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PRECIPITATION HARDENING ANALYSIS OF AN Al-8%Ag ALLOY

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Abstract. Aluminum alloys are important in aerospace industry, due to their mechanical properties, low specific weight and good corrosion resistance. Such properties are achieved due to a heat treatment of solubilization, quenching and aging, in order to precipitate metastables phases, which act as dislocation obstacles, increasing the strength of the alloy. In the present study, the precipitation sequence of Al-8%Ag alloy was analyzed via Vickers hardness and Transmission Electron Microscopy. The size and morphology of the precipitated particles, involved in the stages of precipitation process was characterized. It was determined the microstructure at the peak hardness, which is mainly composed of spherical GP zones with about 6 nm average diameter, which are responsible for the alloy achieve a value of 72 HVN. It was observed that this hardness value does not compete with others well known alloys, like AA 6061 and AA 2024, which can be precipitation hardened. The main reason for the low values of HVN, is because of there is no enough difference between the matrix and the precipitated particles lattice parameters, and don't cause a significant elastic strain by coherence in the matrix lattice, that could produce a substantial hardening. To ascertain this assumption, the aged material was severely plastic deformed, achieving 94 HVN, and the grain refinement and high dislocations density were the major hardening mechanisms, since the precipitates behavior was similar as the matrix, because particles were distorted instead of acting as impediment to material flow.

Introduction

In this study an Al-Ag alloy with low content (8% wt) was used to obtain a precipitation hardened alloy. The precipitation sequence during the phase decomposition of the supersaturated solid solution in the aluminum rich region is [1]:

$$\alpha'_{\text{(supersaturated solid solution)}} \qquad \alpha' + \text{GP Zones} \qquad \alpha' + \gamma' \qquad \alpha + \gamma_{(\text{Ag2Al})}$$

This system has been studied as an example to test the precipitate morphology theory [2,3], the phases involved are spherical GP zones, metastables hexagonal plate-like γ' precipitates, where the difference with the stable phase γ is in the lattice parameters, which are a_0=0,286 nm; c_0=0,4607 nm and a_0=0,2885 nm and c_0=0,4582 nm respectively. Due to this small difference, it is very difficult differentiate the γ' from the γ precipitates in the selected area electron diffraction pattern. The metastable precipitates are coherent with the Al matrix and exhibit well defined crystallographic orientation relationship (111)α//(0001)γ and [1-10]α//[11-20]γ [1,2].

The first attempts to determine the age hardening response in the Al-Ag system were performed by Barton *et al.* [4], who concluded that the mobility of the dislocations was reduced because of the interaction of dislocations with precipitates. Among the whole precipitation strengthening mechanisms, the stacking fault strengthening is the main mechanism acting on Al-Ag system, consequence of the difference in stacking fault energy between the matrix and the precipitates [5,6]. The aim of this work is substantiate the reason why the Al-Ag system does not improve its mechanical response because of the precipitation hardening, in comparison to other well known aluminum alloys, where the effect of the precipitates increases the mechanical strength. The assumptions were based on the lattice parameters and the possible strain caused by a misfit strain on the interphase matrix/precipitates. It has been studied previously the behavior of the alloy with 4 and 5 wt% [7], but the alloy response to the aging was poor, according to Vickers hardness tests.

Experimental Procedure

The alloy was obtained using a commercial Al-1100 alloy, and silver grit with 99,999% of purity, by melting the aluminum in a electric furnace at 1023 K, and gradually adding the silver grit to obtain the alloy by difusional process. The obtained ingot was annealed at 823 K for 48 h. The alloy was solution heat treated at 823 K and immediately quenched at 273 K in iced water, brine and methanol mixture in order to retain the solid solution. The samples were aged at 423 and 473 K, for different times. The Vickers hardness values were determined in a Leitz Wetzlar durometer. Moreover, some samples in the condition of maximum hardness were severely plastic deformed by Equal Channel Angular Pressing (ECAP), with the object of introduce a strain ~1, after 1 ECAP pass at room temperature. Samples for transmission electron microscopy (TEM) were prepared by grinding until 0,1 mm of thickness, and then electropolishing by using a twin jet method in a mixture of 70% methanol and 30% sulfuric acid at 233 K. The TEM characterization was performed in a JEOL 1230 with 100 kV current beam.

Results and Discussion

The TEM results show the evolution of precipitates morphology during the aging sequence at 473 K (Fig. 1). At 700 s of aging time (Fig. 1a) it can be seen the spherical GP zones with average diameter of 6,5 nm. The zones changed to elongated plate-like γ' precipitates after 24 h (Fig. 1b) with a length around 510 nm, and finally at 14 day, the process gave place to form the γ precipitated plates with hexagonal morphology (Fig. 1c).

