Residual Strain in Mechanically Ground Powders of Al–Ti Base Ternary and Quaternary Compounds with L1₂ Single-Phase at 1450 K *¹

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We examined the microstructural characters of L1₂-type (Al, X, Y)₃Ti (X, Y=Mn, Cr, Ag, Fe) phase alloys, with compositions selected from the regions connecting ternary L1₂-type (Al, X)₃Ti and (Al, Y)₃Ti single-phase regions in the quaternary phase diagrams using laser microscopy, scanning electron microscopy and electron probe microanalysis. The residual strain, which was induced during milling of the alloys, was measured by examining the X-ray diffraction peak profile of the alloy powders. Most of the (Al, X, Y)₃Ti alloys heat-treated at 1450 K for 24 h exhibited the L1₂ single-phase structure, while a few alloys containing Ag showed two-phase structures consisting of L1₂ and Ag-rich phases. The lattice parameters of (Al, Mn, Ag)₃Ti and (Al, Cr, Ag)₃Ti increased remarkably with increasing Ag content, while those of (Al, Mn, Cr)₃Ti, (Al, Mn, Fe)₃Ti and (Al, Cr, Fe)₃Ti decreased gradually with increasing Cr or Fe content. The change of the lattice parameter with the contents of ternary and/or quaternary elements could be explained by the atomic size effect. The size and fraction of porosity, which were formed during heat treatment at 1450 K for 24 h, depended on the composition of the alloys. The changes of the size and fraction can be explained by the Kirkendall effect, irrespective of alloy systems. The increase of the residual strain by variation of the composition in the Al–Ti–Mn–Ag and Al–Ti–Cr–Ag systems may be explained by the fact that the decrease of size and fraction of the porosity increased the residual strain. However, in the Al–Ti–Mn–Cr, Al–Ti–Mn–Fe and Al–Ti–Cr–Fe systems, the change could be explained by the fact that the residual strain increased as the sum of the contents of Al and Ti decreased.

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1. Introduction

The intermetallic compound Al₃Ti is considered to be of great promise as a high-temperature structural material because of its strength at high temperatures,¹⁾ high melting temperature, low density and excellent oxidation resistance. However, Al₃Ti shows very brittle deformation behavior at ambient temperature because of the low crystal symmetry of the tetragonal DO₂₂ structure. Recently it was shown that the crystal structure of Al₃Ti is transformed to a cubic L1₂-type structure with higher symmetry with the addition of transition metal M.^{2–4)} The ternary L1₂ compounds, Al₃Ti–M, exhibit a wide single-phase region in the ternary Al–Ti–M phase diagram. Thus, many recent efforts have concentrated on ductilizing Al₃Ti by transforming the DO₂₂ structure to the L1₂ structure. Unfortunately, all these transformed cubic compounds, (Al, M)₃Ti, are still very brittle when tension or bending is applied to them. The maximum tensile ductility does not exceed 1% at room temperature, irrespective of M.

In the isothermal phase diagram of the Al–Ti–V system at 1373 K, the DO₂₂-type Al₃Ti phase exhibits unlimited mutual solid solubility with the DO₂₂-type Al₃V phase.⁵⁾ The deformation characteristics of the ternary DO₂₂ phase Al₃(Ti, V) alloys have been studied by Umakoshi *et al.*,⁶⁾ who showed that the compressive strain at room temperature was remarkably affected by the amount of the third element of Ti or V, and that the lattice parameter was also affected by the solute third element. The latter result suggests that an average interatomic bonding force and/or the atomic arrangement in the

DO₂₂ structure might be changed with the addition of solute third atoms of Ti or V. Hence, the change of the stacking fault energy may be anticipated, and then the effect of the third element of Ti or V on compressive strain can be understandable. The L1₂-type (Al, Mn)₃Ti phase also exhibits unlimited mutual solid solubility with the L1₂-type (Al, X)₃Ti phase,⁷⁾ and L1₂-type (Al, Mn, X)₃Ti is obtained. Hence, an effect of an additive element, X or Mn, on the deformation behavior of the L1₂-type (Al, Mn, X)₃Ti may be anticipated.

