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An initial report on achieving high comprehensive performance in an Al-Mg-Si alloy via novel thermomechanical processing

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ABSTRACT

A novel thermomechanical processing that significantly augments the tensile mechanical strength of an Al-Mg-Si alloy while maintaining excellent tensile ductility is reported. Our thermomechanical processing involved solution treatment followed by low-temperature hold and asymmetric rolling with thickness reduction of 5–10%, and speed ratios between 1.1 and 1.5. The resultant strength and ductility is striking: the tensile strength levels were similar to or higher than the T6 and T8 tempers, while the tensile ductility was similar to that of the T3 temper. The mechanism of the novel thermomechanical processing is thought to be the result of solute clusters, dislocation substructures, and textural microstructural influences.

1. Introduction

Improvements in the characteristics of structural alloys provide new degrees of freedom in mechanical design and enable thinner structural sections that can result in significant weight reduction [1]. The 6xxx system based on Al-Mg-Si alloys contains the most widely used aluminum alloys for automotive body applications, as they possess good corrosion resistance, welding performance, and formability [2]. However, there is still a large gap in overall performance, with the combination of strength and plasticity less than satisfactory in Al-Mg-Si alloys. The main strengthening precipitate phases in Al-Mg-Si alloys are the β'' phase and β′ phase, which are intermediate metastable precursors of the equilibrium β phase (Mg_2Si) [3–6]. Like other heat-treatable aluminum alloys [7,8], the role of pre-precipitate solute clusters rich in Mg and Si are thought to play a significant role in both the hardening and evolution of the microstructure in 6xxx alloys [9]. While nonequilibrium second-phase precipitate particles can increase the strength of the alloy by hindering the movement of dislocations, stress concentrations can be introduced in various ways: at grain boundaries in the case of shearable precipitates and at the precipitate/matrix interface in the case of nonshearable precipitates [10]. As a result, conventional heat treatments (such as T6 and T8 heat treatment) for Al-Mg-Si alloys usually increase strength at the cost of ductility [11]. Stress concentration appears to be less of an issue when solute atom clusters are engineered into the microstructure [7–9,12].

Thermomechanical treatments can open up new opportunities to improve the properties of aluminum alloys. However, excellent combinations of strength and ductility are difficult to achieve [13,14]. Here, we provide an initial report on the effect of a novel thermomechanical processing (NTMP) approach on the microstructure and properties of an Al-1.0 Mg-1.0 Si alloy, the content of which falls into the category of many typical aluminum alloys for automobiles.

2. Materials and methods

A cold-rolled (CR) Al-1.0 Mg-1.0 Si-0.9 Cu-0.2 Zn (wt. %) alloy 2 mm sheet was used in this research. All samples were solution treated (ST) at 530 °C for 1 h and then quenched in water to room

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temperature. The NTMP samples were subsequently aged at 100 °C for different periods of time before being subjected to the asymmetrical rolling (ASR) process at room temperature for a total thickness reduction of 5% or 10% with one rolling pass. All NTMP samples were subjected to a final natural aging (NA) period of 15 days (d) prior to mechanical testing and characterization. In this study, the speed ratios (SR) between the rolls during the ASR were 1.1, 1.3 and 1.5. The T3 (ST, 10% strain, NA for 15 d), T4 (ST, NA for 15 d), T6 peak age (aged at 175 °C for 8 h), T8 peak age (10% strain, aged at 175 °C for 6 h) states are also measured for comparison. Our tensile tests were performed on an MTS 858 instrument. A constant strain rate of 8×10^{-4} s^{-1} was used. Three parallel samples were tested for each data point with the magnitudes of the error bars being the standard deviations. The orientation distribution functions (ODF) were measured on a D/Max 2500 X-ray diffractometer. The SEM observations were performed on a Quanta 200 field emission scanning electron microscope with an operating voltage of 20 kV. The specimens for TEM observation were prepared by the standard twin-jet electropolishing method using a solution of methanol and nitric acid (3:1 by volume). The TEM observations were performed on a TECNAI G220 transmission electron microscopy, with an operating voltage of 200 kV.

3. Results and discussion

The tensile properties of the Al-Mg-Si alloy after NTMP using different routes are provided in Table 1. It is clear that the strength of the Al-Mg-Si alloy significantly increased by using the NTMP featuring the initial low-temperature hold. In the case of NTMP (Route 2 with SR = 1.5), the ultimate tensile strength increased by over 60 MPa compared with that of the T3 alloy. Generally, the strength of aluminum alloys increases at the expense of their ductility. The T4 alloy exhibits excellent ductility but also records the lowest strength, while the T8 alloy possesses the highest strength, but records the lowest ductility. On the other hand, the NTMP techniques increase the strength of the Al-Mg-Si alloy studied here while maintaining excellent ductility. Specifically, the NTMP samples (Route 2 with SR = 1.3) demonstrate an increase in tensile yield strength (YS) to 289 MPa, ultimate tensile strength (UTS) to 371 MPa, and tensile elongation to 21.4%, which is similar to that of the T3 alloy (22.2%).

