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Dynamic recrystallization of Ti-5553 alloy during sub-transus thermomechanical processing: Mechanisms and its role in formation of a bi-modal structure

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ABSTRACT

Formation of a bi-modal structure, which is strongly affected by the dynamic recrystalliaztion induced by the sub-transus thermomechanical processing, is an effective way to improve ultimate tensile strength, plasticity and high cycle fatigue strength of Ti-5Al-5Mo-5V-3Cr (Ti-5553) titanium alloy. By using isothermal compression machine, transmission electron microscopy and electron backscattered diffraction, we have investigated the dynamic recrystallization behavior of lamellar Ti-5553 alloy during the sub-transus thermomechanical processing at 780 °C and the effect of dynamic recrystallization on bi-modal microstructure transformation for Ti-5553 alloy. The critical strain for the dynamic recrystallization of the alloy ranges from 0.2 to 0.3, at which low angle grain boundary has formed in Beta grains. The dynamic recrystallization of Beta phase goes through the following steps, including formation of dislocation tangles ($\varepsilon = 0.1$), formation of cell blocks ($\varepsilon = 0.15$), formation of density dislocation walls ($\varepsilon = 0.2$) and intersection of dense dislocation walls ($\varepsilon = 0.3$). If the compression strain reaches 0.9, low angle grain boundary could transform to high angle grain boundary. The critical strain for the dynamic recrystalliaztion of Alpha phase is approximate 0.2, which is lower than that of Beta phase. Here we have established a link between sub-transus thermomechanical processing and bi-modal microstructure transformation of Ti-5553 alloy: dynamic recrystallization of Beta phase should be an essential condition on bi-modal microstructure transformation. The main results of this work might be beneficial to the homogeneity optimization of large scale bi-modal Ti-5553 alloy bulks.

1. Introduction

Owing to the excellent strength to weight ratio, Ti-5Al-5Mo-5V-3Cr (Ti-5553) Beta-titanium alloy has been used to prepare the heavy sectional component for landing gear assemblies on the 787 airframe since 2005 [1,2]. As a kind of light-weight structural material, the mechanical properties of Ti-5553 alloy are strongly influenced by its structures [3,4]. Microstructures of high strength Ti-5553 alloy are classified into lamellar microstructure and bi-modal microstructure, which is based on the morphology of Alpha phases and the dimension of equiaxed Beta grains [5,6]. Although the ultimate tensile strength (UTS) of lamellar Ti-5553 alloy is higher than 1200 MPa, the alloy exhibit obvious tensile embrittlement and low high-cycle-fatigue (HCF) strength [3,5,7]. Conversely, bi-modal Ti-5553 alloy processes the excellent combination of UTS, HCF strength, plasticity and receivable fracture toughness [4,7,8]. Furthermore, bi-modal Ti-5553 alloy demonstrates a superior dynamic strength-plasticity performance ($\dot{\varepsilon} = 2000s^{-1}$): the tensile elongation (TE) of bi-modal alloy is almost twice as that of lamellar alloy at the same UTS level [9]. Although bi-modal Ti-5553 alloy processes outstanding mechanical properties in quasi-static condition and dynamic condition, the evolution mechanism of bi-modal microstructure has not been completely understand yet.

Fig. 1 schematically shows the processing route for bi-modal Ti-5553 alloy, which includes sub-transus thermomechanical processing (TMP), solutionising and ageing. It should be noted that the initial states of Ti-5553 alloy might be equiaxed microstructure, Beta processed microstructure or lamellar microstructure, but the

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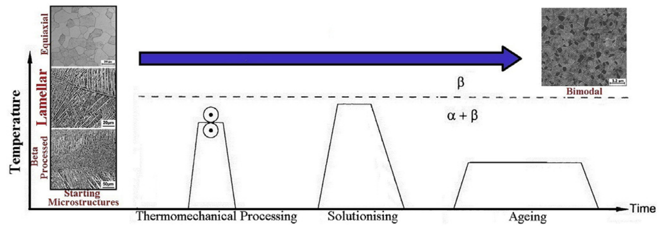

Fig. 1. The schematic processing route for bi-modal microstructure transformation of Beta titanium alloys. The starting microstructures of Ti-5553 alloy may be equiaxed microstructure, Beta annealed microstructure or Beta processed microstructure, and the target microstructure is bi-modal microstructure.

lamellar microstructure is more common. There are two TMP parameters that could influence the bi-modal microstructure transformation of Ti-5553 alloy, which are the TMP temperature and the TMP strain [7,10]. The recent investigation suggests that the proper TMP temperature is in the range of 750 °C–810 °C [7,10]. If the TMP strain is higher than 0.3, it is reported that dynamic recrystallization of Beta phase might take place and the alloy could transform to bi-modal microstructure after solutionising and ageing [10]. Temperature is the only critical parameter for solutionising and ageing. The typical solutionising temperature range is 790 °C–815 °C, and the proper ageing temperature range is 560 °C–677 °C [6,10].

Conversion of lamellar microstructure to bi-modal microstructure is a complex microstructure evolution for Ti-5553 alloy. Although much work has been done on transformation of bi-modal structure, quite few studies have been conducted on the microstructure evolution of lamellar alloy during sub-transus TMP. In addition, little attention has been paid to the relationship between bi-modal microstructure formation and dynamic recrystalliaztion that occurs during the sub-transus TMP. In this paper, we have studied the dynamic recrystallization mechanism of Ti-5553 alloy as well as the effect of the dynamic recrystallization of Beta phase on the formation of bi-modal structure. The aim of this work is to develop a deeper understanding on bi-modal microstructure transformation and to provide a guideline for the preparation of large scale bi-modal Ti-5553 alloy bulks.

