Correlation of Tensile Properties, Deformation Modes, and Phase Stability in Commercial β-Phase Titanium Alloys

S. HANADA and O. IZUMI

The effect of plastic deformation mode on tensile properties of quenched commercial β-phase titanium alloys has been investigated at approximately constant grain size and oxygen content. In addition, stability of β-phase has been estimated from ω-reflections or diffuse streaking in electron diffraction patterns in a manner similar to the previous works on binary β-phase titanium alloys. Dominant mode of plastic deformation is $\{332\}\langle 113\rangle$ twinning in the alloys with large instability of β-phase, such as Ti-11.5Mo-6Zr-4.5Sn and Ti-15Mo-5Zr, and is crystallographic slip in the alloys with small instability of β-phase, such as Ti-15Mo-5Zr-3Al, Ti-3Al-8V-6Cr-4Mo-4Zr, Ti-15V-3Cr-3Al-3Sn, Ti-8Mo-8V-2Fe-3Al, and Ti-13V-11Cr-3Al. Twinning leads to low yield strength and large elongation, while slip results in high yield strength and small elongation in agreement with binary and ternary β-phase titanium alloys.

I. INTRODUCTION

AT present there is no satisfactory method which predicts mechanical properties of β-phase multicomponent titanium alloys. Therefore, commercial β-phase titanium alloys with superior cold deformability are being developed experimentally or empirically. Recently, tensile properties of metastable β-titanium binary alloys Ti-V,[1] Ti-Mo,[1] and Ti-Cr[2] have been found to show significant dependencies on modes of plastic deformation as follows: low yield stress and large elongation in the β-titanium alloys with large instability result from $\{332\}\langle 113\rangle$ mechanical twinning. Twinning probability decreases and $\langle 111\rangle$ slip appears as alloying elements increase. Deformation only by slip in the β-titanium alloys with small instability leads to high yield stress and small elongation. It has also been shown that the stability of β-phase can be estimated by measuring positions of ω-reflections or diffuse streaking in electron diffraction patterns of as-quenched alloys.[3]

These findings in binary alloys may be used to predict that β-phase multicomponent titanium alloys with superior cold deformability might be designed by controlling the stability of β-phase.

This research was undertaken to confirm the possibility of this prediction in ternary alloys and commercially developed alloys.

II. EXPERIMENTAL PROCEDURE

β-phase ternary alloys Ti-20V-3Sn, Ti-20V-0.15O, and Ti-20V-6Al, and commercial β-alloys Ti-11.5Mo-6-Zr-4.5Sn, Ti-15Mo-5Zr, Ti-15Mo-5Zr-3Al, Ti-3Al-8V-6Cr-4Mo-4Zr, Ti-15V-3Cr-3Al-3Sn, Ti-8Mo-8V-2Fe-3Al, and Ti-13V-11Cr-3Al, were prepared by arc melting in an argon gas using nonconsumable laboratory furnace. Since no weight change took place during melting, the compositions of the alloys were expressed in nominal wt pct. Approximately 100 g buttons were obtained by multiple melting operations to achieve homogeneity. The alloys were prepared from high purity titanium sponge containing about 200 ppm oxygen. The oxygen doped alloy was prepared by using a Ti-2 pct O alloy which was made from titanium sponge and TiO_2 powder. The alloys were hot rolled at temperatures from 850 to 950 °C to approximately 3 mm thickness, scalped, and then cold rolled to about 1.5 mm thick plates.

Tensile samples with a gage size of 1.5 mm × 3 mm × 16 mm were prepared by spark machining, where their tensile axes were parallel to the sheet rolling direction. The samples were encapsuled in a quartz tube, solution heat treated at various temperatures to develop a β grain size of about 0.2 mm, and then quenched in iced water. The capsule was broken immediately after quenching. Oxide layer formed on the samples during heat treatment was removed by mechanical polishing on emery paper. Oxygen and nitrogen contents of the tensile samples were analyzed after tensile tests. The result is given in Table I. Tensile tests were performed on Instron type testing machine at an initial strain rate of $5.2 \times 10^{-4}\ s^{-1}$. d^*_{0002}/d^*_{222} was determined with a photometer by measuring distances from center to center of the spots or diffuse streaks in selected area diffraction patterns.

