Journal of Alloys and Compounds 714 (2017) 294–302

Contents lists available at ScienceDirect

Journal of Alloys and Compounds

journal homepage: http://www.elsevier.com/locate/jalcom

Precipitation of secondary alpha in competition with epitaxial growth of primary alpha in two-phase titanium alloys

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ARTICLE INFO

Article history:
Received 10 March 2017
Received in revised form
16 April 2017
Accepted 19 April 2017
Available online 22 April 2017

Keywords:
Titanium alloy
Sympathetic nucleation
Competitive growth
Thermal history
Heat treatment

ABSTRACT

The significant variation of microstructure features, i.e. volume fraction, size of secondary α phase, due to β→α phase transformation in heat treatment of two-phase titanium alloys can directly affect the mechanical properties. Understanding the nucleation mechanism and modeling the precipitating of secondary alpha phase with existed primary equixed-alpha phase are critical to precisely control the microstructure. In the present work, sympathetic nucleation mode is proposed by crystallographic orientation analysis and validated by calculating the nucleation energy barrier. Then, it is found that the growth of primary α phase competes with precipitation of secondary α phase by consuming the matrix supersaturation related to the driven force for precipitating. By considering the effects of the growth of primary α phase and thermal history on diffusion field around secondary α phase, and introducing a ledge growth mechanism-related factor, a physically founded fast-acting model is developed to predict the precipitation process during cooling. Good agreements between the experimental and computed results are obtained when applied to TA15 alloy with different original microstructures.

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1. Introduction

Two-phase titanium alloys have been gaining extensive applications in aerospace industry due to their excellent properties such as high specific strength, good thermal stability and welding performance [1]. The microstructure morphology of constituent phases, i.e. lamellar, equiaxed, bi-modal and tri-modal structures, has a significant influence on mechanical properties of titanium alloys [1–3]. Due to the combination of good high-temperature strength and fracture toughness, bi-modal structure is the preferable microstructure and can be usually obtained by a series of thermo-mechanical processing and heat treatment steps in the forming process of these alloys [4].

It has been found that the microstructure morphology, such as volume fraction, thickness, aspect ratio of secondary α phase, is very sensitive to processing history [1]. Moreover, the facture toughness could be determined by the characteristics of secondary α phase in titanium alloys [5]. Therefore, it is crucial to predict the secondary α phase evolution laws to control the microstructure morphology and improve the mechanical properties. In the heat treatment of titanium alloys, β→α solid-state phase transformation involving nucleation and growth plays a key role among secondary α phase evolution. However, the nucleation mode of secondary α phase is usually related to the thermal history, which determines the matrix supersaturation and thus the drive force for its formation [6–8]. As for its diffusion-controlled thickening, diffusional interaction between adjacent alpha plates is complex due to its ledge-growth mode [9]. In addition, the precipitation of secondary α phase can be influenced significantly by the epitaxial growth of primary alpha under the non-isothermal condition [10,11]. As a result, it is difficult to precisely model the precipitation process of secondary alpha due to the ambiguous nucleation mode and complex microstructure evolution associated with the interaction of adjacent alpha plates and the effect of primary alpha phase.

By now, a lot of investigations have been made on the formation mechanism of alpha plates during cooling in titanium alloys. Appolaire et al. [12] assumed that secondary alpha mainly nucleated at the grain boundary α and proposed a transformation model of sympathetic nucleation for near β titanium alloys.

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http://dx.doi.org/10.1016/j.jallcom.2017.04.209
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However, this model can be limited when applied to cooling from two-phase region due to the existence of primary alpha. By the conventional TEM observations and analysis, Li et al. [6] also revealed that the crystallographic orientation of secondary alpha was slightly different from that of pre-existing alpha lath during the air-cooling process in TC21 alloy. However, when secondary alpha forms at the primary equiaxed α (α_p) at slow cooling rate, the nucleation mode needs to be considered carefully. Meanwhile, Salib et al. [13] found that secondary α inherited the orientations taken by grain boundary α in Ti-17 alloy. Similar results were also found due to interface instability nucleation in near α Ti alloys [14]. The above studies provided a good basis for the present work. However, the nucleation mechanism proposed by Sun et al. [14] depended on SEM images which could not fully exhibit orientation relationships between primary and secondary α phases.

The physically based internal state variable modeling describes the underlying phenomena in terms of a small number of internal state variables [15]. The microstructure development, such as diffusion-controlled growth of precipitates, can be captured by a simultaneous set of differential equations. Thus, they are suitable for processes with complex thermo-mechanical history. Lengthening of plate-shaped precipitates were usually treated by the modified Zener–Hillert equations [16]. Appolaire et al. [10] described the growth of grain boundary alpha with the assumption of no overlapping around precipitates. However, in the finite matrix, the interaction of diffusion fields occurs and soft impingement needs to be considered carefully. According to Aaronson's analysis [17], the coherent broad face could only grow perpendicular to itself by nucleation and migration of ledges. Enomoto [18,19] studied the migration of an array of steps using finite difference scheme and showed that the thickening kinetics followed nearly a parabolic law at long reaction time under isothermal conditions. However, this method is complex and the effect of thermal history is not considered when applied to predicting the thickening process of α_s phase. Moreover, the influence of primary α on precipitation of secondary α is not coupled into these models.

