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Mechanical behaviors of Ti–V–(Al, Sn) alloys with α′ martensite microstructure

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A B S T R A C T

The mechanical properties and deformation behavior of Ti–V–(Al, Sn) alloys with a α′ martensite microstructure are examined in this work. In as-quenched Ti–V–Al alloys consisting of α′ martensite microstructure with compositions near α − (α + β), the cold-rolling ability drastically decreases from a reduction of more than 75% to less than 40% with an increase in Al content. While, excellent cold-rolling ability (more than 75% reduction) is seen in the constituent phases of α′ in the compositional region near β − (α + β), and α″ and β. On the other hand, in as-quenched Ti–V–Sn alloys, excellent cold-rolling ability with a reduction of more than 80% that is independent of phase constituents is seen. In α′ martensite Ti–V–Al–Sn alloy with compositions in α − (α + β) that is tensile-deformed at a strain of 5%, a homogenous deformation substructure with straight dislocations is seen in alloy that with a low Al content. In contrast, the dislocation substructure begins to exhibit a planar configuration with increasing Al content. This change in deformation substructure due to increasing Al content is considered to cause the low ductility in α′ martensite Ti alloy that contains high Al. On the other hand, in α′ martensite Ti–V–Al alloy with a composition near β − (α + β), plastic deformation is found to be proceeded via the activation of the basal ⟨*a*⟩ slip and the $\{10\bar{1}1\}$ twin with a homogenous deformation substructure. As a consequence, this deformation mode is considered to cause the high ductility in the α′ phase as compared with the equilibrium (α + β) phase.

1. Introduction

Ti alloys are widely used in industrial applications due to their excellent mechanical properties combined with low density. In particular, a Ti–V–Al alloy system, such as Ti–6Al–4V, Ti–6Al–6V–2Sn and others, is most widely used as an industrial Ti alloy. In general, Ti alloys are classified as α, α + β, and β alloys, with further subdivision into near α – and metastable β alloys. In (α + β)-Ti alloys, the alloy-type can be further subdivided into near α-type and near β-type depending on alloy composition whether it is close to α-type or β-type. Metastable β alloys belong to near β − (α + β) type alloy. The microstructures of Ti alloys, which can be controlled by hot forging and heat treatment, are well known to strongly influence on their mechanical properties. The microstructures obtained from supratransus solution treatment depend on the degree of homogenization during solution treatment and the cooling rate from the β field. Quenching suppresses the diffusion-controlled β(B.C.C.)-to-α(H.C.P.) transformation and leads to a martensitic transformation in which the β phase becomes needle-like α′(H.C.P.) martensite or α″ (orthorhombic) martensite, depending on the alloy content.

We have recently presented new types of Ti–V alloys that are composed of fully α′ martensite, with a low Young's modulus and high strength [1,2] for industrial applications such as screw and suspension springs. By optimizing the alloy content, this α′ martensite-type Ti–V alloy exhibits excellent cold-rolling ability of more than 90%. These results suggest that this new type of α′ martensite Ti alloy can be expected to replace the typical α, α + β and β Ti alloys.

So far, the α′ martensite has been utilized for microstructural control of the nucleation sites of α in (α + β) Ti alloy, though this microstructural control is not so common in the industrial production of Ti alloy [3]. Therefore, there have been few reports that systematically examine the characteristics of the martensite (α′ or α″) phase in Ti alloys in relation to mechanical properties. In this study, a systematic evaluation of the microstructures and mechanical properties of Ti–V–(Al, Sn) alloys with a α′ martensitic structure is conducted with changing alloying content and heat treatment conditions.

