Variations in the Microstructure and Hardness with Solution Treating and Aging Conditions in New $\alpha+\beta$ Titanium Alloy Ti–4.5%Al–6%Nb–2%Fe–2%Mo

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Variations in the microstructure and hardness with solution treating and aging conditions in new $\alpha+\beta$ titanium alloy Ti–4.5%Al–6%Nb–2%Fe–2%Mo were investigated. Solution treating (ST) temperatures were varied from 1 048 to 1 173 K with both cooling conditions of water quenching and air cooling. Cooling rate after solution treating was varied from 1 to 60 K/s using a hot working simulator. Age-hardening was investigated at temperatures ranged from 573 to 823 K and aging time periods were varied in the range from 0.36 to 32.4 ks. An extremely fine two-phase microstructure with $\alpha$ grain size of around 1 $\mu$m was obtained at ST temperature of 1 048 K. The large amount of retained $\beta$ phase was obtained by solution treating at temperatures below 1 098 K, and the amounts of $\alpha'$ martensite and athermal $\omega$ phase increased with an elevation of ST temperature above 1 098 K. For solution treating with water quenching, no hardness variation with solution treating temperature was observed. Age-hardening at the temperatures above 723 K took place at an extremely short aging time. Peak-age was observed in such a short aging time period as 0.36 ks, and hardness value at peak-age increased markedly and continuously with an elevation of solution treating temperature. Age-hardening was confirmed to be caused by very fine a precipitates, and age-hardening behavior noted above was explained by athermal $\omega$ phase playing a role of a nucleus for a precipitate.

KEY WORDS: $\alpha+\beta$ titanium alloy; solution treating; age-hardening; two-phase microstructure; transformed products; retained $\beta$ phase; athermal $\omega$.

1. Introduction

Most $\alpha+\beta$ or $\beta$ titanium alloys are strengthened by solution treating and aging, of which optimum conditions are determined primarily by consideration of objective mechanical properties such as strength-ductility balance, fatigue strength or fracture toughness. Refinement of $\alpha+\beta$ two-phase microstructure evolved by $\alpha+\beta$ processing is the most favorable for excellent strength-ductility balance and fatigue strength in $\alpha+\beta$ titanium alloys, while high fracture toughness is obtained by $\beta$ heat treatment.1–3) The microstructure evolved in $\alpha+\beta$ titanium alloys is primarily controlled by solution treating (noted as ST hereafter) temperature and cooling rate in solution treating,4–6) where the former determines $\alpha/\beta$ volume fractions. Strength and ductility obtained by solution treating and aging in titanium alloys are controlled by aging temperature and aging time, and aging temperature adopted in commercial heat treatment is in general above 673 K, because aging below this temperature leads to formation of aged or thermal $\omega$ phase which causes deterioration of ductility.7–10) The aging time to reach peak-age of hardness is needed to be more than 21.6 ks in most of commercially used titanium alloys, in particular $\beta$ titanium alloys.9–12) Such a long aging time possibly causes high energy consumption and a reduction of productivity, which turns out one of the disadvantages in industrial heat treatment of titanium alloys.

Currently, the authors developed new $\alpha+\beta$ titanium alloy, Ti–4.5%Al–6%Nb–2%Fe–2%Mo (notation of % indicates mass percent, being eliminated hereafter) possesses much improved cold workability compared with most of commercially used $\alpha+\beta$ titanium alloys (noted as new alloy hereafter).13) While new alloy possesses similar mechanical properties to SP-700 alloy,14–16) biocompatibility was improved by optimization of alloy compositions such as elimination of vanadium, which is contained in SP-700 alloy.15) The new alloy with high $\beta$ transus such as Ti–6Al–4V or Ti–6Al–7Nb alloy have relatively low hardenability, and solution treating in these alloys forms $\alpha'$ martensite. On the other hand, new alloy has very low $\beta$ transus of 1 156 K,13) and $\beta$ phase formed by heat treating in $\alpha+\beta$ two-phase region or a single $\beta$ phase region in alloys with such a low $\beta$ transus possibly transforms to $\alpha'$ martensite and athermal $\omega$ phase as well as retained $\beta$ phase.16–19) These features of the transformed microstructures in new titanium alloy may also cause different aging response from those of alloys with high $\beta$ transus.