Recently, we have demonstrated that the ductility of intermetallic compounds of L1₂-type (Al, Mn)₃Ti(V) alloy exhibits a positive linear relationship with the residual strain, as determined by peak profile analysis from the X-ray diffraction of powder specimens after milling of the intermetallic alloy.⁸⁾ X-ray diffraction and the preparation of the intermetallic alloy powder specimen are so easy that the ductility of many intermetallic alloys can be evaluated faster, more cheaply, and with lower labor costs than is possible with mechanical testing such as bending or tensile testing.

On the basis of the above-mentioned information, the microstructure and the residual strain that was induced during milling of the quaternary L1₂-type (Al, X, Y)₃Ti alloys have been examined with varying amounts of X and Y elements. Mn, Cr, Ag and Fe were selected as X and Y, alternately. We describe here the effects of alloy composition, fraction and size of porosity and lattice parameter on the residual strain by considering the positive relationship between residual strain and ductility.⁸⁾

2. Experimental Procedure

The purity of materials used was 99.7% for Mn and Ti, 99.99% for Ag, and 99.9% for Al, Cr and Fe. Button ingots weighing 50 g each were made by nonconsumable electrode

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Fig. 1 Partial isothermal section in the vicinity of L_{12} -type titanium trialuminide phase at 1423 K in a Al–Ti–M (M=Ag, Mn, Cr, Fe) ternary diagram.⁹⁾ The circle and ellipses indicate the L_{12} -type single phase regions.

Fig. 2 The composition diagram of the alloys used in a Al–Ti–Mn–Ag quaternary system. Two gray ellipses exhibit the ternary L_{12} -type (Al, Mn)₃Ti and (Al, Ag)₃Ti single-phase regions. Numerals in the gray ellipses indicate the composition of the ternary alloys used. Solid circles on the linear line connected with the number 2 in both ternary systems exhibit the composition of the quaternary alloys used.

arc-melting under an argon atmosphere on a water-cooled copper hearth. These ingots were remelted three times to promote chemical homogeneity. The weight loss after the arc-melting was less than 0.03% for all the button ingots. The alloy buttons wrapped with tantalum foil were homogenized at 1450 K for 24 h in a vacuum and then cooled in flowing argon gas from 1450 K to room temperature at the cooling rate of about 2.0 K/s.⁷⁾

Figure 1 shows the L_{12} single-phase regions in the isothermal phase diagram of the ternary Al–Ti–M (M=Mn, Cr, Ag, Fe) system at 1423 K.⁹⁾ Here, one circle indicates the L_{12} single-phase region in the Al–Ti–Fe system, and the other three ellipses show the L_{12} single-phase regions in the other ternary systems. The small, middle and large size ellipses correspond to the Al–Ti–Cr, Al–Ti–Mn and Al–Ti–Ag ternary systems, respectively. The ternary alloy composition was se-

Table 1 Nominal compositions of alloys used (mol%).

Alloys	Ti	Al	Mn	
M-1	25.56	64.82	9.62	
M-2	25.14	62.17	12.69	
M-3	26.56	61.88	11.56	
M-4	27.98	61.59	10.43	
M-5	27.56	58.93	13.51	
	Ti	Al	Cr	
R-1	25.38	65.38	9.24	
R-2	24.73	63.84	11.43	
R-3	25.72	63.50	10.78	
R-4	26.71	63.16	10.13	
R-5	26.06	61.62	12.32	
	Ti	Al	Ag	
G-1	25.37	63.98	10.65	
G-2	26.03	60.17	13.80	
G-3	27.67	60.00	12.33	
G-4	29.31	59.83	10.86	
G-5	29.96	56.02	14.02	
	Ti	Al	Fe	
F-1	26.90	66.41	6.69	
	Ti	Al	Mn	Ag
MG-1M	25.51	64.61	7.22	2.66
MG-2M	25.36	61.67	9.52	3.45
MG-3M	26.84	61.41	8.67	3.08
MG-4M	28.31	61.15	7.82	2.72
MG-5M	28.16	58.20	10.13	3.51
MG-1C	25.47	64.40	4.81	5.33
MG-2C	25.58	61.17	6.35	6.90
MG-3C	27.12	60.94	5.78	6.17
MG-4C	28.64	60.71	5.22	5.43
MG-5C	28.76	57.48	6.76	7.01
MG-1G	25.41	64.19	2.41	7.99
MG-2G	25.81	60.67	3.17	10.35
MG-3G	27.39	60.47	2.89	9.25
MG-4G	28.97	60.27	2.61	8.15
MG-5G	29.35	56.75	3.38	10.52
	Ti	Al	Cr	Ag
RG-1R	25.38	65.03	6.93	2.66
RG-2R	25.06	62.92	8.57	3.45
RG-3R	26.20	62.63	8.09	3.08
RG-4R	27.36	62.32	7.60	2.72
RG-5R	27.04	60.22	9.23	3.51
RG-1C	25.38	64.68	4.63	5.33
RG-2C	25.37	62.01	5.71	6.90
RG-3C	26.70	61.75	5.39	6.17
RG-4C	28.01	61.49	5.07	5.43
RG-5C	28.01	58.83	6.16	7.01
RG-1G	25.37	64.33	2.31	7.99
RG-2G	25.70	61.09	2.86	10.35
RG-3G	27.18	60.87	2.70	9.25
RG-4G	28.66	60.66	2.53	8.15
RG-5G	28.99	57.41	3.08	10.52

