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Microstructural stability of long term aging treated Ti−22Al−26Nb−1Zr orthorhombic titanium aluminide

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Abstract: The microstructure development of lamellar structure of an orthorhombic Ti_2AlNb-based Ti−22Al−26Nb−1Zr alloy, including *B*2 decomposition and spheroidization of *O* phase, was investigated. The results show that the lamellar structure is fabricated by heating the samples in the single *B*2 phase field and cooling slowly in the furnace. Aging treatments are conducted in the (*O*+*B*2) phases field by air cooling. After aging at 700 °C for a short time within 100 h, there is no significant change of microstructures, whereas the coarsening of lamellae is observed in the long-term aged microstructure. Ti−22Al−26Nb−1Zr alloy exhibits microstructural instability including the severe dissolution of *B*2 lamella, discontinuous precipitation and spheroidization of *O* phase during the long term aging process at 700 °C up to 800 h. In addition, a pronounced formation of branch-shaped *O* phase lamella is observed for the alloy aged over 100 h.

Key words: orthorhombic titanium aluminide; microstructural evolution; aging treatment; spheroidization

1 Introduction

Ti_2AlNb-based alloys have been regarded as the most promising candidates for use in the aircraft engine materials in the past decades due to their low density and good high-temperature properties, and are verified to possess better mechanical properties over conventional *α*2-based alloys [1−3]. Especially the second generation orthorhombic alloys, such as Ti−22Al−25Nb and Ti−22Al−27Nb (mole fraction, %), with *O*+*B*2 lamellar microstructure preserve superior mechanical properties including high tensile strength, improved creep resistance as well as fracture toughness [4−8].

As a family of high-temperature used alloys, approximately 600 to 750 °C [9], the microstructural stability during aging in air environment is a key focus of Ti_2AlNb-based alloys. Whereas, the creep and mechanical properties of Ti_2AlNb-based alloys are significantly susceptible to microstructural evolution. It has been reported that *B*2 phase in Ti_2AlNb-based alloys is unstable and undergoes a decomposition around 700 °C [10]. The instabilities of *B*2 phase may pronouncedly degrade the creep properties and other high-temperature properties. YANG et al [11] have observed an abnormal acceleration of creep rate and ascribed it to the formation of prismatic dislocations generated from *O*/BCC interfaces by the transformation of *B*2 phase to *O* phase above 700 °C. ROWE et al [12] have found that discontinuous precipitation could lead to high primary creep strains, and influenced the minimum creep rate in the long time creep process. Meanwhile, the work of LIN et al [13] indicates that Ti−22Al−25Nb exhibits a microstructural instability during the tensile deformation, characterized by coarsening, fragmentation, and spheroidization of the *O* phase. More recently, the *O* phase plates with a special type of arrangement, which account for the enhancement of microhardness, have been found by KHADZHIEVA et al [14] in the course of the *β*→*O* transformation during the aging process for 4 h. As multiphase alloys, the microstructural stability during high-temperature exposure always determines the practical applications of Ti2AlNb-based intermetallic compounds. However, to date, only a limited number of reports focus on the microstructural evolution of Ti_2AlNb-based alloys in service temperature in contrast to *γ*-TiAl [15−18]. In order to give immediate guidance in the choice of prolonged aerospace application materials, it is necessary to understand the high-temperature microstructural stability of Ti_2AlNb-based

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alloys at elevated temperatures.

Recently, a Zr-containing Ti_2AlZr phase has been found to be considerably stable, and the alloys with this phase possess high yield strength [19−21]. Additionally, the addition of Zr may also improve the creep resistance of Ti_2AlNb-based alloys [22]. Taking these advantages into consideration, the element of Zr is introduced to investigate its effect on the microstructural stability of Ti2AlNb-based alloys to improve the high temperature properties to some extent. The present work mainly concerns to establish a fundamental understanding of the microstructural changes of a Zr-containing Ti_2AlNb-based alloy during the long-term aging treatment.

2 Experimental

The nominal composition of the test alloy was selected as Ti−22Al−26Nb−1Zr (mole fraction, %). The ingot was melted by vacuum consumable electrode melting from the commercial high-purity starting materials of Al shot, Ti sponge, Zr sponge and Ti−Nb binary alloy. The ingot was melted three times to ensure the chemical homogeneity. Then, the ingot was broken down in the *B*2 phase field, followed by an extra *β* forging plus a subtransus low temperature forging.

