Effects of Alloying Elements on Deformation Mode in Ti–V Based β Titanium Alloy System

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We have discussed the rule by which the predominant cold deformation modes of metastable β phase are governed depending on their chemical compositions through the examinations of the effects of Sn, Al, and Zr on β -quenched and slightly cold rolled microstructures in a Ti–V based alloy system. It seems in Ti–V binary alloys, that orthorhombic martensitic transformation temperatures M_s and M_d are depressed far below room temperature by the athermal ω phase at over about 15% V. Tin and aluminum intrinsically lower these temperatures like vanadium. On the other hand, tin and aluminum simultaneously suppress the athermal ω phase of Ti–16V to first rise M_d and even M_s up to above room temperature, respectively. The deformation mode of β phase consequently depends on both the M_d temperature and the degree of suppression of the athermal ω phase formation. Alloys having a M_d above room temperature undergo stress-induced martensitic transformation. As for alloys having a M_d below room temperature, alloys where the athermal ω phase is sufficiently suppressed undergo slip, whereas alloys where it is not so done $\{332\}\langle 113 \rangle$ twinning. Since aluminum strongly suppresses the athermal ω formation, increased Al additions change the deformation mode of Ti–16V from $\{332\}\langle 113 \rangle$ twinning to stress-induced martensitic transformation via a quenched α' region or that of Ti–16V–4Sn from stress-induced martensitic transformation to slip. On the other hand, since tin does not suppress it so much as aluminum, increased Sn additions first change the deformation mode of Ti–16V from $\{332\}\langle 113 \rangle$ twinning to stress-induced martensitic transformation but subsequently revive $\{332\}\langle 113 \rangle$ twinning again before slip. The deformation mode of Ti–V–Al–Sn alloys can be interpreted by superposing the effects of V, Al, and Sn. Zirconium also depresses martensitic transformation temperatures and Ti–14V–6Zr undergoes stress-induced martensitic transformation. However, Ti–16V based Zr-added alloys undergo $\{332\}\langle 113 \rangle$ twinning in the wide range of Zr content because the athermal ω phase formation is rarely suppressed by zirconium. This interpretation has solved the discrepancy of the transition of deformation modes of β titanium alloys.

KEY WORDS: β titanium alloys; deformation modes; stress-induced martensitic transformation; twinning; slip; athermal ω phase formation.

1. Introduction

The deformation modes of β titanium alloys affect both their cold deformability¹⁾ and α precipitation behavior.²⁾ From the industrial and metallurgical point of view, it is a very important problem to clarify how the deformation modes are governed by their chemical compositions. In general, unusual $\{332\}\langle 113 \rangle$ twinning or stress-induced martensitic transformation takes place in the most unstable alloys.³⁾ The choice of deformation mode is interpreted in terms of the degree of instability of β phase.^{4,5)} When alloys are sufficiently β -stabilized, they are deformed by slip^{4,5)} or $\{112\}\langle 111 \rangle$ twinning.^{6,7)} However, few efforts have been made to systematically understand the reason some alloys undergo stress-induced martensitic transformation and others $\{332\}\langle 113 \rangle$ twinning even though both categories of alloys are sufficiently unstable.

Recently, Ishiyama and co-workers⁸⁾ have discussed this subject in connection with the effects of important

alloying elements Al, Sn, and Zr on the deformation mode of β titanium alloys. They have pointed out that tin and aluminum retard athermal ω phase transformation and move the deformation mode from $\{332\}\langle 113 \rangle$ twinning to stress-induced martensitic transformation in Ti–16V–Sn and Ti–16V–6Zr–Al alloy systems, etc., whereas zirconium maintains the $\{332\}\langle 113 \rangle$ twinning in Ti–16V and Ti–7Cr because there is little suppression of athermal ω phase formation. Additionally, they have suggested that a larger amount of Sn or Al depresses the stress-induced martensitic transformation by their solution hardening effect, leading to slip deformation. Their discussion seems to imply that with decreases in the instability of β phase, the deformation mode of β phase changes from $\{332\}\langle 113 \rangle$ twinning to stress-induced martensitic transformation and subsequently to slip in that order. On the other hand, the co-workers have confirmed that Ti–20V–3Sn⁵⁾ predominantly deforms by $\{332\}\langle 113 \rangle$ twinning even though the athermal ω phase formation must be more heavily suppressed