lected from these four ternary systems. As for the ternary alloys, five compositions were selected from each ellipse region, while only one composition was selected from the circle region for the Al–Ti–Fe system. Thus, a total of sixteen com-

	Ti	Al	Mn	Cr
MR-1M	25.52	64.95	7.22	2.31
MR-2M	25.03	62.59	9.52	2.86
MR-3M	26.34	62.29	8.67	2.70
MR-4M	27.67	61.98	7.82	2.53
MR-5M	27.19	59.60	10.13	3.08
MR-1C	25.46	65.10	4.81	4.63
MR-2C	24.94	63.00	6.35	5.71
MR-3C	26.14	62.69	5.78	5.39
MR-4C	27.34	62.37	5.22	5.07
MR-5C	26.81	60.28	6.76	6.16
MR-1R	25.43	65.23	2.41	6.93
MR-2R	24.83	63.43	3.17	8.57
MR-3R	25.92	63.10	2.89	8.09
MR-4R	27.02	62.77	2.61	7.60
MR-5R	26.44	60.95	3.38	9.23

	Ti	Al	Mn	Fe
MF-1C	26.23	65.61	4.81	3.35
MF-2C	26.01	64.29	6.35	3.35
MF-3C	26.73	64.14	5.78	3.35
MF-4C	27.43	64.00	5.22	3.35
MF-5C	27.23	62.66	6.76	3.35

	Ti	Al	Cr	Fe
RF-1C	26.13	65.89	4.63	3.35
RF-2C	25.82	65.12	5.71	3.35
RF-3C	26.31	64.95	5.39	3.35
RF-4C	26.80	64.78	5.07	3.35
RF-5C	26.48	64.01	6.16	3.35

positions were determined for the ternary alloys.

Figure 2 shows an example of isotherms of the quaternary system, shown in the tetrahedron. Here, the bottom and left side triangles indicate the ternary Al–Ti–Mn and Al–Ti–Ag systems, respectively. The gray ellipse shows the $L1_2$ single-phase region in each ternary system. Numbers in the single-phase region indicate the characteristics of the composition, *i.e.*, the alloys with large numbers have a lower Al content than those with smaller numbers. Figure 2 also shows a designation rule for quaternary alloys, *i.e.*, the alloys with the number 2 on the single-phase region in each ternary system are connected by a straight line, and the line is divided into four portions of equal length. The divided portions indicate the quaternary alloy compositions, as shown in the figure by three solid circles termed MG-2M, MG-2C and MG-2G, respectively. Here, the last letters of M, C, and G indicate the place of the alloy compositions, *i.e.*, the Mn side and center and the Ag side on the connected line, respectively.

Table 1 shows alloy designations and the nominal compositions used in this study. In Table 1, the first letter M, R, G or F in the designation of ternary alloys indicates Mn, Cr, Ag or Fe, respectively. The two letters MG, RG, MR, MF and RF in the quaternary alloys show the combination of the elements in quaternary systems. The number and letter following the hyphen in the designation mean the relative Al content and the position of the alloy composition, respectively, which were previously noted in the designation of the ternary alloy composition, as shown in Fig. 2.

The temperature at which the alloys were homogenized in

this study was 27 K higher than that in the ternary Al–Ti–M isotherm used as a reference.⁹⁾ According to the previous result,⁷⁾ however, all of the homogenized samples at 1450 K exhibited the $L1_2$ single-phase structure. The porosity fraction and the porosity size were measured precisely using laser microscopy.

