

# LIGHTWEIGHT ALLOYS FOR AEROSPACE APPLICATION

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## **ALUMINUM ALLOYS**

Improving Recrystallization Resistance  
in Wrought Aluminum Alloys with  
Scandium Addition

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## **Improving Recrystallization Resistance in Wrought Aluminum Alloys with Scandium Addition**

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### **Abstract**

The 5xxx, 2xxx, and 7xxx aluminum alloys can all benefit from minor additions of Sc to control recrystallization. The addition of Sc rapidly precipitates high volume fraction of homogeneously distributed dispersoids having coherent  $\text{Al}_3\text{Sc}$  ( $\text{L}1_2$ ) structure.  $\text{Al}_3\text{Sc}$  dispersoids impart some improvement in recrystallization resistance to wrought alloys, namely with higher annealing temperatures and volume fractions, compared to alloys employing  $\text{Al}_3\text{Zr}$ . However, the higher coarsening rate of  $\text{Al}_3\text{Sc}$  compared to that of  $\text{Al}_3\text{Zr}$  may limit its usability. When both scandium and zirconium are used in the same alloy  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  dispersoids form, which are more effective than either  $\text{Al}_3\text{Sc}$  or  $\text{Al}_3\text{Zr}$  acting alone. The  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  dispersoid benefits from the rapid precipitation characteristic of Sc in Al, the slow coarsening associated with Zr, and a high volume fraction as an effect of both Sc and Zr. Furthermore the distribution of  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  is homogeneous. It was also established that alloys containing up to 3.5Mg showed improvement in recrystallization resistance with both Sc and Zr present.

## Introduction

### Inhibiting Recrystallization with Dispersoids

The use of dispersoid phases in aluminum alloys is well established as a method for controlling grain growth, recovery, and recrystallization in wrought aluminum alloy products. Several reasons exist for the need to control recrystallization. Strengthening 5xxx alloys by strain hardening is effective if the unrecrystallized wrought state of the alloy can be maintained. In 2xxx alloys recrystallization negatively impacts corrosion resistance. Recrystallization in 7xxx alloys increases weld cracking, decreases fracture toughness, and decreases corrosion resistance.

Dispersoid phases precipitate via a solid state reaction to form a distribution of  $<0.5\mu\text{m}$  diameter particles when the alloy is preheated properly. Dispersoids are known to prevent the motion of subgrain boundaries during annealing, termed "Zener drag", which inhibits recrystallization [1]. Zener's theory shows: increasing volume fraction, decreasing coarsening kinetics, and maintaining a coherent interface between the dispersoid and matrix all increase the resistance to recrystallization. Also, a homogeneous distribution of dispersoids is essential for good recrystallization resistance. Currently the dispersoids  $\text{Al}_{12}\text{Mn}_2\text{Cr}$ ,  $\text{Al}_{20}\text{Mn}_3\text{Cu}_2$ , and  $\text{Al}_3\text{Zr}$  are employed as recrystallization inhibitors in aluminum alloys of the 2xxx and/or 7xxx family. Of these,  $\text{Al}_3\text{Zr}$  is known to be the most potent recrystallization inhibitor [2-4].

The recent literature is replete with references to the commercial potential of adding scandium to wrought aluminum alloys. A recent review contains references to much of this information [5]. Scandium forms  $\text{Al}_3\text{Sc}$  ( $\text{L}_{12}$ ) dispersoids that impart a high degree of recrystallization resistance in many wrought aluminum alloy systems. Al-Mg [6-9] alloys show a large increase in recrystallization resistance when  $\text{Al}_3\text{Sc}$  is present. Al-Mg containing  $\text{Al}_3\text{Sc}$  can sustain a large degree of deformation while retaining an unrecrystallized microstructure. Al-Zn-Mg-(Cu) [10] alloys benefit from  $\text{Al}_3\text{Sc}$  by increased corrosion resistance, decreased weld cracking, and increased fracture toughness [11, 12]. Copper containing alloys, such as Al-Cu [10, 13, 14], and Al-Zn-Mg-Cu benefit from scandium addition although the scandium and copper contents must be closely controlled to avoid deleterious response to alloy performance by formation of W-phase.

### Dispersoid Phase Equilibria

Zirconium, a commonly used dispersoid forming element, is peritectic with Al having a maximum usable solid solution of 0.14wt% which is limited by the break in the liquidus curve. Scandium is known to be eutectic with aluminum having a maximum solid solubility of 0.38wt%Sc and eutectic isotherm at 655 C [15, 16]. Scandium forms coherent stable  $\text{Al}_3\text{Sc}$  dispersoids with aluminum and has an equilibrium  $\text{L}_{12}$  structure.

The literature shows that elemental substitution exists in  $\text{Al}_3\text{Sc}$  and  $\text{Al}_3\text{Zr}$  dispersoids allowing alloys containing both zirconium and scandium to precipitate  $\text{Al}_3(\text{Sc}_{1-x}\text{Zr}_x)$  phase [10]. No reports showing co-existence of independent  $\text{Al}_3\text{Sc}$  and  $\text{Al}_3\text{Zr}$  phases are known. Figure 1 compares the solvus lines of scandium and zirconium in aluminum. Both solvi are in the same temperature and composition region making their respective dispersoids well suited for co-precipitation in aluminum alloys.

The  $\text{Al}_3\text{Sc}$  phase is reported to have rapid precipitation kinetics at 400 C [10, 11] combined with a moderate coarsening rate at this same temperature [17, 18]. The precipitation temperature range of  $\text{Al}_3\text{Zr}$  is similar to  $\text{Al}_3\text{Sc}$ , although nucleation of  $\text{Al}_3\text{Zr}$  is much slower, as shown in

figure 2 [10, 11].