The forged Ti−22Al−26Nb−1Zr alloy was solution-treated at 1150 °C and then water-quenched. In order to obtain the lamellar microstructure, the cooling rate after the solution treatment was chosen as about 2.4 °C/min. Aging treatments were performed at 700 °C for various durations from 24 to 800 h after solution treatment. The microstructures of the specimens were characterized by optical microscopy, scanning electron microscopy (SEM) and transmission electron microscopy (TEM). The phase compositions of alloys were examined by X-ray diffraction, with the continuous scanning mode with 0.03° interval and 1.0 s counting time. The diffraction data were collected at room temperature on a DX−2700 diffractometer using Cu K_α radiation. The voltage and anode current were 40 kV and 30 mA, respectively. The thin foils for transmission electron microscopy were prepared by twin jet polishing in a solution of methanol, butanol and perchloric acid at −20 to −30 °C. Transmission electron microscopy (TEM) was performed using an FEI Tecnai G2 F30 TEM equipped with an energy-dispersive X-ray (EDX) detector and operated at 300 kV.

3 Results

3.1 Microstructure of solution-treated alloy

The microstructures of solution-treated Ti−22Al−26Nb−1Zr alloy are shown in Fig. 1. Figure 1(a) shows an optical image of lamellar microstructure in the prior *B*2 phase. The typical lamellar colony (in section *A*) and Widmanstatten lath (in section *B*) are apparently observed, and equiaxed constituents are avoided. In Fig. 1(b), the dark section can be ascribed to *α*2 phase, the bright and gray sections are *B*2 and *O* phases, respectively, which is consistent with the observation in Ref. [23]. The *α*2 phase locates inside the grain exists as long aligned lath caused by supertransus processing and irregular shape block [24]. However, the formation of irregular shape blocks has not been well determined yet. Figure 2 shows the X-ray diffraction pattern of the as-soluted sample, which confirms the presence of *α*2, BCC and *O* phases, and the ordered *B*2 phase is

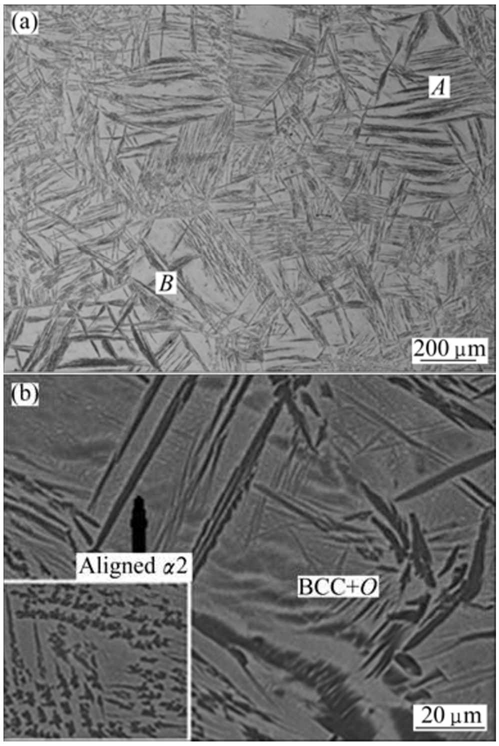

Fig. 1 Optical (a) and back scattered SEM (b) images of Ti−22Al−26Nb−1Zr alloy after heating in the *B*2 phase field for 1 h followed by furnace cooling

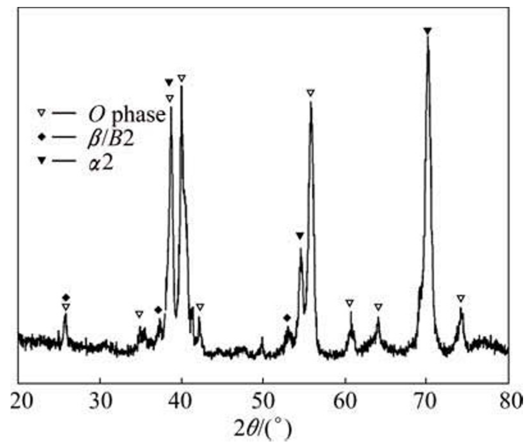

Fig. 2 X-ray diffraction pattern of as-soluted Ti−22Al−26Nb−1Zr alloy

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