by the extra 4% vanadium^{9,10)} in this alloy than in a Ti-16V-3Sn alloy belonging to the strain-transformable region.⁸⁾ Furthermore, we have confirmed, even though by optical microstructural observations only, that Ti-16V-10Sn undergoes mechanical twinning. There is still a problem concerning the alternative of $\{332\}\langle 113\rangle$ twinning and stress-induced martensitic transformation in the most unstable β titanium alloys. It is necessary to clarify the rule by which the deformation modes of β phase are determined.

Hence, we have studied the roles of main alloying elements V, Sn, Al, and Zr on the deformation mode of β phase, using a Ti-V based β titanium alloy system.

2. Experimental Procedure

In this study we examined the effects of V, Sn, Al, and Zr on β -solution treated or slightly cold rolled microstructures in Ti-(10–22)V-(0–12)Sn-(0–7)Al-(0–6)Zr alloys.

Small button ingots were plasma-arc melted in dimensions of about 40 mm in diameter and 23 mm in thickness. They were hot-rolled at 1273 K to a 5 mm thickness, solution treated at 1273 K for 0.6 ksec, water-quenched, machined to a 4 mm thickness for descaling, cold-rolled at 75%, solution treated at 1273 K for 0.6 ksec, and water-quenched. Subsequently, small pieces cut from the sheets were cold-rolled at 3% in order to introduce strain-induced products into the matrices. Vickers hardness was measured for some water-quenched samples. Microstructures of both the undeformed and the cold-rolled samples were observed with a Nomarski optical microscope. Some cold-rolled samples were examined by X-ray diffraction or transmission electron microscopy as well.

3. Results

3.1. Quenched Microstructures

Figure 1 shows microstructures after water-quenching of Ti-V binary alloys with 12 to 15% V. The β phase of Ti-12V was completely transformed to orthorhombic α'' ^{11–13)} or hexagonal α' ³⁾ phase. With increasing V content the martensite phase gradually decreases, and finally disappears at 15%. This shows that V lowers the martensitic transformation finish temperature M_f . The martensitic transformation start temperature M_s exists between M_f and the upper limit temperature of the strain-transformable range M_d . These temperatures M_d and M_s may probably shift in almost the same manner, corresponding to each other. Vanadium should lower these temperatures as well.

Figure 2 shows the microstructural change with Al content in Ti-16V based alloys. A small addition of 0.3% Al promotes α'' martensitic transformation. An addition of 3% Al gives rise to pronounced α'' phase transformation. On the other hand, additions of 5 to 7% Al can completely retain the β phase. Maeda and Flower¹⁴⁾ have suggested that the α'' transformation first promoted by a small amount of Al additions is attributable to the suppression of athermal ω phase

Fig. 1. Quenched microstructures of Ti-V binary alloys with 12 to 15% V.

Fig. 2. Microstructural change with Al content in Ti-16V based alloys.

Fig. 3. Quenched microstructures of Ti-16V based alloys containing 0.5 to 12% Sn.

formation. Aluminum is expected to lower martensitic transformation temperatures.

Figure 3 shows quenched microstructures of Ti-16V based alloys containing 0.5 to 12% Sn. Unlike Al additions, any percent of Sn addition never causes the α'' transformation during quenching in the range from 0.5 to 12% for Ti-16V. Figure 4 shows quenched microstructures of Ti-16V-1Sn-3Al, Ti-16V-4Sn-1Al, and Ti-16V-4Sn-3Al. An addition of 1% Sn reduces the α''

Fig. 4. Quenched microstructures of Ti-16V-1Sn-3Al, Ti-16V-4Sn-1Al, and Ti-16V-4Sn-3Al.