The present study was conducted to investigate variations
in the microstructure and hardness with both conditions of solution treating and aging in newly developed $\alpha + \beta$ titanium alloy. The heating temperature and cooling rate in solution treating were primarily varied, and the phase map of new alloy was drawn based on these results. In aging experiments, the temperature and the time period of aging were varied from 573 to 823 K and from 3.6 to 32.4 ks, respectively. Peak-age at 748 K was found to reach in the aging time period of around 3.6 ks, and this rapid aging response is different from commonly observed age-hardening behavior in most $\alpha + \beta$ and $\beta$ titanium alloys. Thus, aging response in shorter aging time periods of less than 3.6 ks was investigated, and the cause for this very rapid aging response was discussed based on TEM observation and XRD analysis.

2. Experimental Procedures

Alloy ingots used in this study were prepared using argon arc melting furnace with a non-consumable electrode, of which chemical composition was Ti–4.3Al–5.5Nb–2.1Fe–2.1Mo. The ingot size was 140 mm $L \times 40 \text{mm}^2 \times 18 \text{mm}$, where $L$, $B$ and $T$ are, length, width and thickness, respectively, and this ingot was hot forged by two-step process. The first step of forging was conducted by heating at 1 373 K and forging to a round bar with a diameter of 20 mm. In the second step, the forged bar was reheated at 1 073 K in $\alpha + \beta$ two-phase region, followed by forging to 8 mm diameter bar with forging ratio of 5. Many coupon specimens with a diameter of 8 mm and a thickness of 2 mm were machined from the bar, and were subjected to heat treatment.

ST temperatures were varied from 1 048 to 1 173 K in the heating time period of 3.6 ks, followed by both cooling conditions of water quenching and air cooling. Aging temperatures were varied from 573 to 823 K in the heating time period of 3.6 ks, and ST temperatures in this aging experiment were 1 073 K, 1 123 K and 1 173 K, followed by water quenching. Aging time periods at the temperature of 748 K were varied from 3.6 to 32.4 ks under the same ST temperatures as described above. Since aging response of new alloy was found to occur so rapidly from the aging experiments, an additional aging experiment was conducted in shorter aging time periods ranged from 0.36 to 3.6 ks at aging temperatures of 723 K, 773 K and 823 K.

Variations in the microstructure and hardness with cooling rate after solution treating were investigated using the hot working simulator (THERMEC MASTER-Z, Fuji Electronic Industry Co.) with systems of an induction heating and He gas cooling which enabled controlled cooling. Specimens with a diameter of 6 mm and a height of 10 mm were heated at the temperature of 1 073 K or 1 123 K for 3.6 ks, and were cooled by various cooling rates ranged from 1 to 60 K/s. After cooling, specimens were aged at 723 K for 3.6 ks.

Hardness measurements of as solution treated and aged specimens were performed using a Vickers hardness tester with a load of 1 kgf. The microstructural observation was conducted using an optical microscope and a scanning electron microscope (SEM). The $\alpha$ grain size and $\alpha$ volume fraction were quantitatively evaluated using the optical and scanning electron micrographs employing an image analysis software (Scion image). Transmission electron microscopic (TEM) observations were performed for solution treated or aged specimens of new alloy, and thin foil specimens for TEM observations were prepared by a twin-jet electro-polishing method at 243 K. The electrolyte was a solution consisted of perchloric acid, 2-butoxyethanol and ethanol with a composition of 6 : 37 : 57 in volume. The TEM observations including bright and dark field images were conducted using a Hitachi H-800 electron microscope operated at 200 kV. X-ray diffraction (XRD) analysis (Philips, X’Pert) was performed using CuK$\alpha$ at voltage of 40 kV.

