THE EFFECT OF SCANDIUM ON THE AGE-HARDENING BEHAVIOR OF AN Al-Cu ALLOY

Masaru Nakayama[1] and Yasuhiro Miura[2]
1. Kurume National College of Technology, Department of Materials Science and Metallurgical Engineering, Kurume, 830 Japan
2. Kyushu University, Department of Materials Science and Engineering, Fukuoka, 812 Japan

Abstract

Effect of scandium on characterristics of age-hardening and precipitation behaviors in an Al-2.5mass%Cu-0.23maass%Sc ternary alloy was investigated. An Al-2.5mass%Cu alloy with the addition of 0.23mass%Sc was aged at temperature from 423K to 623K following solution treatment and quenching. Also an Al-Cu alloy and an Al-Sc binary alloy were aged under the same condition. In the Al-Cu-Sc ternary alloy, the maximum hardness, Hv≈80, was obtained by aging at 573K for $6x10^3$s. In the Al-Cu binary alloy, more than 10^6s is required at 423K in order to get the same hardeness. Precipitates which occur during aging in the Al-Cu-Sc alloy were θ'-Al_2Cu and Al_3Sc. They precipitate independently in the matrix and don't apparently interact as for heterogeneous precipitation under the experimental condition of the present study. In aging at 573K the age-hardening curve of the Al-Cu-Sc alloy well agree with the sum of the respective curves of the Al-Cu alloy and the Al-Sc alloy.

Introduction

Aluminum with the addition of small amount of scandium is strengthened by the precipitation of coherent spherical $L1_2$ type Al_3Sc intermetallic phase[1],[2]. Al_3Sc precipitates are distributed uniformly and finely in the matrix. The authors have reported that hardness markedly increases by aging at temperature from 523K to 623K in an Al-0.23mass%Sc alloy[3]. Effects of scandium on age-hardening behavior have been reported for Al-Mg alloys[4] and Al-Li alloys[5]. Yield stress and hardness of these alloys increase remarkably with the additon of about 0.2mass%Sc. The addition of scandium to Al-Cu alloys, typical age-hardening aluminum alloys, may produce alloys of higher strength by the synergistic age-hardening effects of copper and scandium. In this study, characteristics of age-hardening and precipitation behaviors of an Al-2.5mass%Cu-0.23 mass%Sc ternary alloy are investigated on the basis of hardness test and TEM observation.

Materials and experimental procedure

An Al-2.6mass%Cu binary alloy, an Al-0.23mass%Sc binary alloy and an Al-2.5mass%Cu-

0.23mass%Sc ternary alloy were prepared by melting aluminum ingot with purity of 99.99%, scandium ingot with purity of 99.9% and Al-40mass%Cu ingot. These alloy ingots were hot-rolled at 673K to plates of 2mm thick. Specimens for aging treatment were cut from the rolled plates. After surface-finished and encapsuled in glass tubes with Ar gas, the specimens were solution treated at 873K for 2.16×10^4s, water-quenched and aged at temperature from 423K to 623K for the time up to 6×10^5s. Vickers hardness was measured and TEM observation was made for the aged specimens using a Hitachi 600AB TEM, operating at 100kV. Scandium content at and around the precipitate sites was investigated for some of the aged specimens by an analytical electron microscope(JEM-200CX equipped with Kevex 7000).

Results and discussion

Age-hardening curves

Figure 1 shows measured age-hardening curves for Al-2.6mass%Cu alloy and Al-2.5mass% Cu-0.23mass%Sc alloy aged for various time up to 6×10^5s at temperature from 423K to 623K. In the Al-Cu alloy, with increasing aging temperature, age-hardening curves move toward low hardness and toward short aging time. On the other hand, in the Al-Cu-Sc alloy age-hardening curves at 423K and 473K are nearly same as the curves of Al-Cu alloy, though at the temperature more than 523K, hardness greatly increases with increasing temperature, and difference between hardness in the Al-Cu alloy and hardness in the Al-Cu-Sc alloy increases as the time proceeds.