Table I. Oxygen and Nitrogen Contents in the Alloys

Alloys	Oxygen (Wt Pct)	Nitrogen (Wt Pct)
Ti-20V	0.028	0.016
Ti-20V-3Sn	0.031	0.015
Ti-20V-0.15O	0.152	0.023
Ti-20V-6Al	0.027	0.012
Ti-11.5Mo-6Zr-4.5Sn	0.029	0.0026
Ti-15Mo-5Zr	0.027	0.0023
Ti-15Mo-5Zr-3Al	0.023	0.0009
Ti-3Al-8V-6Cr-4Mo-4Zr	0.038	0.0074
Ti-15V-3Cr-3Al-3Sn	0.036	0.0092
Ti-8Mo-8V-2Fe-3Al	0.023	0.0061
Ti-13V-11Cr-3Al	0.024	0.011

S. HANADA, Instructor, and O. IZUMI, Professor, are with The Research Institute for Iron, Steel and Other Metals, Tohoku University, Sendai 980, Japan.

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III. RESULTS

A. *Ternary β-Phase Alloys*

Williams *et al.*[4] have shown that ternary additions of aluminum, oxygen, tin, and zirconium to metastable Ti-Mo and Ti-V β-phase alloys retard ω-phase transformation. Also, Froes *et al.*[5] reported that increasing oxygen content suppresses ω-phase formation in Ti-11.5Mo-6Zr-4.5Sn. In agreement with their results, ω-phase reflections in as-quenched alloys become weak or diffuse with additions of tin, oxygen, and aluminum to Ti-20V, as shown in Figure 1. Figure 2 shows true stress–true strain curves during uniform elongation for the as-quenched β-phase ternary alloys, where that for the binary alloy Ti-20V is also illustrated for comparison. Yield stress is higher in Ti-20V-0.15O and Ti-20V-6Al than in Ti-20V-3Sn, while work hardening rate is higher in Ti-20V-3Sn and Ti-20V-0.15O than Ti-20V-6Al. In Ti-20V-3Sn and Ti-20V-0.15O stress-induced plate-like markings were observed throughout straining. These plates were determined to be mainly $\{332\}\langle 113\rangle$ twins by using transmission electron microscopy. An example indicating $\{332\}\langle 113\rangle$ twinning is shown in Figure 3, where electron diffraction patterns are taken from the boundary region of a stress-induced plate (a), the matrix (b), and the plate (c) of Ti-20V-3Sn strained approximately 2 pct, respectively. One can find that the alloys, Ti-20V-3Sn and Ti-20V-0.15O, having the same deformation mode of $\{332\}\langle 113\rangle$ twinning, show a similar rate of work hardening (Figure 2). The deformation mode of Ti-20V has been found to be intrinsically $\{332\}\langle 113\rangle$ twinning[3] but ω-reflections in the twins become stronger than those in matrix, leading to higher rate of work hardening (Figure 2).

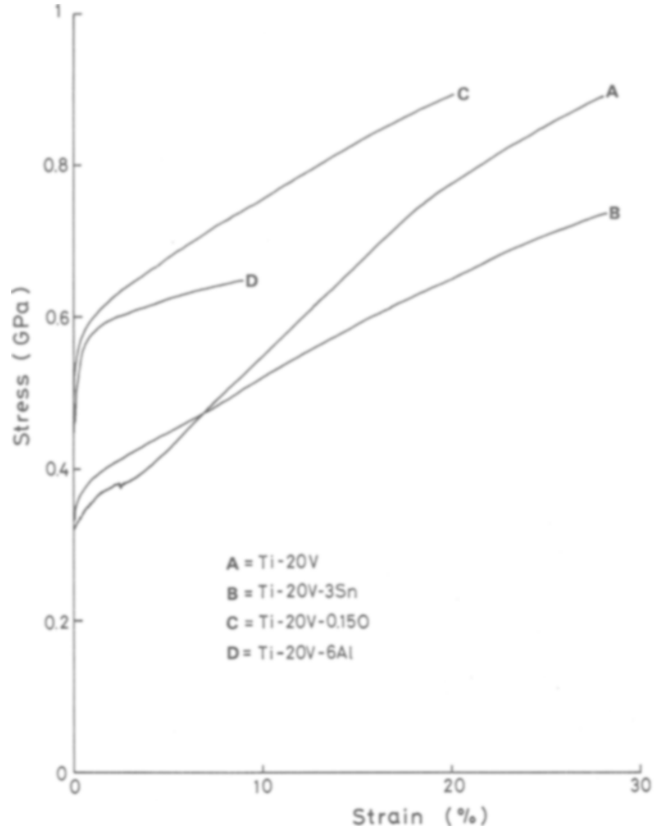

Fig. 2—Stress-strain curves for as-quenched β-phase ternary alloys. For comparison the curve for as-quenched Ti-20 pct V is illustrated.

