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Plastic deformation mode of retained β phase in β-eutectoid Ti–Fe alloys

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Plastic deformation mode of β-eutectoid Ti–Fe alloys has been investigated at 300 and 77 K in a retained β single phase (containing athermal ω phase). Surface analysis and transmission electron microscopy show that $\{332\}\langle 113\rangle$ twinning and $\langle 111\rangle$ slip appear to be dependent on orientation, composition and deformation temperature. The $\{332\}\langle 113\rangle$ twinning appears only in metastable β regions adjacent to the M_s curve in good agreement with previous work in β-isomorphous alloys. Orientation dependence for occurrence of the preferential $\{332\}\langle 113\rangle$ twinning among the twelve equivalent twinning systems can be explained in terms of the Schmid factor and the polarization of twinning shear. It is concluded that the $\{332\}\langle 113\rangle$ twinning is common for β-titanium alloys and related to the instability of the β phase.

1. Introduction

The addition of transition metal elements to titanium is known to stabilize the bcc(β) phase at low temperatures in these systems. Titanium alloys composed of the metastable bcc phase have been practically used due to their superior ductility at room temperature and subsequent age-hardenability. It has been confirmed in many metastable β phase alloys [1–9] that the superior ductility results from the occurrence of unusual $\{332\}\langle 113\rangle$ mechanical twinning, since $\{112\}\langle 111\rangle$ twinning is found extensively in other bcc metals and alloys. Recent studies [10, 11] on temperature and composition dependence of deformation mode in β phase titanium alloys have revealed that the $\{332\}\langle 113\rangle$ twinning occurs in a metastable β phase region adjacent to the curve showing composition dependence of M_s (martensitic start) temperature. For an alloy with its composition within this region, ω phase forms easily on ageing at relatively low temperatures. Therefore, the β phase is considered to be very unstable. By increasing the content of the alloying element, away from the M_s curve, the deformation mode was found to change from twinning to slip [10, 11]. The β phase is now hard to decompose into the ω phase on ageing at low temperatures. Therefore, the $\{332\}\langle 113\rangle$ twinning may be related to thermal instability of the β phase.

It is, however, in β-isomorphous alloys such as Ti–Mo [1–9], Ti–V [4, 10] and Ti–Nb [11] that the $\{332\}\langle 113\rangle$ twinning has been reported. In β-eutectoid alloys such as Ti–Fe, Ti–Cr and Ti–Mn, a metastable β phase can also be retained at low temperatures by rapid cooling. The β phase is known to decompose on ageing in a similar manner to that in β-isomorphous alloys [12]. Therefore, the objective of the present paper is to investigate the plastic deformation mode in a metastable β-eutectoid titanium alloy for a further understanding of $\{332\}\langle 113\rangle$ twinning.

2. Experimental procedure

Ti–4, 4.5, 5, 6 and 10 wt % Fe alloys were prepared by arc-melting of sponge titanium (> 99.8 wt %) and electrolytic iron (> 99.9 wt %) in an argon atmosphere. The arc-melted buttons were hot-rolled at 1200 K to approximately 3 mm thick plates. Bars with a square cross-section of 2.5 mm × 2.5 mm were obtained from the plates by cutting and grinding. They were subjected to strain annealing for 36 ksec at 1473 K for grain growth. Single crystals of 2.5 mm × 2.5 mm × 5 mm were spark-cut from the bars having large grains of bamboo type structure. The single crystals were sealed in a vacuum quartz tube, homogenized at 1273 K for 3.6 ksec and quenched into iced water. After mechanical and chemical polishing, the samples were deformed at 300 and 77 K using an Instron type testing machine at a strain rate of $1.7 \times 10^{-3}\,\mathrm{sec}^{-1}$. The deformation mode was determined by two surface trace analysis and transmission electron microscopy (TEM) techniques.