Hardness at 573K for 10^5s is about Hv30 in the Al-Cu alloy, while about Hv80 in the Al-Cu-Sc alloy. Hardness at the peak-aging is the highest at 573K and the value is about Hv90. Hardness decreases gradually as the aging time goes, in the over-aging stage. Age-hardening curve of Al-0.23mass%Sc alloy aged at 573K is also shown(Fig.1d). In this binary alloy decrease of hardness is very small in the over-aging stage. The gradual decrease of hardness in this stage is probably due to fact that Al_3Sc precipitates distribute finely in the matrix after long aging time, which is probed by TEM observation. In aging at 623K, the time for peak-aging is shorter and the maximum hardness is lower than in aging at 573K. It can be seen from the above results that effect of scandium on age-hardening of Al-Cu alloy is most remarkable at 573K. In this case, the hardness for peak-aging is Hv90, and the time for peak-aging is 6×10^3s. In Al-2.6mass%Cu alloy, more than 10^6s is required at 423K in order to get the hardness as high as the maximum hardness in the Al-Cu-Sc alloy at 573K. Though higher aging temperature is required in order to obtain higher hardness in the Al-Cu-Sc alloy, from the viewpoint of aging time it takes much shorter time compared with the case of Al-Cu allloy. Thus scandium is one of the effective elements for further strengthening Al-Cu binary alloys.

TEM observation

Figure 2a-d shows bright field TEM images for the stages of under-aging(1.2×10^3s), peak-aging(6×10^3s) and over-aging(6×10^4s) in the Al-Cu-Sc alloy aged at 573K. The direction of electron beam is paralell to [100] of the matrix in all TEM micrographs in Fig.2. Plate-like θ'-Al_2Cu precipitates are present in under-aging stage(Fig.2a). Volume fraction of θ' precipitate increases as aging time goes. At the over-aging stage, spherical precipitates, which show isotropic elastic strain field contrast, are seen to scatter between θ' precipitates(Fig.2c, arrowed). These precipitates are considered to be Al_3Sc. Though Al_3Sc precipitates grow after

