Microstructure and Hardness of Scandium Trialuminide with Ternary Rare-Earth Additions

Yoshihisa Harada^{1, a} and David C. Dunand^{2, b}

¹Advanced Manufacturing Research Institute, National Institute of Advanced Industrial Science and Technology (AIST), 1-2-1 Namiki, Tsukuba, Ibaraki 305-8564, Japan

²Department of Materials Science and Engineering, Northwestern University, 2220 Campus Drive, Evanston, IL 60208-3108, USA

^aharada.y@aist.go.jp, ^bdunand@northwestern.edu

Keywords: Microstructure, Micro-hardness, Lattice parameter, Al₃Sc, Scandium, Rare-earth, SEM/EDS, X-ray diffraction.

Abstract. The microstructure of ternary Al₃(Sc_{1-y}RE_y) intermetallic compounds (where RE is one of the rare-earth elements La, Ce, Nd, Sm, Eu, Yb or Lu), was investigated as a function of RE concentration for 0 < y ≤ 0.75. Alloys with La, Ce, Nd, Sm or Eu additions consist of a L1₂ phase containing a dendritic second phase with D0₁₉ (La, Ce, Nd, Sm) or C11_b (Eu) structure. Alloys with Yb or Lu additions show a single L1₂ phase. The RE solubility limits at 1373 K in the L1₂-Al₃(Sc_{1-y}RE_y) phase are very low for La, Nd, Ce and Eu (0.08-0.41 at.% or y=0.0032-0.0164), low for Sm (3.22 at.% or y=0.1288) and complete for Yb and Lu. The lattice parameter of the L1₂ solid-solution increases linearly with RE concentration and the magnitude of this effect is correlated with the atomic size mismatch between Sc and the RE elements. The Vickers micro-hardness of the L1₂ solid-solution increases linearly with increasing RE concentration.

Introduction

Scandium trialuminide, Al₃Sc, is of interest because of its low density (3.03 Mg m⁻³), relatively high melting point (1593 K) and potentially useful high-temperature strength. Al₃Sc has the cubic L1₂ structure which is expected to show enhanced ductility, because of the high crystallographic symmetry and thus, the high number of possible slip systems. Despite its L1₂ structure, polycrystalline Al₃Sc is brittle at room temperature [1]. Partial substitution of Sc by other elements may increase the ductility of Al₃Sc and also decrease the cost of the alloy. However, very little is known about the alloying behavior of Al₃Sc, with the exception of a recent study investigating the solubility limit of the transition metals (TM) Y, Ti, Zr, Hf, V, Nb and Ta [2]. It was found that solid-solution strengthened Al₃(Sc_{1-y}TM_y) showed improved creep properties as compared to binary Al₃Sc [3]. The alloying behavior of Al₃Sc is also important for the rational design of Sc-containing aluminum alloys, in which small volume fractions of coherent Al₃Sc particles are precipitated [4]. In multi-component alloys, these Al₃Sc precipitates often contain other alloying elements in solid-solution [5], which can improve the strength and coarsening resistance of the precipitates. One study indeed exists, showing that Gd, Ho and Er added to Al-Sc alloys lead to increase hardness [6]. In the present study, the microstructure and hardness of ternary Al₃(Sc_{1-y}RE_y), where scandium is replaced substitutionally by seven rare-earth elements (La, Ce, Nd, Sm, Eu, Yb, Lu), is investigated as a function of alloying element concentration for 0 < y ≤ 0.75.

Experimental Procedures

Button ingots of ternary Al₃(Sc,RE), where RE was chosen from the rare-earth elements La, Ce, Nd, Sm, Eu, Yb or Lu, were cast by non-consumable electrode arc-melting of the pure elements on a water-cooled copper hearth under a purified argon atmosphere. A total of 21 different ternary compositions were prepared with stoichiometry Al₃(Sc_{1-y}RE_y) where y=0.02, 0.04 and 0.1 for RE=Ce

or Nd, $y=0.1$, 0.25 and 0.5 for RE=La, Sm or Eu, and $y=0.25$, 0.50 and 0.75 for RE=Yb or Lu. The resulting ingots showed a weight loss under 1 % as compared with the initial charge. Finally, the ingots were homogenized in argon for 24 hours at 1373 K for La, Ce, Nd, Sm, Eu and Lu, or 96 hours at 1123 K for Yb.

