Calorimetric Study of Precipitation Process in Al-Mg Alloys

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The precipitation process was examined by measuring specific heat vs temperature (S - T) curves of aged Al-alloys containing 7.6~12.5 at% Mg. Age-hardening was also studied with Al-12.5 at% Mg alloy. (1) Four heat-absorptions ('A', 'B', 'C' and 'D') and one heat-evolution ('P') were observed on the S - T curves: 'A' was attributed to the decomposition of δ'' (short-range ordered Al_3Mg or AlMg) and 'B' to that of δ' (long-range ordered Al_3Mg or AlMg); 'C' and 'D' were due to the redissolution of β' and β phases into the matrix respectively; and 'P' was ascribed to the precipitation of β phase. (2) The precipitation in Al-Mg alloys was a four-stage process below 323 K: Solid solution $\rightarrow \delta'' \rightarrow \delta' \rightarrow \beta' \rightarrow \beta$. (3) Temperature limits for the formation of δ'' and δ' were determined as a function of the Mg content of alloys. Below the solvus temperature of the solute Mg atom, there existed no temperature limit for β' formation. (4) Age-hardening of Al-Mg alloys was associated with the precipitations of δ' , β' and β phases. The formation of δ'' did not cause hardening of the alloys. (5) The δ'' particles were already present in the as-quenched state from 723 K in the alloys containing more than 10.5 at% Mg. (6) The T-T-T diagram determined with Al-12.5 at% Mg alloy shows a double C-curve separated into two parts at 363 K.

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I. Introduction

Aging behavior of binary Al-Mg alloys has been extensively examined by previous workers through various experimental techniques except for thermal analysis. From the results of the early work, Kelly *et al.*⁽¹⁾ concluded that a possible aging sequence is: Solid solution $\rightarrow$ intermediate phase $\beta' \rightarrow$ equilibrium phase β (Al_3Mg_2), where the β phase is a hexagonal or complex fcc structure. The β' phase differs in crystal structure from the β phase, but has a composition close to the β phase⁽²⁾. An increase in electrical resistivity was observed during aging of alloys containing 1.5~7⁽³⁾, 6~12⁽⁴⁾ and 10~12⁽⁵⁾ mass% Mg, at 213~313 K. This resistivity increase was considered good evidence for clustering of solute atoms (or formation of G. P. zones). Pollard⁽⁶⁾ showed from tensile testing that the characteristics of room-temperature aging were consistent with pre-precipitation of G. P. zone, and found that

tensile properties of a room-temperature aged alloy could be restored to the as-quenched properties by a reversion treatment above 323 K. More recent studies with transmission electron microscopy⁽⁷⁾⁽⁸⁾ and diffraction of X-ray⁽⁷⁾, electron⁽⁷⁾⁽⁸⁾ and neutron⁽⁹⁾ have indicated that the zones are internally ordered and have the composition of 50⁽⁷⁾ or 25⁽⁸⁾⁽⁹⁾ at% Mg.

The present paper reports on results of measurements of specific heat versus temperature curves of aged Al-Mg alloys. These curves give information about the thermal stability of various precipitates formed and the precipitation process during aging. The process of age-hardening in the alloys also is described in this paper.

II. Experimental Procedure

Al-alloys containing 7.6~12.5 at% Mg were prepared from 99.99% aluminum and 99.9% magnesium, and were melted in alumina crucibles of high purity under an argon atmosphere. The cast ingots of 25 mm $\times$ 25 mm $\times$ 70 mm were annealed at 723 K for 324 ks under an argon stream for homogenization. Table I shows the results of chemical analysis of the alloy-ingots. The shape of specimens for

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Table 1 Mg content of alloys (at%).

Nominal content	7.6	10.5	12.0	12.5
Analyzed content	7.63	10.45	12.02	12.53

specific heat measurements is of a disc, 30 mm in diameter and 3 mm in thickness.