A bulk sample weighing about 4 g was crushed with a hammer and ground to a powder of under 270 mesh using an agate mortar. The powder size ranged from 270 to 325 mesh, and the five kinds of (111)–(222) peak profiles were used in the X-ray diffraction measurement to obtain the residual strain. X-ray intensity data were collected by a Rigaku RINT-2500 diffractometer system with $\text{CuK}\alpha$ radiation monochromatized with curved graphite at 40 kV and 300 mA. To the analysis of the residual strain, we applied Wilson's analysis method¹⁰⁾ using diffraction profiles from a standard sample. After powdering of the homogenized ternary M-3 alloys, the powder was annealed at 1223 K for 4.8 ks and used as the standard sample. Multiple regression analysis was applied to obtain the relationship between the residual strain and the microstructure by using six independent variables, as described in the following section.

3. Results and Discussion

3.1 Lattice parameter

The shrinkage of the crystal lattice may suggest an increase of the inter-atomic bonding force between adjacent atoms in the crystal lattice. In such a case, a ductility increase of the crystal lattice may be anticipated, because the increase of the bonding force may increase the fracture stress. Therefore, we examined the relationship between the lattice parameter and the amount of the fourth element for the quaternary alloys. The results obtained are shown in Figs. 3(a)–(e). The lattice parameters of the Al–Ti–Mn–Ag and Al–Ti–Cr–Ag alloys increase remarkably with an increase in the amount of Ag, as shown in Figs. 3(b) and (c), respectively. On the other hand, the lattice parameters of the Al–Ti–Mn–Fe and Al–Ti–Cr–Fe alloys decrease with increasing Fe content, as shown in Figs. 3(d) and (e), respectively. However, as shown in Fig. 3(a), the change of the lattice parameter in the Al–Ti–Mn–Cr system is not so obvious as that in the other quaternary systems. According to the table for determining the atomic radius by E. Teatum *et al.*,¹¹⁾ Ag and Fe atoms are the largest and the smallest among the elements used, respectively, and the atomic radius of Mn is not so different from that of the Cr atom. Hence, the changes in the lattice parameters shown in Figs. 3(a)–(e) can be explained from the atomic size effect. We believe, therefore, that the lattice parameter shrinkage occurring with the addition of the Fe atoms does not contribute to the improved ductility of the $L1_2$ phase, because the solution does not change the electron configuration around the constituent atoms and hence does not affect the bonding force between neighboring atoms.

3.2 Porosity fraction and size

In the microstructure of $L1_2$ -type Al_3Ti -based alloys heat-treated at high temperatures, porosity formation due to the Kirkendall effect was reported to be without the exception.¹²⁾ The effects of the amount and size of the porosity on ductility

Fig. 3 Relation between the content of the fourth element, Y, and lattice parameter for the $L1_2$ -type $(Al, X)_3Ti(Y)$ quaternary alloys. The lattice parameters increase with the amount of the fourth element having a larger atomic size than the host elements, Al, Ti and X, while they decrease with the amount of the element Y having a smaller atomic size.

of the $L1_2$ -type alloys at ambient temperature are not small,¹³⁾ so we examined the size and fraction of the porosity of all the alloys.

Figures 4(a)–(e) show the laser optical microstructures of the Al-Ti-Cr-Ag alloys heat-treated at 1450 K for 24 h. The porosity fractions in these microstructures differ from each other clearly, and they tend to decrease with the number in the alloy designation. As the increase of the number in the designation corresponds to the decrease of Al content, as described previously, the porosity fraction of these alloy decreases with decreasing Al content.

Mabuchi *et al.*¹³⁾ reported in detail that the porosity depends on the amount of interdendritic second phase observed in the cast alloys and formed by the Kirkendall effect when the cast alloys are heated for homogenization treatment. We have confirmed a tendency in quaternary alloys⁷⁾ similar to that in the ternary alloys reported by Mabuchi *et al.*

Figures 5(a)–(e) show the microstructures of five different kinds of quaternary alloys with the designation number 2. Figures 5(b) and (c) show the microstructures of alloy heat-treated with Ag. The porosity of Al-Ti-Mn-Ag and Al-Ti-Cr-Ag alloys is much greater than that of the other three alloys. In the two cast alloys, a significant volume of interdendritic second phase is observed. The porosity increased greatly after the homogenizing heat treatment, suggesting that the porosity in the heat-treated Al-Ti-Mn-Ag

Fig. 4 Laser optical micrographs viewing the mechanical-polished surface of the Al-Ti-Cr-Ag alloys heat-treated at 1450 K for 24 h. The slanting ellipse indicates the $L1_2$ single-phase region. The amount of porosity (black dots in the figure) decreases as the Al content decreases.