3. Results

3.1. Variations in the Microstructure and Hardness with Solution Treating Conditions

Variations in the microstructure with ST temperature are shown in Figs. 1(a) to 1(h), which were subjected to water quenching (1(a) to 1(d)) and air cooling (1(e) to 1(h)) from ST temperature. Because of extremely fine $\alpha + \beta$ two-phase microstructure, the microstructures of the specimens for solution treated at the temperatures from 1 048 to 1 148 K were observed by SEM. Grains with a dark color and a white color in scanning electron micrographs are $\alpha$ and $\beta$ phases, respectively, and $\alpha$ grain size tends to coarsen and $\alpha$ volume fraction decreases with an elevation of ST temperatures. As seen from Fig. 1, very little difference in two-
phase microstructures was found between water quenched and air cooled specimens at ST temperatures of 1048 and 1098 K. The microstructure obtained at ST temperature of 1148 K with air cooling reveals numerous straight lines inside primary β phase, which is confirmed to be acicular α by TEM observation. Figures 1(d) and 1(h) show the optical micrographs of the specimens solution treated at 1173 K, which is above the β transus of new alloy. Both microstructures of water quenched and air cooled specimens show markedly coarsened β grains accompanied with disappearance of primary α phase. The transformed β microstructure evolved by water quenching was confirmed to be α′ martensite by TEM observation and XRD analysis, and formation of acicular α as well as very thin grain boundary α is observed in the air cooled specimen.

Figure 2 shows variations in α grain size and α volume fraction with ST temperature. Very small α grain size of around 1 μm is obtained at the lowest ST temperature of 1048 K, and α grain size increases with an elevation of ST temperature. Grain size difference between water quenched and air cooled specimens is very tiny, although air cooled specimens yielded slightly coarser grain size. The β transus of new alloy obtained from the β-approach curve shown in Fig. 2(b) is 1156 K and the volume fraction of both phases appears to become equal at the temperature of around 1030 K.

Hardness variations with ST temperature are shown by open marks in Fig. 3. Water quenched specimens yield almost constant hardness value of approximately 300 in Hv in the whole ST temperature range including above the β transus, although the volume fractions of both α and β phases are widely varied in this temperature range (see Fig. 2). On the other hand, hardness in air cooled specimens increases sharply at ST temperatures above 1148 K, while hardness below this ST temperature is mostly same with that of water quenched specimens. This increase in hardness at the higher ST temperatures is caused by precipitation of acicular α phase as observed in Figs. 1(g) and 1(h). Variations in hardness after aging are shown by solid marks in Fig. 3. Aging was conducted at the temperature of 723 K for 3.6 ks. An elevation of ST temperature from 1048 to 1173 K continuously increases hardness, indicating that the increase of β volume fraction in two-phase microstructure formed at ST temperature yields the higher hardness after aging. The highest hardness value of around 490 in Hv is obtained at ST temperature of 1173 K above the β transus. Difference of hardness value between water quenched and air cooled specimens is not so large, and this difference becomes smaller with an elevation of ST temperatures, approaching to the same hardness value in both cooling conditions at ST temperature of 1173 K.

Hardness variations with cooling rate were investigated in a wider range of cooling rate using a hot working simulator, of which results are shown in Fig. 4. Open and solid marks in this figure show hardness in as solution treating and after aging at 723 K for 3.6 ks, respectively, and β volume factions evolved at ST temperatures of 1073 K and 1123 K were around 60% and 80%, respectively as seen from Fig. 2. In specimens solution treated at 1073 K, no variation in hardness values with cooling rate is observed. In contrast, in specimens solution treated at 1123 K, peak of hardness is observed at cooling rate of 10 K/s. For the latter, precipitation of acicular α phase in β matrix with a reduction of cooling rate was observed. The fine α precipitates formed at cooling rate of around 10 K/s cause peak
hardness, and the further decrease of cooling rate may coarsen precipitates, resulting in a reduction of hardness. On the other hand, in the specimens solution treated at 1073 K, no $\alpha$ precipitate was observed in the whole range of cooling rate, indicating that $\beta$ phase formed at this ST temperature is more thermally stable compared with that of ST temperature of 1123 K. Hardness after aging continuously increases with the increase of cooling rate for both ST temperatures, and the higher ST temperature yields higher hardness similarly to the result shown in Fig. 3. It is concluded from both results shown in Figs. 3 and 4 that harness obtained under a fixed aging condition is primarily controlled by both ST temperature and cooling rate.

