Influence of Small Additions of Sc and Ag on the Microstructure and Properties of Al Alloy 7017

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Abstract

Individual influence of small additions of Sc and Ag on the aging behaviour, microstructure and stress corrosion cracking resistance of an Al-Zn-Mg-Mn-Zr base 7017 alloy has been examined. It is shown that both Sc and Ag additions reduce the time taken to attain the peak aged strength to one-fourth, and modify the nature of the precipitates present both within the grain as well as at the grain boundary. Both Sc and Ag additions have further effects of improving the stress corrosion cracking (SCC) resistance of the base alloy, as determined by standard tests involving alternate immersion in 3.5 % NaCl solution.

1. Introduction

The commercial 7xxx series AI alloys are based on medium strength AI-Zn-Mg and high strength AI-Zn-Mg-Cu systems. Unlike the Cu bearing alloys, the Cu-free, medium strength alloys are weldable and these alloys find wide applications where joining of components by welding is an essential step [1]. On the other hand, the Cu-free alloys have relatively inferior SCC resistance, and this property cannot be improved by over aging treatments as it is otherwise achieved in high strength, commercial 7xxx series alloys that contain Cu [1]. One known way to improve the SCC resistance of Cu-free AI-Zn-Mg alloys is to utilize Zr (instead of Cr) that is realized, for example, in the alloy AA 7017 [2]. Improvements in the SCC resistance of these alloys could further be realized through Ag additions, because Ag additions reduce the width of the "soft" precipitate free zones adjacent to the grain boundary and modify the distribution of the strengthening precipitates [3]. Recent studies have further shown that Sc additions improve the SCC resistance of high strength AI-Zn-Mg-Cu-Zr alloys [4]. It is the purpose of the present work to examine the individual effects of small additions of Sc and Ag on the aging behaviour, microstructure and SCC resistance of the alloy AA 7017.

2. Experimental Procedure

The chemical compositions of the alloys investigated in the present work are shown in Table 1. The reason for the selection of the particular Zn and Mg contents in the base alloy is to maintain close to an equiatomic proportion of Zn:Mg so as to improve the SCC resistance [5]. Zr and Mn contents are maintained in appropriate amounts in order to minimize recrystallization during alloy processing. The reason for the selection of the

particular Sc content is to primarily obtain the desired grain refinement action in the ascast microstructure [6]. Addition of 0.4 wt % i.e. about 0.1 at % Ag was typically made to reproduce the known microstructural effects that Ag has on these alloys [3]. The alloys were prepared in an induction furnace under an argon atmosphere. The as-cast alloy slabs $(450\times300\times100 \text{ mm})$ were homogenized, scalped, hot rolled to plates of 15 mm thickness, solution treated, water quenched, controlled stretched and peak aged (T651) by artificially aging at 120° C for 48 h for the base alloy, and 12 h for base alloys containing Ag and Sc.

Alloy	Zn	Mg	Zr	Mn	Sc	Ag	Fe	Si
7017	5.2	2.1	0.13	0.30	-	-	0.06	0.02
7017+Sc	5.2	2.1	0.13	0.30	0.25	-	0.06	0.02
7017+Ag	5.2	2.1	0.13	0.30	-	0.40	0.06	0.02

Table 1: Chemical compositions (wt %) of the alloys (balance AI) investigated.

Age hardening response of the alloys was monitored by Vickers hardness tests using a 10 Kg load. Tensile properties of the alloys were evaluated in longitudinal and long transverse directions, and the data obtained in the long transverse direction are presented in this paper. 50 mm gauge length specimens were pulled in tension at a nominal strain rate of 3×10^{-4} sec⁻¹. SCC resistance (K_{1scc}) of peak aged alloys was evaluated using compact tension specimens in L-T orientation through alternate immersion in 3.5 % NaCl solution in accordance with ASTM standard E-1681 [7]. Microstructural studies by transmission electron microscopy (TEM) were carried out on a Philips 430 T electron microscope operating at 300 kV. X-ray EDS (energy dispersive spectroscopy) microanalysis facilities attached to the TEM was used to detect Ag in association with the g.b. precipitates in the alloy containing Ag; the alloy was over aged (48 h at 120°C) in order to coarsen the g.b. precipitates and to facilitate the microanalysis.