Fig. 5 Laser optical micrographs illustrating porosity variations with alloy composition. The quaternary alloy containing silver, shown in (b) and (c), exhibits a large amount of porosity, while the other alloys, shown in (a), (d) and (e), show a small amount of porosity.

and Al-Ti-Cr-Ag alloys was generated by the Kirkendall effect. The Al-Ti-Mn-Ag and Al-Ti-Cr-Fe alloys exhibit the largest and the smallest amount of porosity, respectively, suggesting that the amount of porosity depends not only on Al contents but also on the kinds of the ternary and quaternary elements. Therefore, the contributions of the content of Al and the other elements to the porosity fraction should be distinguished. For that reason, we examined the relationship between the porosity fraction and the composition using linear regression analysis.

First, we selected an element content from all of the elements and used it as a variable for the linear regression analysis. In the next step, we selected a pair of elements from all of the elements and used the sum of contents of the two elements as a variable for the linear regression analysis. We used a similar method for the combination of three elements. In each step, Al content was excluded because only three elements in the quaternary system are independent variables and the effect of Al content on the porosity fraction has already been confirmed. The results obtained are shown in Figs. 6(a)–(e). In each part of the figure, the abscissa indicates the variable that exhibits the highest correlation with the porosity fraction among the selected variables. The porosity fraction in Fig. 6(a) decreases gradually as the sum of the contents of three elements, Ti + Mn + Cr, increases. The porosity fraction in Figs. 6(b) and (c) decreases remarkably with an increase in the sum of the contents of two elements, (Ti + Mn) and (Ti + Cr), respectively, while the decrease slows down in the vicinity of 35%. The porosity fraction in Figs. 6(d) and (e), decreases monotonically with the increase of Ti content in both alloy systems. Thus, the suitable variable for the presentation of the porosity fraction depends on the alloy systems. The Ti content is considered to be the factor controlling the porosity fraction in Figs. 6(d) and (e); however, in Figs. 6(a)–(c), the factor is not simple, because these vari-

Fig. 6 Relation between the content and porosity fraction for the Al-Ti-X-Y alloys. The abscissa in the figure indicates the variable of linear regression analysis that exhibits the highest correlation with the porosity fraction among the selected variables. Parts (d) and (e) show the relationship with the content of a single element, parts (b) and (c) show the relationship with the sum of the contents of two elements and part (a) shows the relationship with the sum of the contents of three elements.

ables are the sum of the contents of three or two elements, i.e., Ti + Mn + Cr in Fig. 6(a), Mn + Cr in Fig. 6(b) and Ti + Cr in Fig. 6(c). To simplify the factor, we ignored the variable that exhibited the smallest change in the composition ranges among the three or two elements. Thus, the selected

Fig. 7 Relation between the content and porosity size for the Al–Ti–X–Y alloys. The abscissa in the figure indicates the number of variables as in Fig. 6, that is, parts (d) and (e), the content of a single element, parts (b) and (c), the sum of the contents of two elements and part (a), the sum of the contents of three elements.

variable decreased to Mn + Cr in Fig. 6(a), Mn in Fig. 6(b), and Cr in Fig. 6(c), respectively. That is, we considered the factors controlling the porosity fraction to be mainly the sum of the contents of Mn and Cr in Fig. 6(a), to be mainly Mn content in Fig. 6(b) and to be mainly Cr content in Fig. 6(c). Therefore, the porosity fraction in Al–Ti–Mn–Cr, Al–Ti–Mn–Ag and Al–Ti–Cr–Ag can be considered to depend mainly on the content of Mn and/or Cr, while the porosity fraction basically depends on the content of Al. On the other hand, the porosity fraction in Al–Ti–Mn–Fe and Al–Ti–Cr–Fe depends on the content of Ti and Al. Mabuchi *et al.*⁹⁾ reported for ternary ($\text{Ti}_{27}\text{Mn}_{14}\text{Al}_{59}$) and ($\text{Ti}_{25}\text{Cr}_{14}\text{Al}_{61}$) alloys that the porosity fraction is extremely decreased with increasing Mn or Cr content over 14 mol%. Therefore, the composition dependence of the porosity fraction obtained for Al–Ti–Mn–Cr, Al–Ti–Mn–Ag and Al–Ti–Cr–Ag in this study shows a similar tendency to that in the ternary alloys reported by Mabuchi *et al.* The similarity of the porosity fraction change in the quaternary and ternary alloys may suggest that the porosity formation mechanism due to the Kirkendall effect is mainly controlled even in the quaternary alloys by the amount of Mn and Cr.