Scandium diffusion in Al has  $D_0=5.31 \text{ cm}^2/\text{s}$  and a high activation energy for diffusion of  $Q=173\text{kJ/mol}$  [19]. For comparison, zirconium has  $D_0=6800 \text{ cm}^2/\text{s}$  and  $Q=241.6 \text{ kJ/mol}$  [20]. Figure 3 compares the diffusivity of Sc and Zr in Al between 350 -600 C. Diffusivity of Sc in Al can be several orders of magnitude greater than Zr, which may lead to rapid coarsening of  $\text{Al}_3\text{Sc}$  compared to  $\text{Al}_3\text{Zr}$ .

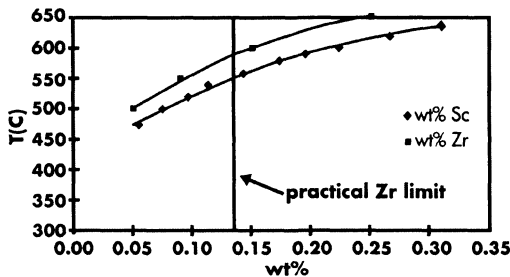


Figure 1 — Solvus lines of Sc and Zr in Al.

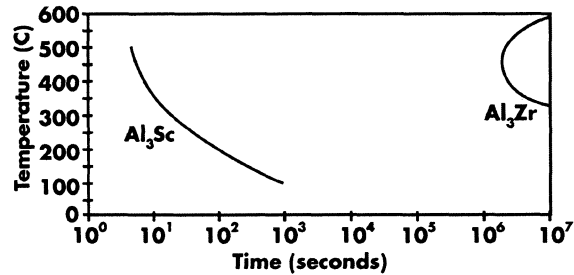


Figure 2 — Precipitation curves for  $\text{Al}_3\text{Sc}$  and  $\text{Al}_3\text{Zr}$ .

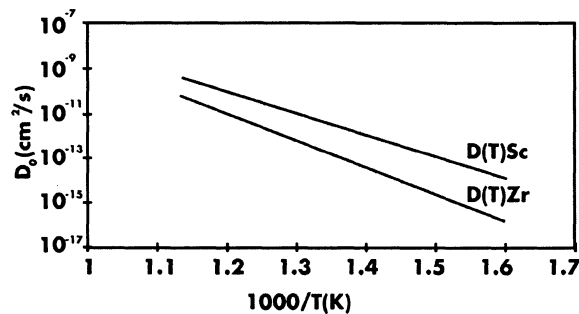


Figure 3 — Diffusivity of Sc and Zr in Al.

Thermodynamics and kinetics of Sc in Al are compatible with current preheating practices of aluminum alloys with and without Zr. Furthermore, rapid precipitation kinetics of  $\text{Al}_3\text{Sc}$ , high volume fraction, and homogeneous distribution of  $\text{Al}_3\text{Sc}$  are favored over the slow precipitation, limited volume fraction, and often heterogeneous distribution of  $\text{Al}_3\text{Zr}$  [21, 22]. However, faster coarsening kinetics of  $\text{Al}_3\text{Sc}$  compared to  $\text{Al}_3\text{Zr}$  limits the direct replacement of Zr by Sc in wrought Al alloys. This makes scandium an attractive addition to Al alloys.

## Experimental

### Ingots

High purity aluminum (99.99%) was provided by Alcoa, portions of an Al-2Sc master alloy from Ashurst Technology, Ltd., Al-6.3Zr from Alcoa, and Al-6Mg from Alcoa were used to cast ingots. The composition of all alloys are reported in weight percent (wt%). Portions of these master alloys were added to clay bound graphite crucibles coated with boron nitride and placed in a resistance furnace at 720 C until completely dissolved. The liquid was stirred and poured into room temperature steel book molds coated with boron nitride to produce ingots having dimensions of 15 x 7 x 2cm.

Ingots follow one of two thermo-mechanical treatments after casting. The following describes the two treatments.

**TMP 1:** The ingot was first preheated from room temperature to 450 C at 50 C/hr and held for 4 hours, followed by quenching in cold water. After preheating the ingot was hot rolled in three stages. The hot rolling temperature, 425 C, was reached within 15 minutes by heating the ingot in a preset resistance furnace. The passes through the rolling mill reduced the ingot to 40, 63, and finally 80% of the original thickness then quenched in cold water. Ingots were heated to the rolling temperature between each pass.

**TMP 2:** Following casting the ingots were immediately cold rolled at room temperature to 50% reduction of the original thickness in one pass through the mill. Next, the ingots were solution heat treated above the solvus temperature for the alloy's composition for one hour then quenched in cold water.

Table 1 describes alloy compositions and thermo-mechanical conditions for alloys used in this research.

Table 1 — Alloy compositions and thermo-mechanical treatments of alloys in this study.