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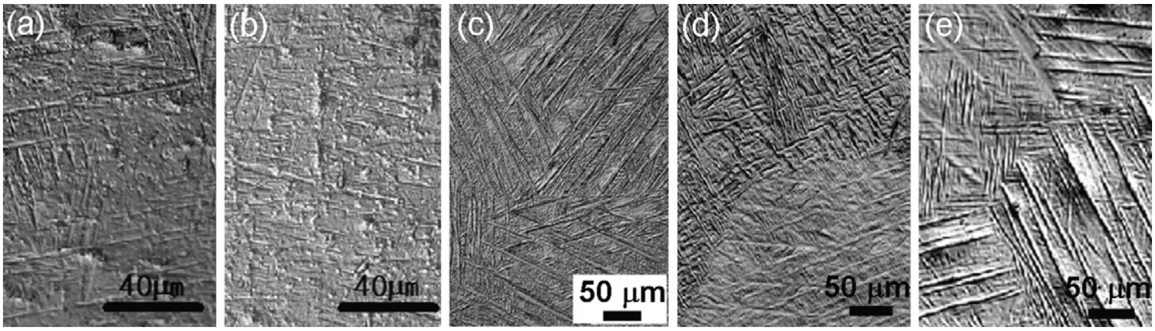

Fig. 1. Optical micrographs of STQ alloys of (a) (Ti–4%V)–1%Al, (b) (Ti–4%V)–4%Al, (c) (Ti–12%V)–2%Al, (d) (Ti–8%V)–4%Sn, and (e) (Ti–12%V)–6%Sn.

The Al and the Sn are well recognized as major alloying elements in Ti–V based alloy due to the effects of strengthening of the alloy and suppression of ω-phase formation. Therefore, three kinds of alloy systems of Ti–V–Al, Ti–V–Sn and Ti–V–Al–Sn were selected in this work. In this paper, the previous results of Young's modulus [4] are also demonstrated, and we discuss the mechanical properties of α' Ti alloys in relation to the effects of the alloying element and content, with a comparison with those of equilibrium $(\alpha+\beta)$ Ti alloys. Based on these results, this paper then discusses the possibilities of new type α' Ti alloys and a microstructural control technique that utilizes the α' martensite phase.

2. Experimental procedures

(Ti–(4–25) mass% V)–(0–6) mass% Al alloys and (Ti–(4–25) mass% V)–(0–6) mass% Sn alloys were prepared by arc melting in an argon atmosphere using high purity Ti, V, Al and Sn. The Al or the Sn was added to Ti–V alloys so that the ratio of the Ti-to-V content was kept constant at Ti–(4–25 mass%) V. For example, 2, 4, or 6 mass% Al was added to (Ti–4 mass% V), keeping (Ti–4 mass% V) constant. The impurity content was about 600 mass ppm for oxygen and 80 mass ppm for nitrogen in the Ti–V–Al and Ti–V–Sn alloys. Similarly, Ti–(0, 2, 4, 6) mass% Al–6 mass% V–2 mass% Sn alloys were prepared by arc melting in an argon atmosphere. Since the changes in weight before and after arc melting were less than 50 mg, the alloy compositions will be denoted hereafter by their nominal compositions.

Arc-melted buttons with a weight of about 90 g were homogenized at 1423 K for 86.4 ks in an argon atmosphere and hot rolled at 1123 K to a thickness of 5 mm and 7 mm. The hot-rolled plates were homogenized at 1423 K for 86.4 ks, followed by cooling in an argon atmosphere (Hereafter, these homogenized samples will be called 1423 K HT). In addition, the hot-rolled plates were solution treated at 1223 K (for near $\beta-(\alpha+\beta)$-type Ti–V–Al and Ti–V–Sn alloys), 1323 K (for near $\alpha-(\alpha+\beta)$-type Ti–V–Al and Ti–V–Sn alloys) or 1373 K (for Ti–(0, 2, 4, 6) mass% Al–6 mass% V–2 mass% Sn alloys) for 7.2 ks in an evacuated quartz tube, and then quenched in ice water (Hereafter, quenched samples will be called STQ). The STQ samples were cold rolled to a final thickness of 1 mm (Hereafter, the cold rolled samples will be called CR). The STQ sample with a composition of (Ti–12%V)–2%Al was heat treated at 823 K for 86.4 ks, followed by cooling in an Ar atmosphere to obtain the equilibrium $(\alpha+\beta)$ phase (Hereafter, this heat treated sample will be called 823 K HT).