3. Results and discussion

Recently, Yamane and Ito [13] measured the M_s temperature of Ti–Fe alloys. According to them, the martensitic transformation can be suppressed by rapid cooling in Ti–Fe alloys containing $\gtrsim$ 4 wt % Fe. Electron microscopic observations were performed on thin foils obtained from as-quenched compression samples of Ti–4, 4.5, 5, 6, 10% Fe alloys. In this study martensitic transformation was suppressed in all the samples. Fig. 1 shows the $\{011\}_\beta$ diffraction patterns of as-quenched Ti–10% Fe, Ti–6% Fe and Ti–4% Fe. While weak and diffuse ω reflections are present in the as-quenched Ti–10% Fe alloy, clear and intense ω reflections are present in the Ti–6% Fe and Ti–4% Fe alloys. Dark field micrographs obtained from ω reflections in Figs. 1b and c show the presence of a uniform dispersion of fine ω particles in the β matrix (Fig. 2). The ω particles seem to be ellipsoidal in contrast to

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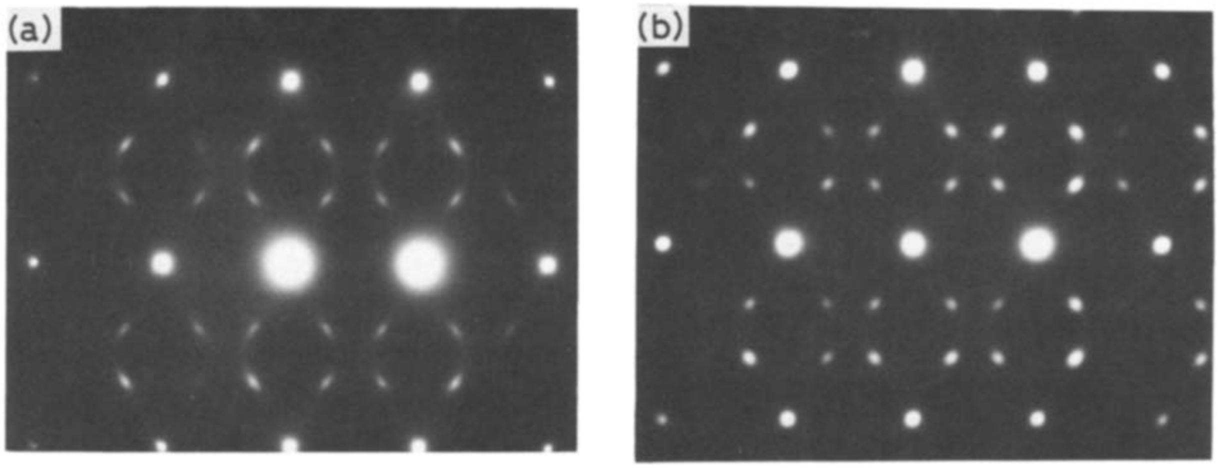

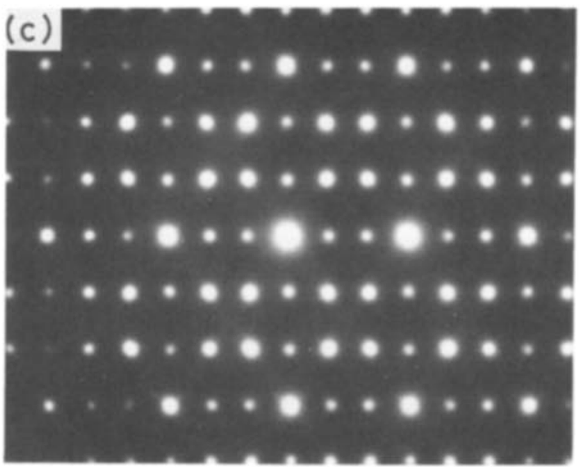

Figure 1 Electron diffraction patterns of quenched (a) Ti–10% Fe, (b) Ti–6% Fe and (c) Ti–4% Fe (Zone axis $[0\,1\,1]_\beta$).

those of aged β Ti–Fe alloys containing cuboidal particles [14]. This result may be related to the fact that ω phase morphology is controlled by the lattice misfit between the precipitate and the β matrix [14], since the misfit in β Ti–Fe alloys increases remarkably on ageing [15].