The specimens were solution-annealed at 723 K for 3.6 ks under an argon stream, quenched into iced water and then aged at 248~263 K in alcohol; at room temperature (R.T.) in air; at 273~473 K (except for R.T.) in silicon oil; and at 523~573 K in a molten salt. After the aging treatment, the specimens were quenched into iced water and then mounted in an adiabatic calorimeter. Measurements of specific heat vs temperature curves were made during continuous heating of the specimens over a range from 283 to 773 K in the same manner as in previous work⁽¹⁰⁾. The rate of heating was about 0.045 K/s.

The micro-vickers hardness was determined on aged specimens of 6 mm × 6 mm × 3 mm with an Al-12.5 at% Mg alloy under a load of 0.245 N.

III. Experimental Results

1. Specific heat vs temperature curves of as-quenched alloys

Alloys, as quenched from 723 K, give specific heat vs temperature curves shown in Fig. 1. About 1.2 ks (20 min) was unavoidably required to begin the heating of alloys for the specific heat measurement after quenching. In Fig. 1, the dotted line is the base line which is specific heat (ΣC_p) vs temperature (T) curves calculated on the assumption of Kopp-Neumann's rule, and the solid curve is the observed specific heat (S) vs T curve. The difference ($\pm \Delta C_p$) between S and ΣC_p cannot be detected on the S - T curve of the 7.6 at% Mg alloy (Fig. 1(a)). In general, $-\Delta C_p$ is due to the evolution of heat associated with precipitate formation and $+\Delta C_p$ due to absorption of heat accompanied by a redissolution of the precipitate into the matrix solid solution.

With the Al-10.5 at% Mg alloy (Fig. 1(b)), a heat evolution 'P' occurs in a temperature range

Fig. 1 Specific heat vs temperature curves of Al-Mg alloys as-quenched from 723 K: (a) Al-7.6 at% Mg alloy, (b) Al-10.5 at% Mg alloy and (c) Al-12.5 at% Mg alloy.

543 to 608 K and is followed by a heat absorption 'D' in a range 608 to 658 K. Another heat absorption 'A' takes place in a low temperature range 293 to 333 K. The reactions 'A', 'P' and 'D' occur also in the Al-12.5 at% Mg alloy (Fig. 1(c)). The energy of a heat-evolution or -absorption is determined from the area enclosed by the S - T curve and the base line. The energy [P] of 'P' is equal to [D] in Fig. 1(b) and (c). Here, we designate the [P]- and [D]-values for the as-quenched alloy as [P]₀ and [D]₀. The position (temperature) of the maximum of $+\Delta C_p$ in a heat-absorption is referred to as a "temperature (T_p) of heat-absorption-peak" and shown by an arrow on the S - T curve. The T_p is 323 K for the 'A' (T_A) and 648 K for the 'D' (T_D), with Fig. 1(c).

2. Specific heat vs temperature curves of aged alloys

The shape of S - T curves of Al-7.6 and -10.5 at% Mg alloys changed so slowly with aging time below 383 K that we employed mainly the Al-12.5 at% Mg alloy in the present work.

Figure 2 shows the S - T curves of Al-12.5 at% Mg alloy aged, (a) at room temperature (R.T.) for 154.8 ks and (b) at 328 K for 1.278 Ms. Heat absorption 'A' occurs also in Fig. 2(a). Another heat-absorption 'B' having the T_B of 363 K is detected in a temperature range 328 to 378 K (Fig. 2(b)). Reactions 'P'

Fig. 2 Specific heat vs temperature curves of Al-12.5 at% Mg alloy aged, (a) at room temperature (R.T.) for 154.8 ks, and (b) at 328 K for 1.278 Ms.