Metallographic preparation consisted of mounting and polishing with SiC paper and 0.05 μm Al_2O_3 . Energy dispersive X-ray spectroscopy (EDS) and scanning electron microscopy (SEM) were then performed to obtain the compositions of the bulk samples and individual phases. Portions of the homogenized ingots were pulverized and the resulting powders, sieved to less than 250 mesh size (75 μm), were analyzed by X-ray diffraction (XRD) using Cu-K α radiation. Lattice parameters were calculated by the least-squares method using the JADE program (Rigaku, Tokyo). After polishing, the Vickers micro-hardness of the L1_2 matrix phase was measured using a 200 g load and an indentation time of 10 sec, with the average of 5 measurements reported.

Results and Discussion

Microstructure. The binary Al_3Sc was found by X-ray diffraction to have the L1_2 structure with a lattice parameter $a = 4.103(1)$ Å, with previous results [2]. The bulk composition was 74.8 at.% Al - 25.2 at.% Sc as measured by wet chemical analysis and this value was used to calibrate the EDS detector [2]. The microstructures of homogenized $\text{Al}_3(\text{Sc}_{1-y}\text{RE}_y)$ alloys for $y=0.02$ (Ce, Nd), 0.1 (La, Sm, Eu) are shown in Figs. 1(a)-(e). These alloys consist of a majority L1_2 $\text{Al}_3(\text{Sc}, \text{RE})$ phase and a low volume fraction of a dendritic $\text{Al}_3(\text{RE}, \text{Sc})$ phase. The latter phases was found by EDS to be $\text{D0}_{19}\text{-Al}_3(\text{La}_{0.96}\text{Sc}_{0.04})$, $\text{D0}_{19}\text{-Al}_3(\text{Ce}_{0.82}\text{Sc}_{0.18})$, $\text{D0}_{19}\text{-Al}_3(\text{Nd}_{0.79}\text{Sc}_{0.21})$ or $\text{D0}_{19}\text{-Al}_3(\text{Sm}_{0.91}\text{Sc}_{0.09})$, respectively,

independent of the overall composition (Table 1). In the case of Eu addition (Fig. 1(e)), the alloy exhibits a L1_2 matrix and low volume fractions of an Al-phase and a second phase found to be $\text{C11}_b\text{-Al}_4(\text{Eu}_{0.975}\text{Sc}_{0.025})$, due to non-existence of a trialuminide phase in the Al-Eu system. As shown in Figs. 1(f) and (g), $\text{Al}_3(\text{Sc}_{1-y}\text{RE}_y)$ alloys for $y=0.25$ (Yb, Lu) show a single L1_2 phase with some Kirkendall

Fig. 1 Typical microstructures of homogenized $\text{Al}_3(\text{Sc}_{1-y}\text{RE}_y)$ where RE as (a) La for $y=0.1$, (b) Ce for $y=0.02$, (c) Nd for $y=0.02$, (d) Sm for $y=0.1$, (e) Eu for $y=0.1$, (f) Yb for 0.25 and (g) Lu for $y=0.25$.

porosities, formed during homogenization of the as-cast structure.

Fig. 2 shows typical X-ray diffraction patterns of $\text{Al}_3(\text{Sc}_{1-y}\text{RE}_y)$ with RE=Yb, Sm or La. The X-ray diffraction patterns evolve with increasing ternary RE concentrations, in agreement with the metallographic (Fig. 1) and EDS observations (Table 1). For Yb additions (Fig. 2(a)), the Bragg peaks of the L1_2 phase are shifted to lower angles with increasing Yb content. For $y = 0.25$, only the L1_2 phase with Al_3Sc composition is present, while for $y = 0.50$ and 0.75, the L1_2 phase with Al_3Yb composition is formed (see peak 221, 310). It is clear that this system shows complete solubility over the composition range studied. In the cases of Lu, we find similarly complete solubility. For Sm additions (Fig. 2(b)), the Bragg peaks of the L1_2 phase are slightly shifted to lower angles with