2. Materials and methods

The Ti-5Al-5V-5Mo-3Cr (in wt.%) alloy ingot, with the dimension of Φ 160 mm × 420 mm, was prepared by a consumable electrode vacuum furnace. The Beta-transus temperature of the alloy is 845 °C, which is measured by metallographic methods. The ingot was first forged at 1180 °C and then forged at 900 °C with an approximate 50% height reduction by a 12.5 MN hydraulic press. After the Beta field forging, Ti-5553 alloy plate (120 mm × 120 mm × 15 mm) was machined from the billet. The plate was first solutionized at 900 °C for 1 h in a preheated furnace following water quenching and then solutionized at 780 °C for 3 h in a preheated furnace following water quenching. Finally, cylinder specimens (Φ 8 mm × 12 mm) were machined from the billet to perform the sub-transus TMP.

Before the sub-transus TMP the microstructure of the alloy was observed by ZEISS Axiovert-A1 optical microscopy (OM) and Zeiss Supra-55 scanning electron microscopy (SEM). The sub-transus TMP was performed on Gleeble-3800 testing machine. In order to monitor the temperature of specimens, K-type thermo-couple was welded on the mid-height side of specimens. Graphite grease and tantalum films were used to decrease the friction between the platen and the specimen. Under the clamping load of 1000 N, specimens were heated to 780 °C at the rate of 6 °C s^{-1}. Once at that temperature, specimens were held for 180 s to achieve thermal equilibrium. At the strain rate of 10^{-2} s^{-1} specimens were compressed to strain levels of 0.1, 0.15, 0.2, 0.3, 0.6 and 0.9, respectively. All the deformed specimens were air quenched to the room temperature. After finishing the sub-transus TMP, specimens with different strain levels were first solutionized at 810 °C for 3 h in a pre-heated furnace following water quenching and then aged at 600 °C for 8 h in a pre-heated furnace following water quenching.

Microstructure evolution of specimens during the sub-transus TMP is investigated by JEM 200CX transmission electron microscopy (TEM) and Zeiss Supra-55 SEM equipped with electron backscatter diffraction (EBSD) detector. The TMPed specimens are cut in half along the compression direction to prepare TEM samples and the EBSD samples. TEM specimens are located in the center region of the specimen, and the normal of the foil is perpendicular to the compression direction. The EBSD scanning region is also located in the center part of specimens. The operating voltage, the tilting angle and the scanning step are 20 kV, 70° and 100 nm, respectively. Zeiss Supra-55 SEM was used to observe the microstructure of Ti-5553 alloy specimens after sub-transus TMP, solutionising and ageing.

3. Results and discussion

Fig. 2 (a) shows the OM photograph of Ti-5553 alloy before the sub-transus TMP. The boundary of equiaxed prior Beta grains (PBG) exhibits a bright contrast, and the grain size of PBG is approximate 200 μm. Fig. 2 (b) illustrates the secondary electron SEM image that is recorded inside PBG. Comparing with the darker Alpha plate, we found that the Beta matrix processes a brighter contrast. Before sub-transus TMP the specimens were first solutionized at 900 °C for 1 h following water quenching and were then solutionized at 780 °C for 3 h following water quenching. After the solutionising at 900 °C the alloy consists of equiaxed Beta grains, and each grain processes a unique orientation [7]. During the solutionising at 780 °C Alpha plates could form in Beta grains. Therefore, the orientation of Beta matrix (Fig. 2 (b)) should be unique in each PBG. It could also be observed from Fig. 2 (b) that Alpha precipitations, which are approximate 5 μm in length, distribute homogeneously inside the Beta matrix. In summary, our OM and SEM results suggest that the microstructure of Ti-5553 alloy should be lamellar before the sub-transus TMP. The diameter of large scale Ti-5553 billets is more than 450 mm, and the initial microstructure of billets is commonly lamellar before the engineering sub-transus TMP [2,7,10]. Although our specimens are in small scale (Φ 8 mm × 12 mm), they are appropriate for investigating the dynamic recrystallization behavior of large scale Ti-5553 alloy bulk owing to the microstructure consistency.

Fig. 3 shows the inverse pole figure (IPF) map of specimens with different TMP strain levels, and the insertions in each sub-figure illustrate the misorientation profile of the line inside the black rectangle region. Fig. 3 (a) demonstrates the IPF map of the specimen with $\varepsilon = 0.2$ TMP. The Beta phase is indexed as green region, and the range of the misorientation profile is less than 2°. It seems that the Beta matrix still processes a unique orientation. These results indicate that the dynamic recrystallization of the Beta phase does not occur in the specimen with $\varepsilon = 0.2$ TMP. Fig. 3 (b) shows the IPF map of the specimen with $\varepsilon = 0.3$ TMP. The IPF map consists of three regions that originally belong to different lamellar units. In the rectangle region, several irregular sections (indexed with different colours) could be found, and there are a few dark interfaces among these sections. These results suggest that sub-grains have form in the Beta matrix of the specimen with $\varepsilon = 0.3$ TMP. Therefore, we conclude that dynamic recrystallization of the Beta phase occurs if the TMP strain reaches 0.3. In addition, the misorientation between the recrystallized grains is approximately 8°, indicating that the sub-grains process a low angle grain