The relationship between the SR and tensile properties is provided in Fig. 1. Higher strength levels were achieved with higher SR values of 1.3 or 1.5. However, for both 5% and 10% ASR reduction, the elongation was the highest when the SR was 1.3. Following previous research that correlated variations in the tensile properties in materials processed with different SRs with changes in texture [15,16], we have also aimed to investigate changes in the microstructural texture.

The ODFs of the Al-Mg-Si alloy are shown in Fig. 2. It can be observed in Fig. 2-(a) that the dominant orientations in the conventional T6 temper are the Goss texture ({011}⟨100⟩), R-Goss texture ({011}⟨110⟩), and {001}⟨610⟩, which deviates slightly from the cube texture ({001}⟨100⟩). For samples in the conventional T8 temper (Fig. 2-(b)), there is more of a Brass texture ({011}⟨211⟩) component, due to the rolling deformation applied following ST. The former {001}⟨610⟩ orientation also continues to rotate towards an R-cube texture ({001}⟨110⟩), forming {001}⟨310⟩ as a result of the minor shear stress in conventional rolling. For the sample from Route 2 with SR = 1.3 (Fig. 2-(d)), the volume of the R-cube texture is significantly higher. We attribute this to the fact that the shear stress is greatly increased during the ASR process [17–19]. The intensity of the rolling textures (mainly the Brass texture) is also higher, indicating a larger equivalent strain despite the identical thickness reduction (10%) [20].

The texture decomposition was performed by means of the particle swarm optimization method [21], and the Taylor theory of crystal plasticity was used to calculate the Taylor factor (M), in which the average Taylor factor is the superposition of each textural component (M_i) weighted by its volume fraction (f_i) [22], as in Equation (1):

$$M = \Sigma f_i M_i, i \in (1, n) \quad (1)$$

The major decomposed components of the ODFs as well as the Taylor factor (M) are listed in Table 2.

The β-fiber components, which are textures with high M values (the Brass, S ({123}⟨634⟩) and Copper ({112}⟨111⟩) textures), are greater in Route 2 with SR = 1.3, however, the cube and R-cube components are also high; thus the overall M is only slightly increased. The change in texture components is partly in agreement with a previous work, in which the texture components are significantly affected by SR, and the β-fiber textural components reach a maximum with an SR approximately 1.3 [16].

Fracture surface morphologies of the Al-Mg-Si alloy sample post-tensile testing were investigated using SEM. The results comparing NTMP (both Route 2 with SR = 1.3 and Route 3), T6 and T8 tempers are provided in Fig. 3.

As shown in Fig. 3-(a, b), the NTMP sample fractures are mainly composed of tough dimples, which is typical for ductile fracture. In Fig. 3-(c, d), both shear surfaces and tearing ridges were observed, indicating that the fracture mechanism is likely the combination of quasi-cleavage fracture and ductile fracture; in Fig. 3-(d), the features of quasi-cleavage fracture are more pronounced. The ductility of the conventionally treated samples is lower because the quasi-cleavage fracture is usually considered as a form of brittle fracture [23].

TEM images of the Al-Mg-Si alloy under NTMP (Route 2 with SR = 1.3 and Route 3), are compared with overview microstructures of the T4 and T6 at conventional conditions in Fig. 4.

Table 1
Tensile mechanical properties of the Al-Mg-Si alloy after different treatments.

Process	Preaging	Deformation	Final aging	YS [MPa]	UTS [MPa]	Elongation [%]
T3	–	CR 10%	NA15 d	206 ± 3	312 ± 2	22.2 ± 1.8
T4		–	NA15 d	194 ± 3	309 ± 3	25.8 ± 1.2
T6		–	175 °C –8 h	212 ± 1	341 ± 7	15.4 ± 0.1
T8		CR 10%	175 °C –6 h	313 ± 3	361 ± 3	7.6 ± 0.7
Route 1	100 °C –16 h	ASR5%,1.1SR	NA15 d	241 ± 9	338 ± 7	24.4 ± 0.8
		ASR5%,1.3SR		259 ± 3	359 ± 2	25.0 ± 0.8
		ASR5%,1.5SR		249 ± 6	355 ± 5	22.5 ± 1.5
Route 2	100 °C –16 h	ASR10%,1.1SR	NA15 d	273 ± 5	352 ± 3	19.0 ± 0.8
		ASR10%,1.3SR		289 ± 3	371 ± 1	21.4 ± 0.8
		ASR10%,1.5SR		289 ± 5	373 ± 4	20.2 ± 0.8
Route 3	100 °C –16 h	CR 10%	NA15 d	222 ± 6	343 ± 5	20.2 ± 1.5