3.2. Variations in Hardness with Aging Conditions

In this section, the effects of aging temperature and aging time period on hardness were investigated in detail. Variations in hardness with aging temperature under a fixed aging time period of 3.6 ks are shown in Fig. 5. ST temperatures of 1073 K, 1123 K and 1173 K formed $\beta$ volume fraction of approximately 60%, 80% and 100%, respectively. Both peak hardness value and the temperature yielding peak hardness increase with an elevation of ST temperature or the increase of $\beta$ volume fraction. Aging temperature yielding peak hardness shifts from 673 to 748 K with an elevation of ST temperature.

The effect of aging time period at temperature of 748 K on hardness is shown in Fig. 6. ST temperatures were varied similarly to Fig. 5, and aging time was varied from 3.6 to 32.4 ks. Hardness increases very rapidly in such a short aging time period as 3.6 ks at all ST temperatures, followed by almost constant hardness values in aging time periods extended up to 32.4 ks at a respective ST temperature. Consequently, no over-age behavior in aging time vs. hardness curves at this aging temperature is observed in the range of aging time periods investigated here. The higher ST temperature yields the higher hardness value, and difference of ST temperatures between 1073 K and 1173 K results in hardness difference of approximately 85 in Hv in the plateau region in aging time vs. hardness curves.

The further investigation was conducted regarding on such a very rapid aging response in new alloy as seen in Fig. 6. Age-hardening behavior in aging time periods of less than 3.6 ks was investigated at various aging temperatures. Hardness variations with aging time at temperatures of 723 K, 773 K and 823 K are shown in Figs. 7(a), 7(b) and 7(c), respectively. Aging time vs. hardness curves shown in these figures are very similar to those shown in Fig. 6, the aging time scale in a horizontal axis shifting to a shorter period by one order compared with Fig. 6. Hardness increases rapidly during very short aging time less than 0.36 ks, and the further extension of aging time up to 3.6 ks almost yields a plateau of hardness. An elevation of ST temperature simply shifts aging response curves to higher values of plateau harness at all aging temperatures.

Microstructural observations of aged specimens by means of optical microscope or SEM showed two-phase microstructure consisting of primary $\alpha$ phase and strongly etched transformed $\beta$ phase or a single phase of the trans-
formed \( \beta \). TEM observations were conducted to identify constitutional phases or products in the transformed \( \beta \) phases which are shown in Discussion.

4. Discussion

4.1. Transformation of \( \beta \) Phase and the Phase Map of New Alloy

Transformed products from \( \beta \) phase obtained by solution treating in \( \alpha+\beta \) titanium alloy vary depending on thermal stability of \( \beta \) phase. These change from acicular \( \alpha \), \( \alpha' \) martensite to \( \alpha' \) martensite with the increase of \( \beta \) phase stability, finally retained \( \beta \) phase being obtained with the highest stability of \( \beta \) phase. It is also known that formation of \( \alpha' \) martensite often accompanies with athermal \( \omega \) phase formation.\(^6,18,19,21\) As seen in Fig. 3, neither variation in hardness value nor formation of acicular \( \alpha \) precipitates was observed for water quenching from all ST temperatures. This appears to be attributed to relatively high contents of \( \beta \) stabilizing elements in new alloy, which can be inferred from very low \( \beta \) transus compared with that of other commercially used \( \alpha+\beta \) titanium alloys. Stability of \( \beta \) phase in \( \alpha+\beta \) titanium alloy is also varied by the volume fraction of \( \beta \) phases or the ST temperature in two-phase region, because the amount of \( \beta \) stabilizing elements partitioned into \( \beta \) phase varies by \( \beta \) volume fraction in two-phase microstructure. For ST temperature of 1 073 K, no variation in hardness was observed in a wide range of cooling rate as shown in Fig. 4, and this appears to be attributed to enhanced \( \beta \) stability for low \( \beta \) volume fraction. The increase in hardness and precipitation of acicular \( \alpha \) were observed at ST temperature above 1 148 K followed by air cooling as shown in Figs. 1 and 3. This is caused by a reduction of \( \beta \) stability accompanied with the increase in \( \beta \) volume fraction, resulting in a variation in the transformed products depending on cooling rate. This is also confirmed in the result shown in Fig. 4, where the increase of ST temperature from 1 073 to 1 123 K yielded hardness variation with cooling rate.