increasing Sm content. For $y = 0.1$ and 0.25 , only the L_{12} Al_3Sc phase is found, while for $y = 0.50$, the $D0_{19}$ Al_3Sm phase is also visible in the spectra. For all La additions (Fig. 2(c)), the Al_3La phase with $D0_{19}$ structure is found in addition to the L_{12} phase. Its fraction, as inferred from the peak height, increases with increasing La concentrations. The XRD patterns for the $Al_3(Sc, Ce)$, $Al_3(Sc, Nd)$ and $Al_3(Sc, Eu)$

alloys are similar to that of the $Al_3(Sc, La)$ alloy. They confirm that $Al_3(Ce_{1-y}Sc_y)$ and $Al_3(Nd_{1-y}Sc_y)$ have the $D0_{19}$ phase, and $Al_4(Eu_{1-y}Sc_y)$ the $C11_b$ structure.

Fig. 2 Typical X-ray diffraction patterns of $Al_3(Sc_{1-y}RE_y)$ where RE as (a) Yb ($y=0.25, 0.5, 0.75$), (b) Sm ($y=0.1, 0.25, 0.5$) and (c) La ($y=0.1, 0.25, 0.5$).

Table 1 Phase composition (measured by EDS quantitative analysis) and lattice parameters (measured by XRD) for $Al_3(Sc_{1-y}RE_y)$ with RE=La, Ce, Nd, Sm, Eu, Yb and Lu.