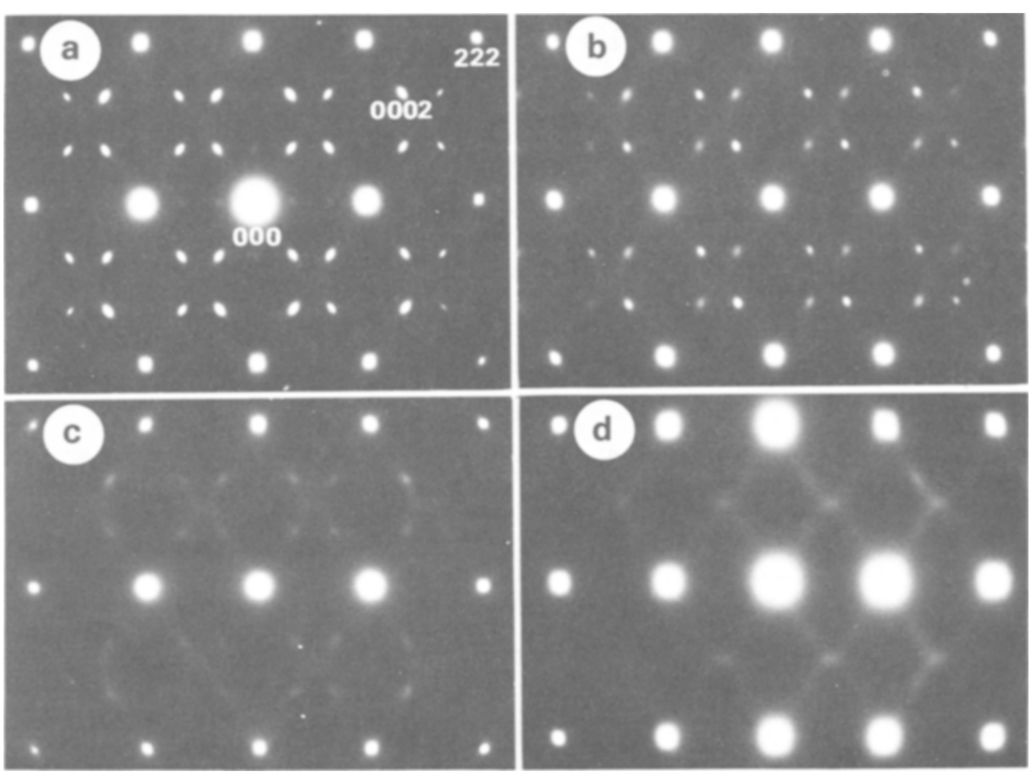

Fig. 1—$(110)_\beta$ electron diffraction patterns from as-quenched alloys. The reflections of two of the four ω-variants other than β are present in the $(110)_\beta$ section. (*a*) Ti-20 pct V. (*b*) Ti-20 pct V-3 pct Sn. (*c*) Ti-20 pct V-0.15 pct O. (*d*) Ti-20 pct V-6 pct Al.