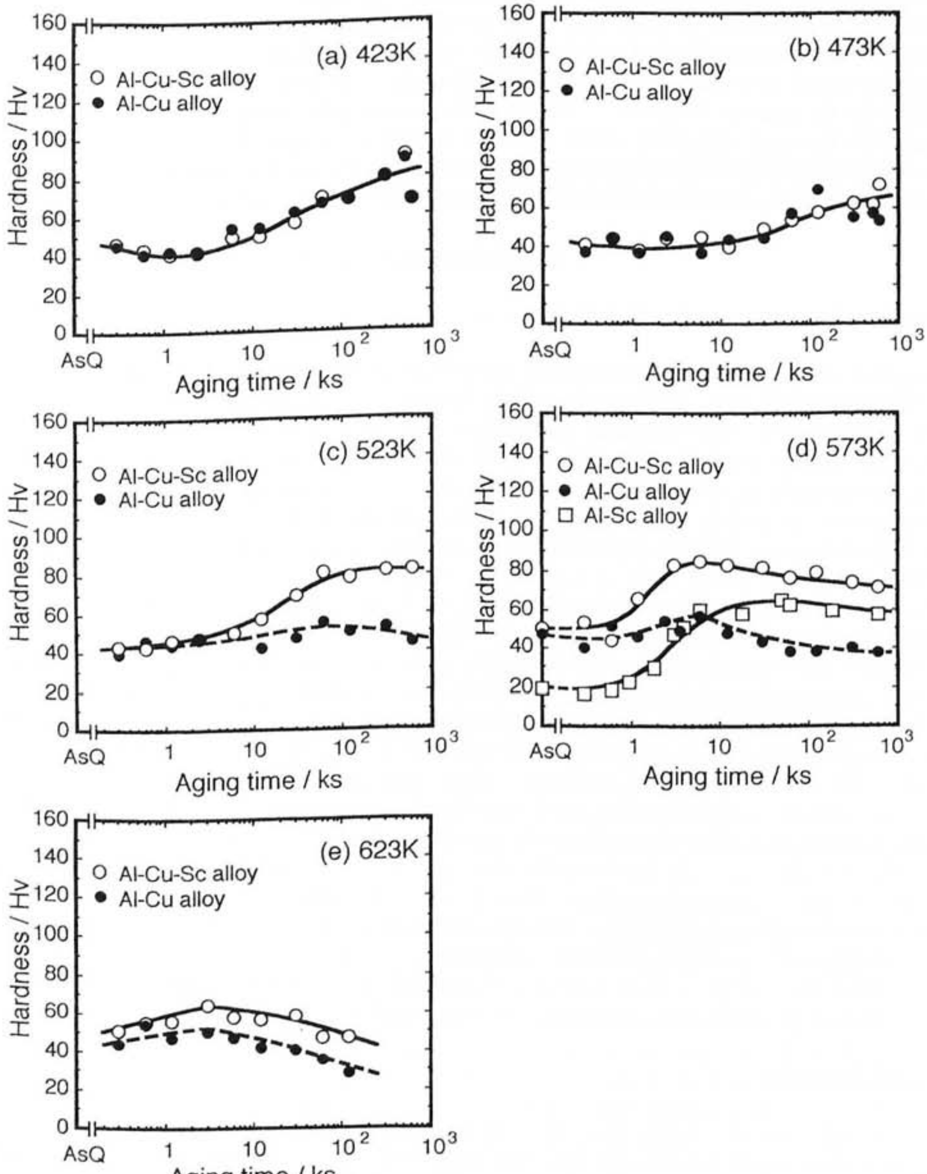

Fig.1 Age-hardening curves of Al-Cu-Sc alloy, Al-Cu alloy and Al-Sc alloy aged at : (a) 423K, (b) 473K, (c) 523K, (d) 573K and (e) 623K.

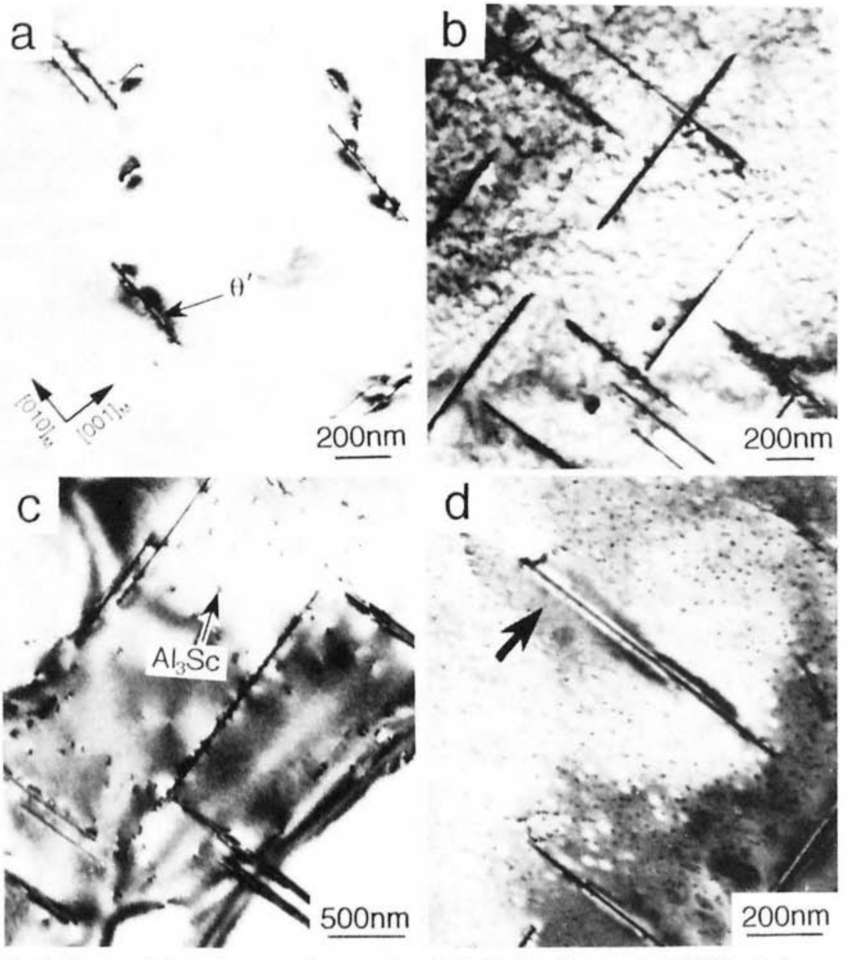

Fig.2 Transmission electron micrographs of Al-Cu-Sc alloy aged at 573K. Aging time is: (a) $1.2\text{x}10^3$s, (b) $6\text{x}10^3$s, (c) $6\text{x}10^4$s, (d) $6\text{x}10^5$s. Electron beam is parallel to [100] of the matrix.