3. Results and Discussion

3.1 As-Cast and Peak Aged Grain Structure

Figures 1(a) and (b) represent the grain structure in the as-cast base alloy 7017 and ascast alloy 7017 + Sc, respectively. The finer grain size of the Sc-bearing alloy is noticeable, and this is consistent with the known influence of Sc on the as-cast grain refinement in Al alloys [6,8]. Figures 1(c) and (d) represent the grain structure in peak aged (i.e. T651 temper) 7017 and 7017 + Sc alloys, respectively. The grain structure of the base alloy is partly recrystallized. Whilst, the grain structure of the Sc-bearing alloy could be best described as unrecrystallized. The grain structure of the peak aged alloy containing Ag was found to be partly recrystallized; TEM revealed the presence of well recovered subgrain structure in the unrecrystallized regions of such an alloy, Figure 1(e).

3.2 Peak Aged Tensile Properties and Hardness Data

Table 2 represents the peak aged tensile properties and hardness data revealing that compared to the base alloy, the time required to attain peak hardness at 120°C is reduced to one-fourth for the alloys containing Sc and Ag. Ag addition further increases the peak hardness by 12%.

Table 2: Peak aged (PA) tensile properties (in long transverse direction) and hardness data.

Alloys	0.2 % YS	UTS (MPa)	Elongation	Hardness	Time (h)
-	(MPa)	. ,	%	(VHN)	taken to attain
	× ,			~ /	PA hardness
7017	400	475	12	160	48
7017 + Sc	415	490	12	168	12
7017 + Ag	440	500	12	180	12



Figure 1: Light micrographs showing grain structure in (a) as-cast 7017, (b) as-cast 7017 + Sc, (c) peak aged 7017, and (d) peak aged 7017 + Sc alloys. (e) Transmission electron micrograph showing development of subgrain structure in peak aged 7017 + Ag alloy.

3.3 TEM of Peak Aged Alloy

Figures 2(a) through (f) represent transmission electron micrographs showing distribution of precipitates in peak aged materials.





Figure 2: Transmission electron micrographs showing development of microstructure in peak aged (a) & (b) 7017, (c) & (d) 7017 + Sc, and (e) & (f) 7017 + Ag alloys. The grains diffracting strongly in (a), (b), (c), (e) and (f) are in close to $<110>_{Al}$ orientation. The two-beam orientation in (d) was obtained near $<100>_{Al}$ zone.

Consideration of Figures 2(a) & (b), 2 (c) & (d), and 2 (e) & (f) corresponding to base alloy, base alloy + Sc, and base alloy + Ag, respectively shows that (1) the size of the grain boundary (g.b.) precipitates, recognized to be η (MgZn₂), and the width of the precipitate free zones (PFZs) adjacent to the g.b. are the greatest for the base alloy, Figure 2(a), (2) the width of the precipitate free zones (PFZs) adjacent to the g.b. precipitates in the Sc containing alloy are relatively

coarser and form discontinuously compared to those in the alloy containing Ag, Figure 2(c) and that (4) the alloy containing Sc contains a uniform distribution of fine $AI_3Sc_XZr_{1-X}$ precipitates that formed during the homogenization treatment, Figure 2(d).

