a solution containing 100 ml CH_3OH, 100 ml HNO_3, 100 ml HCl and 4 drops of HF. While the HF solution revealed detailed information on the second phase particles, the grain structure of the AA 6061 aluminum alloy was more clearly revealed by the Weck's solution. The electron microscopic investigation of the deformation behavior of the impacted specimens was done using JOEL JSM-6010LV scanning electron microscope (SEM). Secondary electron imaging technique and an acceleration voltage of 15 V was used. Synchrotron light radiation at the Canadian Light Source (CLS) in Saskatoon was used for a spectroscopic investigation of the transformed shear bands observed in the impacted alloy. Typical optical and scanning electron micrographs of the alloy etched with the HF solution are presented in Fig. 1. The optical micrographs show that the alloy contains coarse grains which are irregular in shapes and sizes. The second phase particles are evenly distributed within the continuous aluminum phase. The HF etching solution preferentially attacked the second phase particles and scanning electron micrographs reveal them in groves. They are mostly rectangular in shape (plate-like) and vary in sizes. Since the HF solution better reveals the morphology and distribution of the second phase particles in the AA 6061 alloy, all the microstructures presented in this article are for the specimens that were etched with this solution.

3. Experimental results

3.1. Hardness and quasi-static compression tests

The results of compressive (quasi-static) and hardness tests on the AA 6061 aluminum alloy are presented in Fig. 2 and summarized in Table 1. Whereas the hardness of the T8 alloy is only about 2.8% higher than that of the T6 alloy, the hardness of the T6 alloy is about 23% higher than that of T4 alloy. Although the difference in the hardness of T6 and T8 alloys is marginal, the yield strength of the T8 is almost double that of the T6 while the T4 alloy showed the lowest yield strength. GP zones consisting of very fine precipitates were observed by Gracio et al. [17] in an Al–Mg–Si alloy (AA 6022) during natural aging. This caused the hardness of the alloy in T4 temper to be slightly higher than that of the alloy in the as-quenched condition. Dislocations are reported to easily cut through these GP zones at low stress. Thus the alloy in T4 temper condition will begin plastic deformation at lower flow stress com-