long aging time, their diameter is less than 10 nm and they distribute finely and uniformly in the martix(Fig. 2d).

It is recognized from the selected area diffraction pattern that Al_3Sc precipitates are coherent with the matrix even after aging time for a long time of $6\text{x}10^5$s. There are some areas where Al_3Sc precipitates are not seen in the vicinity of the θ' precipitates(see the arrow in Fig. 2d). Those areas are precipitate free zones(PFZ), which were presumably formed by the migration of Sc atoms in Al_3Sc to θ' precipitates. Formation of PFZs has been reported in some aluminum alloys, such as δ'-Al_3Li-PFZ in the vicinity of β'-Al_3Zr and δ'-PFZ around θ' in Al-Cu-Li-Zr allo alloys[7]. In the present case, little scandium was detected in θ'-plates by the energy dispersion X-ray analysis. This suggests that solubility of scandium in θ' is extremely low. In aging at 623K, Al_3Sc precipitates with isotropic elastic strain fields is observed at peak-aging stage($3x10^3$s), as shown in Fig.3a. Maximum hardness of the alloy aged at 623K is lower than that at 573K. Deformation process in the over-aged aluminum crystals containing a small amount of Al_3Sc precipitates is controlled by Orowan bypassing mechanism[8]. Thus the decrease in hardness at higher aging temperature(623K) is due to increase of interparticle distance by the growth of Al_3Sc and θ' precipitates.

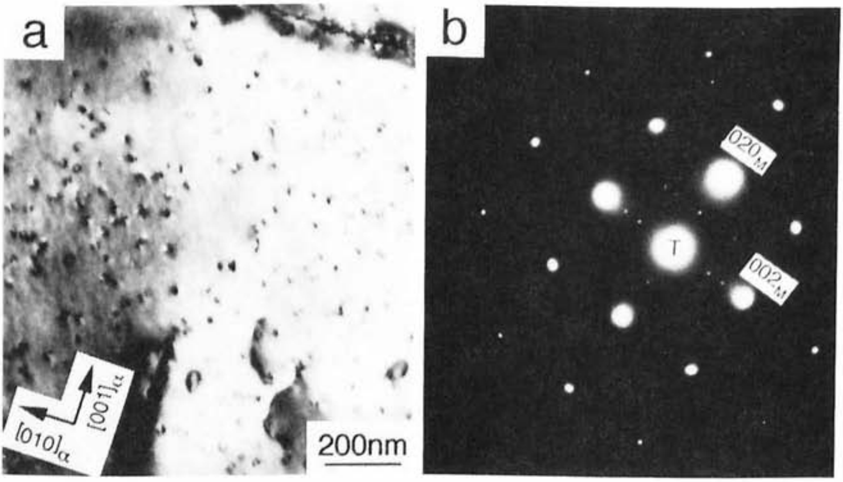

Fig.3 TEM image and selected area diffraction pattern of Al-Cu-Sc alloy aged at 623K for $3x10^3$s. Electron beam is parallell to [100] of the matrix.

