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Enhanced grain refinement in AA7050 Al alloy by deformation-induced precipitation

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ABSTRACT

In this study, an improved double-step thermomechanical treatment was proposed for producing fine grain structure of high-strength Al–Zn–Mg–Cu series alloy based on deformation-induced precipitation. The new method composed of a low temperature pre-deformation and a traditional hot rolling. Effects of this proposed treatment on microstructure and properties of the alloy were studied using optical microscopy, transmission electron microscopy, X-ray diffraction, and tensile tests. The microstructural analysis results show that the pre-deformation can accelerate both formation and spheroidization of the fine precipitates. Those deformation-induced precipitates could exert more drag force on migrations of the grain boundaries. So, recovery and coarseness of those deformed grains can be suppressed during subsequent heating and hot rolling, contributing to the final fine-grain structure of the AA7050 alloy. The properties test results reveal that the alloy processed by the double-step thermomechanical treatment displays superior strength and elongation than the conventional hot-rolled alloy because of the fine grain.

1. Introduction

As one of the key materials, the Al–Zn–Mg–Cu alloys are widely used in the aircraft and automotive industries for their excellent combination of high strength to weight ratio, toughness, corrosion resistance and so on [1]. And further improvement of properties and performance are greatly desired to exert the materials potential for safety, light-weighting and fuel efficiency.

Grain refinement often results in favorable changes of mechanical properties in Al alloys, which can meet the increasing performance demand of these alloys in aerospace and automobile industries [2]. So, many efforts have been conducted to achieve fine grain for the high-strength aluminum alloy by improving thermomechanical treatment (TMT) [3–6] and developing severe plastic deformation (SPD) techniques [7–18] as well composition adjustment [19].

For structural applications, Al–Zn–Mg–Cu alloys are generally required of reasonable dimensions in substantial quantities and low fabrication cost. Thermomechanical processing which involves the deformation and annealing of bulk alloys is usually considered to be the optimum method for producing such fine-grained alloys. Traditional thermomechanical treatment of alloys often results in grain sizes about 30–250 μm [20]. In order to achieve further fine grains with size less than 30 μm, Wert et al. developed a new or improved TMT scheme [4]. In their TMT, great number of precipitates larger than ~0.75 μm should be formed first. Those large particles stimulate recrystallization of deformed grains during subsequent annealing (which was known as particle-stimulated nucleation of recrystallization, PSN). Thus, the alloy grains could be refined to the scale of about 10 μm suiting for superplastic forming. However, this improved TMT caused an increase of cost, and producing finer grain structures needed careful controls of the material and processing parameters. Furthermore, those large precipitates could not redissolve into Al alloy completely during post heat treatment, leading to the loss of tensile strength.

SPD has its own advantages in producing stable fine or ultrafine grain for the simultaneous increase of strength and ductility. However, those SPD methods could hardly be used directly to produce large-size Al alloy parts, especially those ultra-thick plate parts used in aerospace industry [21].

Up to now, innovative thermomechanical treatment is still preferred method for grain refinement of aluminum alloys during industrial production and is also of great practical importance.

Here, we propose a new double-step thermomechanical method to produce fine-grained Al–Zn–Mg–Cu alloys according to deformation-induced precipitation (DIP) widely used in advanced steel production [22]. And the effect of such method on microstructural evolution and mechanical properties was studied.

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Table 1
Chemical composition of AA7050 alloy (in wt%).

Elements	Zn	Mg	Cu	Zr	Fe	Si	Mn	Cr	Ti	Al
wt%	5.7–6.7	1.9–2.6	2.0–2.6	0.08–0.15	<0.15	<0.12	<0.10	<0.04	<0.06	Bal.

Table 2
Description of prescribed procedures.

Designation	Processing regimes
HQ	The SQ-treated sample was heated to 573 K with a holding of 120 s
DIP	The SQ-treated sample was heated to 573 K, held for 120 s, then deformed with a reduction of 60%
CHR	Conventional hot rolling. The SQ-treated sample was heated to 673 K, held for 120 s, deformed with a reduction of 80%, one-pass deformation
DHR	Double-step hot rolling. The SQ-treated sample was heated to 573 K, held for 120 s, then pre-deformed with a reduction of 60% (first-step, DIP treatment), and sequentially heated to 673 K with a holding of 10 s at 673 K followed by deforming with a 50% reduction of DIP sample (second-step), the total reduction of double-step hot rolling is also 80%

2. Experimental

The investigations were carried out on a commercial AA7050 aluminum alloy, and the nominal composition of the alloy was shown in Table 1. The alloy was solutionized at 753 K for 16 h and 756 K for 8 h, and then rapidly quenched into room-temperature water (SQ). Cylindrical samples with 10 mm in diameter and 15 mm in height were machined from the SQ-treated alloy. All compressing deformations were carried out under the procedures as listed in Table 2 on the Gleeble-1500 thermomechanical simulator (GTMS) with a heating rate of 2 K/s and a strain rate of 10 s^{-1}. The deformed specimen was immediately water quenched to room temperature for preserving the hot-deformed microstructures.