Fig. 5. Quenched microstructures of Ti-13V-6Zr and Ti-14V-6Zr.

transformation of Ti-16V-3Al. Moreover, an addition of 1% Al promotes the α'' transformation even in Ti-16V-4Sn.

This means that an addition of 4% Sn suppresses the athermal ω phase formation less than that of 0.3% Al in Ti-16V. The β phase is completely retained in Ti-16V-4Sn-3Al. These facts conclusively prove that tin lowers martensitic transformation temperatures like Al but suppresses the athermal ω phase formation less than Al.

Figure 5 shows quenched microstructures of Ti-13V-6Zr and Ti-14V-6Zr. As Fig. 1 showed, Ti-13V and Ti-14V partly undergo the α'' transformation during quenching. On the other hand, the α'' transformation is depressed by an addition of 6% Zr. Ishiyama *et al.*⁸⁾ have confirmed that zirconium rarely suppresses the athermal ω phase formation. However, zirconium also lowers M_s .

3.2. Deformation Modes of Ti-V Base Metastable β Alloys

We have examined stress-induced products of 3% cold rolled samples for alloys where all β phase could be obtained by water-quenching.

Figure 6 shows optical microstructures of 3% cold rolled samples for Ti-16V-5Al, Ti-16V-7Al, Ti-18V-3Al, and Ti-22V-3Al. Deformation-induced products are clearly observed in all the samples. One of the authors of this paper²⁾ previously reported with the optical microstructural characteristics of slightly deformed β titanium alloys, that stress-induced α'' tends to be produced in groups of martensite plates aligned in parallel by cold rolling, deformation twins are frequently accompanied by the internal twins, and that slip brings about a striated structure of planar slip bands with the localized dislocations. However, these features are typical and it is very difficult to identify the deformation-induced products from the optical microstructural observations. So, we have studied the microstructures by X-ray diffraction or transmission electron microscopy. Figure

Fig. 6. Optical microstructures of the 3% cold rolled samples for Ti-16V-5Al, Ti-16V-7Al, Ti-18V-3Al, and Ti-22V-3Al.

7 shows X-ray profiles of the samples. Orthorhombic α'' phase is detected in Ti-16V-5Al, Ti-16V-7Al, and Ti-18V-3Al. Therefore, the predominant deformation mode of these three alloys is stress-induced martensitic transformation. On the other hand, only diffractions from β phase are obtained in Ti-22V-3Al. Figure 8 shows a transmission electron microstructure of the sample of Ti-22V-3Al, together with a diffraction pattern commonly obtained from the positions 1, 2, and 3. Dislocations are introduced with a tendency to form slip bands in the matrix. Slip is the predominant deformation mode of Ti-22V-3Al.

Figure 9 shows optical microstructures of 3% cold rolled samples for Ti-16V-8Sn, Ti-16V-12Sn, Ti-20V-4Sn, and Ti-22V-4Sn. In Ti-16V-8Sn mechanical twins occasionally accompanied by internal twins are observed, whereas in Ti-16V-12Sn only vague striations are observed. Transmission electron microstructures of the two samples shown in Figs. 10 and 11 realize that the stress-induced products of Ti-16V-8Sn and Ti-16V-12Sn are $\{332\}\langle 113 \rangle$ twins and dislocations, respectively. Figure 12 shows X-ray profiles of the samples for Ti-20V-4Sn and Ti-22V-4Sn. Diffraction peaks of orthorhombic α'' phase are unmistakably observed in Ti-20V-4Sn. Ti-20V-4Sn undergoes stress-induced martensitic transformation. In the optical microstructure of Ti-22V-4Sn in Fig. 9 a few sharp stress-induced products exist in a vague striated matrix. The α'' phase is not detected by X-ray analysis, as shown in Fig. 12. Figure 13 shows a transmission electron microstructure of Ti-22V-4Sn. The diffraction pattern identifies the sharp products to be $\{112\}\langle 111 \rangle$ twins.