Fig. 3 Specific heat vs temperature curves of Al-12.5 at% Mg alloy aged at 273 K for (a) 7.92 ks, (b) 864 ks and (c) 9.51 Ms.

and 'D' in Fig. 2(a) and (b) are identical to those in Fig. 1(c). Figures 3 and 4 show S - T curves below 413 K, for the alloy aged at 273 or 313 K for times up to 9.51 Ms or 1.188 Ms. Both $[A]$ and T_A increase with aging time at 273 K (Fig. 3). Figure 4 reveals that $[A]$ and T_A

Fig. 4 Specific heat vs temperature curves of Al-12.5 at% Mg alloy aged at 313 K for (a) 10.8 ks, (b) 61.2 ks, (c) 259 ks and (d) 1.188 Ms.

Fig. 5 Specific heat vs temperature curves of Al-12.5 at% Mg alloy aged at 383 K ((a)~(c)), and at 423 K ((d)~(f)). The aging time is shown in the figure.

increase while 'B' appears on aging at 313 K, in such a way that the 'A' overlaps with 'B' and then disappears; T_B seems to be independent of aging time. A comparison of Fig. 2(b) with 4(c) or 4(d) indicates that the T_B is also independent of aging temperature (T_a). Neither 'A' nor 'B' is observed on S - T curves in Fig. 5. Figure 5(a) shows that the temperature range of 'P' for an alloy aged at 383 K for 418 ks (448~573 K) is lower than that for the alloy as-quenched (Fig. 1(c)) (488~583 K). The energy $[P]$, however, is yet nearly equal to $[D]$, which is slightly larger than $[D]_0$. With increasing aging time (Fig. 5(b)), a hump 'q' appears in the 'P', and the $[P]$ decreases while the $[D]$ increases. Another heat absorption 'C' having a T_C of 518 K occurs in an alloy aged at 383 K for a longer time, e.g. 13.6 Ms (Fig. 5(c)). Energies $[C]$ and $[D]$ increase with aging time at 423 K (Fig. 5(e) and (f)). A heat absorption regarded as 'C' having a higher T_C (543 K)

Fig. 6 Specific heat vs temperature curves of Al-12.5 at% Mg alloy aged at 473 K((a)~(c)), and at 523 K((d)~(g)). The aging time is shown in the figure.

occurs on the S - T curve of an alloy aged at 473 K for 50.4 ks (Fig. 6(b)). Temperature T_C is 558 K for an alloy aged at 523 K for 7.6 ks (Fig. 6(e)). The increase in energy with aging time is more predominant in 'D' than in 'C' during 523 K-aging (Fig. 6(e) and (f)). With further aging at this temperature, 'C' disappears (Fig. 6(g)). 'A' similar to that observed in the as-quenched alloy (Fig. 1(c)) occurs again in an alloy aged at 573 K for a short time (0.72 ks) (Fig. 7(a)). $[P]$ is equal to $[D]$, but $[P]$ (or $[D]$) is smaller than $[P]_0$ (or $[D]_0$), for this alloy. Figure 7(b) shows that 'A' disappears when $[P]$ becomes smaller than $[D]$ with aging time. A reaction 'C' having a T_C of 588 K occurs (Fig. 7(c)) and then disappears (Fig. 7(d)) with further aging at 573 K.

3. Temperatures of heat-absorption-peaks

Temperatures T_p 's (T_A , T_B , T_C and T_D) are plotted in Fig. 8 as a function of aging temperature T_a , with Al-12.5 at% Mg alloy. T_B and T_D are independent of T_a , and are denoted by T' and T_e respectively. The T_A increases to 326 K with increasing T_a ; and the T_C to 588 K. The lowest T_A and T_C obtained in the present work are plotted as a function of Mg content of the alloys in Fig. 9 (curves A and C). Curves B and D in Fig. 9 show the plots of the $T_B(T')$ and $T_D(T_e)$ against Mg content of the alloys.

Fig. 7 Specific heat vs temperature curves of Al-12.5 at% Mg alloy aged at 573 K. The aging time is shown in the figure.

Fig. 8 Temperature T_p of peaks of heat-absorptions 'A', 'B', 'C' and 'D' obtained during heating of aged Al-12.5 at% Mg alloy, as a function of aging-temperature T_a . T_e : Solvus temperature of stable precipitate β . T' : Temperature limit for formation of δ' .