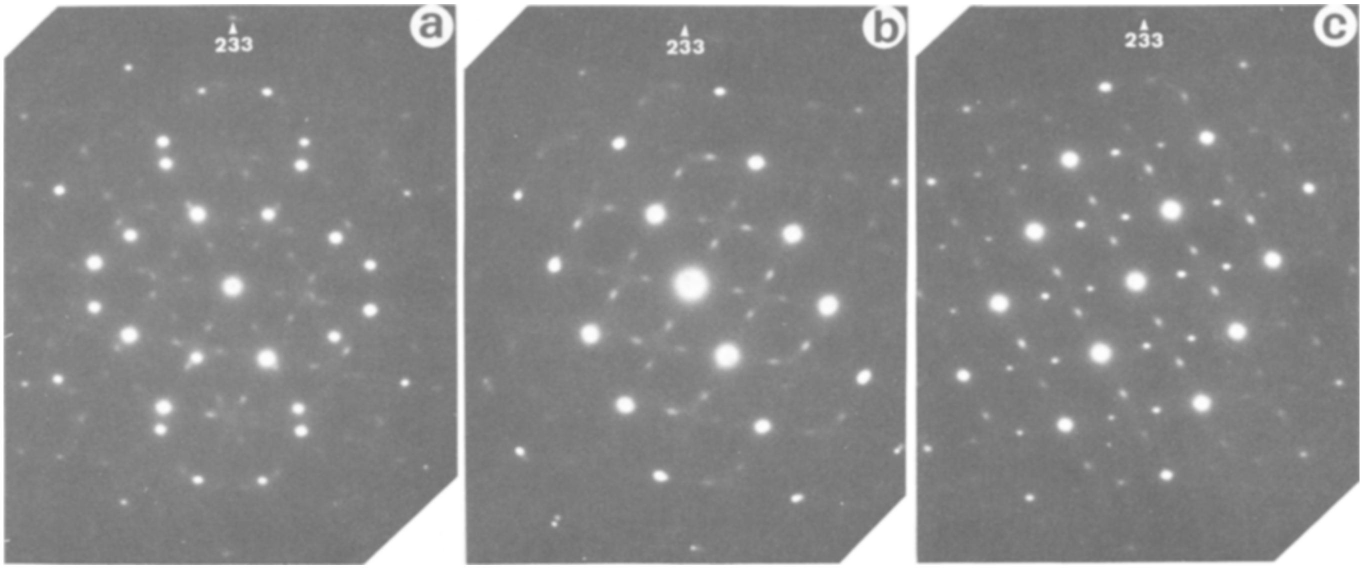

Fig. 3 — Electron diffraction patterns from tensile strained Ti-20 pct V-3 pct Sn. (*a*) {332}⟨113⟩ twin boundary region. (*b*) Matrix. (*c*) Twin.

On the other hand, plastic deformation mode of Ti-20V-6Al was only slip. In Ti-V binary alloys, deformation only by slip was found to occur in alloys containing more than 28 pct vanadium.[1] It appears that aluminum is an effective β-stabilizing element on $\beta \rightarrow \beta + \omega$ transformation.

Thus, it is concluded that ternary additions of tin, oxygen, and aluminum to Ti-20V suppress stress-induced ω-phase transformation in the {332}⟨113⟩ twins or change plastic deformation mode from {332}⟨113⟩ twinning to crystallographic slip. The stability change due to the ternary additions was estimated from Figure 1 in the same manner as used in the binary alloys.[2,3] Namely, ratios of distances of 0002 ω-phase and 222 β-phase reflections in reciprocal space were measured. The obtained results are summarized in Table II, together with tensile properties. It is evident that {332}⟨113⟩ twinning occurs in the alloys with large d^*_{0002}/d^*_{222} ratio (>0.660), and the occurrence of twinning leads to low yield stress and large elongation when the alloys contain an equal content of oxygen. In Ti-20V-0.15O yield stress is considerably high, although dominant mode of plastic deformation is mechanical twinning. This implies that a small addition of oxygen results in significant solution hardening, although it does not lower d^*_{0002}/d^*_{222} ratio significantly.

B. *Commercial β-Phase Alloys*

Figure 4 shows $(110)_\beta$ electron diffraction patterns from as-quenched commercial β-phase alloys. Weak ω-phase reflections are observed in Ti-11.5Mo-6Zr-4.5Sn and Ti-15Mo-5Zr, while diffuse streaking is seen in other alloys. Figure 5 shows true stress–true strain curves during uniform elongation for the as-quenched commercial alloys. The stress-strain curves can clearly be classified into two groups: (I) Ti-11.5Mo-6Zr-4.5Sn and Ti-15Mo-5Zr, (II) Ti-15Mo-5Zr-3Al, Ti-3Al-8V-6Cr-4Mo-4Zr, Ti-15V-3Cr-3Al-3Sn, Ti-8Mo-8V-2Fe-3Al, and Ti-13V-11Cr-3Al. The alloys in the group (I) show low yield stress, high rate of work hardening, and large uniform elongation in comparison with those in the group (II).

Straight plates containing a high density of dislocations were frequently observed in Ti-15Mo-5Zr strained approximately 2 pct, as shown in Figure 6. An electron diffraction pattern taken from the boundary region of a plate (inserted in Figure 6) indicates that the plate is a {332}⟨113⟩ mechanical twin. The same result was obtained in Ti-11.5Mo-6Zr-4.5Sn, as shown in Figure 7. This figure is quite similar to Figure 3. Thus, it is apparent that dominant deformation mode of Ti-15Mo-5Zr and Ti-11.5Mo-6Zr-4.5Sn is {332}⟨113⟩ twinning. On the contrary, only dislocations were observed in the group (II) alloys.

Stability of the commercial β-phase alloys at room temperature was estimated from Figure 4 in the same manner as described above. The results are summarized in Table III, together with tensile properties. This table clearly shows that {332}⟨113⟩ twinning occurs in the alloys with large d^*_{0002}/d^*_{222} ratio (>0.660), and the occurrence of twinning leads to low yield stress and large elongation in good agreement with the results of binary and ternary alloys.

Table II. d^*_{0002}/d^*_{222} Ratios and Tensile Properties for As-Quenched Binary and Ternary β-Phase Alloys

Alloys	Deformation Modes	d^*_{0002}/d^*_{222}	Yield Stress	Elongation
Ti-20V	twinning**	0.667 ± 0.002	328 MPa	37 pct
Ti-20V-3Sn	twinning	0.662 ± 0.002	350	39
Ti-20V-0.15O	twinning	0.661 ± 0.002	555	27
Ti-20V-6Al	slip	0.651 ± 0.003	501	14

Twinning** is accompanied by stress-induced ω-transformation.