There are many dislocations in addition to the twins. The predominant deformation mode of Ti-22V-4Sn may therefore be slip.

Figure 14 shows an X-ray profile of Ti-16V-4Sn-3Al. Orthorhombic α'' phase is detected in this sample. **Figure 15** shows transmission electron microstructures of Ti-20V-4Sn-3Al, Ti-16V-4Sn-5Al, and Ti-16V-8Sn-3Al. Dislocations are observed in all the samples.

Figures 16 and 17 show an optical microstructure and an X-ray profile of Ti-14V-6Zr, respectively. In this

alloy, an addition of 6% Zr depresses M_s to below room temperature. The X-ray profile indicates that Ti-14V-6Zr undergoes stress-induced orthorhombic α'' transfor-

Fig. 7. X-ray profiles of the 3% cold rolled samples for (a) Ti-16V-5Al and Ti-16V-7Al, and (b) Ti-18V-3Al and Ti-22V-3Al.

Fig. 8. Transmission electron microstructure of the 3% cold rolled sample of Ti-22V-3Al, together with a diffraction pattern commonly obtained from the positions 1, 2, and 3.

Fig. 9. Optical microstructures of the 3% cold rolled samples for Ti-16V-8Sn, Ti-16V-12Sn, Ti-20V-4Sn, and Ti-22V-4Sn.

Fig. 10. Transmission electron microstructure of the 3% cold rolled sample of Ti-16V-8Sn.

mation.

3.3. Transition of Deformation Modes

The occurrence or absence of stress-induced martens-

itic transformation is necessary to correspond to whether or not the alloy has M_d above room temperature. As for the Ti-V binary system, alloys deform by $\{332\}\langle 113\rangle$ twinning in the range from 16 to 22% V and by slip for more than 24% V at room temperature.¹⁵⁾ As was shown in Fig. 1, Mf, and in turns, M_d and M_s may sluggishly decrease with increasing V content from 12 to 14%. On the other hand, M_d is considerably depressed at 16% V.

Fig. 11. Transmission electron microstructure of the 3% cold rolled sample of Ti-16V-12Sn.

Fig. 12. X-ray profiles of the 3% cold rolled samples for Ti-20V-4Sn and Ti-22V-4Sn.

Fig. 13. Transmission electron microstructure of the 3% cold rolled sample of Ti-22V-4Sn.

Fig. 14. X-ray profiles of the 3% cold rolled samples for Ti-16V-4Sn-3Al.

Fig. 15. Transmission electron microstructures of Ti-20V-4Sn-3Al, Ti-16V-4Sn-5Al, and Ti-16V-8Sn-3Al.

Fig. 16. Optical microstructure of the 3% cold rolled sample of Ti-14V-6Zr.

Fig. 17. X-ray profile of the 3% cold rolled sample of Ti-14V-6Zr.

Hanada *et al.*¹⁵⁾ have reported that Ti-16V does not undergo stress-induced martensitic transformation but predominantly deforms by $\{332\}\langle 113 \rangle$ twinning even at 77 K. This implies that an addition of 16% V depresses Md down to below 77 K. Compared with the gradual decrease in Mf from 12 to 14% V, Ms seems to be remarkably lowered by the increasing V content from 14 to 16%. Figure 18 shows an X-ray profile of the quenched Ti-16V alloy where the athermal ω phase is unmistakably detected. This remarkable depression may therefore be expected to be due to the athermal ω phase formation, as has also been suggested by Maeda and Flower.¹⁴⁾ This presumption is supported by the fact that a small amount of Al addition promotes the α'' transformation during quenching.¹⁴⁾ Figures 16 and 17 did not show that a Ti-14V based metastable alloy which is β -stabilized by 6% Zr undergoes $\{332\}\langle 113 \rangle$ twinning. This does not mean that the β phase of Ti-14V-6Zr is unstable enough to form the athermal ω phase.^{5,6,15)} Zirconium rarely suppresses the athermal ω phase formation.⁸⁾ The athermal ω phase formation may not be accelerated so much at 14% V. With increases in V content, therefore, Md and Ms are gradually lowered down to about 14% but are more markedly depressed downwards out of the V content dependence of intrinsic transformation temperatures by the activated athermal ω phase formation beyond 14%. The athermal ω phase formation, however, may be reduced gradually.^{9,10)} The athermal ω phase formation is suppressed enough for slip deformation at about 24%.¹⁵⁾ Md, Ms,

Fig. 18. X-ray profile of the quenched sample of Ti-16V.