Fig. 9 Temperature T_p of heat-absorption-peaks as a function of Mg content of alloys. Δ and ∇ : Temperature limits for formations of δ'' and δ' respectively. $\square$: Redissolution temperature of β' phase formed during 383 K-aging. —: β -solvus due to Eickhoff *et al.* (1953). Curve D: β -solvus due to the present work.

Fig. 10 Age-hardening curves at various temperatures for Al-12.5 at% Mg alloy. $\uparrow$: Time to begin age-hardening. $\downarrow$: Time to attain the maximum hardness.

4. Age-hardening curves

Changes in micro-vickers hardness with aging time is shown for Al-12.5 at% Mg alloy in Fig. 10. Double and single arrows in the figure indicate the times to begin age-hardening and to attain the maximum hardness, respectively. The age-hardening behaviors above 373 K resemble those of Al-10% Mg alloy⁽¹¹⁾. It is found that the hardening behaviors at 313 K and below differ markedly from those above 373 K.

IV. Discussion

1. Interpretation of heat-absorptions and -evolutions

Figure 9 shows that the $T_D(T_e)$ is close to the solvus temperature due to Eickhoff *et al.*⁽¹²⁾ Thus the 'D' can be attributed to the redissolution of the stable precipitate β into the matrix. Since $[P]$ is equal to $[D]$ in the as-quenched alloys (Fig. 1(b) and (c)), 'P' can be associated with the precipitation of β phase. The T_A (318~323 K) shown in Fig. 9 are close to the reversion temperature (323 K) for the precipitates of G.P. zone type⁽⁶⁾, or to the temperature limit (~ 313 K) above which no increase in electrical-resistivity is observed during aging^{(3)~(5)(13)}. Thus 'A' can be attributed to the redissolution of G.P. zones (or clusters) into the matrix. Figure 4 suggests that 'B' is closely related to 'A'. If 'A' is concerned with G.P. zones as described above, then 'A' and 'B' are probably due to the redissolution of two types of G.P. zones such as G.P.1 and G.P.2 respectively designated by Hardy⁽¹⁴⁾ for

the G.P. zones appearing in Al-Cu alloys. This interpretation however leads to conclusions that the reversion temperature of G.P. zones is 363 K instead of 323 K and that even above 323 K the G.P. zones—i.e. G.P.2—are formed during aging. Since no increase in resistivity is observed on aging above 313 K^{(3)~(5)(13)}, the G.P.2 formation is considered to cause no increase in resistivity.

Bernole *et al.*⁽⁷⁾ have obtained transmission electron micrographs of the zones which are of spherical-shape with 8~10 nm in diameter, in Al-10% Mg alloy aged at room temperature for 3 years (94.6 Ms). They concluded from X-ray diffraction study that the zones have the composition of 50 at% Mg(AlMg) and structure of CuAu-type superlattice. Dager *et al.*⁽⁸⁾⁽⁹⁾ have shown from X-ray and neutron diffraction studies that the spherical zones have the structure of Cu₃Au-type superlattice and the composition of 25 at% Mg(Al₃Mg). This morphology (sphere) of the G.P. zones is considered to be unusual, according to Guinier's classification⁽¹⁵⁾: The difference in atomic size between the solute and solvent is about 10% for Al-Mg alloys and is large enough to prefer G.P. zones of plate-shape to sphere. That is, there is some doubt as to whether the precipitates, which are redissolved into the matrix at 323 K and 363 K, are G.P. zones or particles of metastable superlattice (Al₃Mg or AlMg) differentiated from G.P. zones. Moreover, the shape of the *S-T* curves for 'A' (Figs. 3 and 4(b)) differs from that for the redissolution of G.P. zones into the matrix solid solution of Al-Cu alloys⁽¹⁶⁾: This shape for 'A' is not of "λ-type" which is characteristic of the redissolution of G.P. zones based on their solvus curve—also, the λ-type of *S-T* curve is characteristic of disordering of long-range order in three-dimensional superlattice. Bethe's theory⁽¹⁷⁾ indicates that a disordering of short range order in one- or two-dimensional crystals gives a *S-T* curve having a characteristic shape which is designated by the present authors as "inverse λ-type". The shape for 'A' is of inverse λ-type. Furthermore, the change in the shape of *S-T* curves of 'A' and 'B' with aging time, as shown in Fig. 4, is similar to that occurring with a growth of one- or two-dimensional fine

