EFFECT OF DIFFERENT SOLID SOLUTION CONDITIONS ON THE MICROSTRUCTURE AND TENSILE PROPERTIES OF AA6061 ALUMINUM ALLOY

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Abstract: Aluminum–magnesium–silicon alloys are very used in extruded products and in the automotive industry. This alloys exhibit medium strength, good corrosion resistance, exceptional formability and great age hardening response. The knowledge about the kinetics of solid solution heat treatment and artificially aging is very important to understand its influence on the microstructural and mechanical features. In this study, three different solid solution treatment conditions were applied in a commercial AA6061 alloy in order to analyze the impact on the mechanical properties. Optical and scanning electron microscopy (SEM) was used to examine the microstructure of the alloys. The mechanical properties were evaluated via hardness (HRB) and tensile tests. Peak-aged conditions were reached in this alloy after a 90 min solution treatment at 803 K and 18 h aging treatment at 433 K. The variation of the yield stress, ultimate tensile strength, and ductility with aging time is measured and discussed.

Keywords: Al-Mg-Si alloys, mechanical properties, T6 heat treatment.

1. INTRODUCTION

Al–Mg–Si alloys exhibit medium strength, excellent formability, good corrosion resistance and good age hardening response due to the Mg and Si solutes that are the main responsible for the precipitation hardening and consequently the increase of the strength [1].

T6 heat treatment is very applied in 6xxx series alloys. Therefore, the solid solution treatment followed by artificial aging and water quenching are a usual practice to enhance the strength of the alloy [2]. The solid solution treatment is performed at high temperature to promote the super-saturation of the matrix with the dissociated solutes, such as Si and Mg. Artificial aging consists of heating the alloy to about 423-473 K for various periods of time and leads to the precipitation. The hardness and strength are controlled by the precipitate size, type and density [3].

The literature [4, 5] describes that the precipitation in 6xxx alloys is very complex. The precipitation sequences of these alloys are: SSSS (solution treated) → atomic clusters → GP zones → β'' → β' → β (stable phase).

It is important to find the optimum T6 heat treatment condition to achieve a good strength and to reduce the processing costs. Therefore, in this study, varying the temperature and time, the solid solution and artificial aging conditions were analyzed using hardness and tensile tests, the microstructural features were evaluated via SEM and optical microscopy.

2. EXPERIMENTAL PROCEDURE

Commercial AA6061 alloy in the form of cylindrical bars of 16 mm diameter was used in this study. The chemical composition determined by spectral analysis is shown in Table 1.

Table 1. Chemical composition of commercial AA6061 alloy (wt. %).

Al	Si	Mg	Cu	Zn	Fe	Cr
balance	0.62	0.84	0.38	0.19	0.20	0.18

The heat treatments were performed in an electric furnace without controlled atmosphere. In order to evaluate the matrix super-saturation, three solid solution conditions were used: 768 K/90 min, 768 K/180 min and 803 K/90 min. To all of the solid solution treated alloys, the artificial aging was performed at 433 K for different times and submitted to water quenching after solution and aging treatment. The scheme of the solution treatment procedure is shown in Figure 1.

Samples were prepared using conventional metallography procedures, etched with Keller's and the microstructural characterization were made by optical microscopy and scanning electron microscopy (SEM).

Rockwell B hardness were measured five times to each condition alloys. The tensile tests were performed according to ASTM E8/E8M and it was carried out in a universal test tensile machine at room temperature with a strain rate of 2 mm/min. The samples corresponding to each processing conditions were tested three times.

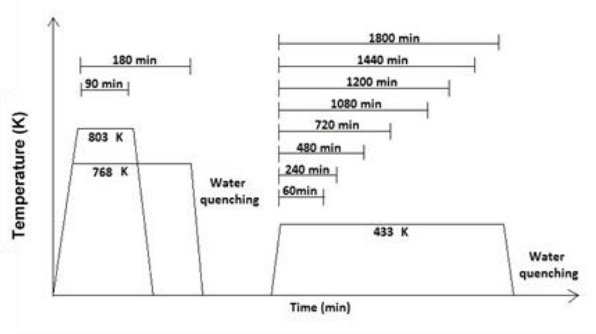

Figure 1. Scheme of the T6 heat treatments procedure to AA6061 alloys.

3. RESULTS AND DISCUSSION

Figure 2 shows the hardness with artificial aging time variation in 6061 alloys previously solid solution treated at 768 K/90 min (T1), 768 K/180 min (T2) and 803 K/90 min (T3) artificially aged at 433 K.