Furthermore, the morphology of the matrix precipitates, η' (recognized in centered dark field images obtained using n'reflection), is the finest in the alloy containing Ag [compare Figures 2(b) & (c) with Figure 2 (f)]. This would be consistent with Ag influencing the formation of more number of stable G-P zones (through the known Ag-Mg-vacancy interaction) that in turn contributes toward the formation of an increased number density of η' precipitates. This explains the increased strength attained by the alloy containing Ag. Xray EDS analysis of n precipitates formed at the q.b. in the Aq-bearing alloy (see Figure 3) further revealed that the η precipitates dissolve Aq. The apparent acceleration in aging observed in the alloy containing Sc, on the other hand, appears to have been resulted due to an increase in the total grain & subgrain boundary area in that the strength increment due to dispersoid hardening and substructural strengthening due to Sc addition has partly been negated by undesirable precipitation of alloy phases on the increased g.b. regions that act simultaneously as vacancy sinks to adversely affect the morphology of precipitates within the grains [9]. This is reflected by only a slight increase in strength of the Sc-bearing alloy over that of the base alloy [see Table 2] and results of our unpublished work that 0.4 wt % Ag addition improves the peak aged strength of the Sc-bearing alloy to higher than that attained by the 7017 + Ag alloy.



Figure 3: (a) Transmission electron micrograph showing g.b. precipitates in the alloy containing Ag when aged at 120°C for 48 h. X-ray EDS spectra obtained (with same vertical scale) from (b) an η (MgZn₂) particle [marked by an arrow in (a)] at the g.b., and (c) g.b. region apparently free from any particle. An enrichment of the η (MgZn₂) phase with Ag is observed.

3.4 Stress Corrosion Cracking Studies

Table 3 presents the stress corrosion crack growth status of the ongoing SCC studies on various 7017 base alloys. It may be noted that both Ag and Sc have the effect of improving the SCC resistance of the base alloy.

Table 3: Stress corrosion crack growth, Δa (mm), status at different initial applied stress intensity factors, K (MPam^{1/2}), for peak aged (T651) alloys (CG – crack growth). Alloy K = 10 MPa m^{1/2} K = 15 MPa m^{1/2} K = 20 MPa m^{1/2}

Alloy	$K = 10 \text{ MPa m}^{1/2}$	$K = 15 \text{ MPa m}^{1/2}$	$K = 20 \text{ MPa m}^{1/2}$
7017	no CG in 5000 h	∆a ≈ 0.55 mm in 3800 h	$\Delta a \approx 0.8 \text{ mm}$ in 3800 h
7017 +Ag	-	no CG in 5000 h	no CG in 5000 h
7017 + Sc	-	no CG in 5000 h	no CG in 5000 h

The superior SCC resistance of the base alloy due to Ag and Sc additions could be due to a combination of several factors such as (1) narrow pfzs adjacent to the g.b. in Ag and Sc bearing alloys, thus reducing areas that could be anodic to the adjacent matrix, or could be softer where strain could be concentrated [1], (2) presence of Ag in the g.b. precipitate, η , and hence the possibility of alteration of the degree of anodic nature i.e. reducing the activity of such precipitates, (3) presence of a uniform distribution of unshearable Al₃Sc_xZr_{1-x} precipitates that homogenizes deformation together with the presence of smaller grain size that reduces the concentrations of stress (due to dislocations) at the g.b. in the alloy containing Sc [4], and finally (4) due to accelerated aging, solutes are removed more effectively from the solid solution in alloys containing Ag and Sc, thus reducing the corrosion potential gradient between the matrix and the grain boundary and henceforth the cause of selective corrosion.

It is the purpose of our ongoing work to verify whether the SCC resistance of the weldable alloy 7017 containing Ag or Sc could be improved to the level equivalent to that of the recently developed weldable alloy AA 2519 which combines the strength of the alloy 7017 and the SCC resistance of the alloy 5083 ($K_{1SCC} = 23 \text{ MPam}^{1/2}$) [10].

4. Summary and Conclusions

Small additions of Ag and Sc to the weldable alloy AA 7017 reduce the time required to attain the peak aged strength to one-fourth, and modify the nature and morphology of the precipitates formed within the grain as well as at the grain boundary to improve the SCC resistance considerably.

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