# Microstructure Property Relationships in a High Strength AI-Zn-Mg-Cu-Zr Alloy

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### Abstract

The use of Al-Zn-Mg-Cu-Zr base AA 7055 / AA 7449 extrusions and plates for strength critical aerospace applications is known. The present work examines the microstructure and properties of an AA 7449 alloy in the form of sheet. It is shown that selection of suitable alloy chemistry and heat treatments gives rise to a combination of high strength and stress corrosion cracking (SCC) resistance in such materials. Quantitative Varestraint test results are further presented to demonstrate that small additions of either Sc or Ag to the 7449 alloy greatly reduce the hot cracking susceptibility during solidification of welds produced by gas tungsten arc welding process.

### 1. Introduction

The high strength, Al-Zn-Mg-Cu-Zr base 7xxx series alloys having upper limits of Zn contents equal to or more than 8.4 wt % (designated AA 7055, AA 7449 etc.) have been designed during the last decade for upper wing skin materials in commercial aircrafts [1-3]. These alloys are produced in the form of extrusions as well as plates, and thermally processed using proprietary heat treatment cycles to obtain T76 level of exfoliation and SCC resistance while maintaining yield strength level comparable to that obtained in the T6 temper [1]. It is the purpose of this article to describe the process details by which an AA 7449 alloy could be readily processed in the form of sheets, and thermally treated thereafter to develop desired combinations of yield strength and SCC resistance. Such an alloy, however, like other high strength Al alloys of 7xxx series and unlike the Cu-free, weldable Al-Zn-Mg alloys, gives rise to hot cracking during solidification of welds produced by conventional gas tungsten arc welding process. It is demonstrated that suitable micro / trace alloying additions to the alloy greatly reduce the hot cracking susceptibility during solidification of welds.

## 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 Cu and Mg contents of the base alloy is not to allow the undesirable S ( $Al_2CuMg$ ) phase to survive commercial homogenization treatments [4], thereby improving workability as well as avoiding degradation of mechanical properties and SCC resistance of the wrought alloy [3,5].

Small additions of Sc (0.25 wt %) and Ag (0.4 wt %) were made in order to determine the beneficial effects of these elements on the weldability of the base alloy (i.e. like these elements have on the weldability of the Al alloy 7010 [6,7]). The weldable Al alloy 7020 was utilized as control to compare its resistance to solidification cracking with that of the base alloy and base alloys containing Sc and Ag. Al alloy 7010 (that is extensively used for aerospace applications) was utilized for thermal analysis by differential scanning calorimetry (DSC) in order to compare the major transformations that occur during heating of as-cast alloys 7010 and 7449.

Alloy	Zn	Mg	Cu	Zr	Sc	Ag	Mn	Cr	Fe	Si
7449	8.2	1.9	1.7	0.18	-	-	-	-	0.06	0.02
Base alloy										
7449+Sc	8.2	1.9	1.7	0.18	0.25	-	-	-	0.06	0.02
7449+Ag	8.2	1.9	1.7	0.18	-	0.40	-	-	0.06	0.02
7020	4.59	1.3	-	0.13	-	-	0.20	0.18	0.08	0.04
7010	6.3	2.3	1.5	0.14	-	-	-	-	0.06	0.02

Table 1: Chemical Compositions (wt %) of Alloys (balance Al) Investigated.

Except the alloy 7020 that was available commercially, the other alloys were prepared in an induction furnace under an argon atmosphere. The 7449 base alloy slabs (450  $\times$  300  $\times$ 100 mm) were homogenized, scalped, hot-rolled to sheets of 5 mm thickness, solution treated, water quenched, controlled stretched (in order to reduce the dimensional instabilities) and heat treated to the T651 (peak aged), and T7751 i.e. retrogressed and reaged (RRA) tempers. A combination of DSC and light microscopy was used to determine the optimum homogenization schedule. DSC, hardness and electrical conductivity tests were conducted to find out the optimum aging heat treatments that in turn were carried out using a combination of air circulating oven and salt bath. For the T651 temper, the allow was subjected to a two-step aging involving 8 h at 100°C followed by 24 h at 120°C in an air-circulating oven. For the T7751 temper, the T651 treated alloy was retrogressed at 205°C for 4 minutes in a salt bath, water guenched at room temperature followed by reaging to the peak aged condition using an air circulating oven. The solidification cracking susceptibility of the alloys (except the alloy 7010), processed to T651 sheets, was examined by guantitative Varestraint test [8] using conventional gas tungsten arc (GTA) welding.