To examine the composition dependence of the porosity size, we applied the linear regression analysis in a similar way as we had done in the case of the porosity fraction. The results obtained are shown in Figs. 7(a)–(e). The variables on the abscissa in Fig. 7 are the same as those in Fig. 6, except for Fig. 7(e). However, the variable Cr in Fig. 7(e) is essentially the same as Ti in Fig. 6(e), because the content of Cr in Al–Ti–Cr–Fe alloys is inversely proportional to that of Ti. Hence, the composition dependence of the porosity size in

Figs. 7(a)–(e) is explained by the same factors as those in the case of the porosity fraction. Therefore, we concluded that the same mechanism operates with regard to porosity size as with regard to the porosity fraction.

3.3 Effect of compositions on residual strain

In general, the residual strain stored in crystal lattice during plastic deformation depends on the density of dislocations and vacancies introduced by plastic deformation.¹⁴⁾ Several investigators have proposed that a very small difference between the deformation stress and the fracture stress may be one of the reasons for the poor ductility of ternary L1_2 (Al, X)₃Ti compounds at ambient temperature.^{15–17)} Their proposals regarding the poor ductility may be correct, since they are based on their experimental evidence for the fracture mode change from transgranular to intergranular with increasing temperatures.¹⁷⁾ The dislocation density induced in powder samples during milling of L1_2 (Al, X)₃Ti compounds is less than that in metallic materials, and in turn the change of the dislocation density during preparation processing of powder specimens may be small. Powder samples of this type of compound are easily obtainable by a little deformation. On the other hand, the particle size of powder is less than 50 μm and is considerably smaller than the grain size of about 1000 μm , and thus the influence of intergranular fracture on the residual strain in the powder may not be part of the consideration, and we can conclude that the residual strain is almost proportional to the dislocation density derived from transgranular fracture in L1_2 (Al, X)₃Ti compounds. The present authors had previously reported that the tensile ductility of L1_2 (Al, X)₃Ti compounds at room temperature was proportional to the residual strain in powder specimens of the compounds.⁸⁾ Thus, the residual strain provides important information regarding the tensile ductility of L1_2 (Al, X)₃Ti compounds.

To examine the composition dependence of the residual strain in each alloy system, we applied the linear regression analysis in a similar way as we did in the case of the porosity fraction. The results obtained are shown in Figs. 8(a)–(e). As described previously in Fig. 6, the abscissa also indicates the variable that exhibits the highest correlation with the residual strain. The residual strain exhibits a tendency to increase with increasing variables, irrespective of alloy systems. The variables in Figs. 8(a), (d) and (e) are different from those in (a), (d) and (e) of Figs. 6 and 7, while the variables in Figs. 8(b) and (c) are the same as those in (b) and (c) of Figs. 6 and 7, respectively. That is, the residual strain in Figs. 8(b) and (c) changes with the same variables for the size and fraction of porosity. This result suggests that the residual strain in the Al–Ti–Mn–Ag and Al–Ti–Cr–Ag alloys might be mainly controlled by the amount of porosity. Figures 9(a)–(e) show the relationship between the residual strain in powder specimens and the porosity fraction of the ternary and quaternary alloys used. The increase of porosity fraction decreases the residual strain as shown in Figs. 9(b) and (c). However, these data in Figs. 9(a), (d) and (e) are focused in a circular region. The results indicate that the residual strain in Figs. 9(a), (d) and (e) does not depend on the porosity. Thus, the residual strain in the Al–Ti–Mn–Cr, Al–Ti–Mn–Fe and Al–Ti–Cr–Fe alloys may be controlled by some other factors. The abscissa in Figs. 7(b) and (c) can be replaced by the content of Mn

and Cr, respectively, and the residual strain increases if the variables are increased, as shown in Fig. 8. The increase of the content of the ternary and/or quaternary elements means the decrease of the content of Al and/or Ti. Therefore, one of

Fig. 8 Relation between the content and residual strain for the Al-Ti-X-Y alloys. The abscissa in the figure indicates the variable of linear regression analysis that exhibits the highest correlation with the residual strain among the selected variables. All of the abscissas in the figure show the sum of the contents of two elements.

