Beta-Ti Alloys

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β-Ti alloys are known for their excellent corrosion resistance, biocompatibility (in Ti-Nb alloys), and high strength-to-weight ratios, and some grades have a relatively low Young's modulus (E). These favorable properties have led to the use of these alloys in the automotive, aerospace, biomedical, and industrial sectors.

Keywords: β-Ti alloys ; Ti-Nb alloys ; Texture ; Thermomechanical Processing ; EBSD

1. Introduction

To develop the final products, β-Ti alloys are subjected to different thermomechanical processing (TMP) after casting. TMP includes various processes, such as the homogenization of cast ingots, hot (HR)/cold rolling (CR), heating the alloys to different temperatures (above and below the β-transus temperature) followed by final cold rolling (CR) deformation, and aging treatment [1]. It is well documented that phase stability in Ti alloys depends on their composition or alloy additions, since α-Ti consists of α-stabilizers (Al, N, O, Sn) and β-Ti consists of β-stabilizers (V, Nb, Cr, Fe, Mo) [2][3]. Based on the alloy's constituents, β-Ti alloys are further divided into near β, metastable β, and stable β-Ti alloys. A high fraction of β-stabilizers stabilizes the BCC microstructure even at RT, whereas a decrease in the fraction of β-stabilizers may develop α-phase within the β-matrix. Ti-4Al-7Mo-3Cr-3V (Ti-4733) [4], Ti-7Mo-3Nb-3Cr-3Al (Ti-7333) [5], Ti-5Al-4Zr-8Mo-7V (Ti-5487), and Ti-15Nb-5Zr-4Sn-1Fe (Ti-15541) [6] are a few β-Ti alloys possessing high strength, a low Young's modulus (E), fatigue resistance, high toughness, and superplasticity. Plastic deformation such as uniaxial tension/compression, or the hot/CR of various β-Ti alloys triggers complex deformation mechanisms in the β-matrix [7]. These are associated with the evolution of deformation-induced products such as dislocations/sub-grain formation, stress-induced phase transformation (β→α″ (SIM) and β→ω), deformation twins, and kink bands [8]. Annealing treatment is another method to develop different types of microstructures in deformed β-Ti alloys. Usually, shear bands (SBs), strain-induced martensite (SIM, α″), and deformed grain boundaries (GBs) are regions of high stored energy and hence preferential nucleation starts at these locations [9][10].

2. Beta-Ti Alloys

2.1. Deformed Microstructure and Texture Evolution

The low-strain-rate deformation of metallic materials often involves mechanisms of slip and twinning deformation. Dislocation slip bands were observed at low strain rates when deforming β-Ti alloys [11][12]. During the tensile loading of two-phase Ti alloy [13], it was found that fracture was initiated at GBs, and GB sliding through dislocation slip was responsible for intergranular cracks at slow strain rates. On the other hand, at high strain rates (>10^3 s^{-1}) most metallic material failures are associated with the formation of adiabatic SBs. The compressive deformation (10^{-3} s^{-1} and 10^3 s^{-1}) behavior of a solution-treated Ti-25Nb-3Mo-2Sn-3Zr alloy (near β-Ti alloy) was studied in [14]. The evolution of adiabatic SBs, α″ martensite, and slip bands inside the β-matrix were observed. The α″ martensite phase consisted of fine needle twinned domains intersecting each other at 120° or 90° [14].

RT rolling/CR leads to the evolution of various deformation-induced products as well as the development of α- and γ-fiber texture components. These are well-known texture fibers, in which <100>//RD refers to an α-fiber and <111>//ND indicates a γ-fiber texture. In the case of Ti-15V-3Al-3Sn-3Cr alloys (Ti-15333), unidirectional rolling (UDR) leads to strong α- and γ-fiber textures, whereas multistep cross-rolling (MSCR) shows a strong rotated cube component {100}<110> [15]. Lan et al. [16] reported the texture evolution in cold-rolled Ti-32.5Nb-6.8Zr-2.7Sn biomedical β-Ti alloy. The transition from an α-fiber texture to γ-fiber texture took place during 90% CR. With the increase in the reduction ratio (RR), an increase in micro-strain, dislocation density, and grain refinement was observed [16]. The formation of SBs in an 85% cold-rolled Ti-35Nb alloy was reported in [17]. Its crystallographic texture shows that the β phase had a BCC rolling texture combination of a cube {100}<001>, and rotated cube, with a strong α-fiber and weak γ-fiber texture at low RRs (<65%), whereas the 85% rolled sample showed a strong γ-fiber and weak α-fiber texture [17]. In another study, Cojocaru et al. [18] reported crystallographic texture evolution in a Ti-29Nb-9Ta-10Zr alloy after cold-rolling with different amounts of RR, up to 60%. The major texture components developed during cold-rolling were γ-fiber, {112}<111>, {001}<010> and {010}<001> texture components [18]. A decrease in Young's modulus (E) after different cold-rolling was observed, which was mainly due to the α″ phase formation. At 60% RR, an E close to 45 GPa was obtained, coupled with an average Vickers microhardness close to 279 Hv [18].

In contrast to CR, where only external stresses induce plastic deformation, hot rolling involves the additional factor of temperature. Hence, differences in microstructural and crystallographic texture can be observed between hot-rolled and cold-rolled conditions. In one such study, heterogeneities in the crystallographic texture evolution in a Ti-15Mo-3Al-2.7Nb-0.2Si alloy were observed during hot-rolling (84 and 97% RR) [19]. The development of shear strain in the surface region leads to weakening of texture intensity, which in turn gives rise to significant through-thickness texture gradients. Dynamic recrystallization (DRX), during hot-rolling, occurs and weakens the deformation texture [19].