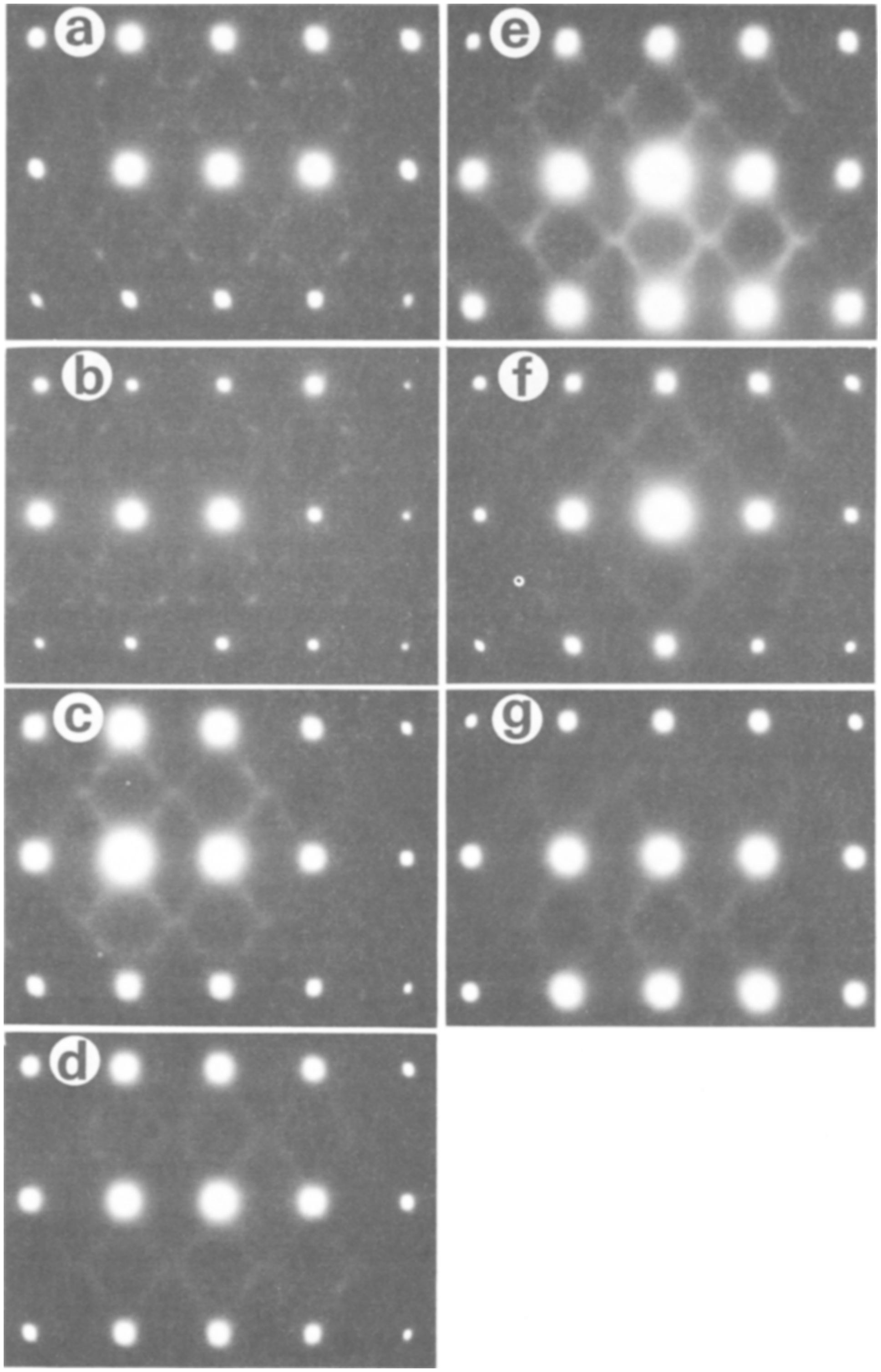

Fig. 4—$(110)_\beta$ electron diffraction patterns from as-quenched β-phase commercial alloys. (*a*) Ti-11.5 pct Mo-6 pct Zr-4.5 pct Sn. (*b*) Ti-15 pct Mo-5 pct Zr. (*c*) Ti-15 pct Mo-5 pct Zr-3 pct Al. (*d*) Ti-3 pct Al-8 pct V-6 pct Cr-4 pct Mo-4 pct Zr. (*e*) Ti-15 pct V-3 pct Cr-3 pct Al-3 pct Sn. (*f*) Ti-8 pct Mo-8 pct V-2 pct Fe-3 pct Al. (*g*) Ti-13 pct V-11 pct Cr-3 pct Al.