Interaction between θ' and Al_3Sc as for heterogeneous precipitation

Generally in some aluminum alloys several kinds of precipitates appear during aging. In such cases, one precipitate is liable to nucleate heterogeneously on the other precipitates. For example in the Al-Cu alloy with a small amount of tin, θ' precipitates heterogeneously on β-Sn which occurs at early stage of aging[9]. In the Al-Cu-Li-Zr alloys, θ' or δ' precipitates preferentially on β' which remains undissolved within the matrix after solution treatment[10], [11]. In the Al-Li alloy with 0.1mass%Zr, δ' nucleates on β' which has precipitated during solution treatment and in case of two step aging of the same alloy, δ' precipitates during the second aging on β' which has occurred in the first aging[12],[13]. In this study θ' and Al_3Sc occur during aging at 573K in the Al-Cu-Sc alloy. Large (100) plane of θ' and Al_3Sc are both coherent with the matrix. Lattice parameters of θ', Al_3Sc and the matrix are 0.404-0.408 nm, 0.410nm and 0.404nm, respectively. Therefore, the misfits at matrix/θ' and matrix/Al_3Sc are 0-0.01and 0.015, respectively. Interface energy is nearly proportional to the square of misfit, if difference of density between precipitate and the matrix is neglected[14]. The energy of interface at matrix/Al_3Sc is hence slightly higher than that at matrix/θ'. Thus the interface energy may decrease when θ' precipitates on Al_3Sc, while it may increase when Al_3Sc precipitated on θ'. Therefore heterogeneous precipitation of Al_3Sc on θ' is difficult to take place against possible heterogeneous precipitation of θ' on Al_3Sc. Spherical dark contrasts are observed in contact with θ'(Fig.2c and Fig.3a). It is considered that these are not precipitates, but probably coherent strain contrast between θ' and the matrix. It can be accordingly concluded that heterogeneous precipitation of θ' on Al_3Sc doesn't occur under the aging condition of this study.

In the age-hardening curves of the Al-Cu alloy, the Al-Sc alloy and the Al-Cu-Sc alloy aged at 573K, time when hardness begins to increase is almost same in each of three alloys(Fig.1). It is considered from this result that θ' and Al_3Sc begin to precipitate nearly at the same time. In order for one phase to precipitate heterogeneously on the other, it is necessary that the other phase should grow up to a certain critical size to provide the energy for the nucleation of one phase. In the present Al-Cu-Sc alloy, Al_3Sc particles are very fine even at the over-aging stage, probably smaller than the critical size. This could be the reason why the precipitation of θ' on Al_3Sc have not been observed.

Large Al_3Sc particles(~100 nm in diameter), probably undissolved during solution treatment are rarely observed, as shown in Fig.4. θ'-plates precipitate preferentially on spherical Al_3Sc in contact with each other with an orientation relationship, $(100)_M // (100)_{\theta''}$, $[010]_M // [100]_{\theta''}$. The presence of the large Al_3Sc was onfirmed in an as-quenched specimen by TEM observation.

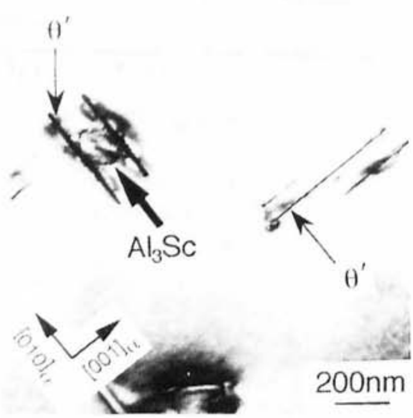

Fig.4 TEM image showing heterogeneous θ' precipitation on the Al_3Sc particle which have existed before aging in the Al-Cu-Sc alloy aged 573K for $1.2x10^3$s. Electron beam direction is parallel to [100] of the matrix.

Additivity of age-hardening curves

Age-hardening curves at 573K in Fig.1d are reproduced in Fig.5. In this figure, the curve A+B presents simple addition of the age-hardening curves for each binary alloy (curve A in the Al-Cu alloy and curve B in the Al-Sc alloy). It is in good agreement with the curve of the Al-Cu-Sc alloy in aging at 573K. When two binary phases independently precipitate in some Al-X-Y ternary alloys, for example Al-Cu-Si alloy, age-hardening curve of ternary alloy is approximately expressed as simple sum of curves of Al-X and Al-Y binary alloy[15]. This is consistent with the present TEM observation that, in 573K aging, only θ' and Al_3Sc phases have been identified and the binary precipitates are distributed independently in the matrix.