Alloys	Structure	Chemical composition [at.%]			Volume of L_{12} f	Lattice parameter [Å]	
		Al	Sc	RE		a	c
Al_3Sc	L_{12}	74.8 (1)	25.2 (1)	-	1.00	4.103 (1)	-
$Al_3(Sc_{0.98}La_{0.02})$	L_{12}	75.08 (11)	24.85 (13)	0.07 (4)	0.87 (6)	4.103 (2)	-
	$D0_{19}$	74.88 (34)	1.01 (7)	24.11 (30)	0.12 (3)	-	-
$Al_3(Sc_{0.78}La_{0.22})$	L_{12}	75.02 (9)	24.90 (15)	0.08 (6)	0.70	4.103 (4)	-
	$D0_{19}$	75.00 (33)	0.67 (8)	24.33 (26)	0.29 (9)	-	-
$Al_3(Sc_{0.5}La_{0.5})$	L_{12}	75.07 (16)	24.81 (16)	0.12 (2)	0.49	4.103 (3)	-
	$D0_{19}$	74.84 (44)	0.92 (2)	24.24 (45)	0.51	-	-
Al_3La [7]	$D0_{19}$	-	-	-	-	6.662	4.609
$Al_3(Sc_{0.98}Ce_{0.02})$	L_{12}	75.02 (11)	24.96 (12)	0.02 (2)	0.96 (2)	4.103(2)	-
	$D0_{19}$	75.09 (22)	4.47 (25)	20.44 (25)	0.03 (7)	-	-
$Al_3(Sc_{0.98}Ce_{0.02})$	L_{12}	75.06 (10)	24.87 (14)	0.06 (6)	0.90 (1)	4.103(2)	-
	$D0_{19}$	75.18 (64)	7.20 (34)	17.62 (44)	0.10 (9)	-	-
$Al_3(Sc_{0.9}Ce_{0.1})$	L_{12}	75.18 (12)	24.74 (13)	0.08 (5)	0.83 (2)	4.103(8)	-
	$D0_{19}$	75.16 (29)	7.10 (5)	17.74 (30)	0.16 (7)	-	-
Al_3Ce [7]	$D0_{19}$	-	-	-	-	6.545	4.609
$Al_3(Sc_{0.98}Nd_{0.02})$	L_{12}	75.15 (14)	24.54 (16)	0.31 (4)	0.97 (9)	4.105(1)	-
	$D0_{19}$	75.10 (41)	5.29 (6)	19.61 (45)	0.02 (1)	-	-
$Al_3(Sc_{0.98}Nd_{0.02})$	L_{12}	74.96 (13)	24.74 (13)	0.30 (2)	0.95 (3)	4.106(2)	-
	$D0_{19}$	75.09 (31)	7.16 (5)	17.74 (35)	0.04 (6)	-	-
$Al_3(Sc_{0.9}Nd_{0.1})$	L_{12}	75.01 (14)	24.63 (20)	0.36 (6)	0.85 (6)	4.106(2)	-
	$D0_{19}$	75.07 (38)	9.22 (10)	15.71 (45)	0.14 (4)	-	-
Al_3Nd [7]	$D0_{19}$	-	-	-	-	6.47	4.603
$Al_3(Sc_{0.98}Sm_{0.02})$	L_{12}	75.11 (11)	22.13 (13)	2.76 (7)	0.92 (7)	4.117(1)	-
	$D0_{19}$	74.94 (49)	2.23 (8)	22.82 (55)	0.07 (2)	-	-
$Al_3(Sc_{0.78}Sm_{0.22})$	L_{12}	74.98 (10)	22.15 (9)	2.87 (3)	0.77 (4)	4.124(1)	-
	$D0_{19}$	75.19 (14)	2.36 (5)	22.45 (13)	0.22 (5)	-	-
$Al_3(Sc_{0.5}Sm_{0.5})$	L_{12}	75.03 (11)	21.75 (9)	3.22 (2)	0.45 (8)	4.130(1)	-
	$D0_{19}$	75.08 (41)	2.85 (9)	22.07 (32)	0.54 (1)	-	-
Al_3Sm [7]	$D0_{19}$	-	-	-	-	6.38	4.597
$Al_3(Sc_{0.9}Eu_{0.1})$	L_{12}	74.79 (10)	24.89 (12)	0.33 (4)	0.90(7)	4.103 (1)	-
	$C11_b$	79.60 (29)	0.50 (8)	19.90 (25)	0.05(5)	-	-
$Al_3(Sc_{0.78}Eu_{0.22})$	L_{12}	74.69 (12)	24.89 (15)	0.41 (10)	0.83(7)	4.103 (4)	-
	$C11_b$	79.59 (21)	0.50 (7)	19.91 (26)	0.13(4)	-	-
$Al_3(Sc_{0.5}Eu_{0.5})$	L_{12}	74.92 (12)	24.78 (17)	0.30 (7)	0.48(5)	4.103 (9)	-
	$C11_b$	79.49 (21)	0.20 (3)	20.31 (24)	0.39(7)	-	-
Al_3Eu [7]	L_{12}	79.7	0.3	0	0.11(8)	-	-
	$C11_b$	-	-	-	-	4.398	11.17
$Al_3(Sc_{0.78}Yb_{0.22})$	L_{12}	74.79 (13)	17.30 (15)	7.91 (4)	0.98(2)	4.116 (3)	-
	L_{12}	74.70 (19)	13.80 (37)	11.5 (22)	0.98(5)	4.131 (5)	-
$Al_3(Sc_{0.25}Yb_{0.75})$	L_{12}	74.80 (58)	6.30 (20)	18.9 (72)	0.95(3)	4.169 (3)	-
	L_{12}	-	-	-	-	4.204	-
$Al_3(Sc_{0.78}Lu_{0.22})$	L_{12}	74.80 (14)	17.30 (13)	7.90 (6)	0.99(2)	4.115 (1)	-
	L_{12}	74.90 (36)	13.20 (24)	11.91 (33)	0.96(1)	4.121 (1)	-
$Al_3(Sc_{0.25}Lu_{0.75})$	L_{12}	74.69 (22)	4.80 (29)	20.50 (33)	0.95(9)	4.143 (2)	-
	L_{12}	-	-	-	-	4.191	-

Solubility Limits for the Rare-Earth Elements. XRD from powder samples and EDS analysis of individual phases from bulk samples can be used to estimate the RE solubility limit in Al_3Sc . The lattice parameters of Al_3Sc with La, Ce, Nd, Sm, Eu, Yb or Lu additions are listed in Table 1. The lattice parameters of binary Al_3La , Al_3Ce , Al_3Nd , Al_4Eu , Al_3Yb and Al_3Lu are experimental values reported by Ref. [7]. Furthermore, Table 1 lists the composition of the L_{12} - Al_3Sc phase measured by EDS as a function of the RE concentration. The lattice constant of the L_{12} phase increases with increasing Nd, Sm, Yb or Lu concentrations, but remain near constant with increasing La, Ce or Eu concentrations. For the $Al_3(Sc, Yb)$ and $Al_3(Sc, Lu)$ alloys, a decrease in Sc content in the L_{12} phase is compensated for by an increase Yb or Lu concentration, while the Al concentration remains constant. We conclude that Yb and Lu replace Sc in the L_{12} phase and make a complete L_{12} solid-solution. In the case of Sm, the lattice constant of the L_{12} phase increases with Sm addition for $y=0.1$ and 0.25 , but the effect becomes less for $y=0.5$. The L_{12} phase exhibits a decrease in Sc content which is again nearly compensated for by an increase in Sm content, with the Al content remaining constant. As shown in Fig. 1(d) and 2(b), the $D0_{19}$ phase appears already at the lowest Sm concentration ($y = 0.1$). This suggests that both the L_{12} and $D0_{19}$ phases have a low solubility limit for ternary elements: 3.22