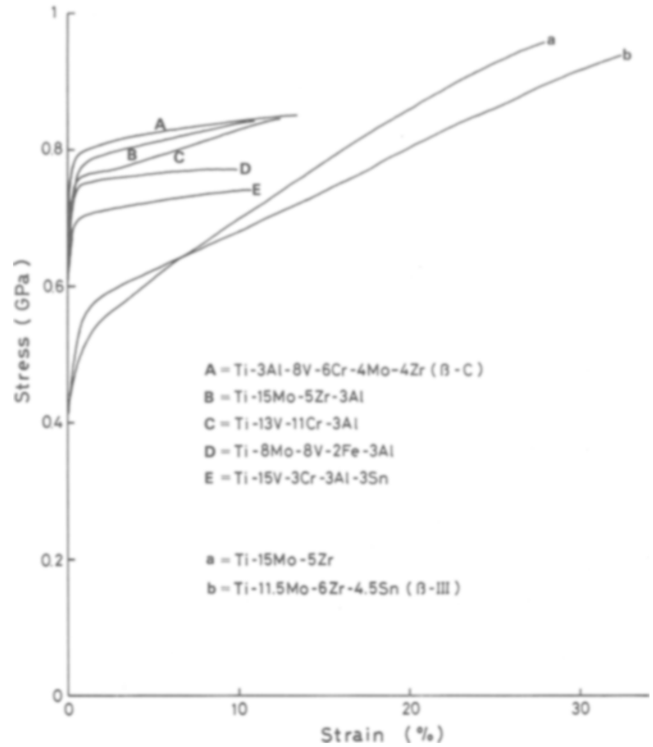

Fig. 5 — Stress-strain curves for as-quenched commercial β-phase alloys.

IV. DISCUSSION

Several commercial β-phase titanium alloys have been developed over the last decade. Compared with $\alpha + \beta$ titanium alloys, they possess the outstanding room temperature fabricability in the solution treated condition. There are a large variety of additional elements and compositions in these alloys. In addition, the reported mechanical properties on the alloys were obtained from the samples which were subjected to different heat treatments, had different grain sizes, and contained a different amount of oxygen. It is very difficult, therefore, to discuss the causes of differences in the reported mechanical properties of respective alloys, although the knowledge of the causes is important in developing new β-phase titanium alloys. In this work we prepared tensile samples having not only the same

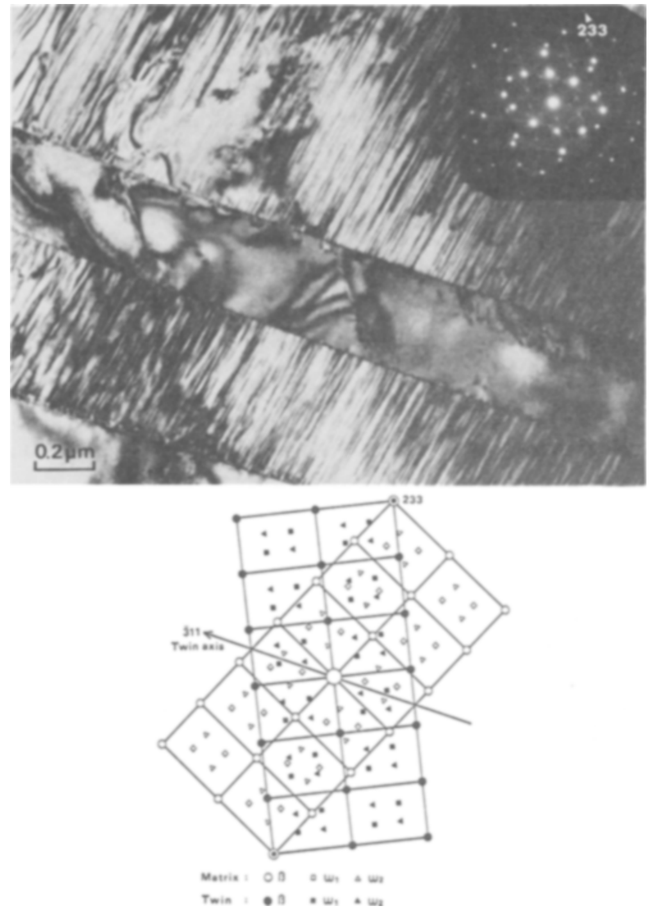

Fig. 6 — Electron micrograph of {332}⟨113⟩ twins in Ti-15 pct Mo-5 pct Zr.