Fig. 19. Schematic changes of Md, Ms, and ω_s with V content in Ti-V binary alloy system. Presumable regions of deformation mode are indicated by question marks (?).

and the athermal ω phase formation temperature ω_s may therefore be expected to change with the V content as Fig. 19 schematically shows. The regions of each predominant deformation mode are also indicated including the quenched α'' region (α''_q). Presumable regions of deformation mode were also indicated in Fig. 19 accompanied by question marks (?), in the same way as the other schematic displays below.

The effects of the other alloying elements on the deformation mode of β phase also have to be discussed in terms of both the changes of Md and ω_s . Komatsu *et al.*¹⁶⁾ and Ikeda *et al.*¹⁷⁾ have demonstrated from the examinations of resistivity of quenched Ti-Fe and Ti-Fe-Al alloys, that the hardness of the alloys lowers with the decreases in the athermal ω phase. This probably means that the athermal ω phase plays a role in precipitation hardening. The athermal ω phase should prevent shear deformation required for martensitic transformation. Even a very small amount of Al additions promotes martensitic transformation, as was shown in Fig. 2. Supposing that Md or Ms should be depressed downwards out of the intrinsic Md or Ms temperatures when ω_s substantially exists above them and depressed more with increases in the athermal ω phase, aluminum must suppress ω_s more abruptly than Md or Ms, as

Fig. 20. Schematic changes of Md, Ms, and ω_s with Al content in Ti-16V-Al alloy system. Presumable regions of deformation mode are indicated by question marks (?).

Fig. 21. Schematic changes of Md, Ms, and ω_s with V content in Ti-V-3Al alloy system.

schematically shown in Fig. 20. At 16% V, Md and Ms are depressed to below 77 K.¹⁵⁾ A small amount of Al addition first recovers Md and Ms up to above room temperature due to the suppression of the athermal ω phase formation.¹⁴⁾ Md and Ms, however, may begin to lower intrinsically when ω_s goes below them. Consequently, Ti-16V based alloys containing 0.3 to 3% Al give rise to the α'' transformation during quenching and alloys containing 5 to 7% Al undergo stress-induced martensitic transformation, as was shown in Fig. 7. Probably, alloys with less than 0.3% Al and alloys with more than 7% Al undergo stress-induced martensitic transformation and slip, respectively. Ti-V-Al ternary alloys containing 3% Al change the deformation mode from stress-induced martensitic transformation to slip beyond 18% V. This may be because the athermal ω phase is sufficiently reduced by V when Md goes below room temperature, as shown in Fig. 21.

As mentioned above, tin does not suppress the athermal ω phase formation so much as Al. Tin, however, really suppresses the athermal ω phase formation, as can be understood from an X-ray profile of the quenched Ti-16V-4Sn sample shown in Fig. 22 where the athermal ω phase is not detected. As Fig. 23 schematically shows, ω_s probably decreases more sluggishly in Ti-16V based Sn-added alloys than it does in the Al-added alloys, compared with the decreases in Md or Ms. The β phase

Fig. 22. X-ray profile of the quenched sample of Ti-16V-4Sn.