The Ti–Fe alloys in the single phase β (containing only athermal ω phase) were deformed below 1% plastic strain prior to two surface trace analysis and electron microscopy. In Ti–10% Fe, slip appeared independent of the crystallographic orientation of the compression axis (shown as a, d, e . . . in Fig. 3) and deformation temperature. Fig. 3 shows the result at 300 K, indicating that the observed slip planes (a′, d′, e′ . . .) correspond with the maximum resolved shear stress planes. Compression axes have a tendency to be concentrated around [0 0 1], presumably by causing a recrystallization texture during strain annealing heat treatments. The observed slip planes at 77 K were found to deviate from the maximum resolved shear stress plane to $(\bar{1}\,\bar{1}\,2)$. These results are in good agreement with previous work in Ti–V [16]. In Ti–6% Fe, only slip also appeared on deformation at 300 and 77 K.

On the other hand, stress-induced products (SIP) as well as slip appeared in Ti–5% Fe depending on the orientation of compression axis. Small discontinuous serrations accompanied by significant work hardening occurred in a stress–strain curve when SIP appeared, while continuous parabolic work hardening was seen when slip occurred. Typical optical micrographs of SIP in Ti–5% Fe deformed at 77 K is shown in Fig. 4a. SIP was found to form approximately along $(\bar{3}\,3\,2)$. If the SIP is mechanical twinning in the same type as that in β-isomorphous alloys such as Ti–V, Ti–Mo and Ti–Nb, the twinning system is regarded as $(\bar{3}\,3\,2)[\bar{1}\,1\,\bar{3}]$. Then thin plates having the orientation illustrated in Fig. 4b were spark-cut from the compressed sample and thinned electrolytically for electron microscopy. Fig. 5a shows an electron micrograph taken from a region containing the $(\bar{3}\,3\,2)$ boundary in the thin foil. Figs. 5b and c show an electron diffraction pattern at the boundary and the corresponding key diagram, respectively, indicating that the SIP is the mechanical twin of $(\bar{3}\,3\,2)[\bar{1}\,1\,\bar{3}]$ twinning system. The microstructure is clearly divided

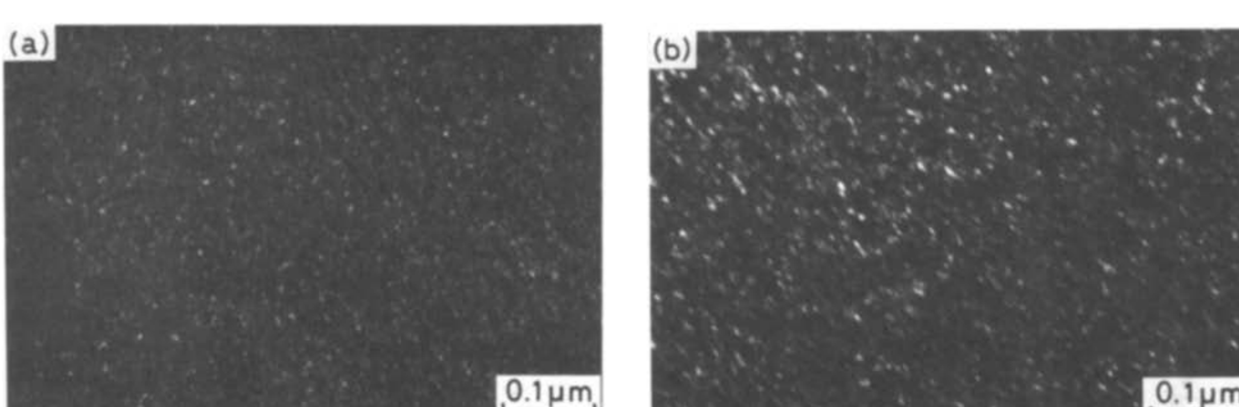

Figure 2 Dark-field micrographs showing ω particles in a b c c matrix of quenched (a) Ti–6% Fe and (b) Ti–4% Fe.