particles (δ'') to large spherical particles (δ') of metastable superlattice Al₃Li precipitated in Al-Li alloys⁽¹⁰⁾. The formation of δ''-particles causes an increase in resistivity, whilst that of δ'-particles causes a decrease in resistivity⁽¹⁸⁾. Thus we prefer an alternative interpretation for 'A' and 'B': These reactions are due to decomposition of the particles of metastable superlattice (Al₃Mg or AlMg) which should be differentiated from G.P. zones—i.e. 'A' is due to the decomposition of one- or two-dimensional fine particles (δ'') of short-range ordered Al₃Mg or AlMg and 'B' to that of large spherical particles (δ') of long-range ordered Al₃Mg or AlMg.

Heat-absorption 'C' shown in Figs. 5~7 should be due to the redissolution of β' particles, because the previous workers have recognized the formation of them besides the β particles during aging at 373~473 K⁽¹⁾⁽¹⁹⁾⁽²⁰⁾. The hump 'q' appearing in Fig. 5(b) is probably due to the effect that a heat-evolution 'P' is counter-balanced by a heat-absorption 'C'.

The *T_p*-value is in order of *T_A* < *T_B* < *T_C* < *T_D*. This result implies that the precipitates increase their thermal stability in the sequence of δ'', δ', β' and β.

2. State of as-quenched alloys

An equality between [*P*]₀ and [*D*]₀ in Fig. 1(b) or (c) indicates that no β-precipitate is present in the as-quenched alloys. The appearance of 'A' reveals that the δ'' particles are already present in the as-quenched alloy, because no heat-evolution other than 'P' is observed. These particles may be produced either on the quenching from 723 K or on aging at room temperature for ~1.2 ks which is unavoidably required to begin the heating of alloys for the specific heat measurements after the quenching from 723 K. This consideration leads us to a supposition that a solute-rich region is present in the solid solution annealed at 723 K and assists the formation of δ'' particles on the quenching from 723 K or on aging at room temperature for ~1.2 ks after the quenching.

3. Aging behavior

Energy [*A*] and temperature *T_A* increase

while $[P]$ holds its equality with $[D]$ during aging at room temperature, as seen from a comparison between Fig. 1(c) and 2(a). This result indicates that the amount and thermal stability of δ'' -particles increase with aging time, while no precipitate other than δ'' forms in the initial stage of aging at room temperature. Similarly, the δ'' particles alone are present in the alloys aged at 273 K for times up to 9.51 Ms (2642 h) (Fig. 3). Figure 4 indicates that the δ'' particles form initially and then disappear by their transition to δ' on aging at 313 K which is lower than the temperature limit (323 ± 3 K) for the formation of δ'' in the Al-12.5 at% Mg alloy. No 'A' but 'B' is observed in Fig. 2(b). This figure shows that the δ' particles alone are present in the alloy aged for 1.278 Ms at 328 K which is slightly higher than the temperature limit for the formation of δ'' .