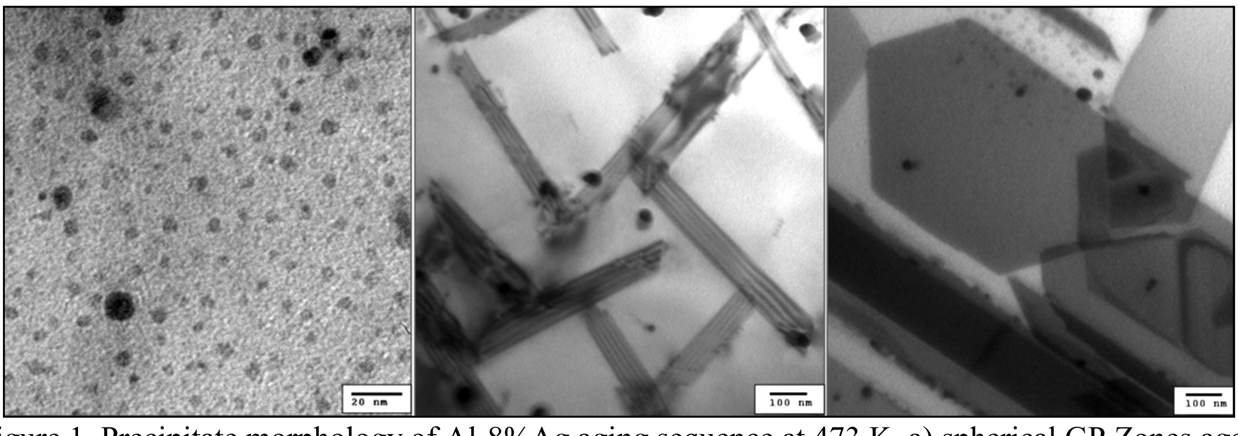

Figure 1. Precipitate morphology of Al-8%Ag aging sequence at 473 K. a) spherical GP Zones aged by 700 s, b) plate-like γ' precipitate aged by 24 h, and c) hexagonal plate-like precipitate plates aged by 14 days.

The Vickers hardness values of the precipitation process at 423 and 473 K are shown in the Fig. 2a, it can be seen that the peak hardness at 423K (72HVN) was reached after 9 days of aging time, and the process at 473 K (59 HVN) is reached in 24 h. The microstructure of the alloy at the peak hardness at 423 K aging is shown in the Fig. 2b. Spherical GP zones are the principal metastable phases found, with average diameter of 6 nm. According to Gerold and Hartmann [5], and Nembach [6], who have studied the hardening of Al by (Ag-Al) spherical zones, the major strengthening mechanism in the first stages of precipitation, is the stacking-fault strengthening, which occurs when there is a difference between the stacking fault energy of the precipitates and the matrix when these are either both fcc or hcp in structure [8].

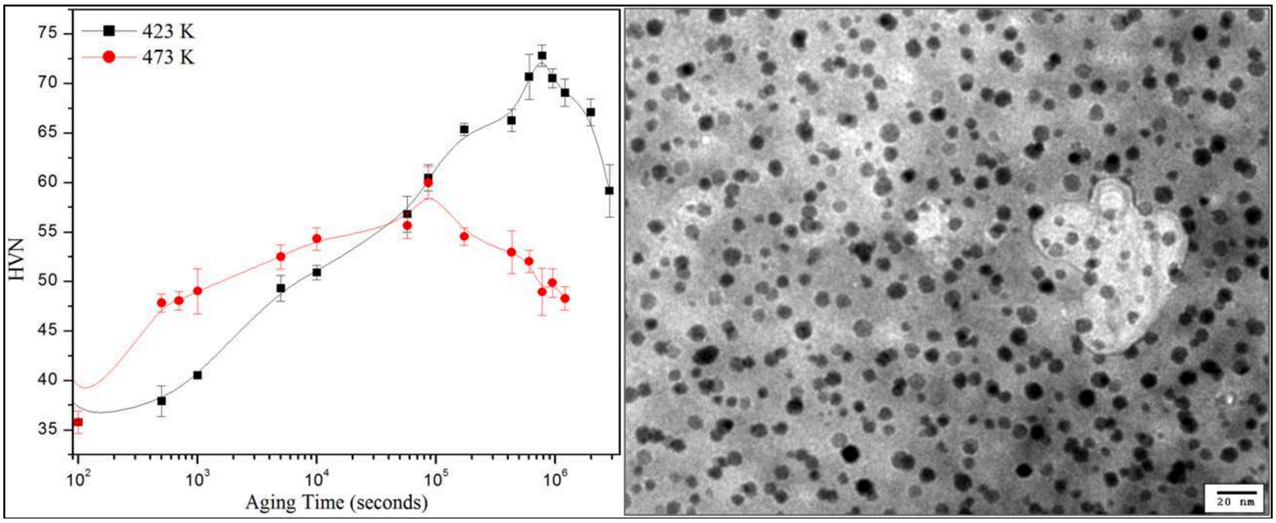

Figure 2. a) Vickers hardness vs aging time curve of Al-8%Ag alloy aging process at 423 K, b) main microstructure of the Al-8%Ag alloy in the peak hardness condition.