Alloy #	Alloy composition (wt%)				TMP
	Zr	Sc	Mg	Al	
1	0.12	-	-	bal.	TMP 1
2	0.12	-	2	bal.	TMP 1
3	-	0.2	-	bal.	TMP 1
4	-	0.2	-	bal.	TMP 2
5	-	0.3	-	bal.	TMP 2
6	-	0.2	2	bal.	TMP 1
7	0.12	0.2	-	bal.	TMP 1
8	0.12	0.2	-	bal.	TMP 2
9	0.12	0.1	2	bal.	TMP 1
10	0.12	0.2	2	bal.	TMP 1
11	0.12	0.2	3.5	bal.	TMP 1

#### Sample preparation, microscopy, and stereology

Samples for optical microscopy were sectioned from the ingots and mounted in bakelite having the long transverse direction available for inspection. Polishing to a 0.04 $\mu$ m finish was performed in the usual manner. Samples were anodized using a solution of 0.5L H<sub>2</sub>O, 2.3mL fluoboric acid, and 3.5g boric acid with an applied voltage of 12VDC at room temperature. A Reichert-Jung MeF3a with a polarized lens was used for optical microscopy. Recrystallization data were taken from midline in the sample.

TEM samples were prepared by electrothinning of pre-thinned 3mm diameter samples using a Struers Tenupol 3. The thinning solution was 75% methanol, 25% HNO<sub>3</sub> at -40 C using 13VDC. A JEOL 100C operated at 100keV was used for TEM work concerning general characterization, select area diffraction (SAD), and coarsening measurements. Measurements of particles using TEM were calibrated with a magnification calibration standard in the appropriate range. At least 500 dispersoids, photographed in dark field, were used for each datum point to ensure statistical significance of results. Figures that include TEM photographs were recorded in either bright field (BF) or dark field (DF) and indicated accordingly. EDS spectra were collected using a Hitachi HF-2000 TEM operated at 200keV equipped with a Noran EDS detector.

## Results

### Dispersoid Coarsening

#### *Coarsening Rates*

Coarsening data for Al-0.2Sc, Al-0.3Sc, and Al-0.12Zr-0.2Sc alloys having been isothermally aged are presented in  $r^3$  vs.  $t$  form in figure 4. Standard deviation, which was measured experimentally, however, is omitted in figure 4 to increase the clarity of the presented data. Standard deviation is considered in the next paragraph. Coarsening rate of  $\text{Al}_3\text{Sc}$  in Al-0.2Sc is moderate at 450 C but increases dramatically as temperature is increased to 500 C. A strong dependence of coarsening on volume fraction of  $\text{Al}_3\text{Sc}$  is also noted when comparing the trends of Al-0.2Sc and Al-0.3Sc at 450 C. An increase from 0.2Sc to 0.3Sc substantially increases the coarsening rate of  $\text{Al}_3\text{Sc}$ . In Al-0.12Zr-0.2Sc alloys, dispersoid coarsening is significantly slower than Al-Sc alloys, even though the volume fraction of dispersoids present is higher.

A functional relationship exists between average particle radius,  $R$ , and standard deviation,  $s$ , and is presented in figure 5 a,b,c. When plotted as standard deviation vs. average radius, regardless of thermo-mechanical treatment, each alloy composition shows strong linear behavior. A linear regression fit yielded a  $s=mr+b$  relationship for each data set. When the particle has  $r=0$  the standard deviation  $s=0$  so the value of  $b$  in all cases is forced to be zero in the linear regression solution. Thus, the linear regression is reduced to  $s=mr$  form where  $m$  is the slope of the line. The linear fit line, including the equation, for each alloy composition is presented in figure 5d. Note in figure 5a some Al-0.2Sc alloys were aged directly from the 50% cold rolled condition (CR50). Alloys of this study contain minor solute additions. Therefore, the slopes of  $s$  vs.  $r$  should vary only slightly between compositions. The data reflect this trend.

#### *Coherent to non-coherent transformation*

Using SAD it was determined that the dispersoids of this study,  $\text{Al}_3\text{Sc}$  and  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$ , precipitate as coherent  $\text{L1}_2$  spheres. In both cases the distribution of dispersoids is homogeneous as a result of preheating and homogeneous nucleation characteristic of scandium [23]. Figures 6 a, b show well distributed coherent dispersoids in Al-0.2Sc and Al-0.12Zr-0.2Sc in alloys thermo-mechanically treated using TMP 2, respectively.  $\text{Al}_3\text{Sc}$  dispersoids remain coherent up to 25nm radius, at which point coherency to the matrix is lost. The dispersoids lose the Ashby-Brown contrast, indicative of coherent particles.  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  dispersoids coarsen more slowly than  $\text{Al}_3\text{Sc}$  at a given temperature. The size at which the coherent to non-coherent transition occurs for  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$ , if it occurs at all, was not reached in the present coarsening study.  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  dispersoids remain as coherent spheres throughout all the coarsening studies of this research. The largest dispersoids of  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  produced in this study were 28nm radius after 100 hours at 500 C.

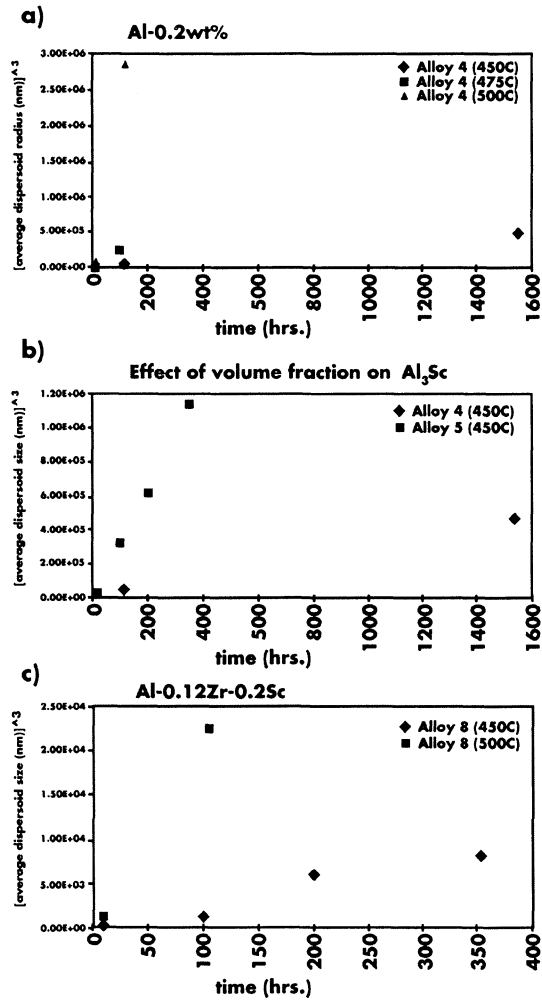


Figure 4 —  $r^3$  vs  $t$  coarsening data. Aging temperature is in parenthesis.