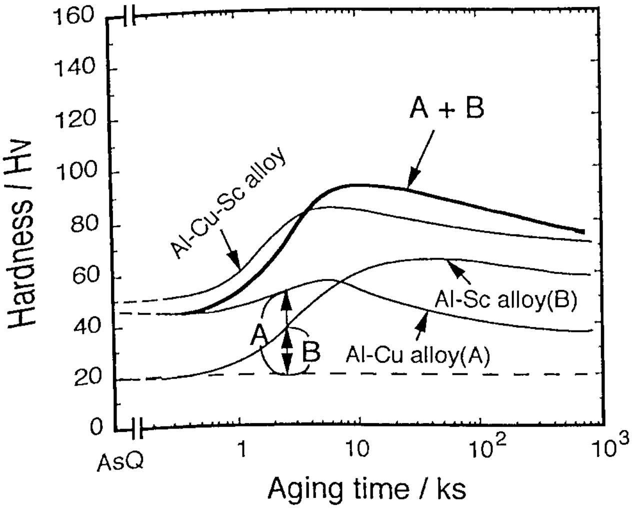

Fig.5 Comparison of the age-hardening curve of Al-Cu-Sc alloy with the sum of the age-hardening curves of Al-Cu alloy and Al-Sc alloy aged at 573K.

Summary

The effect of scandium on age-hardening and precipitation behavior of an Al-Cu alloy were investigated on the basis of the results from hardness test and TEM observation using Al-2.6 mass%Cu alloy, Al-0.23mass%Sc alloy and Al-2.5mass%Cu-0.23mass%Sc alloy. There are no any substantial differences in the age-hardening behavior between the Al-Cu binary alloy and the Al-Cu-Sc ternary alloy at temperatures below 473K. The effect of scandium on age-hardening is most remarkable at 523K and the maximum hardeness is obtained at 573K for $6x10^3$s. In aging at 573K the age-hardening curve of the Al-Cu-Sc ternary alloy agrees with the simple sum of the curves of the Al-Cu and Al-Sc binary alloy. Precipitates which occur during aging in the Al-Cu-Sc alloy are θ'-Al_2Cu and/or Al_3Sc. They precipitate independently in the

matrix and don't apparently interact as for heterogeneous precipitation under the experimental condition of the present study.

References

1. N. Blake and M.A. Hopkins, J. Mater. Sci. 20, (1985), 2861.
2. N. Sano et al., J. Japan Inst. Light Metals 39, (1989), 444.
3. M. Nakayama et al., Abstracts of Japan Inst. Light Metals Spring Meeting, (1991), 273.
4. P. Villars and L.D. Calvert, Person's Handbook of Crystallographic Data for Intermetallic Phases (Metals Park, OH: American Society for Metals, 1985).
5. M.E. Drits et al., Phys. Dokl. 26, (1981), 344.
6. S. Saji et al., Abstracts of Japan Inst. Light Metals Spring Meeting, (1990), 87.
7. M.H. Tosten et al., Metall. Trans., 19A, (1988), 51.
8. Y. Miura et al., Proc. 7th JIM Int. Symp. on Aspects of High Temperture Deformtion and Fracture in Crystalline Materials, ed. Y. Hosoi et al. (Nagoya, Japan: Japan Inst. Metals, 1993), 255.
9. M. Kanno and H. Suzuki, J. Japan Inst. Light Metals 29, (1979), 168.
10. J.M. Galbraith et al., J. Mater. Sci. 22, (1987), 27.
11. M. Kanno and B. Ou, J. Japan Inst. Light Metals 40, (1990), 672.
12. T. Sato and T. Takahashi, "Aging phenomena and Mechanical Properties of Al-Li Alloys and Al-Li-Zr Alloys" (Report No.21, Al-Li alloys, Japan Inst. Light Metals, 1989), 84.
13. H. Wang et al., J. Japan Inst. Light Metals 25, (1975), 27.
14. D.A. Porter and K.E. Easterling, Phase Transformations in Metls and Alloys (England, Van Nostrand Reinhold Co.Ltd., 1987), 110.
15. S.Matsuo and T. Hirata, J. Japan Inst. Light Metals 25, (1975), 18.