Figure 5(a) shows that both 'A' and 'B' disappear as long as the equality of $[P]$ with $[D]$ is held. This result indicates that there is no detectable amount of the precipitates in the alloy aged at 383 K for 418 ks. Figure 5(b) shows that $[D]$ is larger than $[P]$, indicating that β phase forms in the alloy aged at 383 K for 1.012 Ms. However, since the hump 'q' appearing in the 'P' is considered to be due to overlap of 'C' on 'P', the difference $[D-P]$ should be written as $[C+D-P]$, which gives the amount of β' plus β phases present in the alloy. Figure 5(c) shows the occurrence of 'C' and disappearance of 'P', taking zero as the $[P]$ -value in $[C+D-P]$. The comparison between Fig. 5(e) and (f) suggests that even when the stable phase β is already formed, the metastable phase β' precipitates newly, in good agreement with the observation by transmission electron microscopy for Al-8 and -10% Mg alloys⁽²¹⁾. T_C of 518 K shown in Fig. 5(c), (e) and (f) (or Fig. 8) is not the solvus temperature of β' , because the β' -precipitate can form during the aging above 518 K, as seen from Figs. 6(e) and 7(c). The fact that the T_C increases from 518 K to 588 K with increasing T_a (Fig. 8) suggests that the β' phase has no definite thermal-stability (i.e. no definite temperature-limit above which no β' -phase forms during aging). On the other hand, the particles of

metastable superlattice have a definite thermal-stability, namely neither δ'' and δ' forms above 363 K (T' in Fig. 8). The comparison between Fig. 5(e) and (f) shows also that the thermal stability of β' -phase (T_p of 'C') is less sensitive to the aging time, in contrast with the highly sensitive stability of δ'' to the aging time (Figs. 3 and 4). Figures 6(e) and (f) show that the increase in amount of precipitate with aging time is more predominant for β than for β' phase at 523 K. The 'A' observed in Fig. 7(a) is due to the decomposition of δ'' particles. It is difficult to expect that the δ'' particles are formed during aging at 573 K which is considerably higher than the temperature limit (323 K) for the δ'' formation. These δ'' particles may be produced either on quenching from 573 K or on aging at room temperature for ~ 1.2 ks which is unavoidably required after the quenching from 573 K. Therefore a solute-rich region similar to that produced in the alloy annealed at 723 K for 3.6 ks is present in the alloy aged at 573 K for a short time (0.72 ks). The small $[P]$ and its equality with $[D]$ in Fig. 7(a) suggest that the presence of the solute-rich region assisting the formation of δ'' particles retards the precipitation of β phase on heating of the alloy. A further study however is necessary to discuss the nature of the solute-rich region in some detail.

The sum $[A+B]$ gives the amount of metastable superlattice particles (δ'' plus δ') present in the aged alloy, and is plotted in Fig. 11 as a function of aging time below 323 K. Also, the $[C+D-P]$ is plotted in Fig. 11 for aging at

Fig. 11 Energy $[A+B]$ or $[C+D-P]$ in aged 12.5 at% Mg alloy, as a function of aging time at various temperatures. $\uparrow$: Time to being age-hardening. $\downarrow$: Time to attain the maximum hardness.

383~573 K. $[C+D-P]$ increases with aging time after an incubation period, whereas $[A+B]$ is considered to increase without any incubation period. Arrows shown in Fig. 11 indicate the times defined in Fig. 10. The incubation time for the increase in $[C+D-P]$ is close to the time to begin age-hardening at 383 K or above. Thus the age-hardening at these temperatures is attributed to the $(\beta'+\beta)$ precipitation. The particles of δ'' alone are present in the alloys aged at room temperature for times—e.g. 154.8 ks (Fig. 2(a))—less than 1.8 Ms (500 h) at which the age-hardening begins. The δ'' formation therefore does not contribute to the age-hardening. The time ~ 108 ks (30 h) to begin the age-hardening at 313 K is considered to be comparable to the time after which 'B' occurs significantly on the S - T curve (Fig. 4). Thus the age-hardening at 313 K is due to the δ' formation.