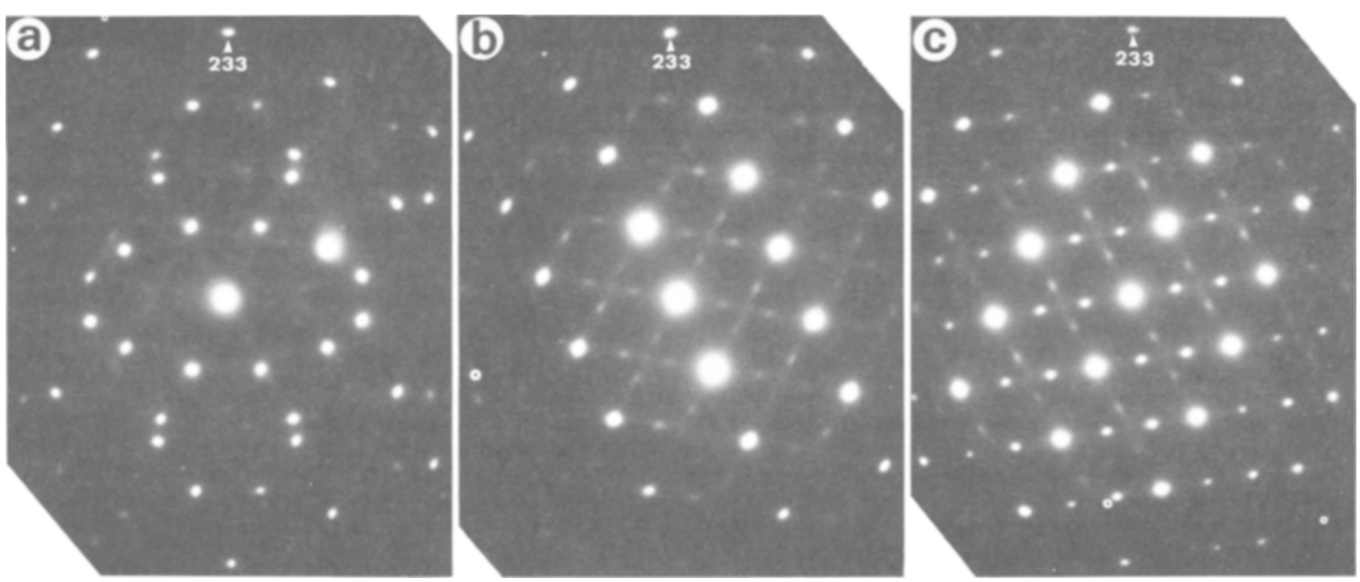

Fig. 7 — Electron diffraction patterns from tensile strained Ti-11.5 pct Mo-6 pct Zr-4.5 pct Sn. (*a*) {332}⟨113⟩ twin boundary region. (*b*) Matrix. (*c*) Twin.

Table III. d^*_{0002}/d^*_{222} Ratios and Tensile Properties for As-Quenched Commercial β-Phase Alloys

Alloys	Deformation Modes	d^*_{0002}/d^*_{222}	Yield Stress	Elongation
Ti-11.5Mo-6Zr-4.5Sn	twinning	0.664 ± 0.002	471 MPa	44 pct
Ti-15Mo-5Zr	twinning	0.664 ± 0.002	451	40
Ti-15Mo-5Zr-3Al	slip	0.658 ± 0.002	769	16
Ti-3Al-8V-6Cr-4Mo-4Zr	slip	0.650 ± 0.003	697	11
Ti-15V-3Cr-3Al-3Sn	slip	0.650 ± 0.003	665	17
Ti-8Mo-8V-2Fe-3Al	slip	0.649 ± 0.003	680	16
Ti-13V-11Cr-3Al	slip	0.643 ± 0.003	721	18

compositions as the commercial β-phase alloys but also an approximately constant grain size (0.2 mm) and oxygen content (<400 ppm). Consequently, the present results are considered to be suitable for comparison of the relation between the tensile properties and the stability of β-phase in the commercial β alloys, although the tensile properties are slightly different from the reported values on the commercial alloys. It was found in the present experiments that dominant mode of plastic deformation was {332}⟨113⟩ mechanical twinning in Ti-11.5Mo-6Zr-4.5Sn and Ti-15Mo-5Zr. The twinning has been reported to be dominant in both alloys produced in commercial quantities (containing more than 1000 ppm oxygen).[6,7,8] This fact suggests that the increase in oxygen content does not lower d^*_{0002}/d^*_{222} ratio significantly if the instability of β-phase estimated by d^*_{0002}/d^*_{222} affects plastic deformation mode, although it leads to remarkable solid solution hardening. Actually, d^*_{0002}/d^*_{222} of Ti-15Mo-5Zr produced in commercial quantities have been measured to be 0.663.[3] This oxygen effect can also be seen in Table II, where dominant deformation mode in Ti-20V and Ti-20V-0.15O is {332}⟨113⟩ twinning in spite of large difference in yield strength. This is considered to result from a slight decrease of d^*_{0002}/d^*_{222} ratio (from 0.667 ± 0.002 to 0.661 ± 0.002). It has been reported that addition of more than 0.21 pct oxygen to Ti-20V is needed to change plastic deformation mode from twinning to slip.[11]