According hardness results, it is possible state that the peak obtained (72 HVN) arise 30 HVN from the initial value (without aging). However, this value is lower in comparison to others in several aluminum alloys like AA 2024 (140 HVN) or AA 6061 (107 HVN). This behavior is due to the small difference in lattice parameters of the precipitates and the matrix, which could cause a more significant misfit strain in the matrix, despite the great volume fraction of precipitates obtained. It is apparent from Table1, that lattice parameter a_0 of the α matrix and c_0 of γ´and γ have similar values. Besides, the strain field in the matrix around the precipitates in the TEM micrographs was not possible to elucidate, unlike others systems where the strain field in the matrix is easier to observe, and the precipitation hardening response is greater [9].

Table 1. Lattice parameters of the phases involved in the aging sequence of Al-Ag system.

Phase	**Lattice Parameter a_0 [nm]**	**Lattice Parameter c_0 [nm]**	**Crystalline Structure**
(Al) α Matrix	0,404	-	fcc
γ' Precipitate	0,286	0,460	hcp
γ Precipitate	0,288	0,458	hcp
(Ag)	0,408	-	fcc

The effect of severe plastic deformation on the hardness of the alloy can be seen from Table 2, where other hardness values in the Al-Ag system are presented. It is apparent that the aging temperature plays an important role on the mechanical properties of the Al-8%Ag alloy, because the process at 423 K had a higher Vickers hardness number than the 473 K aging. In a subsequent work, a TEM micrograph of the alloy in the peak hardness condition and ECAP processed, shown spherical GP zones as previously mentioned, mixed with regions where the precipitates seems to flow instead to be sheared or fragmented, as C. Xu reported [10] in an AA 7034 alloy, where $MgZn_2$ precipitates were fragmented from rod-like to small spheres during ECAP processing. From these results it is evident that the precipitates do not contribute to harden the alloy as much as it is expected, despite the high volume fraction achieved, substantiating the previous assumptions on the Al-Ag alloy.

Table 2. Vickers hardness values of different thermal treatments and composition of Al-Ag system in the peak hardness.

Condition	HVN
Al-8%Ag as cast	41
Al-8%Ag aged at 423 K	72
Al-8%Ag aged at 473 K	60
Al-8%Ag aged at 423 K and ECAP	94
Ohashi y col. [11] Al-10,8%Ag aged at 373 K	79
Ohashi y col. [11] Al-10,8%Ag ECAP and aged at 373 K	90

Additionally, the hardness values obtained can be compared with those of Ohashi *et al*. [11] (Table 2), whose results are similar. Ohashi *et al* have processed first by ECAP followed by aging treatment. The hardness results show that there is no great difference in the sequence of the process (Aging + ECAP vs. ECAP + aging), since the hardness value in peak hardness of the alloy is similar, being 72 and 79 HVN respectively with a conventional T6 aging, taking in count that the composition of our alloy is 2,8% lower than Ohashi *et al*. After aging and ECAP processing, the hardness values was 94 HVN, and Ohashi *et al* was 90 HVN result of ECAP processing and aging, both values are close, leading to the assumption that there is no great difference in the sequence of processing, and further strengthening in our system, conducted by ECAP after aging until peak hardness condition, is mainly due to strain hardening, grain refinement and high dislocation density introduced into the alloy, well know phenomena occurring during ECAP processing [8-12]. Further work is in progress in this theme.

Conclusions

The precipitation hardening of Al-8%Ag was characterized by means of Vickers hardness and Transmission Electron Microscopy. The morphology evolution of precipitates were determined, having spherical GP zones in the firsts stages on precipitation, elongated plate-like γ' precipitates in the precipitate growth and hexagonal γ plates on the coarsening stage. The main microstructure in the peak hardness in the 423 K aging process was composed of spherical GP zones with 6 nm diameter. The values of hardness cannot compete with commercial alloys due to the fact that there is not much difference on lattice parameters between the matrix and the precipitates, which could cause a significant misfit strain on the matrix lattice. The stacking fault strengthening is not a such good dominating strengthening mechanism that could improve the mechanical properties of the Al-Ag alloy, unlike coherency or modulus strengthening mechanisms, which domain in others systems. After processing by severe plastic deformation, was seen that the precipitates tend to flow as equal as the matrix, instead acting as obstacles, concluding that the extra hardening obtained by ECAP process in the alloy, is awarded mainly to cold work, grain refinement and the high dislocation density introduced.

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