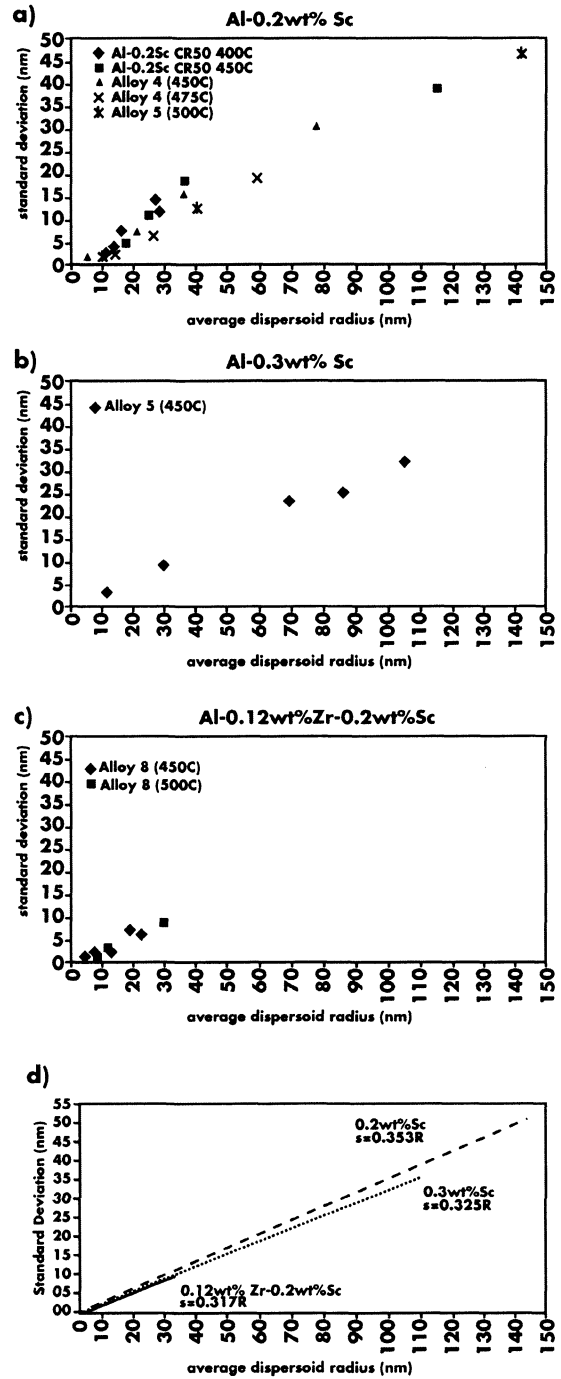


Figure 5 — Standard Deviation vs. Radius. a),b),c) alloys grouped by composition. d) trend lines for a, b, c.

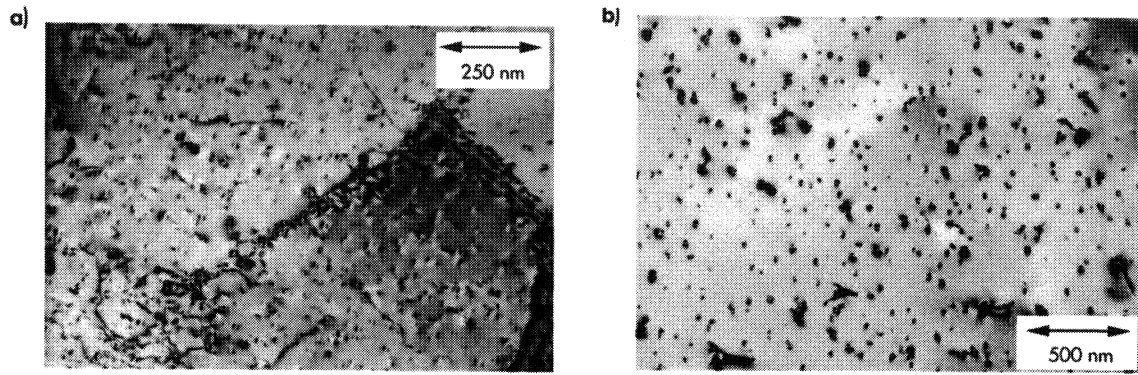


Figure 6 — Well distributed a)  $\text{Al}_3\text{Sc}$  dispersoids in Al-0.2Sc, b)  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  dispersoids in Al-0.12Zr-0.2Sc.

### Recrystallization

#### *Recrystallization in Mg-free wrought Al alloys*

Alloys containing 0.12Zr and/or 0.2Sc with the balance being Al have been preheated and hot deformed to 80% total reduction (TMP 1), as described in the experimental procedure. Following this, the alloys were annealed for 1 hour at 400, 450, 500, 550, or 590 C. Figure 7 presents volume fraction recrystallized for these alloys. Al-0.12Zr has no ability to resist recrystallization even at 400 C, the lowest temperature studied in this work. Al-0.2Sc is able to retard recrystallization up to 550 C, while Al-0.12Zr-0.2Sc completely suppressed recrystallization up to 590 C.