Fig. 23. Schematic changes of Md, Ms, and ω_s with Sn content in Ti-16V-Sn alloy system.

of Sn-added Ti-16V based alloys seems to be still unstable enough for $\{332\}\langle 113 \rangle$ twinning when Md is lowered down to below room temperature. Ti-16V based Sn-added alloys may therefore be expected to undergo stress-induced martensitic transformation at 2 to 6% Sn,⁸⁾ $\{332\}\langle 113 \rangle$ twinning at about 8% Sn, and slip at more than 12% Sn. Since vanadium probably depress Md, Ms, and ω_s intrinsically in a Ti-V-Sn alloy system as well, the Sn content dependencies of these temperatures of Ti-16V-Sn alloys in Fig. 23 should be shifted downwards as a whole by the extra 4% V, as shown in Fig. 24. Ti-20V-3Sn⁵⁾ has been reported to deform by $\{332\}\langle 113 \rangle$ twinning. It then seems that Ti-20V based alloys containing up to 3% Sn have a Md below room temperature but are still so unstable to the athermal ω phase formation that they undergo $\{332\}\langle 113 \rangle$ twinning. Stress-induced martensitic transformation, however, occurs in Ti-20V-4Sn, as was shown in Fig. 12. Ti-22V-4Sn is deformed by slip. This may be because both Md and ω_s are depressed enough. Presumably, with an increase in Sn content also, alloys may be deformed by slip.

As was understood from Figs. 14 and 15, Ti-16V-4Sn-3Al changes the deformation mode from stress-induced martensitic transformation to slip when the contents of each alloying element are increased. These transitions of deformation mode can be interpreted by superposing their effects on Md and ω_s , as shown in

Fig. 24. Schematic changes of Md, Ms, and ω_s with Sn content in Ti-20V-Sn alloy system. Presumable regions of deformation mode are indicated by question marks (?).

Fig. 25. Schematic changes of Md, Ms, and ω_s with V content in Ti-V-4Sn-3Al alloy system.

Fig. 26. Schematic changes of Md, Ms, and ω_s with Sn content in Ti-16V-Sn-3Al alloy system.

Figs. 25, 26, and 27. The β phase of Ti-16V-4Sn-3Al is fairly stabilized by 3 % Al and 4 % Sn. A further addition of V, Al or Sn must depress not only Md but also ω_s too much to cause $\{332\}\langle 113\rangle$ twinning.

Zirconium does not affect the athermal β phase formation behavior substantially. Additions of Zr simply lower the martensitic transformation temperatures, as shown in Fig. 28. Zr-added Ti-16V based alloys consequently undergo $\{332\}\langle 113\rangle$ twinning only in the wide range of Zr content.⁸⁾

Fig. 27. Schematic changes of Md, Ms, and ω_s with Al content in Ti-16V-4Sn-Al alloy system.

Fig. 28. Schematic changes of Md, Ms, and ω_s with Zr content in Ti-16V-Zr alloy system.

4. Discussion

4.1. Contribution of Athermal ω Phase to Precipitation Hardening

It is very important for the occurrence or absence of suppression of martensitic transformation by athermal ω phase to evaluate its precipitation hardening effect. It is still not clear whether or not the athermal ω phase contributes the strength and ductility of β phase. So, we have discussed this issue through the examinations of tensile properties and hardness of quenched phases including α'' .

Figure 29 shows Vickers hardness of the Ti-V binary alloys vs. V content. The samples containing quenched α'' phase are indicated by asterisks (*). Moreover, Fig. 30 shows tensile properties vs. V content. The tensile specimens were prepared by hot rolling and solution treatment at 1173 K using button ingots. As is understood from the both figures, the strength of Ti-V alloys first increases up to 14 % and drops at 15 % V.

Figure 31 shows the change of Vickers hardness after water-quenching with Al content in the Ti-16V-Al alloy system. A small amount of Al addition drastically lowers the hardness of Ti-16V. Maeda and Flower¹⁴⁾ also have confirmed the same tendency using Ti-15V-Al alloys, and suggested that the reduction in hardness of the β phase is attributable to the suppression of athermal ω

Fig. 29. Change of Vickers hardness with V content in Ti-V alloy system. The samples containing quenched α'' phase are indicated by asterisks (*).