Thermal analysis using DSC was undertaken in a DUPONT differential scanning 9900 instrument using standard methods [9]. Age hardening response of the alloys was monitored by Vickers hardness tests using a 10 kg load. Electrical conductivity measurement on the heat-treated samples was carried out using a direct reading type conductivity meter based on eddy current principles. Tensile properties of the base alloys were evaluated in longitudinal and long transverse directions using 25 mm gauge length specimens that were pulled in tension at a nominal strain rate of  $3 \times 10^{-4}$  sec<sup>-1</sup>. Evaluation of SCC resistance (K<sub>1scc</sub>) of artificially aged base alloys was carried out using compact tension specimens in L-T direction through alternate immersion in 3.5 % NaCl solution in accordance with ASTM standard E-1681 [10]. Microstructural studies by transmission electron microscopy (TEM) were carried out on a Philips 430 T electron microscope operating at 300 kV.

## 3. Results and Discussion

Several non-equilibrium invariant melting reactions involving  $\eta$  (MgZn<sub>2</sub>), T (Al<sub>2</sub> Mg<sub>3</sub>Zn<sub>3</sub>) and S (Al<sub>2</sub>CuMg) phases are encountered in high strength Al-Zn-Mg-Cu alloys [11]. Figures 1(a) and (b) represent DSC traces showing major transformations during heating of ascast alloys 7010 and 7449, respectively. The presence of pronounced endotherms in Figures 1(a) and (b), signature of the melting of various phase mixtures present in the ascast microstructure [12], demonstrates that compared to the alloy 7010, the maximum temperature of homogenization for the base alloy *must be restricted* to below 460°C i.e. the initiation temperature of the first endothermic reaction in Figure 1(b). A homogenization temperature of 450°C was accordingly used for the as-cast base alloy. Figures 1(c) & (d) show as-cast and homogenized microstructure, respectively of the base alloy.



Figure 1: DSC traces showing major transformations during heating of as-cast (a) alloy 7010 and (b) base alloy 7449. Light micrographs obtained from the base alloy showing (c) as-cast, and (d) homogenized (35 h at  $450^{\circ}$ C) microstructure.

Figure 2(a) represents the grain structure developed in the 5 mm thick sheet of the base alloy heat treated to the peak aged, T651 temper. The grain structure could be best described as partially recrystallized. Formation of new grains in the alloy mostly occurred at the prior grain boundaries. Figure 2(b) typically shows the subgrain structure developed throughout the unrecrystallized portions of the alloy.

Figure 2(c) shows the morphology of precipitates developed in the peak aged alloy. TEM revealed that most of the intragranular precipitates in the peak aged alloy could be observed in dark field using  $\eta'$  reflections. Figure 2(d) shows the microstructure developed in the base alloy in the T7751 (RRA) temper. Our studies showed that the RRA treatment (1) coarsened the  $\eta'$ plates in the matrix, (2) nucleated  $\eta(MgZn_2)$  precipitates in the matrix, (3) slightly increased the width of the pfzs formed adjacent to the grain boundary, and (4) coarsened the grain boundary precipitates,  $\eta(MgZn_2)$ . The closely spaced, inter-connected

nature of the grain boundary precipitates in the peak aged alloy is largely replaced by coarser and mostly discontinuous grain boundary precipitates in the RRA sample.