at.% Sm in the $L1_2$ phase $Al_3(Sc_{0.8712}Sm_{0.1288})$, and 2.85 at.% Sc in the $D0_{19}$ phase $Al_3(Sm_{0.886}Sc_{0.114})$ (Table 1). For the lattice parameter changes of $Al_3(Sc_{1-y}RE_y)$ which $RE=La, Ce, Nd$ and Eu , the lattice constants remain unchanged with increasing RE concentrations in the alloy. Also, RE, Sc and Al contents of the $L1_2$ phase remain constant. For La additions, the $D0_{19}$ phase appears already at the lowest La concentrations ($y \leq 0.1$) (Fig. 2(c)). These $L1_2$ phases have low solubility limit for RE: about 0.12 at.% La in $Al_3(Sc_{0.9952}La_{0.0048})$, 0.08 at.% Ce in $Al_3(Sc_{0.9968}Ce_{0.0032})$, 0.36 at.% Nd in $Al_3(Sc_{0.9856}Nd_{0.0144})$, 0.41 at.% Eu in $Al_3(Sc_{0.9836}Eu_{0.0164})$. Conversely, the solubility limits of Sc in the $D0_{19}$ phase are about 1.01 at.% Sc in $Al_3(La_{0.96}Sc_{0.04})$, 7.2 at.% Sc in $Al_3(Ce_{0.712}Sc_{0.288})$, 9.22 at.% Sc in $Al_3(Nd_{0.6321}Sc_{0.3688})$ and 0.5 at.% Sc in $Al_3(Eu_{0.98}Sc_{0.02})$. Table 1 summarizes the lattice parameters and the EDS quantitative analyses of binary Al_3Sc and the ternary alloys. From the experimentally-measured phase composition, the volume fraction of the $L1_2$ phase was calculated by mass conservation. Reasonable agreement is found with values found independently from image analysis of optical micrographs.

In summary, rare-earth elements La, Ce, Nd, Sm, Eu, Yb or Lu can be incorporated substitutionally in the Sc sublattice of Al_3Sc (Figs. 1, 2 and Table 1). Binary trialuminides of the rare-earth elements Yb and Lu can crystallize in the $L1_2$ structure, while trialuminides of La, Ce, Nd and Sm have long period $D0_{19}$ structures, which are related closely to the cubic $L1_2$ structure. It has been reported that Er, Tm, Yb and Lu trialuminides crystallize in the $L1_2$ structure which can form a complete $L1_2$ solid-solution with Al_3Sc [8]. On the other hand, the solubility limit in the $L1_2$ phase of Gd, Tb, Dy and Ho has been found to be: $Al_3(Sc_{0.85}Gd_{0.15})$, $Al_3(Sc_{0.57}Tb_{0.43})$, $Al_3(Sc_{0.40}Dy_{0.60})$, $Al_3(Sc_{0.28}Ho_{0.72})$ [9]. These previous results and the present ones indicate that the solid solubility in the $L1_2$ phase varies systematically with the location of the rare-earth element in the periodic table. The late RE (Dy, Ho, Er, Tm, Yb and Lu) have a larger solid solubility in Al_3Sc than the early RE (La, Ce, Nd, Sm, Eu or Gd). The late RE are thus good alloying elements candidates to replace substitutionally Sc in $L1_2$ ternary $Al_3(Sc, RE)$ trialuminides.