It is interesting to note that dominant mode of plastic deformation in β-phase titanium alloys (Table II) and commercial alloys (Table III) varies from twinning to slip at the critical values of $d^*_{0002}/d^*_{222} = 0.660$ in good agreement with that in binary β alloys Ti-V,[3] Ti-Mo,[3] Ti-Fe,[3] and Ti-Cr.[2]

de Fontaine *et al.*[9] have shown that diffuse streaking in the diffraction patterns of the β-phase is changed gradually into the sharp ω-reflections by cooling certain metastable β Ti-V and Ti-Mo alloys to cryogenic temperatures. On the other hand, Terauchi *et al.*[10] have confirmed that the diffuse streaking in the diffraction patterns of the certain β-phase Ti-Mo alloys is changed into the sharp ω-reflections by aging. The composition range where ω transformation occurs readily during cooling or aging is in good agreement with the range where {332}⟨113⟩ mechanical twinning appears during deformation.

Thus, it can be concluded that {332}⟨113⟩ mechanical twinning appears when the β-phase is very unstable and ready to transform into ω-phase. This conclusion seems to disagree with Blackburn and Feeney's explanation[6] that since the atomic movements in the ω-phase during mechanical twinning are easier for {332}⟨113⟩ twinning than {112}⟨111⟩, {332}⟨113⟩ twinning appears when small particles of ω-phase are present. Instead, the present conclusion is most likely to be in accord with Oka and Taniguchi's suggestion.[13] According to them, shuffling of one-half of atoms into the direction different from that of twinning shear, which is needed to create {332}⟨113⟩ twinning,[14,15] is ready to occur in the unstable β-phase.

de Fontaine *et al.*[9] have shown that many features of diffuse intensity patterns in β-phase can be interpreted in terms of elastic energy of the bcc lattice under the harmonic approximation. According to them, d^*_{0002}/d^*_{222} decreases with increasing solute content in agreement with observations of binary titanium[2,3] and zirconium[16] alloys.

The present findings show clearly that d^*_{0002}/d^*_{222} is useful to estimate the stability of β-phase and the resultant plastic deformation modes in the commercial β-phase titanium alloys. As seen in Table III, d^*_{0002}/d^*_{222} ratios of the commercial β-phase titanium alloys developed experimentally or empirically fall on the values between 0.664 ± 0.002 and 0.643 ± 0.003. On the basis of the results, therefore, we can design β-phase titanium alloys having good cold deformability by controlling d^*_{0002}/d^*_{222}. For example, maximum ductility can be obtained by using mechanical twinning at d^*_{0002}/d^*_{222} ratios between 0.667 and 0.660. Moderate ductility is attained by crystallographic slip at d^*_{0002}/d^*_{222} ratios between 0.660 and 0.643. The alloy with d^*_{0002}/d^*_{222} below 0.643 is considered to have disadvantages for age-hardenability or strength-to-weight ratio.

Actually, various properties as well as cold deformability, *e.g.*, strength, fracture toughness, low stress corrosion cracking susceptibility, are required for β-phase titanium alloys. We add many elements to titanium for the purpose of improving the properties. For maximum efficiency, it is now emphasized that the method of measuring d^*_{0002}/d^*_{222} ratio should be introduced.

V. CONCLUSIONS

From comparison of the tensile properties of several β-phase titanium alloys at an approximately constant grain size and oxygen content, the following conclusions can be drawn:

1. Dominant plastic deformation mode of commercial β-phase titanium alloys in quenched condition is {332}⟨113⟩ twinning for Ti-11.5Mo-6Zr-4.5Sn and Ti-15Mo-5Zr, and crystallographic slip for Ti-15Mo-5Zr-3Al, Ti-3Al-8V-6Cr-4Mo-4Zr, Ti-15V-3Cr-3Al-3Sn, Ti-8Mo-8V-2Fe-3Al, and Ti-13V-11Cr-3Al.
2. {332}⟨113⟩ twinning leads to low yield stress and large elongation, while slip results in high yield strength and small elongation.
3. Stability of β-phase is estimated from ω-reflections and diffuse streaking in electron diffraction patterns.

4. {332}⟨113⟩ twinning appears in the most unstable β-phase alloys. As the stability of β-phase increases, deformation mode varies from twinning to slip.
5. These results are in good agreement with those in binary and ternary β-phase titanium alloys.

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