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Change in anisotropy of mechanical properties with β-phase stability in high Zr-containing Ti-based alloys

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Abstract

High Zr-containing β-type Ti-based alloys were designed using electronic parameters to investigate experimentally the effect of β-phase stability on their elastic and plastic properties. Texture structures formed by cold rolling or recrystallization were related closely to the β-phase stability and hence affected the mechanical properties. In tensile tests, as the β-phase stability decreased, non-linearity in the elastic zone was enhanced and the work hardening tended to be diminished. Also, it was found that the lower β-phase stability led to the weaker anisotropy of plastic properties, but to the stronger anisotropy of elastic properties.

Keywords: High Zr-containing Ti alloys; β-Phase stability; Mechanical properties; Texture; Elastic and plastic properties; Anisotropy

1. Introduction

Recently, considerable efforts have been devoted to exploring novel β-titanium alloys for biomedical applications because of their superior properties such as the superelasticity, shape memory, low Young's modulus, satisfactory biocompatibility, and better formability compared to the α and α + β titanium alloys [1–7]. The mechanical properties of the β-titanium alloys depend strongly on the presence of several phases (e.g., ω-phase and martensitic α″-phase) in them. The appearance of these phases could be controlled by either the optimized alloy design [6,7] or the materials processing [4].

Most of the β-Ti alloys possess good workability. It is possible to fabricate a cold-rolled sheet of the alloys by a reduction ratio higher than 90%. In this case strong textures are developed and the anisotropy in elastic and plastic properties is induced inevitably to the sheet, resulting in the modification of alloy properties such as the elastic modulus, elastic strain, Poisson's ratio, strength, ductility, toughness, magnetic permeability and the energy of magnetization [8]. In other words, the elastic and plastic properties of the alloy may be improved by using an orientation effect arising form the textures. It is, therefore, important to examine which kinds of textures can be developed in the β-Ti alloys under the given conditions of thermo-mechanical treatment and to investigate the texture effect on the elastic and plastic properties.

Recently an electronic parameter *e*/*a* (electron-per-atom ratio) and the $\overline{Bo}$–$\overline{Md}$ diagram have been used for the design of new β-type alloys (the so-called GUM metals) [6]. Here, $\overline{Bo}$ is the average bond order between atoms and $\overline{Md}$ is the average d-orbital energy level of the element in the alloy. This diagram has been explained elsewhere [7].

The β-type Ti-based alloys deform by either slip or twin mechanism [1,5]. The stress-induced martensitic transformation also takes place in some alloys upon applying external stress to them [4,5]. These phenomena emerge depending on the β-phase stability and hence will be controlled by alloying. Also, it is known that the slip/twin boundary is close to the β/β + ω + (α″) boundary [7]. At this boundary, the elastic anisotropy factor, $A = C_{44}/C'$, is rather high since the value of the elastic shear modulus, $C' = (C_{11} - C_{12})/2$, is diminishing as the alloy approaches this boundary [9,10]. Also, it has been reported that the $(C_{11} - C_{12})/2$ value approaches zero when the *e*/*a* value is about 4.24 [9]. This is a reason why, in this work the *e*/*a* value was kept at 4.24 in all the designed alloys.

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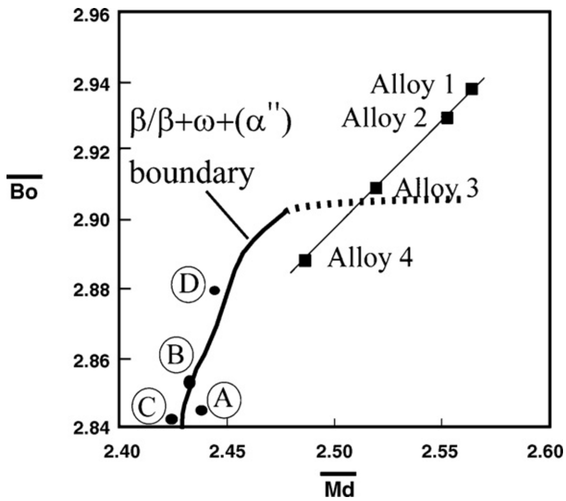

Fig. 1. Extended $\overline{\text{Bo}}$–$\overline{\text{Md}}$ diagram showing the $\beta/\beta+\omega$ $(+\alpha'')$ phase boundary and the location of the designed alloys, 1–4. Also, the alloys A, B, C, and D are 35 mass% Nb–4 mass% Sn, Ti–24Nb–3Al, Ti–35 mass% Nb–7.9 mass% Sn, and Ti–22Nb–6Ta, respectively.

This work is an extension of our previous study [7], and the β-phase stability effect was investigated on the anisotropy of mechanical properties and on the texture developed by cold rolling of four high Zr-containing alloys. Their alloy locations are indicated in the $\overline{\text{Bo}}$–$\overline{\text{Md}}$ diagram shown in Fig. 1. These β-type Ti alloys appear to be located in the positions across the $\beta/\beta+\omega+(\alpha'')$ phase boundary.