Lattice Parameter as a Function of the Rare-Earth Elements. Figure 3 show the relationship between the concentration of various RE in the $L1_2$ phase (as measured by EDS) and the lattice parameter of the $L1_2$ phase (as measured by XRD), together with previous data for $RE=Gd, Tb, Dy, Ho, Er$ [9] and Tm [10]. As shown in these figures, there is a linear relationship between lattice parameter and concentration.

For the low-solubility early RE (La, Ce, Eu and Nd), it is difficult to quantify the effect, but the concentration dependence of the $L1_2$ lattice parameter seems to be low, and similar to that for Sm (Fig. 3(b)). As shown in Fig. 3(b), the $L1_2$ lattice parameter for Yb or Lu addition increases linearly with an increasing ternary concentration, similar to previous data for the late RE (Gd, Tb, Dy, Ho, Er [9] and Tm [10]).

Fig. 3 Lattice parameter of the $L1_2$ phase in $Al_3(Sc, RE)$ as a function of (a) $RE=La, Ce, Nd, Sm, Eu$ and (b) $RE=Yb, Lu, Gd, Tb, Dy, Ho, Er, Tm$.

From the linear relationship between concentration and lattice parameter of the solid-solution $L1_2$ phase (Fig. 3), a best-fit slope, da/dc , is determined. Fig. 4 shows that this slope, i.e. the composition dependence of $L1_2$ lattice parameter together with the previously-determined slopes for transition-metals additions [2], correlates reasonably well with the relative atomic radius mismatch, $dR/R = (R_x - R_{Sc})/R_x$, where R is the metallic hard-sphere radius from Ref. [11] (no significant

difference was found in the correlation when ionic Pauling's radii were used). A similar correlation was found by Yamamoto et al. [12] for the L_{12} quaternary phase $\{(AlMn)_3Ti\}_{(1-n)}X_n$, where X is =Zr, V, Ag or, Ga. This correlation can be understood in terms of the size mismatch associated with the replacement of solvent atoms with solute atoms causing a distortion of the lattice. The resulting strain field may allow solute atoms to interact strongly with dislocations, thus affecting the strength of the solid solution, as discussed in the following section.

Fig. 4 Relationship between dR/R and da/dc in $Al_3(Sc, X)$ with X =transition metals (TM) and rare-earth (RE).

Fig. 5 Vickers microhardness of L_{12} phase in $Al_3(Sc, RE)$ as a function of (a) $RE=La, Ce, Nd, Sm, Eu$ and (b) $RE=Yb, Lu$.

Micro-Hardness as a Function of the Rare-Earth Elements. The Vickers micro-hardness of binary Al_3Sc was found to be 121 HV [2]. This value is very close to the value (111 HV) reported by Nic et al. [13], but lower than the value (140 HV) reported by Schneibel et al. [14] for drop-cast Al_3Sc . Fig. 5 shows the micro-hardness of alloyed Al_3Sc as a function of the RE concentration in the ternary L_{12} phase (as obtained by EDS). In Fig. 5(a), the hardness increase for the early RE is modest for Ce, Nd, Eu or Sm, as expected from their low solubility in the L_{12} phase. A much larger hardening effect is associated with La, despite its similarly low solubility: the hardest L_{12} composition $Al_3(Sc_{0.9952}La_{0.0048})$ is more than three times as hard as binary Al_3Sc . However, the presence of finely-dispersed $DO_{19}-Al_3La$ within the L_{12} phase may be the reason for this anomalous hardness increase. Finally, as shown in Fig. 5(b), the late RE Yb and Lu, with complete solubility in Al_3Sc , show a linear hardness increase with RE content, with hardness doubling from Al_3Sc to $Al_3(Sc_{0.25}RE_{0.75})$.

In an earlier paper [3], we investigated the compressive creep behavior for $Al_3(Sc_{0.74}TM_{0.26})$, where TM is one of the transition-metal of Ti, Y, Zr or, Hf. At 873 K, a decrease in creep rates of one order of magnitude was found for Zr and Hf, and two orders of magnitude for Ti and Y. These trends were in reasonably good agreement with the room temperature hardness of the alloys. The exact mechanism responsible for solid-solution strengthening at ambient and elevated temperatures must await precise observations of dislocations in indented or compressively-deformed specimens, as many factors may be operating, e.g., interaction of solute atoms with vacancies and dislocation jogs, segregation on stacking faults and increase of the Peierls stress. Based on the present hardness measurements, the addition of Sm, Yb or Lu to Al_3Sc can be expected to improve the creep resistance of the ternary intermetallic.