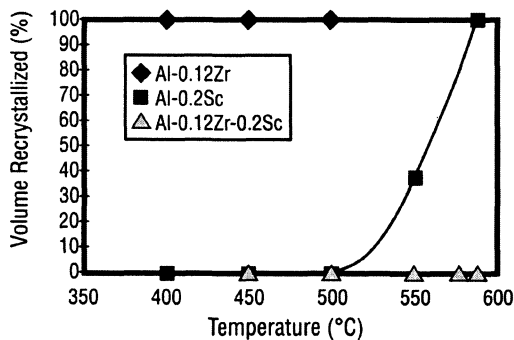


Figure 7 — Volume fraction recrystallized after 1 hour annealing for Al-0.12Zr, Al-0.2Sc, and Al-0.12Zr-0.2Sc. All alloys have been subjected to TMP 1.

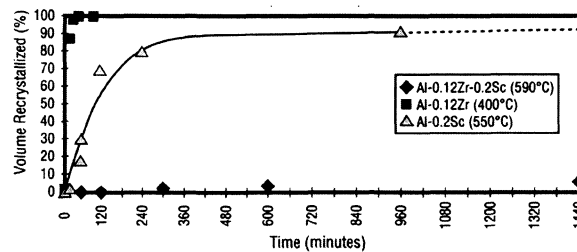


Figure 8 — Volume fraction recrystallized during isothermal annealing up to 24 hours for Al-0.12Zr, Al-0.2Sc, and Al-0.12Zr-0.2Sc. All alloys have been subjected to TMP 1. The annealing temperature for each alloy is indicated.

Annealing was extended to 24 hours for each alloy, namely; Al-0.12Zr at 400 C, Al-0.2Sc at 550 C, and Al-0.12Zr-0.2Sc at 590 C to determine the long term stability of unrecrystallized structures in the Sc containing alloys. Figure 8 presents the findings of the 24 hour annealing study. As presented in figure 7, Al-0.12Zr is fully recrystallized in 1 hour at 400 C. Al-0.2Sc is about 80% recrystallized after annealing at 550 C for 6 hours. However, Al-0.12Zr-0.2Sc remains essentially unrecrystallized throughout the study at 590 C.



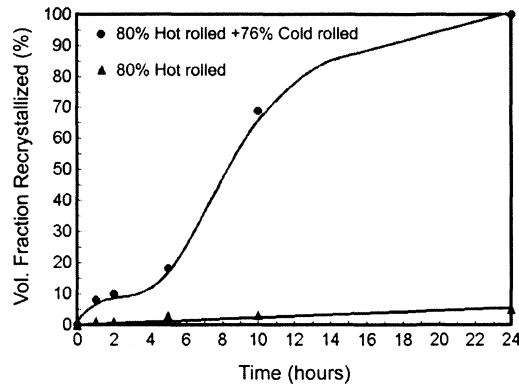


Figure 9 — Volume fraction recrystallized in Al-0.12Zr-0.2Sc after TMP 1 plus an additional 76% cold rolling during isothermal annealing at 590 C.

For the Al-0.12Zr-0.2Sc alloy temperature could no longer be raised in an attempt to induce recrystallization since 590 C is near the melting point of this alloy. An additional 76% cold deformation was added to the Al-0.12Zr-0.2Sc (TMP 1) alloy and recrystallization recorded during a 24 hour isothermal anneal at 590 C. Increasing the level of deformation increases the driving force to recrystallize and may reduce the temperature at which recrystallization begins. Figure 9 presents this data and shows that a full 24 hours at 590 C was necessary to completely recrystallize this highly deformed Al-0.12Zr-0.2Sc alloy.

#### *Recrystallization in Mg-containing wrought Al alloys*

**Ternary Al-Sc-Mg Alloys** The effect of Sc on alloys containing up to 2Mg is illustrated in figure 10. These alloys were preheated and 80% hot rolled (TMP 1) then annealed for 1 hour at 400, 450, 500, or 590 C. Up to 500 C alloys with less than or equal to 2Mg remain unrecrystallized. However, partial recrystallization was observed in both alloys at 550 C, and complete recrystallization at 590 C. Addition of up to 2wt%Mg does not dramatically alter the behavior of Al-0.2Sc when in TMP 1 condition.

#### **Quaternary Al-Zr-Sc-Mg**

**Effect of Mg content** Adding Sc to Al-Zr-Mg has a similar effect on recrystallization as adding Sc to Al-Mg. Al-0.12Zr-0.2Sc alloys containing up to 3.5Mg were preheated, 80% hot rolled (TMP 1), and annealed for 1 hour at 400, 450, 500, 550, or 590 C. Volume fraction recrystallized was recorded and is presented in figure 11. No significant recrystallization occurs when Al-0.12Zr-0.2Sc contains 2Mg. At 3.5Mg partial recrystallization occurs between 400 and 500 C. At 550 C the Al-0.12Zr-0.2Sc-3.5Mg alloy was fully recrystallized within 1 hour.