XRD analysis and TEM observation on specimens water-quenched from various ST temperatures were conducted to investigate transformed products from \( \beta \) phase and also to draw a phase map of new alloy. Figure 8 shows XRD patterns of the specimens water-quenched from ST temperatures of 1 073 K, 1 123 K and 1 173 K, and \( \beta \) volume fractions at a respective temperature are approximately 60%, 80% and 100%. An elevation of ST temperature from 1 073 to 1 173 K continuously decreases peak intensity of retained \( \beta \) phase. Formation of \( \alpha' \) martensite is not detected in the specimen solution-treated at 1 073 K, but it was detected at ST temperatures above 1 123 K and its peak intensity becomes higher with an elevation of ST temperature. Reflection peak intensity of \( \alpha \) phase is attributed to primary \( \alpha \) phase, and it continuously decreases with an elevation of ST temperature. This corresponds to the result of a variation in \( \alpha \) volume fraction with ST temperature shown in Fig. 2(b). A phase map was drawn based on these data of the specimens solution treated at the temperatures from 1 048 to 1 173 K with water quenching. The primary \( \beta \) phase formed at a respective ST temperature was assumed to form constituents consisted of retained \( \beta \) and \( \alpha' \) martensite, and the strongest peak intensity of these phases and their ratio were used for evaluation of the volume fraction of these phases. Figure 9 shows a resultant phase map of new alloy. The \( \beta \) phase formed at ST temperatures below 1 098 K is mostly retained at ambient temperature accompanying with a little transformation to \( \alpha' \) martensite. This result indicates \( \beta \) phase is thermally very stable when primary \( \beta \) volume fraction in two-phase microstructure is below 70%. The amount of \( \alpha' \) martensite sharply increases above 1 098 K, and it becomes over 70% above \( \beta \) transus, resulting in a small amount of retained \( \beta \) phase.

Examples of TEM observation and phase analysis in the specimens with ST temperatures of 1 048 K and 1 123 K are shown in Figs. 10(a) to 10(c) and 10(d) to 10(f), respectively, where a bright field image, selected area diffraction pattern (SADP) of [110] in \( \beta \) phase and its index are shown in a respective TEM observation. TEM micrograph of the specimen solution treated at 1 048 K exhibits retained \( \beta \) phase, and the formation of \( \alpha' \) martensite in the specimen.

Fig. 8. XRD patterns of new alloy solution-treated at (a) 1 073 K, (b) 1 123 K and (c) 1 173 K with water quenching.

Fig. 9. A phase map of new alloy. Notations of \( \beta \), \( \alpha_p \) and \( \alpha' \) are retained \( \beta \), primary \( \alpha \) and \( \alpha' \) martensite, respectively.
solution treated at 1 123 K was confirmed by SADP analysis, and so the micrograph of a bright field image on primary \( B \) phase appears to consist of an extremely fine and dense \( \alpha'' \) martensite and retained \( B \) phase.\(^{14} \) Formation of athermal \( \omega \) phase is identified by diffuse streaks in SADP and its index in both specimens, and spots of \( \omega \) phase in the specimen solution treated at the temperature of 1 123 K is much brighter than those in the specimen with ST temperature of 1 048 K. This indicates the larger amount of athermal \( \omega \) formation in the former specimen. In fact, it was found from TEM observation and SADP analysis of all specimens with various ST temperatures investigated here that the higher ST temperature continuously increased brightness of athermal \( \omega \) spots.