2.2. Heat-Treated Microstructure and Texture Evolution

The annealing treatment of deformed β-Ti alloys leads to the evolution of α and ω phases when applied temperatures are below the β-transus, whereas temperatures above the β-transus lead to nucleations and grain growth of the single β microstructure [20]. The effect of heating rate and temperature (500–600 °C) on the aging behavior of TIMETAL-LCB, VT22, and Ti-15333 β-Ti alloys were reported in [21]. TIMETAL-LCB and VT22 formed fine plate-like α at slow heating rates due to the precipitation of isothermal ω at low temperatures, which serves as nucleation sites for α [21]. However, at high heating rates, the formation of isothermal ω was avoided, leading to coarse, plate-like α microstructures with less desirable properties. Ti-15333, on the other hand, exhibited β phase separation (β′+βmatrix) during isothermal aging rather than isothermal ω formation. It has been reported that like isothermal ω, β′ can also act as a nucleation site for α [21][22]. It is crucial to note that there are three categories of ω phase, depending on the process of formation: (1) a deformation-induced ω phase, (2) athermal ω phase, and (3) isothermal ω phase. The athermal ω phase, resulting from rapid quenching [23], is a diffusionless transformation but is not related to martensitic transformation, while the isothermal ω phase forms during the low-temperature aging of β-Ti alloys. The deformation-induced ω-phase forms under applied stress/strain [23]. Evolving phases during the heat treatment can also affect the mechanical strength of the β-Ti alloys. Precipitation of α or ω precipitates in solution-treated or deformed β-type Ti alloys is useful to obtain improved static and fatigue strength values. On the other hand, α and ω phases have significantly higher intrinsic E (E(ω) ≈ 153 GPa, E(α) ≈ 115 GPa) than the β phase (E(β) ≈ 60–65 GPa) and hence are not useful for bio-implants [24]. A heat-treated microstructure was observed in the hot-compressed Ti-13V-11Cr-3Al alloy samples. GB maps were reported for samples subjected to hot compression tests at 930 and 1030 °C and a strain rate of 0.1 s−1 [25]. EBSD analyses showed that continuous dynamic recrystallization (CDRX) leads to considerable grain refinement through the dissociation of coarse deformed grains [25][26]. It was observed that well-developed subgrains formed by extended dynamic recovery (DRV) were responsible for the grain dissociation. Inhomogeneous compressive deformation, due to the faster evolution of the substructure at regions adjacent to the GBs, can be observed in. Severe deformation causes the evolution of a large number of dislocations near the GBs. The existence of these dislocations at the GB regions favors DRV and leads to accelerating substructure formation. The micrographs also showed the formation of some small substructure-free volumes of the old grains via CDRX, which were surrounded by LAGBs + HAGBs [25]. Grain boundary serrations and nucleations were typical of the propensity for discontinuous dynamic recrystallization (DDRX). However, the decreased fraction of subgrains and the absence of GB serrations as well as DDRX were observed in samples processed at 1030 °C [25]. Since the processing temperature (930 and 1030 °C) was above the β-transus, this caused the presence of a single β phase during compression deformation. In contrast, a low-temperature compression test (700 °C) of Ti-7Mo-3Al-3Nb-Cr (Ti-7333) caused the precipitation of the α phase [27]. A fine α phase was precipitated during isothermal deformation at 700 °C-10^{-3} s^{-1} [27]. The evolution of intragranular α (spherical precipitates) and grain boundary α (αGB) with a combination of HAGBs+LAGBs was observed [27].

The annealing treatment of a cold-rolled Ti-5Al-5Mo-5V-3Cr (Ti-5553) alloy was performed [28]. The as-received Ti-5553 alloy showed a weak texture and its intensity got strengthened near a partial α- and complete γ-fiber texture after 40% RR. Annealing treatment (860 °C–5 min) of a cold-rolled Ti-5553 alloy led to decreased intensity of the deformation texture [28]. Further, aging treatment of the as-received sample (below β-transus) at 670 and 770 °C caused the evolution of a fine α-phase inside the coarse β-grains. The evolution of the α-phase diminishes the texture of the aged samples. In a previous investigation on the annealing treatment of a cold-rolled Ti-15333 alloy, microstructure and crystallographic texture evolution have been discussed [29]. In a study, the effects of cooling rate following β- and α/β-region (920–1000 °C) heat treatment on microstructure and phase transformation were investigated for a Ti-6.5Al-2Sn-4Zr-4Mo-1W-0.2Si (BT25y) alloy [30]. The BT25y alloy was soaked at 920–1000 °C for 10 min, and then cooled at a rate of either 0.15 °C/s–150 °C/s to RT [30]. Microstructure observations indicated that the microstructure of the BT25y alloy was significantly influenced by the cooling rate. When the material was cooled from the β phase field at a lower rate, the grain boundary α (α_{GB}) and Widmanstatten α (α_{WGB}) phases were precipitated [30]. However, increasing the cooling rate greatly restrained the precipitations of α_{GB} and α_{WGB} phases. In this case, acicular martensite (α′) was precipitated inside the β grain. The primary equiaxed-α was retained when the material was cooled down from the α/β phase field [30]. The content and size of equiaxed-α decreased with the increasing solution temperature but were independent of the cooling rate [30]. Texture evolution in a heat-treated Ti-22Nb-6Ta alloy was investigated in [31]. A well-developed {001}<1–10> texture was obtained in the cold-rolled sample and after heat treatment at 600 °C for 10 min. Moreover, a recrystallization texture of {112}<1–10> was developed at 900 °C for 30 min [31].

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