Fig. 30. Tensile properties of Ti-V binary alloys vs. V content.

Fig. 31. Change of Vickers hardness with Al content in Ti-16V-Al alloy system. The samples containing quenched α'' phase are indicated by asterisks (*).

phase formation. However, the hardness of these Al-added alloys is more or less that of the α'' phase. At least, the effects of athermal ω phase on precipitation hardening of β phase can not be discussed comparing the hardness of β phase with that of quenched or stress-induced α'' phases. Figure 32 shows tensile properties of Ti-V alloys with 3% Al vs. V content. These tensile

Fig. 32. Tensile properties of Ti-V-3Al alloys vs. V content.

Fig. 33. Change of Vickers hardness with V content in Ti-V-3Al alloy system. The samples containing quenched α'' phase are indicated by asterisks (*).

specimens were prepared by hot forging and solution treatment at 1273 K using button ingots. The alloys containing 10 to 16% V have quenched α'' phase, and the alloy containing 18% V is strain-transformable. The lattice parameters of orthorhombic α'' phase has a wide variety of lattice parameters depending on its chemical composition.^{3,11,18,19} The difference in tensile properties, especially yield strength, of α'' phase seems to be mainly due to the change of lattice parameters. It is noted that Ti-16V-3Al shows a very low yield strength, compared with Ti-16V. Therefore, the drastic decreases in Vickers hardness of Ti-16V-Al in Fig. 31 seem to be exaggerated by the α'' phase. It is necessary to compare alloys deformed by $\{332\}\langle 113 \rangle$ twinning with alloys deformed by slip in order to evaluate the contribution of athermal ω phase to mechanical properties.

Figure 33 shows Vickers hardness of the Ti-V-3Al alloys vs. V content. As was shown in Fig. 8, Ti-22V-3Al undergoes slip. Comparing this alloy with Ti-16V, Ti-16V shows a far higher hardness than Ti-22V-3Al. Moreover, Ti-16V has a higher yield strength than Ti-24V-3Al, which is expected to deform by slip as well. Strictly, Ti-16V should have a higher yield strength because it must be deformed by $\{332\}\langle 113 \rangle$ twinning

Fig. 34. Change of Vickers hardness with Sn content in Ti-16V-Sn alloy system.

Fig. 35. Change of Vickers hardness with V content in Ti-V-4Sn alloy system.

before yielding. Figures 34 and 35 show Vickers hardness of the Ti-16V-Sn and Ti-V-4Sn alloy systems, respectively. Ti-16V-12Sn and Ti-22V-4Sn also have a far lower Vickers hardness than Ti-16V. These alloys containing Al or Sn are more highly alloyed than Ti-16V. From the viewpoint of solution hardening, this tendency is completely reversed. The athermal ω phase may therefore be expected to play a role in precipitation hardening of β phase.

In Fig. 30, the ductility of Ti-V binary alloys clearly increases with increasing V content beyond 14%. This enhancement of ductility may be due to the suppression of athermal ω phase caused by increased V.^{4,15)}

5. Conclusion

In order to systematically understand the reason some alloys undergo stress-induced martensitic transformation and others $\{332\}\langle 113 \rangle$ twinning, we have discussed the effects of Al, Sn, and Zr additions on quenched microstructure and the deformation mode in Ti-V based metastable β titanium alloys. The results obtained are as follows:

(1) In Ti-V binary β alloys, martensitic transformation temperatures M_d , M_s , and M_f seem to be considerably depressed out of the intrinsic V content dependencies by the athermal ω phase formation beyond about 14% V.

(2) In Ti-16V based ternary alloys, a small amount of Al addition recovers M_s to give rise to martensitic

transformation during quenching. With increasing Al content, however, all β phase can be retained. Since Al sufficiently suppresses the athermal β phase formation before M_d reaches room temperature, Ti-16V based alloys containing 5 to 7% Al undergo stress-induced martensitic transformation.