Microstructures were identified by X-ray diffraction (XRD), optical microscopy (OM), and transmission electron microscopy (TEM). The dynamic Young's modulus at room temperature was measured by the free resonance vibration method using a sample with dimensions of 10 mm × 50 mm × 1 mm. Strength and elongation to fracture at room temperature were evaluated by tensile testing at an initial strain rate of $1.5\times10^{-4}\ \mathrm{s}^{-1}$. In addition, strain rate dependence on 0.2% proof stress, ultimate tensile strength, and elongation to fracture at room temperature were evaluated by tensile testing at strain rates of $1.5\times(10^{-2}, 10^{-3}, 10^{-4}$ and $10^{-5})\ \mathrm{s}^{-1}$. Cold-rolling ability was evaluated by determining the limited cold-rolling reduction at which a cracking on the surface of the plate first appeared under cold rolling. In this evaluation, the samples, which were solution treated and quenched in ice water, were cold rolled from a thickness of 5 mm or 7 mm.

3. Result and discussions

3.1. Microstructures of Ti–V–Al and Ti–V–Sn alloys (STQ and HT)

Fig. 1 shows the optical micrographs of STQ alloys of (a) (Ti–4%V)–1%Al, (b) (Ti–4%V)–4%Al, (c) (Ti–12%V)–2%Al, (d) (Ti–8%V)–4%Sn, and (e) (Ti–12%V)–6%Sn. It can be seen that direct quenching in ice water from 1323 K or 1223 K results in an almost fully acicular martensitic structure for all alloys. The constituent phases of the quenched samples analyzed by XRD and TEM observation are summarized in a Ti–V–Al ternary phase diagram in Fig. 2(a) and a Ti–V–Sn ternary phase diagram in Fig. 2(b), in which the compositional ranges are V from 0 to 30 mass% and (Al or Sn) from 0 to 6 mass%, respectively. From the XRD profiles and TEM observation, it was determined whether the constituent phase was α' (H.C.P.) martensite or α (H.C.P.) phase by confirming the absence of β formation in the microstructure, a needle-like microstructural characteristic, and the formation of a $\{10\bar{1}1\}$ twin in a variant. The formation of a $\{10\bar{1}1\}$ twin under quenching in Ti alloy is recognized as being caused by the martensitic transformation of β/α' [5]. In Fig. 2, we can note the suppression of athermal ω formation with an increase in the alloy content of Sn or Al, and the larger region of single phase of α', α'' and β for Ti–V–Al alloy as compared with Ti–V–Sn alloy. It can also be seen that the addition of A1 to Ti–V alloy is more effective for suppressing the athermal ω formation than the addition of Sn. This difference in the efficiency of the suppression of ω phase can be expected to affect the martensitic transformation. That is, the addition of Al, which strongly retards athermal ω formation, promotes the martensitic transformation, thereby resulting in expansion of the single phase region (α', α'' and β) (Fig. 2(a)). Concerning the phase constituent in quenched Ti–V–(Al or Sn) alloy, the results are in good agreement with the previous reports of Maeda and Flower [6], and Ohyama and Nishimura [7].

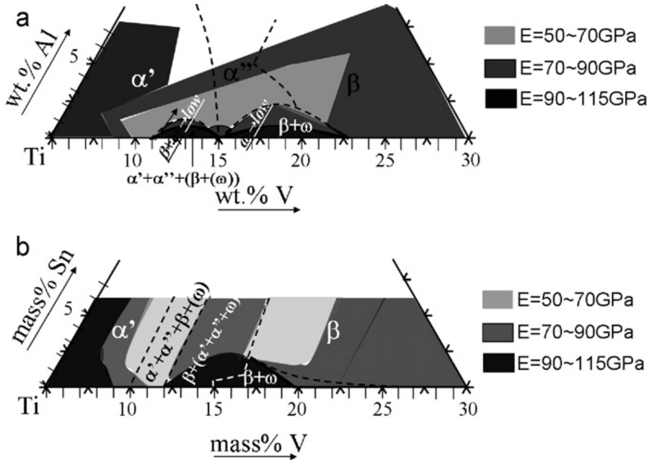

Fig. 2. Young's modulus of STQ (a) Ti–V–Al, and (b) Ti–V–Sn alloys illustrated in ternary phase diagrams.

3.2. Young's modulus of Ti–V–Al and Ti–V–Sn alloys (STQ)

In Fig. 2(a) and (b), the Young's modulus in Ti–V–Al alloy (STQ) and Ti–V–Sn alloy (STQ) is also illustrated. In the compositional