As shown in Fig. 3, hardness in as ST specimens with water quenching was 300 to 310 in Hv in a wide range of ST temperature. That is, hardness of the transformed \( B \) phase obtained by ST in a single \( B \) phase was almost same with that of two-phase microstructures with various volume fractions of both phases. Strength or hardness in \( \alpha + B \) titanium alloys with two-phase microstructure such as Ti–6Al–4V and SP-700 alloys was well analyzed based on the law of mixture for strength which was consisted of the terms of the volume fraction and strength or hardness in a respective phase.\(^{4,22} \) Hardness of primary \( \alpha \) in new alloy was about 310, and this value was varied very little by ST temperature. Fukai et al. also reported that hardness of primary \( \alpha \) in SP-700 alloy was almost constant in a wide range of ST temperature.\(^{22} \) These results indicate that a variation of chemical composition in \( \alpha \) phase with ST temperature exerts very tiny influence on hardness of this phase. As shown in Fig. 9, the increase in the volume fraction of primary \( B \) phase or an elevation of ST temperature increased the amounts of \( \alpha'' \) martensite and athermal \( \omega \), but it did not increase hardness in as ST specimens as seen in Fig. 3. These results indicate that both of \( \alpha'' \) martensite and athermal \( \omega \) did not harden as ST specimens. It has been well confirmed that formation of athermal \( \omega \) did not harden as ST specimens. Consequently, a variation in the volume fraction of \( \alpha'' \) martensite with ST temperature appears not to affect hardness in as ST specimens. The thermal stability of primary \( B \) phase decreases with an elevation of ST temperature, and so a reduction of cooling rate from ST temperature tends to form acicular \( \alpha \), resulting in the increase of hardness in as-cooled specimen as shown in Figs. 3 and 4.

Consequently, \( B \) phase formed by solution treating at lower temperatures in new alloy is thermally stable, resulting in a large amount of retained \( B \) phase at the temperatures below 1 098 K. Therefore, new alloy can be named as \( B \)-rich \( \alpha + B \) titanium alloy similarly to SP-700 alloy or \( B \)-Cez alloy.\(^{14,18} \) The amount of \( \alpha'' \) martensite and athermal \( \omega \) phase increased with an elevation of the ST temperature above 1 098 K, and formation of all of these phases was confirmed not to harden as solution treated specimens.

4.2. Age-hardening Behavior of the New Alloy

Two major features in age-hardening behavior of new alloy were observed in this study. First, age-hardening took place in an extremely short time period at the aging temperatures above 723 K, and peak-age was observed in a time period of less than 0.36 ks. Secondly, peak hardness increased with an elevation of ST temperature or the increase of \( B \) volume fraction in two-phase microstructure formed by solution treating, and the largest age-hardening was observed by solution treating in a single \( B \) phase region. Both phenomena appear to be exerted by the same cause. In general, age-hardening after solution treating in most of \( \alpha + B \) or \( B \) titanium alloys occurs due to very fine \( \alpha \) precipitates in \( B \) matrix, and this precipitation takes place based on nucleation and growth controlled by diffusion of \( \alpha \) stabilizing elements enriched into \( \alpha \) precipitates or \( B \) stabilizing elements exhausted from them. Therefore, age-hardening due to \( \alpha \) precipitate requires very long aging time as noted in Introduction, most of which is spent in the nucleation stage. When \( \alpha \) precipitates nucleate directly in \( B \) phase during aging, the higher volume fraction of retained \( B \) phase or solution treating at lower temperature possibly causes the larger age-hardening due to the increase of the total number of precipitates. However, this is different from the present result, and so, it is needed to find the other explanation for the present age-hardening behavior.

Figures 11 and 12 show TEM observations of aged specimens with ST temperatures of 1 048 K and 1 123 K, respectively. These specimens were aged at 723 K for 3.6 ks. Very fine acicular precipitates observed in dark field images of both figures are identified as \( \alpha \) phase by SADP analysis of \([011]_B\) in \( B \) phase, indicating that age-hardening obtained in this study is caused by \( \alpha \) precipitates. It is also found that the larger amount of \( \alpha \) precipitates are detected in the specimen with ST temperature of 1 123 K compared with that of ST temperature of 1 048 K. The thickness of \( \alpha \) precipitates appears to be less than 10 nm without depending on ST temperature. Therefore, the increase of the peak hardness with an elevation of ST temperature seems to be attributed to formation of the larger amount of \( \alpha \) precipi-
As shown in Fig. 7, marked age-hardening took place during aging time of less than 0.36 ks, and this hardening is caused by \( \alpha \) precipitates, and so hardening taken place during such a short aging time seems to indicate that \( \alpha \) precipitates are formed without incubation time in the nucleation stage. It was found that the amount of athermal \( \omega \) formed in as ST specimens increased with an elevation of ST temperature. If athermal \( \omega \) phase possibly plays a role of a nucleus for \( \alpha \) precipitates during aging, age-hardening behavior observed in this study can be well explained. That is, nuclei for \( \alpha \) precipitates are formed in as ST condition, and \( \alpha \) precipitates can be formed without any incubation time for nucleation at the temperature above 723 K. The very short diffusion distance of \( \alpha \) and \( \beta \) stabilizing elements is required for formation of \( \alpha \) precipitates with the thickness of less than 10 nm. The number of nuclei is associated with the amount of athermal \( \omega \) phase, which increases with an elevation of ST temperature. Then, the increase of the nuclei causes the increase of the number of \( \alpha \) precipitates, resulting in higher hardness.