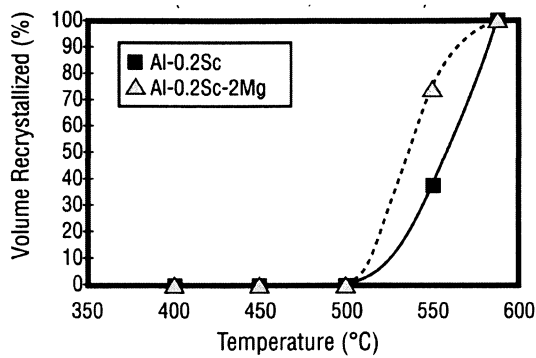


Figure 10 — Effect of 0.2wt%Sc on recrystallization behavior of Al-(0-2)wt%Mg. Alloys were subjected to TMP 1 followed by isothermal annealing for 1 hour.

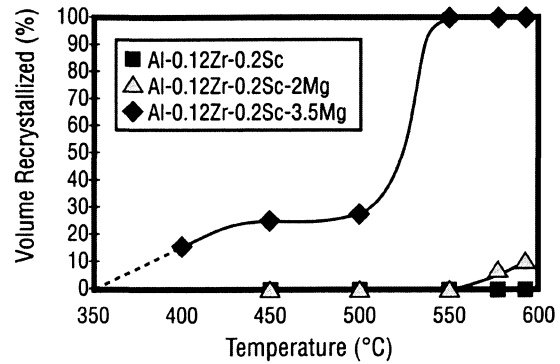


Figure 11 — Effect of 0.12wt%Zr + 0.2wt%Sc on recrystallization behavior of Al-(0-3.5)wt%Mg. Alloys were subjected to TMP 1 followed by isothermal annealing for 1 hour.

**Effect of Sc content** Scandium content less than 0.2wt% still contributed significantly to reducing recrystallization in wrought alloys. The change in recrystallization resistance after 1 hour annealing on Al-0.12Zr-2Mg alloys containing 0, 0.1, and 0.2wt%Sc are presented in figure 12. With no Sc the Al-0.12Zr-2Mg alloy fully recrystallized within 1 hour at 400 C. Adding 0.1Sc to Al-2Mg raised the recrystallization threshold to 550 C. Increasing the Sc content to 0.2wt% raised the temperature at which recrystallization begins to 590 C, near the melting temperature for this alloy.

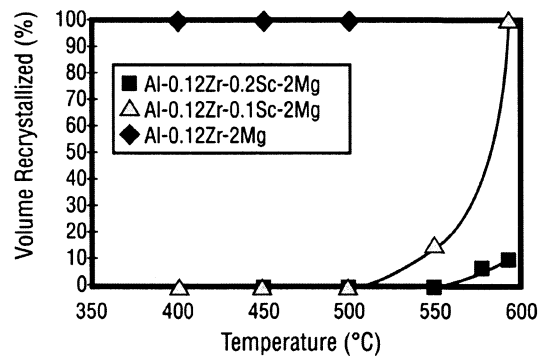


Figure 12 - Effect of Sc level on recrystallization in Al0.12Zr-2Mg alloys. All alloys are subjected to TMP1 followed by isothermal annealing for 1 hour.

### Recrystallized grain morphology

The mechanism of recrystallization is different when comparing alloys containing Zr vs. Sc. Alloys that contain Zr (but no Sc) always recrystallize by nucleation and growth of new grains. When viewed in the optical microscope, recrystallized grains are equiaxed and exhibit no subgrain regions. On the other hand, alloys with Sc (with or without Zr) tend to exhibit extensive recovery. In the Sc containing alloys subgrain bands recover during annealing but do not nucleate new grains. After extensive recovery has taken place the final grains are similar in shape to the wrought grains, elongated in the rolling direction. Figure 13 demonstrates the difference between recrystallization by nucleation and growth in Al-0.12Zr and extensive recovery in Al-0.2Sc.

### Recrystallization vs. Coarsening

A dispersoid's ability to counteract boundary pressure is inversely dependent on its size. At some dispersoid size the particle will become ineffective and recrystallization will occur. Knowing the size limit at which a particular dispersoid becomes ineffective is therefore important when designing recrystallization resistant alloys and their thermo-mechanical treatments. In this study, some alloys of Al-0.2Sc were cold rolled 50% then directly isothermally annealed (below the solvus temperature). Precipitation of  $\text{Al}_3\text{Sc}$  from the wrought solid solution of Al-Sc occurred before recrystallization could take place. During isothermal annealing measurements of dispersoid size and volume fraction recrystallized were made. By comparing dispersoid size and volume fraction recrystallized determination of a size range above which  $\text{Al}_3\text{Sc}$  was no longer effective was made. Figure 14 compares volume fraction recrystallized and dispersoid radius vs. time for annealing performed at 400 and 450 C on 50% cold rolled Al-0.2Sc. In both cases significant recrystallization begins to occur when dispersoids are in the 20-25nm radius range.