Figure 12 shows T_a -dependences of times required to attain an initial energy-increase of 2 kJ/kg and to the saturation of the energy-increase during isothermal aging at a given T_a . The times defined in Fig. 10 are also plotted against T_a . Figure 12 reveals that the C-curve for precipitation is divided into two parts located in temperature ranges above and below 363 K. Figure 13 shows an Arrhenius plot of the incubation time (t) for increase in energy $[C+D-P]$ during aging at 383 K and above. This plot gives an apparent activation-energy

Fig. 12 Times required to attain a certain stage of precipitation, as a function of aging-temperature T_a (Al-12.5 at% Mg alloy). ○: Time to attain an initial energy-increase of 2 kJ·kg⁻¹. ×: Time to finish all of energy-increase. ●: Time to begin age-hardening. ⊗: Time to attain the maximum hardness.

Fig. 13 Arrhenius plot of incubation time (t) for increase in energy $[C+D-P]$ during aging of Al-12.5 at% Mg alloy at 383~523 K.

77.5 kJ/mol for the $(\beta'+\beta)$ precipitation, in fairly good agreement with the value 71.1 kJ/mol obtained by the previous workers from measurements of decrease in resistivity for Al-6~10% Mg alloys⁽¹¹⁾.

V. Conclusions

(1) Four heat-absorptions ('A', 'B', 'C', and 'D') and one heat-evolution ('P') are observed on the S - T curves of quenched and aged Al-Mg alloys. The temperature (T_p) of heat-absorption-peaks is 321 K for 'A', 363 K for 'B', 518 K for 'C', and 648 K for 'D', with an Al-12.5 at% Mg alloy: 'A' is attributed to the decomposition of one- or two-dimensional fine particles (δ'') of short-range ordered Al₃Mg or AlMg and 'B' to that of large spherical particles (δ') of long-range ordered Al₃Mg or AlMg; 'C' and 'D' are due to the redissolution of the β' and β phase respectively into the matrix; and 'P' is ascribed to the precipitation of the β phase.

(2) The precipitation in the Al-Mg alloys is a four-stage process below 323 K: Solid solution $\rightarrow \delta'' \rightarrow \delta' \rightarrow \beta' \rightarrow \beta$. However, the β' and β phases precipitate concurrently and then the

β' phase transforms to β in the high concentration (12.5 at% Mg) alloy.

(3) Temperature limits for the formation of δ'' and δ' particles were plotted against the Mg content of alloys (curves A and B in Fig. 9). The δ'' and δ' particles form only at a temperature below 323 and 363 K respectively. Namely, these particles of metastable superlattice (Al_3Mg or AlMg) have a definite thermal-stability. The β' phase however has no definite thermal-stability: below the solvus temperature of the β phase there exists no temperature-range where no β' -phase forms, although the amount of the β' phase formed decreases with the increasing aging-temperature (T_a). With the Al-12.5 at% Mg alloy, the T_p -value of 'C' (thermal stability of the β' phase) increases from 518 K to 588 K with increasing T_a from 383 K to 573 K. The curve C shown in Fig. 9 is not the solvus curve of β' phase, but gives the redissolution temperature of the β' phase formed at a particular T_a (383 K), as a function of Mg content of alloys.

(4) Age-hardening of Al-Mg alloys is associated with the precipitation of δ' , β' and β phases. The formation of δ'' does not contribute to the age-hardening.

(5) The δ'' particles are already present in the as-quenched state from 723 K, with the alloys containing more than 10.5 at% Mg.

(6) The T-T-T diagram for Al-12.5 at% Mg alloy shows a double C-curve separated into two parts by a temperature 363 K.

(7) An Arrhenius plot of incubation time for the precipitation of β' plus β phases gives an apparent activation energy 77.5 kJ/mol in the temperature range 383~523 K.

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Note added in proof: A recent paper [C. Gault, A. Dager and P. Boch: *Acta Met.*, **28** (1980), 51] has concluded that the temperature limit for G. P. zones formation is 353 K for Al-13 at% Mg alloy. This temperature value is close to the T_B of 363 K found for Al-12.5 at% Mg alloy in the present work.