Consequently, two features of age-hardening behavior observed in new alloy appear to be caused by athermal \( \omega \) formed by solution treating. It is well known that aged or thermal \( \omega \) is also formed with low temperature aging in various titanium alloys, which causes marked age-hardening accompanied with deterioration of ductility. So, age-hardening behavior at low temperature in new alloy was investigated as shown in Fig. 13. Aging temperature is 573 K and aging time periods are varied from 3.6 to 172.8 ks. It is confirmed that age-hardening due to thermal \( \omega \) takes place at extremely slow rate. Hardness at aging time of 172.8 ks is around 400 in Hv, which was varied a very little by ST temperatures. On heating for aging at the temperature above 673 K, the specimen passes the temperature region where thermal \( \omega \) precipitates, but its precipitate kinetics is so slow process. Thus, the present results of age-hardening behavior observed at the temperature above 723 K appears not to be associated with thermal \( \omega \) formation.

Sukedai et al. investigated the detailed microstructures evolved by two-step aging in SP-700 alloy, and found that aged \( \omega \) formed in the first-step aging at the low temperature became nucleation sites for \( \alpha \) precipitates formed in the second-step aging at higher temperature.\(^9\) For confirmation of similar possibility of athermal \( \omega \) playing a role of a nucleus for \( \alpha \) precipitates, it is needed to observe athermal \( \omega \) directly by high resolution transmission electron microscopy and to investigate relationship between athermal \( \omega \) and \( \alpha \) precipitates. Regarding on very rapid aging response observed in new alloy, it is also needed to investigate whether any effect of alloying elements is involved or not in this phenomenon. For example, diffusivity of Fe in \( \alpha \) and \( \beta \) phases contained in new alloy or SP-700 alloy is much higher compared with that of other \( \beta \) stabilizing elements such as Mo or V. Consequently, it appears to be important in the future works to investigate effects of various alloying elements on athermal \( \omega \) formation in solution treating and precipitation kinetics of \( \alpha \) phase in aging.

5. Conclusions

Variations in the microstructure and hardness with both conditions of solution treating and aging in newly developed \( \alpha + \beta \) titanium alloy Ti–4.5%Al–6%Nb–2%Fe–2%Mo
were investigated, and the following results were obtained.

(1) The phase map under solution treated condition in new alloy was drawn, which showed formation of the large amount of retained $\beta$ phase at solution treating temperatures below $1098 \, K$ and the marked increases in the amounts of $\alpha'$ martensite and athermal $\omega$ phase with an elevation of ST temperature above $1098 \, K$.

(2) For solution treating with water quenching, no hardness variation with ST temperature was observed, hardness value being around 300 in Hv. It was confirmed that formation of $\alpha'$ martensite and athermal $\omega$ phase did not increase hardness. Hardness in air cooled specimens increased at ST temperature above $1123 \, K$, which was caused by acicular $\alpha$ precipitates.

(3) Age-hardening at the temperature above 723 K took place at an extremely short aging time, and peak-age was observed to occur in such a short aging time period as 0.36 ks.

(4) Hardness value at peak-age increased continuously with an elevation of ST temperature, and the highest hardness value of 490 in Hv was obtained by solution treating at the temperature above $\beta$ transus.

(5) Age-hardening was confirmed to be caused by very fine $\alpha$ precipitates, and age-hardening behavior noted above appeared to be explained by athermal $\omega$ phase playing a role of a nucleus for $\alpha$ precipitates.

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