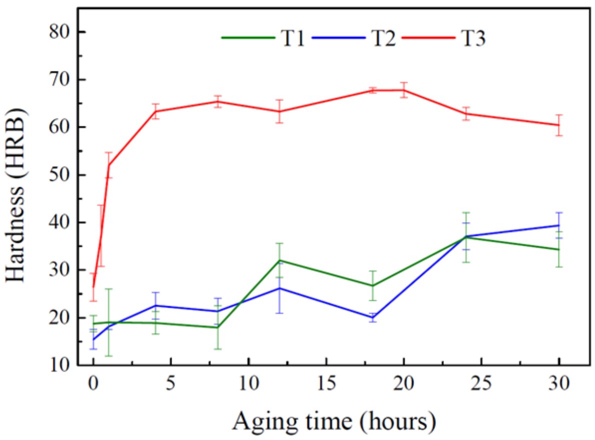

Figure 2. Variation of hardness with artificial aging time to three solid solution condition alloys: 768 K/ 90 min (T1), 768 K/180 min (T2) and 803 K/90 min (T3).

It can be seen by the curves in Figure 2 that the T3 condition alloys (solid solution treated at 803 K) exhibit a superior hardness compared to T1 and T2 (solid solution treated at 768 K). To T3 condition alloys, the hardness increase rapidly and the aged-peak is in 18 hours with the value of 67 HRB; however, in 24 and 30 hours of aging, the hardness starts to decrease. T1 and T2 conditions alloys exhibit a very slow increase of hardness. At 30 hours of artificial aging at 433 K, the hardness is only about 35-40 HRB.

The significant difference between the curves obtained at different solid solution treatment conditions are associated with the dissociation of a dissimilar amount of intermetallic phases, which is responsible for the formation of fine precipitates during the aging treatment [4]. At 803 K (T3) the kinetics of dissociation of the intermetallics phases is significantly faster compared to 768 K (T1 and T2). This result shows that even with a longer time at 768 K, the dissociation of the intermetallics particles are slowly and it is not enough to promote the super-saturation of the matrix, resulting in a poor precipitation during the aging treatment. For this reason, only the T3 solid solution treated alloys will be considered in the next steps of this study since the T1 and T2 condition treatments are not effective to promote both the super-saturation of the elements (mainly Mg and Si) in the matrix and the precipitation during aging, which is responsible for the strength enhancement.

Initially, during the precipitation step in T3 condition, co-clusters of Mg and Si and nm scale precipitates are formed, such as GP zones [4]. This initial precipitation contributes to a slightly increasing of the hardness and yield stress. At 18 hours (T3 curve, see Fig. 2), the peak aging is associated with a high density of β' and β'' precipitates with needle morphology, which act as optimal barriers for dislocations [6]. At 24 and 30 hours of aging (super-aging), a fraction of β' and β'' metastable precipitates remain in the microstructure. At the super-aging the hardness starts to decrease, it may be associated to the formation of the stable phase β-Mg_2Si, which is not coherent with the matrix and does not act as an effective obstacle to the dislocation glide. The association between the hardness and microstructure determined in literature [4, 7] is used here to characterize the state of precipitation in order to relate it to the tensile properties (Figure 5).

The microstructure of the solid solution treated alloys at 803 K/90 min and artificial aged at 433 K/18 hours is shown in Figure 3 (aged-peak of the T3 heat-treated alloys).

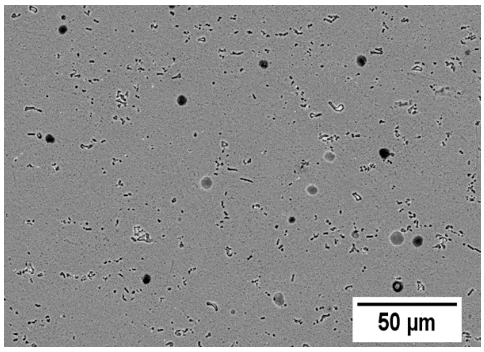

Figure 3. SEM micrograph of solid solution treated alloy at 803 K/90 min and artificial aged at 433 K/18 hours (aged-peak).

The precipitates that normally are displayed in the microstructure of T6 heat-treated AA6061 alloy are fine and their density is very high. There are coarsened precipitates on the aluminum matrix (Figure 3); however, in Figure 4(a) with a higher magnification, it is possible to observe precipitates finely distributed throughout the matrix. In this condition, there is an ideal balance between the super-saturation of the elements in the matrix and the diffusion rate obtained in the T3 heat treatment condition. The high saturation level is favorable to promote the nucleation of precipitates throughout the matrix and the high diffusion rate contributes to the growth of the nuclei already formed, resulting in the formation of small size precipitates highly dispersed throughout the matrix [8].

The typical microstructure of 6xxx alloys exhibits particles of Al_3Fe, β-AlFeSi and Mg_2Si [9]. The fine dispersed precipitates shown in Figure 4(a) may be β-Mg_2Si phase due to the high amount of this element, however, due to its small size it did not appeared in the SEM chemical map. On the other hand, it is possible to see through SEM chemical map that some precipitates exhibit iron in its composition (Figure 4(b) and (c)), which indicate that it may be Al_3Fe phase, once the morphology of β-AlFeSi exhibit needle-shape and there is 0.2 wt.% of iron in the alloy composition, see Table 1.