The microstructure was studied using optical microscopy (OM), transition electron microscopy (TEM) and high resolution electron microscopy (HREM). Specimens for OM observations were prepared by the standard metallographic procedures using SiC abrasive paper, 2.5 μm diamond polishing paste and Keller's reagent. The thin foils for TEM and HREM studies were prepared by twinjet electropolishing in 33% nitric acid solution in methanol. TEM and HREM studies were performed on Hitachi H-800 TEM and FEI-TF20 HREM. X-ray diffraction (XRD) analyses were carried out with Rigaku X-ray diffractometer (D/MAX-RB) using Cu Kα radiation (λ = 0.15406 nm).

Mechanical property tests were carried out on a MTS-810 test machine at room temperature with a nominal strain rate of 10^{-3} s^{-1}. For tensile tests, both CHR-treated and DHR-treated samples were solution treated at 748 K for 30 min and aged at 393 K for 24 h (T6 treatment). All tensile test samples had 3 mm width and 1.5 mm thickness and a 15 mm gauge length. The accepted value for tensile strength or elongation was the average of three measurements.

3. Results and discussion

3.1. Microstructural evolution

Fig. 1(a) shows the optical graph of the SQ-treated materials. Obviously, after a long time holding at the high temperature, the recrystallized grains are dominant in the SQ-treated alloy. And the recrystallized grains were grown to a large size of 600 μm in length and 50–100 μm in transverse. The TEM graph of such SQ-treated alloy shows the complete dissolution of the soluble second particles of η($MgZn_2$), T($Al_2Mg_3Zn_3$) or S(Al_2CuMg) phases usually present in Al–Zn–Mg–Cu alloys, and the large size subgrain of about 8 μm (Fig. 1(b)).

Fig. 2(a) revealed grain microstructures of the DIP-treated alloy. As one could see that plenty of dislocations were formed, piled up and tangled together when the SQ-treated alloy was processed during DIP stage. Those dislocations formed small elongated cells with different sizes rather than polygonizing to well-ordered sub-grains because of low deformation temperature and high deformation rate [20]. The formation of dislocation cells indicated that the hot deformation was predominant during DIP stage. Simultaneously, large number of fine precipitates were formed and uniformly distributed throughout the alloy. Most precipitates were globular and had a size of 20 nm in diameter, only a few of spherical precipitates were about 40 nm in diameter (Fig. 2b). XRD analysis, shown in Fig. 2(c), indicated that fine precipitates were mainly $MgZn_2$.

For discovering the effect of low temperature deformation on precipitates, the SQ-treated alloy was also HQ-treated. TEM graphs

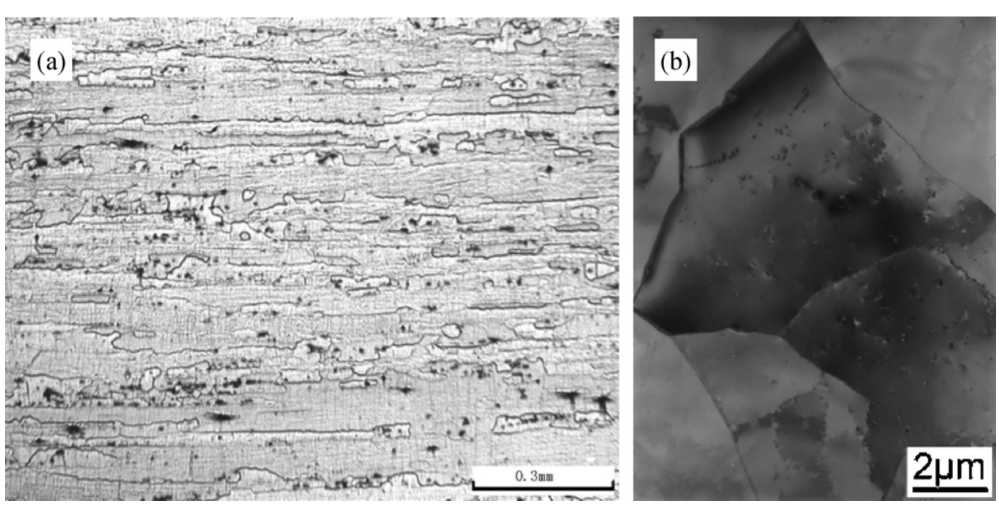

Fig. 1. Microstructures of the SQ-treated alloy (a) OM image and (b) TEM image.