In view of above mentioned literature, the evolution of secondary α phase during cooling is multi-factors-controlling complex process which has not yet been clarified. Therefore, theoretical analyses and microstructure observations should be further carried out and the effects above mentioned (i.e., temperature path, epitaxial growth of primary α) should be considered when modeling the secondary α evolution in order to tailor the microstructure.

The objective of the present work is to obtain an understanding about nucleation mode of secondary α phase in heat treatment of two-phase titanium alloys. Considering the effects of thermal history and primary α phase on the diffusion field around secondary α and assuming a ledge growth mechanism, a fast-acting model is developed to predict precipitation process on the basis of classical nucleation and growth theories.

2. Materials and experimental procedures

The TA15 titanium alloy used in the present work was received in hot forged bar form with measured chemical composition of Ti-6.69Al-2.25Zr-1.76Mo-2.25V-0.14Fe-0.12O and β-transus temperature of 1258 K. The characteristic microstructure of as-received material consisted of approximately 50% primary equiaxed α phase within transformed β matrix and the size of primary α phase is ~10 μm, as given in Fig. 1.

Cylinder specimens with 10 mm in diameter × 15 mm in height were machined from the as-received bar. Prior to heat treatment, a type-K thermocouple was attached to each sample at the mid-length, mid-diameter position to monitor and control the temperature. Then, they were treated via solution treatment at 960 °C for 30min followed by cooling at a rate of 10 °C/min,26 °C/min or 50 °C/min, and finally water quenched after reaching a prescribed temperature of 930 °C, 870 °C, 810 °C or 750 °C. Fig. 2 showed the schematic of corresponding processing route. These samples obtained were denoted as A-L. Our previous work conducted heat treatment experiments of TA15 alloy at a cooling rate of 9.5 °C/min, 60 °C/min or 165 °C/min with the original α_p size of 7.6 μm [10]. The microstructure information (i.e., volume fraction and mean size) would be obtained by quantitative metallographic image analysis and can be subsequently used to validate the precipitation model of secondary α phase.

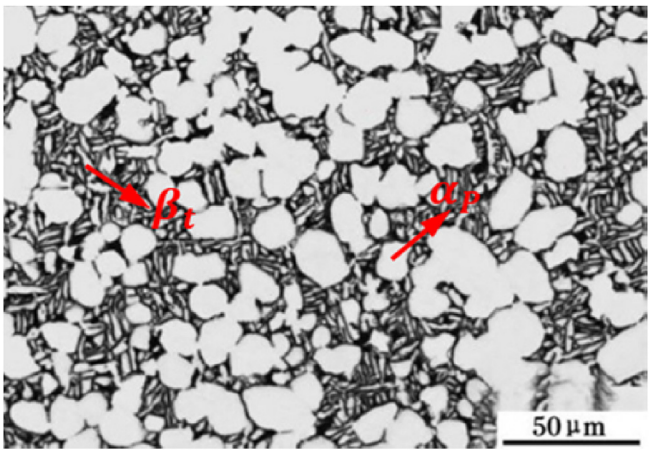

Fig. 1. Original microstructure of the billet.

Following heat treatment, samples C and D were sectioned axially for EBSD observation in the central portion of each specimen. Furthermore, the specimen cooled at 60 °C/min and water quenched after reaching 810 °C with original size of 7.6 μm, which was indicated as 1, was also selected for the observation. These specimens were electropolished in a solution of 5% perchloric acid, 65% methyl alcohol, and 30% butanol at 28 °C with a voltage of 25 V

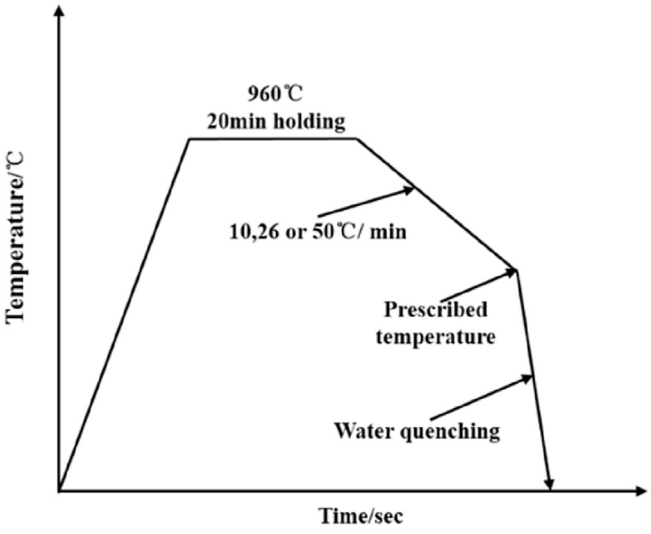

Fig. 2. The schematic of processing route.

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