2. Experimental procedure

As explained above, four high Zr-containing alloys were designed and the chemical compositions are listed in Table 1. In this paper, all the compositions are given in atomic percent units unless otherwise noted. These alloys were prepared by the arc-melting of an appropriate mixture of pure metals (purity: 99.99%) under a high purity argon gas atmosphere. The button-shaped specimens with average 7.5 mm in thickness were cut and homoginized at 1273 K for 7.2 ks, and then cold rolled to the plate with 4.5 mm thick, followed by the solution treatment at 1223 K for 1.8 ks. This primary solution-treated specimen is called STBCR specimen. Subsequently, the specimen was cold rolled by 30 or 60 or 90% reduction in thickness. The cold-rolled specimen is called CR specimen hereafter. The 90% CR specimen was then solution treated at 1223 K for 1.8 ks. This finally solution-treated specimen is called ST specimen hereafter.

The phases existing in the specimen were identified by the conventional X-ray diffraction (XRD) using a Ni-filtered Cu Kα radiation. Electron back scattered diffraction (EBSD) analysis was also made using a HITACHI S-3000H scanning electron microscope (SEM) equipped with a OXFORD INCA Crystal EBSD detector, operated at an acceleration voltage of 20 kV and a tilt angle of 71°. The microstructural characterization was performed using the optical microscope (OM) and the scanning electron microscope (SEM). The micro-Vickers hardness was measured at a load of 20 N. The tensile test was carried out with both the ST and 90% CR specimens where the tensile axis was set to be parallel to either the rolling direction (RD) or the direction of 45° inclined to RD (45°). A cross-head strain rate used was fixed at 1.6×10^{-4} s^{-1}. Also, tensile loading and unloading tests were repeated at the stage of 1 and 2% strain, then loading was continued until fracture. The tensile strain was detected by both a standard strain gauge and a CCD camera.

Table 1
Chemical compositions (at.%) of the designed alloys

	V	Cr	Mo	Nb	Ta	Zr
Alloy 1	4		2	9	7	30
Alloy 2			3	15	3	25
Alloy 3		1		8	14	15
Alloy 4				4	20	5

3. Results and discussion

3.1. Change in β-phase stability with four alloys

The β-phase stability increases with increasing content of the β-stabilizing elements. Shown in Fig. 2 are the X-ray diffraction patterns taken from the four designed alloys in the STBCR condition, the 60% cold rolled (60CR) condition and the ST condition. In all the conditions, a single β-phase was predominant in the alloys 1, 2, and 3. Only in the alloy 4, the martensite α″-phase coexisted with the β-phase, as shown in Fig. 2(d). Therefore, the $\beta/\beta+\alpha''$ boundary shown in Fig. 1 is located between the alloys 3 and 4 as indicated by a dotted curve. So, the alloy 3 is the least stable single β-phase alloy which is defined as the alloy containing a least amount of the β-stabilizing elements to get a β-single phase. According to the $\overline{\text{Bo}}$–$\overline{\text{Md}}$ diagram shown in Fig. 1, the alloy closer to the $\beta/\beta+\alpha''$ boundary has the lower β-phase stability, so the β-phase stability in these alloys decreased in the order, alloy 1 > alloy 2 > alloy 3 > alloy 4. The stability was highest in the alloy 1 and lowest in alloy 4.

3.2. Textures developed by cold rolling

It is known that the cold rolling texture is formed in conventional β-type Ti alloys. As a result, the measured X-ray peak intensity ratio of the cold-rolled specimen, I_{Rcr}, defined as $I_{\text{Rcr}} = I_{\{200\}\beta}/I_{\{110\}\beta}$, changed with cold rolling. Here, $I_{\{200\}\beta}$ and $I_{\{110\}\beta}$ are the X-ray peak intensities of the $\{200\}_\beta$ and $\{110\}_\beta$ reflections, respectively. The I_{Rcr} increased with the reduction ratio of cold rolling. As is evident from Fig. 2 and Table 2, when the alloys were cold rolled by 60%, I_{Rcr} tended to increase with increasing β-phase stability. In other words, a cold rolling texture was developed in the way that the {2 0 0} planes aligned parallel to the rolling plane preferentially. This texture was formed more readily in the order, alloy 1 > alloy 2 > alloy 3 > alloy 4, in agreement with the order of the β-phase stability.

Fig. 3a–c shows {2 0 0}, {1 1 0} and {1 1 2} pole figures obtained from a 90% cold rolled (CR) specimen of the alloy 2. The center of the pole figures corresponds to the direction normal to the specimen surface (ND). The right and the top of the pole figures corresponding to the rolling direction (RD) and the