(3) Tin does not suppress the athermal ω phase formation so much as Al compared with the depression of M_d . The β phase is still unstable when M_d of Ti-16V-Sn alloys reaches room temperature. Therefore, Ti-16V based Sn-added alloys change the deformation mode from stress-induced martensitic transformation to slip via $\{332\}\langle 113 \rangle$ twinning.

(4) The deformation mode of Ti-V-Al-Sn alloys can be interpreted by superposing the effects of V, Al, and Sn.

(5) Zirconium depresses martensitic transformation temperatures and Ti-14V-6Zr undergoes stress-induced martensitic transformation. Ti-16V based Zr-added alloys undergo $\{332\}\langle 113 \rangle$ twinning in the wide range of Zr content because the athermal ω phase formation is rarely suppressed by Zr.

REFERENCES

- 1) H. Ohyama, T. Nishimura and Y. Ashida: Advanced Materials—New Process and Reliability—, ed. by T. Kishi, N. Takeda and Y. Kagawa, SAMPE, (1993), 1790.
- 2) H. Ohyama, H. Nakamori, Y. Ashida and T. Maki: *ISIJ Int.*, **32** (1992), 222.
- 3) J. C. Williams: Titanium Science and Technology, ed. by R. I. Jaffee and H. M. Burte, AIME, Plenum Press, New York, NY, (1973), 1433.
- 4) S. Hanada and O. Izumi: *Metall. Trans. A*, **17A** (1986), 1409.
- 5) S. Hanada and O. Izumi: *Metall. Trans. A*, **18A** (1987), 265.
- 6) M. Oka and Y. Taniguchi: *J. Jpn. Inst. Met.*, **42** (1978), 814 (in Japanese).
- 7) N. E. Paton and J. C. Williams: Proc. of Second Int. Conf. on Strength of Metals and Alloys, ASM, Metals Park, OH, (1971), 108.
- 8) S. Ishiyama, S. Hanada and O. Izumi: *ISIJ Int.*, **31** (1991), 807.
- 9) J. C. Williams, D. de Fontaine and N. E. Paton: *Metall. Trans.*, **4** (1973), 2701.
- 10) M. Ikeda, S. Y. Komatsu, T. Sugimoto and K. Kamei: Sixth World Conf. on Titanium, ed. by P. Lacombe, R. Tricot and G. Beranger, Les Editions de Physique, Paris, (1989), 313.
- 11) H. M. Flower, R. Davis and D. R. F. West: Titanium and Titanium Alloys, Scientific and Technological Aspects, ed. by J. C. Williams and A. F. Belov, Plenum Press, New York, NY, (1982), 1703.
- 12) M. Oka, C. S. Lee and K. Shimizu: *Metall. Trans.*, **3** (1972), 37.
- 13) T. Maeda and H. M. Flower: Sixth World Conf. on Titanium, ed. by P. Lacombe, R. Tricot and G. Beranger, Les Editions de Physique, Paris, (1989), 1629.
- 14) T. Maeda and H. M. Flower: Sixth World Conf. on Titanium, ed. by P. Lacombe, R. Tricot and G. Beranger, Les Editions de Physique, Paris, (1989), 1589.
- 15) S. Hanada, A. Takemura and O. Izumi: *Trans. JIM.*, **23** (1982), 507.
- 16) S. Komatsu, M. Ikeda, T. Sugimoto, K. Kamei and K. Inoue: *J. Jpn. Inst. Met.*, **55** (1991), 491 (in Japanese).
- 17) M. Ikeda, S. Komatsu, T. Sugimoto, K. Kamei and K. Inoue: *Tetsu-to-Hagane*, **80** (1994), 866 (in Japanese).
- 18) R. Davis, H. M. Flower and D. R. F. West: *J. Mater. Sci.*, **14** (1979), 712.
- 19) H. Sasano, T. Suzuki, O. Nakano and H. Kimura: Titanium '80 Science and Technology, ed. by K. Kimura and O. Izumi, TMS-AIME, Warrendale, Penn., (1980), 717.