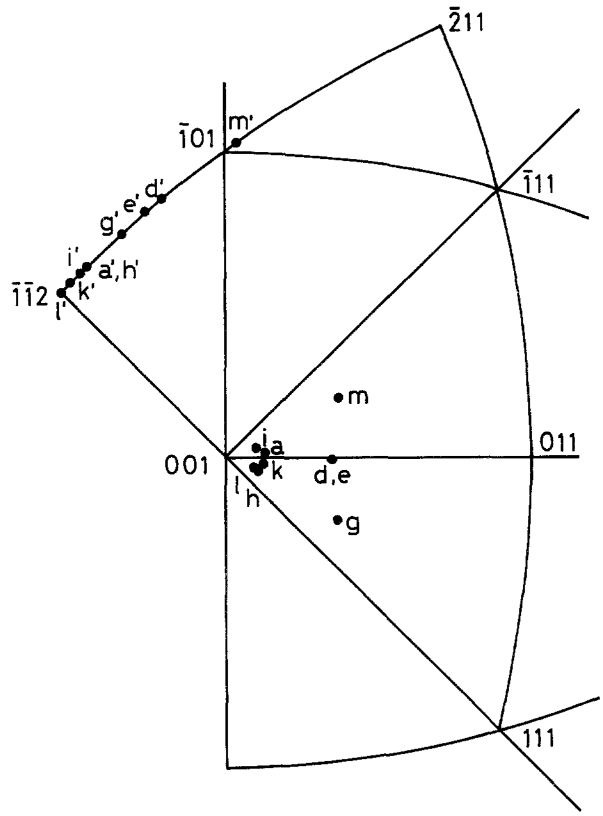

Figure 3 Observed slip planes in Ti–10% Fe deformed at 300 K. a′, d′, . . . represent the slip planes in samples with the compression axes a, d, . . . , respectively.

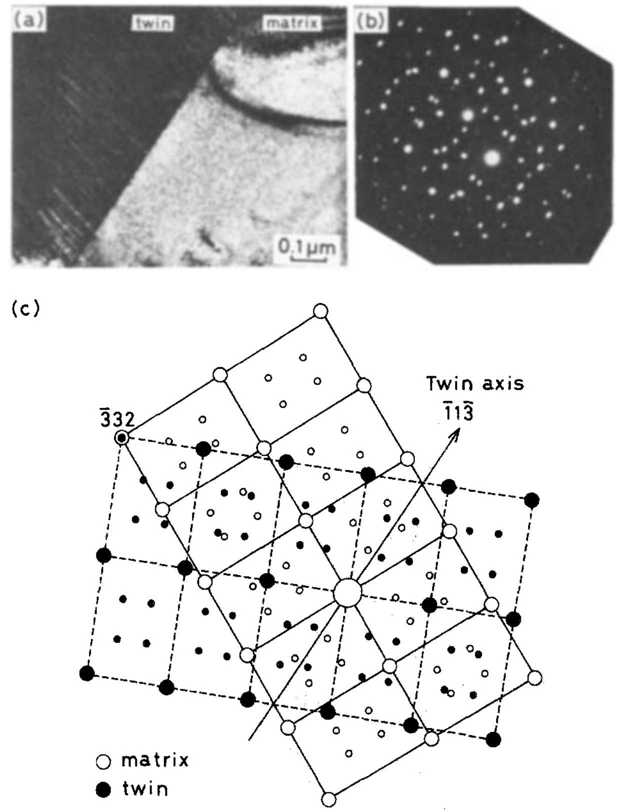

Figure 5 Bright-field micrograph showing {3 3 2} twin boundary in Ti–5% Fe deformed at (a) 77 K, (b) selected area diffraction pattern at the boundary and (c) key diagram.

by the boundary; that is, the twin contains a high density of dislocations, while the matrix with a very low density of dislocations is characterized by mottled contrast resulting from the ω phase. A highly damaged region has been found in a matrix adjacent to the {3 3 2} twin boundary of Ti–Mo [1] and Ti–Nb [11] alloys in contrast to the present results. The Ti–Mo and Ti–Nb alloys contained very diffuse ω phase and showed sharp load drops on deformation, being different from the present alloys. Therefore, the