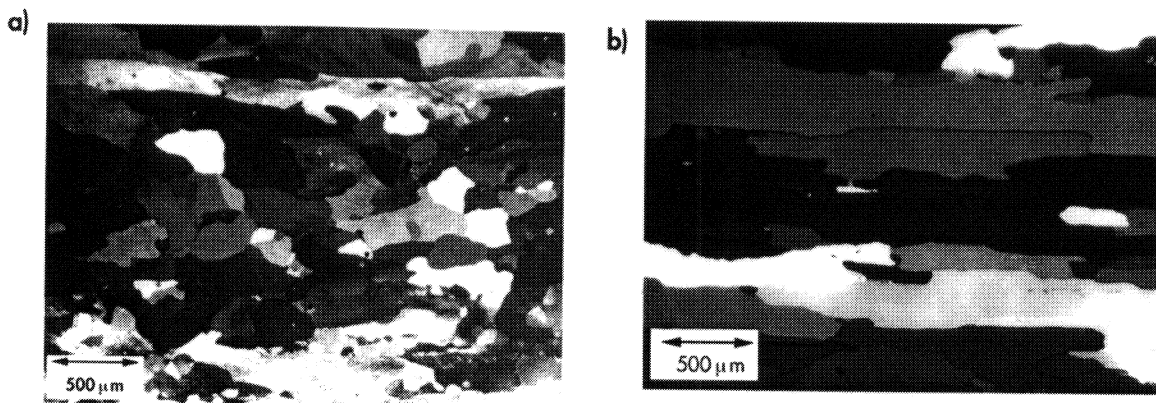


Figure 13 — Typical recrystallization behavior by a) nucleation and growth in Al-0.12Zr and b) extensive recovery in Al-0.2Sc.

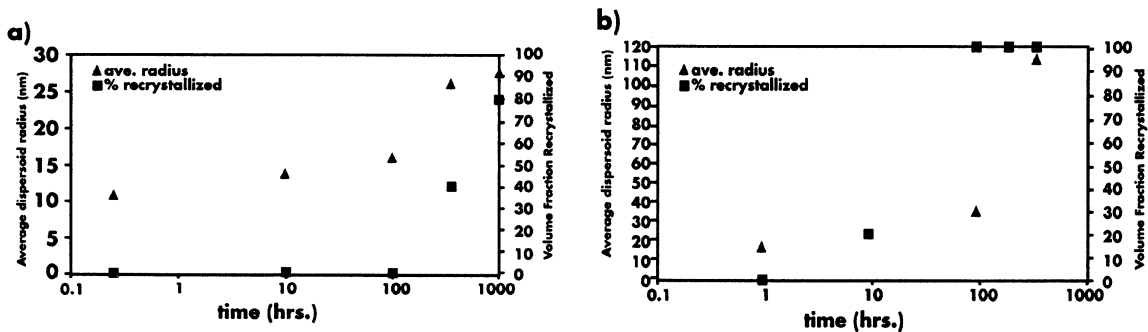


Figure 14 — Comparison of volume fraction recrystallized and dispersoid radius vs. time for 50% cold rolled Al-0.2wt%Sc isothermally annealed at a) 400 C, b) 450 C.

## Discussion

### Dispersoid precipitation behavior

Precipitation of  $\text{Al}_3\text{Sc}$  and  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  from solid solution is rapid. The rapid decomposition of the solid solutions is an effect of Sc addition to the alloy. Cold rolled alloys containing solid solution Sc precipitate  $\text{Al}_3\text{Sc}$  or  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  rapidly enough to retard recrystallization. This observation is not typical of any other dispersoid forming elements. In comparison, the decomposition of Zr forming  $\text{Al}_3\text{Zr}$  requires a significant incubation and requires a significant quantity of other elements to aid in its formation [22, 23]. Moreover, the distribution of  $\text{Al}_3\text{Sc}$  and  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  is always found to be homogeneous, regardless of preheating condition, unlike the heterogeneous nature of  $\text{Al}_3\text{Zr}$ . Assuming the solidus and liquidus arms of the Al-Sc phase diagram are straight then the value of the solute distribution coefficient,  $k$ , is constant and defined as  $C_s^*/C_l^*$ . When calculated,  $k$  has a value of 0.81 for Al-Sc. Given the moderate cooling rate from the melt and a high solute distribution coefficient (indicating possibility for significant coring effects) rejection of Sc to the grain boundary is not severe, as indicated by the homogeneous distribution of  $\text{Al}_3\text{Sc}$  or  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  from both cold rolled and solution heat treated microstructures.

Although  $\text{Al}_3\text{Sc}$  has a very limited, nearly stoichiometric, compositional range it does have solubility for Zr [24]. In fact, when both Zr and Sc are present in the same alloys of this study the resultant dispersoid that forms is always  $\text{Al}_3(\text{Sc}, \text{Zr})$ . In no cases were separate  $\text{Al}_3\text{Sc}$  and/or  $\text{Al}_3\text{Zr}$  phases found when both Sc and Zr were included in an alloy. This was confirmed using numerous EDS scans on dispersoids in alloys containing both Sc and Zr.

### Coarsening Kinetics

Careful attention was given to measurements of dispersoid size for the coarsening study. Coarsening was measured in alloys that were first fully recrystallized and solution heat treated prior to isothermal annealing. This minimized competing and influential mechanisms that could have been acting on dispersoids outside of coarsening. It also ensured that dispersoid forming elements were well distributed before precipitation occurred.

Although not presented here, the coarsening data ( $r^3$  vs.  $t$ ) can also be re-plotted as  $\ln(r)$  vs.  $\ln(t)$ . The slope of this line indicates the limit of the coarsening reaction. If the slope is 1/2, interface kinetics limit coarsening. A slope of 1/3 indicates diffusion limited coarsening. Given this criterion, alloys of this study agree satisfactorily with diffusion limited coarsening.

Several important coarsening trends prevail in this study. Coarsening rate of  $\text{Al}_3\text{Sc}$  increases with increasing temperature from 450 to 500 C. However, comparatively slower coarsening rates were associated with  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  dispersoids in Al-0.12Zr-0.2Sc at 450 and 500 C. Increasing the volume fraction of dispersoids in Al-Sc increases coarsening rate. For the data shown, at a coarsening temperature of 450 C the Al-0.2Sc and Al-0.3Sc equilibrium volume fraction of  $\text{Al}_3\text{Sc}$  is 0.005 and 0.008, respectively.