Fig. 9 Relation between the porosity fraction and the residual strain of the L_{12} -type $(Al, X, Y)_3Ti$ alloys. Plots of data are scattered in an ellipse in (b) and (c), while they gather in a circular region in (a), (d) and (e).

some other factors, which affects on the residual strain, can be considered to be a sum of Al and Ti contents. That is, the decrease of the sum of Al and Ti contents increases the residual strain. The decrease of the sum of Al and Ti contents in the L_{12} alloy may induce a decrease in directionality of atomic bonding between the d-electron of Ti and the p-electron of Al atoms.¹⁸⁾ The important factors for increasing the ductility of the L_{12} alloy are to decrease the sum of the Al and Ti contents and to decrease the fraction and/or the size of porosity, because the tensile ductility of L_{12} $(Al, X)_3Ti$ compounds at room temperature is proportional to the residual strain.

3.4 Residual strain and lattice parameter

If the lattice parameter of L_{12} $(Al, X, Y)_3Ti$ is shrunk by the addition of a quaternary element, the bonding force between near-neighbor atoms in the L_{12} -type crystal lattice may be increased. In that case an increase of the residual strain might be anticipated, because the increase of the bonding force may increase the fracture stress.

Figures 10(a)–(e) show the relationship between the lattice parameter and the residual strain of all the alloys used. Plotted data gather in a circular region in Figs. 10(a), (d) and (e), while they are scattered in all ellipses in Figs. 10(b) and (c). From the former result, it appears that the residual strain in the Al-Ti-Mn-Cr, Al-Ti-Mn-Fe and Al-Ti-Cr-Fe alloys is independent of the lattice parameter. However, from the latter result, it appears that the residual strain in the Al-Ti-Mn-Ag and Al-Ti-Cr-Ag alloys tends to decrease as the lattice parameter increases. As described in §3.1, the change of the lattice parameter was explained by the atomic size effect irrespective of alloy systems, and then the change of the electron configuration around constituent atoms was not considered. Therefore, the reason for the decrease of the residual strain

Fig. 10 Relation between the lattice parameters and residual strain of the L_{12} -type $(Al, X, Y)_3Ti$ alloys. Plots of data gather in a circular region in (a), (d) and (e), while they are scattered in an ellipse in (b) and (c).

with the increase of lattice parameter in the Al–Ti–Mn–Ag and Al–Ti–Cr–Ag alloys should be explained. When the lattice parameters in Figs. 3(b) and (c) are plotted by using the variables shown in Figs. 6(b) and (c), the change of the lattice parameter also exhibits the same tendency as in Figs. 8(b) and (c). That is, the composition dependence of the lattice parameter is quite similar to that of the residual strain. Thus, the decrease of the residual strain with the increase of the lattice parameter can be explained by the composition effect, and the decreasing tendency of the residual strain due to the lattice parameter can be considered to be a pretence behavior. From the above discussion, it can be concluded that the residual strain in all of the alloys used is independent of the lattice parameter.

4. Summary

To estimate the room temperature ductility of $L1_2$ -type single-phase $(Al, X, Y)_3Ti$ ($X, Y = Mn, Cr, Ag, Fe$) quaternary alloys, the effect of alloy composition on the residual strain was quantitatively examined by examining the relationship between the residual strain and the ductility of alloys. The microstructure was examined by X-ray diffraction, scanning electron microscopy and laser optical microscopy. The following conclusions were obtained:

- (1) The change of the lattice parameter with additive elements, X or Y, can be explained from the atomic size effect.
- (2) The change of the fraction and size of the porosity with compositions can be explained by the Kirkendall effect during the homogenization of cast alloys.
- (3) The change of the residual strain with composition in the Al–Ti–Mn–Ag and Al–Ti–Cr–Ag alloys can be explained based on the size and fraction of the porosity. However, in

the Al–Ti–Mn–Cr, Al–Ti–Mn–Fe and Al–Ti–Cr–Fe alloys, the change was not explained by the porosity effect. That is, the residual strain increased with decreasing of the sum of the contents of Al and Ti.

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