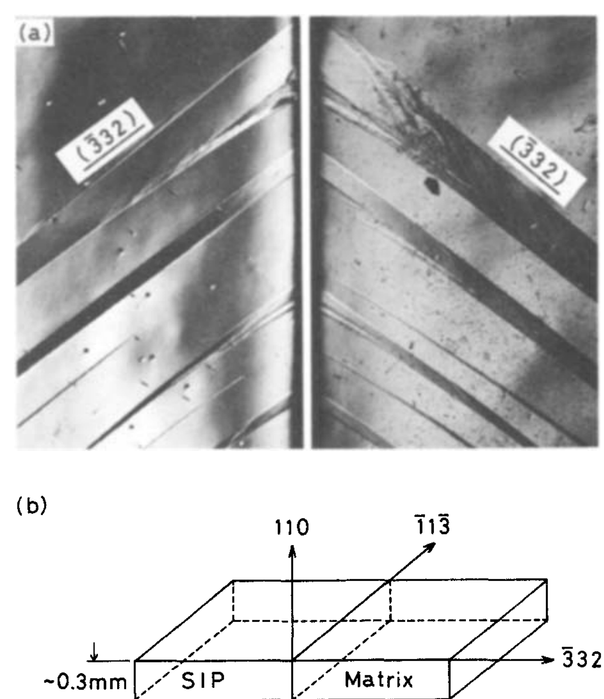

Figure 4 Optical micrographs showing stress-induced products in Ti–5% Fe deformed at (a) 77 K and (b) schematic diagram of the thin plate obtained from the compressed sample for electron microscopy.

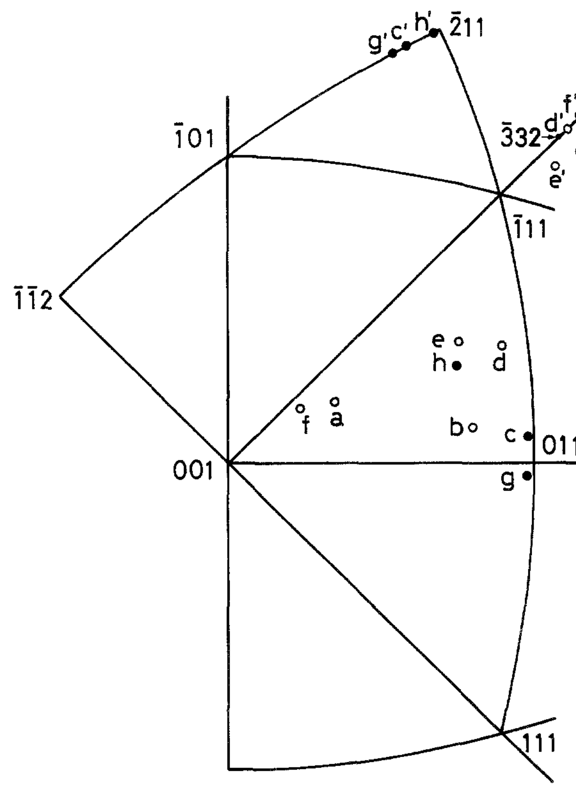

Figure 6 Observed slip and twinning planes in Ti–5% Fe deformed at 77 K. a′, b′, c′ . . . represent the slip planes (closed circles) or twinning planes (open circles) in samples with the compression axes a, b, c . . . , respectively.