Figure 2: (a) Light micrograph showing grain structure in the peak aged (T651) base alloy. Transmission electron micrographs showing (b) subgrain structure developed in the peak aged base alloy, (c) a uniform and fine distribution of  $\eta'$  precipitates in the matrix, and a network of closely spaced,  $\eta(MgZn_2)$  precipitates at the grain boundary in the peak aged base alloy, and (d) relatively coarser yet a uniform distribution of  $\eta'$  precipitates in the matrix, formation of  $\eta$  (a couple marked by arrows) in the matrix and presence of coarse and mostly discontinuous  $\eta$  precipitates at the grain boundary in the retrogressed and re-aged (T7751) base alloy; (c) & (d) are in close to <112><sub>Al</sub> orientation.

Temper-	0.2%Y.S.	UTS (MPa)	% Elongation	Electrical Conductivity
Direction	(MPa)		-	(%IACS)
T651-L	585	625	12	31.0
T7751- L	575	610	11	36.3
T651-LT	555	600	11	31.0
T7751-LT	560	585	11	36.3

Table 2: Tensile properties of T651 and T7751 treated 7449 sheets (L-longitudinal; LT- long transverse).

Table 2 presents the tensile properties and the electrical conductivity values of the base alloy in T651 and T7751 tempers. It may be noted that the differences between properties

obtained in peak aged and RRA tempers are minimal. On the other hand, the electrical conductivity value in the RRA temper has considerably increased compared to that in the peak aged alloy, suggesting improved SCC (and exfoliation [13]) resistance of the RRA treated alloy. This is supported by the results of our ongoing SCC tests revealing crack growth after 260 h for the peak aged alloy when loaded at stress intensity factor, K, value of 23 MPa m<sup>1/2</sup>. Whilst, no crack growth could be observed even after 1500 h for the RRA treated alloy when loaded at the same K value. These results are consistent with the beneficial effects of the RRA treatment on the microstructure in that it (1) results in the coarsening of matrix precipitates together with the formation of equilibrium precipitates in the matrix causing a change in the dislocation-precipitate interaction from one of shearing to the one of bypassing of precipitates, thus homogenizing the slip, (2) gives rise to coarse, discontinuous grain boundary precipitates, thus providing resistance against stress corrosion cracking, and (3) reduces corrosion potential gradient between the matrix and the grain boundary regions, thus reducing the cause of selective corrosion [1,14].



Figure 3: (a) Total crack length vs. augmented strain (%) profiles for the base alloy, base alloy containing 0.25 wt % Sc, base alloy containing 0.4 wt % Ag and the 7020 Al alloy, as determined by Varestraint test. Light micrographs showing grain structure developed in (b) base alloy weld, and (c) Ag & (d) Sc containing base alloy welds.

Figure 3(a) represents the Varestraint test results showing total crack lengths as functions of augmented strains for the alloys investigated for the weldability tests. It may be noted that the hot cracking susceptibility of the "weldable" alloy 7020 is the least, whilst, under similar test conditions, the 7449 alloy proved to be the most susceptible to hot cracking.

Figure 3(a) further shows that small additions of either Ag or Sc considerably reduce the hot cracking susceptibility of the base alloy. Figures 3(b)-(d) show the development of columnar grain structure in the base alloy weld, whilst equiaxed and fine grain structure in Ag and Sc bearing alloy welds, respectively. The beneficial effects of Sc and Ag on the *refinement of weld grain structure* and hence the weldability are, therefore, similar to those observed for the alloy 7010 containing these elements [6,7]. These results are consistent with the view [15] that a critical minimum amount of strain is required for hot tearing, and accordingly the strain experienced by the liquid film at a grain boundary will have decreased as the number of grains increases, thus lowering the possibility of tearing.

### 4. Summary and Conclusions

A combination of high strength and improved SCC resistance has been achieved in a 7449 Al alloy when processed in the form of sheet in T7751 temper. Small additions of either Sc or Ag to this alloy greatly improve its weldability in terms of improving the resistance of the alloy to hot cracking.

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