Summary

The microstructure of the ternary trialuminide $Al_3(Sc_{1-y}RE_y)$ - where RE is one of the rare-earth elements La, Ce, Nd, Sm, Eu, Yb, Lu - was investigated as a function of concentration in the range $0 < y \leq 0.75$. Cast alloys with La, Ce, Nd, Sm or Eu additions exhibit a microstructure consisting of a L_{12}

solid-solution phase containing a dendritic second phase with the $D0_{19}$ (La, Ce, Nd, Sm) or $C11_b$ (Eu) structure. Alloys with Yb or Lu additions show a single majority $L1_2$ phase. After long-time annealing at xx °C, the solubility limits in the $L1_2$ - $Al_3(Sc_{1-y}RE_y)$ phase was about 0.12 at.% ($y=0.0048$) for La, 0.08 at.% ($y=0.0032$) for Ce, 0.36 at.% ($y=0.0144$) for Nd, 3.22 at.% ($y=0.1288$) for Sm and 0.41 at.% ($y=0.0164$) for Eu. Conversely, Yb and Lu exhibited a complete solid solubility in the $L1_2$ phase. The lattice parameter of the $L1_2$ - $Al_3(Sc_{1-y}RE_y)$ increases linearly with increasing concentration of Nd, Sm, Yb or Lu and remains constant with additions of La, Ce or Eu, due to their very small solubility. The concentration-dependence of the lattice parameter correlates well with the Sc/RE atomic size mismatch. The Vickers micro-hardness of $L1_2$ - $Al_3(Sc_{1-y}RE_y)$ increases linearly with increasing ternary concentration.

Acknowledgments – Partial support from grant DE-FG02-98ER45721 from the US Department of Energy is acknowledged.

References

- [1] J.H. Schneibel and P.M. Hazzledine: J. Mater. Res. Vol. 7 (1992), p. 868.
- [2] Y. Harada and D.C. Dunand: Mater. Sci. Eng. Vol. A329-331 (2002), p. 686.
- [3] Y. Harada and D.C. Dunand: Acta Mater. Vol. 48 (2000), p. 3477.
- [4] C.B. Fuller, D.N. Seidman and D.C. Dunand: Scr. Mater. Vol. 40 (1999), p. 691.
- [5] L.S. Toropova, D.G. Eskin, M.L. Kharakterova and T.V. Dobatkina: *Advanced Aluminum Alloys Containing Scandium* (Gordon and Breach Science Publisher, Amsterdam 1998).
- [6] R.R. Sawtell and J.W. Morris: *Dispersion Strengthened Aluminum Alloys* (TMS, Phoenix, AZ 1988), p. 409.
- [7] *The Powder Diffraction File and Related Products* (ICDD, Newtown Square, PA 2005).
- [8] P. Villars and L.D. Calvert: *Pearson's Handbook of Crystallographic Data for Intermetallic Phases* (ASM, Metal Park, OH 1985).
- [9] O.I. Zalutskia, V.G. kontsevii, N.I. Karamishev, V.R. Ryalov and I.I. Zalutskii: Dopovidi Akad. Nauk Ukr. RSR Vol. A (1970), p.751.
- [10] O.I. Zalutskia, E.E. Cherkashin and I.I. Zalutsky: *Strukt. Faz, Fazovye Prevrashch. Diag. Sostoyaniya Met. Sist.* (Publ. Nauka, Moscow 1974).
- [11] E.T. Teatum, K.A. Gshneidner and J.T. Waber: *Report of Los Alamos Scientific Laboratory of the University of California, LA-2345* (Department of Commerce, Washington, DC 1968).
- [12] Y. Yamamoto, K. Hashimoto, T. Kimura, H. Moriya, N. Nobuki and N. kohno: J. Jpn. Inst. Metals Vol. 62 (1998), p. 844.
- [13] J.P. Nic, S. Zhang and D.E. Mikkola: *High Temperature Ordered Intermetallic Alloys IV* (MRS, Pittsburgh, PA 1990).
- [14] J.H. Schneibel and E.P. George: Scr. Metall. Mater. Vol. 24 (1990), p. 1069.