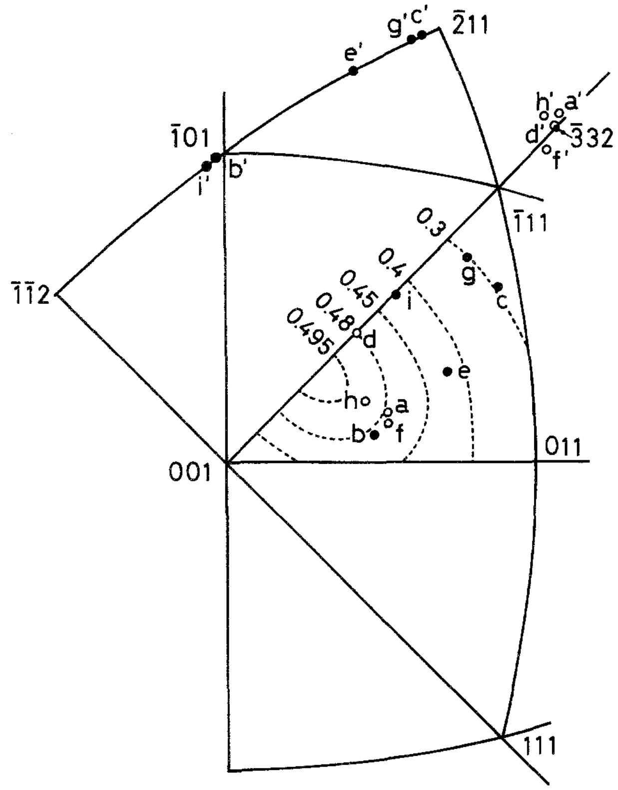

Figure 7 Observed slip and twinning planes in Ti–5% Fe deformed at 300 K. Letters and symbols represent the same meaning as in Fig. 6. Dotted contours indicate the Schmid factors for $(\bar{3}\,3\,2)[\bar{1}\,1\,\bar{3}]$ twinning.

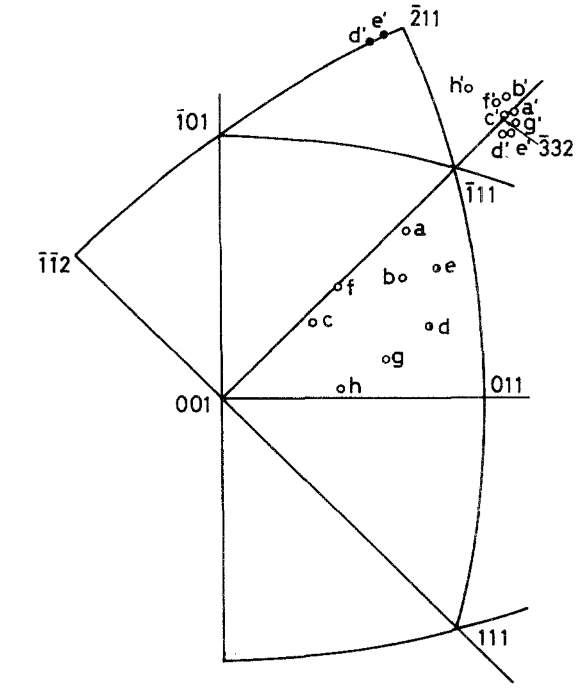

Figure 8 Observed twinning and slip planes in Ti–4.5% Fe deformed at 300 K. a′, b′, c′ represent the twinning planes in samples with the compression axes a, b, c . . . , respectively. ◑ represents coexistence of twinning and slip.

microstructure at the twin boundary may be related to the difference of accommodation of $\{3\,3\,2\}\langle 1\,1\,3\rangle$ twinning. Fig. 6 shows the orientation dependence of deformation mode in the Ti–5% Fe deformed at 77 K. $(\bar{3}\,3\,2)[\bar{1}\,1\,\bar{3}]$ twinning is found to be preferential, although there is a considerable scatter in the observed twinning planes. Slip seems to appear in samples having a compression axis close to [0 1 1]. Fig. 7 shows the orientation dependence of deformation mode in the Ti–5% Fe alloy deformed at 300 K. Compared with Fig. 6, slip is preferential to twinning. Recently, it has been shown that the operative twinning system among the twelve $\{3\,3\,2\}\langle 1\,1\,3\rangle$ systems is controlled by the Schmid factor and the polarization of twinning shear [10]. The result in Fig. 7 can be explained qualitatively on the basis of this concept, namely the $(\bar{3}\,3\,2)[\bar{1}\,1\,\bar{3}]$ twinning system is favourable to operate under compressive stress in all orientations within the [0 0 1]–[0 1 1]–$[\bar{1}\,1\,1]$ triangle [10]. The Schmid factors for the $(\bar{3}\,3\,2)[\bar{1}\,1\,\bar{3}]$ system are illustrated by dotted contours in Fig. 7. One can see that twinning occurs in the samples with the large Schmid factors for $(\bar{3}\,3\,2)[\bar{1}\,1\,\bar{3}]$, excluding sample b (the reason why twinning did not occur in the sample b is uncertain at present, but a critical value of the Schmid factor for producing the twinning may be close to 0.48). Any other systems which are favourable under compressive stress, do not have so large a value as ~ 0.48. On the other hand, samples g and c have large Schmid factors (~ 0.48) for $(2\,3\,3)[\bar{3}\,1\,1]$. However, this twinning system can operate under tensile stress [10].

