EFFECT OF AGING TREATMENT ON PRECIPITATION STRENGTHENING PHASE AND MECHANICAL PROPERTIES OF AL-CU-LI ALLOY

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1. Introduction

Recently, Al-Li alloy has become a more promising alloy compared with conventional aluminum alloy. Because of its combination of low density, high elastic modulus, high specific strength, high fracture toughness, good corrosion resistance and high and low temperature performance, it can replace the traditional aluminum alloy in many aspects [1, 2]. Al-Li alloy is widely used in the field of aerospace, which reduces the weight of components by 10% to 15% and saves fuel consumption [3, 4]. Wherein, Al-Cu-Li alloy and Al-Cu-Mg-Li alloy account for more than half of Al-Li alloy for aviation [5]. As an age-hardened alloy, aging treatment is an effective method to improve the comprehensive properties of Al-Cu-Li alloy. The type, size, volume fraction, morphology and distribution of the precipitation phases determine the mechanical properties of the alloy. The precipitates of Al-Cu-Li alloy mainly include Al_2CuLi phase (T_1) and Al_2Cu phase (θ') and Al_3Li phase (δ') [6].

T_1 phase is considered to be the main strengthening precipitation phase of Al-Cu-Li alloy during aging. The plate-like T_1 phase with a close-packed hexagonal structure is semi-coherent with the matrix Al, and the lattice constants are a = b = 0.4964 nm and c = 0.9345 nm. The T_1 phase is formed on the $(111)_{Al}$ habit planes, and has an orientation relationships of $(0001)_{T1}//(111)_{Al}$, $<1010>_{T1}//<110>_{Al}$ and $<1210>_{T1}//<112>_{Al}$. When the size of T_1 phase is greater than the critical diameter and thickness, it will not be sheared by dislocation during deformation. This significantly hinders dislocation movement and can avoid coplanar slip caused by massive precipitation of δ' phase, which effectively enhance the strength and plasticity of the alloy.

2. Materials and methods

The material used in this experiment was an Al-Cu-Li alloy, whose nominal composition is Al-2Cu-2Li-0.5Mg-0.3Zn-0.1Zr-0.1Ce (wt.%). The material was cast by crucible resistance furnace. The as-cast alloy was homogenized at 500°C for 24 h and cooled in the furnace. Subsequently, the ingot was solution-treated at 530°C for 1 h and water quenched. Then it was hot-rolled at 500°C with the thickness reduction from 7 mm to 2 mm. Finally, the as-rolled samples were aged under three parameters, 150°C, 200°C, and two-stage of 150°C/12 h and 200°C.

Microhardness tests were measured by HXS-1000Z Vickers microhardness tester. The test load was 200 g, and the dwell time was 15 s. For each sample, at least ten indentations were performed to obtain an average value of microhardness. Tensile tests were measured by MTS-858 test machine with a strain rate of 1×10^{-3} s^{-1} at room temperature. Three tensile parallel samples were tested and the average value of the result was adopted. The dimensions of the tensile sample were shown in Figure 1. The tensile direction was parallel to the rolling direction.

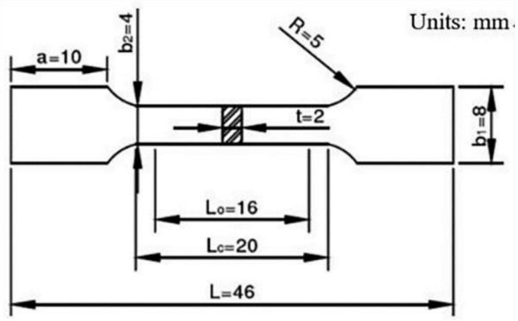

Figure 1. Schematic diagram of the dimensions of the tensile sample.

Microstructures of the alloys were observed by transmission electron microscopy (TEM, FEI Talos F200X G2). The samples for TEM observations were prepared by mechanical thinning and ion beam thinning (GATAN, PIPS II 695). The fracture surface morphologies of the samples after tensile testing were observed by scanning electron microscope (SEM, TESCAN AMBER).

3. Results

3.1 Aging hardening curves

Figure 2 shows the microhardness of Al-Cu-Li alloy with different aging temperature as a function of aging time. The microhardness of the as-rolled alloy is 83 ± 2 HV. At the initial stage of aging, the microhardness values of the alloy with different aging temperature decrease. The Guinier-Preston (GP) zones of the samples with different aging temperature are dissolved. It results in the diffusion of solute atomic clusters and reduces the degree of lattice distortion, which make microhardness of the samples be lower than that of as-rolled alloy [1]. After the short period of the initial stage, the microhardness values increase with aging time. During aging at 150°C, the microhardness value reaches the peak value (114 ± 2.0 HV) at 96 h. For the samples aged at 200℃, the time to reach the peak hardness (122 ± 1.4 HV) is 48 h, significantly shortening the peak aging time with a higher hardness peak value. For the two-stage aging alloy, the microhardness of the samples increases significantly after pre-aging at 150℃ for 12 h, and the peak hardness reaches 129 ± 2.1 HV which is higher than that of the samples aged at 150℃ and 200℃. And it can be observed that the softening is retarded compared with that of the samples aged at 150℃ and 200℃. It could be attributed to type, size and volume fraction of precipitates. During two-stage aging process, T_1 phases have a higher number density, which suppress coarsening of T_1 phases, resulting in the softening of two-stage aging is retarded.

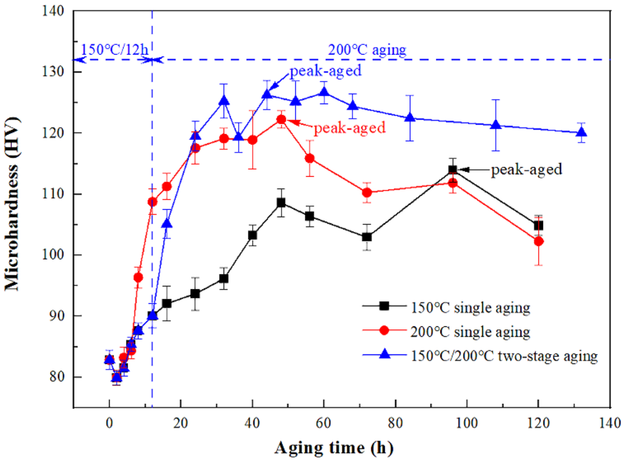

Figure 2. Microhardness of Al-Cu-Li alloy with different aging temperature for various aging times.

Conclusions

(1) Two-stage aging has the fastest aging hardening response. The softening is retarded compared with that of the samples aged at 150℃ and 200℃. It is attributed to the fact that T_1 phase has a higher number density, which competes with δ’ phase for Li atoms and suppresses coarsening of δ’ phase.

(2) The thickness of T_1 phase in two-stage aging sample is 1.2 ± 0.3 nm. The thickness of T_1 phase of the peak-aged sample at 200℃ is 2.9 ± 0.5 nm, which is more than twice that of two-stage aging. It indicates that two-stage aging restrain T_1 phase from thickening.

(3) The diameter distribution range of T_1 phase is relatively wide, ranging from 0 nm to 800 nm, which exceeds critical transition value in strengthening mechanisms from precipitates shearing to by-passing. This result in that the strengthen mechanism of T_1 phase at 200℃ peak aging and two-stage aging is a mixed strengthen mechanism of shear mechanism and Orowan bypass mechanism in this work

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