### Limit of Dispersoid Coherency

Coherency to the Al matrix is preserved with  $\text{Al}_3\text{Sc}$  dispersoids up to 25nm radius. This result compares well with Drits, who showed  $\text{Al}_3\text{Sc}$  maintains coherency up to 19nm radius in Al-0.3at%Sc (Al-0.5wt%Sc) [6]. However, this does not compare well with other data presented by Drits showing complete coherency of  $\text{Al}_3\text{Sc}$  with Al up to 58nm radius [25]. However, that

study used alloys of Al-(4.4-8.3)Mg-(0.25-0.3)Sc. A calculation of relative lattice parameters between Al and  $\text{Al}_3\text{Sc}$  as a function of Mg content was included in previous work [26]. This calculation showed the difference between Al and  $\text{Al}_3\text{Sc}$  lattice parameters decreases with increasing Mg content. This should stabilize  $\text{Al}_3\text{Sc}$  against becoming incoherent by reducing misfit strain, prolonging its ability to remain coherent. Long incubation times necessary for recrystallization to occur in Al-Mg-Sc alloys of this study suggest this may be the case. Therefore, it is plausible that without the presence of magnesium  $\text{Al}_3\text{Sc}$  dispersoids may lose coherency to the matrix at a smaller radius. Further investigation is necessary to confirm this suggestion. It is not known whether change in the lattice parameter of  $\text{Al}_3\text{Sc}$ , and thus dispersoid stability, is affected by substitution of Zr in the dispersoid.

The relatively slow coarsening rate of  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$ , compared to  $\text{Al}_3\text{Sc}$ , suggests slower diffusion to the dispersoid interface and a lattice parameter closer to that of the Al matrix. Also, Sc addition resulted in a uniform distribution of dispersoids. Taking advantage of the slow diffusion, sustained coherency to the matrix, homogeneous distribution, and rapid precipitation kinetics associated with  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  is a useful approach for alloy development when recrystallization control is concerned. Since Zener drag is inversely proportional to particle size, keeping small, well dispersed particles prolongs the effectiveness of the dispersoids to arrest migrating boundaries. Thus,  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  is more effective than  $\text{Al}_3\text{Sc}$  due to a slower coarsening rate. A greater potential volume fraction exists for  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  compared to  $\text{Al}_3\text{Zr}$  and  $\text{Al}_3\text{Sc}$ . Conventional wrought aluminum alloys may also benefit from the increased volume fraction effect not obtainable using Zr or Sc alone, due to solubility limits of each in Al.

#### Recrystallized Grain Morphology

Aluminum alloys containing  $\text{Al}_3\text{Zr}$  dispersoids tend to recrystallize by nucleation and growth of new grains. The recrystallized microstructure of Al-Zr alloys has equiaxed grains. This microstructure suggests recovery may be limited, especially in a heavily deformed matrix, leaving recrystallization the dominant path to reduce stored mechanical energy. However, aluminum alloys having  $\text{Al}_3\text{Sc}$  dispersoids allow substantial recovery to occur, which significantly reduces the driving force available for the nucleation of new grains. Recovery continues in wrought Al-Sc until the final grains have a flat, uniform color when anodized and viewed under polarized light in the optical microscope. The well recovered grains contain no subgrain bands indicating all strain hardening has been eliminated. In effect, the microstructure is comparable to a recrystallized material. A notable difference with Al-Sc alloy "recrystallization" compared to Al-Zr recrystallization is that the final grain structure has grains with an aspect ratio elongated in the rolling direction, reminiscent of the prior rolled grains. When Sc is added to Al-Zr the tendency towards extensive recovery dominates. This may result from the contribution of Sc prolonging the recovery process and increased volume fraction of dispersoids when both Sc and Zr are present, allowing very highly recrystallization resistant microstructure to persist during high temperature annealing.

## Conclusion

Scandium addition to wrought aluminum is compatible with existing thermo-mechanical processes established for these alloys. The addition of Sc rapidly precipitates high volume fraction, homogeneously dispersed dispersoids having coherent  $\text{Al}_3\text{Sc}$  ( $\text{L1}_2$ ) structure.  $\text{Al}_3\text{Sc}$  dispersoids impart some improvement in recrystallization resistance to wrought alloys, namely with higher annealing temperatures and volume fractions, compared to alloys employing  $\text{Al}_3\text{Zr}$ . However, the moderate coarsening rate of  $\text{Al}_3\text{Sc}$  may limit its usability. When both scandium and zirconium are used in the same alloy  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  dispersoids form, which are more effective than either  $\text{Al}_3\text{Sc}$  or  $\text{Al}_3\text{Zr}$  acting alone. The  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  dispersoid benefits from the rapid precipitation characteristic of Sc in Al, the slow coarsening associated with Zr, and a high volume fraction as an effect of both Sc and Zr. Furthermore the distribution of  $\text{Al}_3(\text{Sc}_{1-x}, \text{Zr}_x)$  is homogeneous.

It was also established that alloys containing up to 3.5Mg showed improvement in recrystallization resistance with both Sc and Zr present.

The use of scandium should be considered a potent and compatible solution for increasing recrystallization resistance in some wrought aluminum alloys. With scandium addition, thermal-mechanical processing of certain important industrial alloys can be extended to higher temperatures, longer times, and more severe deformation resulting in semi-finished product having better mechanical properties that remain stable. Scandium is currently an expensive consideration as an alloy additive. If combined with Zr, substantial benefits from Sc can still be realized when used at levels below the maximum solubility.

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