Fig. 8 shows the result of Ti–4.5% Fe deformed at 300 K. The $(\bar{3}\,3\,2)[\bar{1}\,1\,\bar{3}]$ twinning is seen to be preferential in all the orientations. In samples d and e slip was coexistent with the twinning. The $(\bar{3}\,3\,2)[\bar{1}\,1\,\bar{3}]$ twinning was also preferential in Ti–4.5% Fe deformed at 77 K. The deformation mode could not be determined in Ti–4% Fe, since the alloy often failed

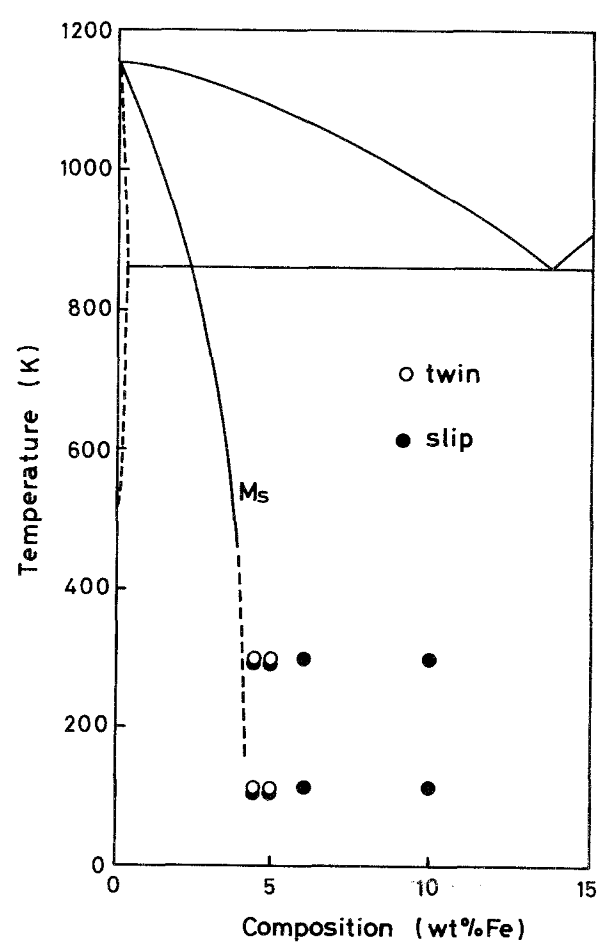

Figure 9 Temperature and composition dependence of plastic deformation mode in β-Ti–Fe alloys.

without any macroscopic plastic strain. The results obtained are summarized in Fig. 9. This figure clearly indicates that the $\{332\}\langle 113\rangle$ twinning appears in the β region adjacent to the M_s curve. This result is in good agreement with that of β-isomorphous alloys [10, 11]. Thus, it is concluded that the $\{332\}\langle 113\rangle$ twinning is common for β-titanium alloys and related to the instability of the β phase.

It is not possible at present to explain why the unusual twinning system is commonly observed in metastable β-titanium alloys. To understand the mechanism of the $\{332\}\langle 113\rangle$ twinning, further studies should be performed on detailed electron microscopic observations of microstructures inside a twin in relation to a high density of dislocations and the ω phase, and accommodation microstructures in a matrix adjacent to the twin boundary. This work is in progress.

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