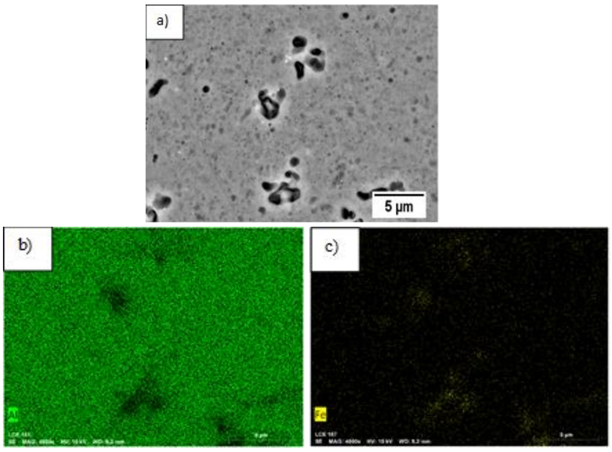

Figure 4. SEM chemical map of the elements present in the micrograph of solid solution treated at 803 K/90 min and artificial aged at 433 K/18 hours. (a) SEM micrograph and chemical map of (b) Al and (c) Fe. The others elements did not display correspondence.

Figure 5 shows the engineering stress-strain curves of the solid solution treated alloys at 803 K/90 min in the conditions of solid solution treated, artificial aged at 433 K for 1 hour (over-aged), 18 hours (peak-aged) and 30 hours (super-aged). To comparison purpose, it is also shown the stress-strain curve of the annealed condition alloy (803 K/1 h, cooled at the furnace). These condition alloys were selected to be tensile tested such to cover the full range of microstructures described above based on the hardness measurements of T3 curve in Figure 2, i.e. over-aged, peak-aged and super-aged conditions.

The annealed alloy (A) and solid solution treated alloy (B) exhibit low resistance compared to T6 heat-treated alloys once that they have not fine precipitates that act as an effective barrier to the movement of the dislocations. The formation of Mg and Si clusters followed by precipitation leads to an increase of the yield stress [7], such as can be seen in the over-aged (C) condition. However, it is possible to observe that it has some influence on the yield stress, but does not change meaningfully the development of the dislocation glide.

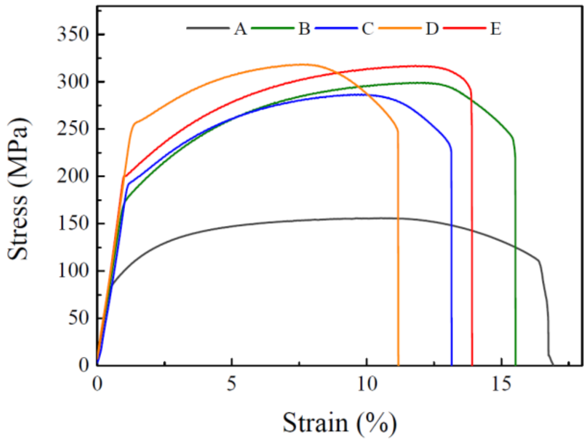

Figure 5. Measured engineering stress-strain curves of AA6061 alloys in different processing conditions: annealed alloy (A), solid solution treated (B), artificial aged for 1 hour (C), 18 hours (D) and 30 hours (E).

Since the microstructure consist of nanoscale β” precipitates homogenously distributed in the matrix [6], at the peak-aged alloy (D) the yield stress increase speedily along the aging time, see Figure 5.

In the super-aged condition (E), the resistance decreases due to the formation of coarsened β-Mg_2Si. As the particles loose coherency with the matrix, such as annealed (A) and super-aged (E) alloys, the resistance decrease due to a shift from precipitate sharing to Orawan looping [7]. Moreover, the elongation at fracture exhibits an inverse behavior of the strength. Increasing the hardness and the strength, the elongation decreases and the formability of the alloys consequently reduces.

4. CONCLUSIONS

In this study, commercial AA6061 alloy was solid solution treated and artificial aged at different conditions and the effect of aging time on mechanical properties were determined. The subsequent conclusions are obtained:

(1) Solid solution temperature of 768 K is not enough to promote super-saturation of the matrix even with a higher time of treatment due to the slow kinetics to dissociate the intermetallic particles. However, to this particular alloy, the temperature of 803 K is effective to dissociate the intermetallic particles.

(2) Peak-aging conditions are reached after solution treatment at 803 K/90 min and artificial aging at 433 K/18 hours.

(3) The nature of the precipitates changes the hardness values and the tensile properties of the alloys, such as the yield stress and ductility.

(4) It is important to find a good relationship between the resistance and formability with the T6 heat treatment based on the alloy application.

5. REFERENCES

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