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Effect of growth rate on microstructure parameters and microhardness in directionally solidified Ti–49Al alloy

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ABSTRACT

Intermetallics Ti–49Al (at.%) alloy was directionally solidified with different growth rates (V = 5 μm/s–30 μm/s) at a constant temperature gradient (G = 12.1 K/mm) by using a Bridgman type directional solidification furnace. The primary dendritic spacing (λ), interlamellar spacing (λ_L), and microhardness (HV) were measured. Effect of V on HV, λ and λ_L was experimental investigated. The dependencies of λ, λ_L and HV on the growth rate were determined by using linear regressing analysis. According to the result, the values of λ and λ_L decrease with the increasing of V, and the values of HV increase with the increasing of V and with the decreasing of λ and λ_L. The results were compared with previous similar experimental results for TiAl-based alloys.

1. Introduction

Intermetallic γ-TiAl based alloys are attractive materials for high temperature structural applications because of their low-density, good oxidation resistance and high temperature strength [1–4]. Of the many microstructures that can formed in TiAl-based alloys, the fully lamellar microstructure consisting of TiAl (γ-phase) and Ti_3Al (α_2-phase) has been studied widely because it displays a good combination of room temperature toughness and elevated temperature strength [5]. The application of TiAl alloys was, however, limited by major obstacles such as the low room-temperature ductility, the difficulty in processing then to fabricate a component and poor oxidation resistance above 800 °C [1–4].

Previously, Johnson et al. [5–7] have studied the TiAl alloy with PST (polysynthetically twinned crystals) structure and found that the full-lamellar TiAl-based alloys with aligned lamellar structures have a good combination of strength and ductility in a wide range of temperature. Yamaguchi et al. [8] indicated that the lamellar orientation can be controlled by directional solidification. Thus, to achieve columnar grain materials of TiAl alloys with the lamellar orientation aligned parallel to the growth direction, appropriate processing techniques have been developed [6]. One way is by directional solidification with a seed material, and the other is controlling the solidification path by directional solidification without a seed material. Recently, extensive works have been carried out to achieve columnar grain materials of TiAl alloys with the lamellar orientation aligned parallel to the growth direction by directional solidification with [9–11] or without seeding technology [8,12,13]. These studies have focused on the control of lamellar orientation by optimizing solidification parameters during directional solidification of TiAl alloys. However, the effect of solidification parameters on the microstructural characteristics and mechanical properties has been rarely reported.

During directional solidification, the solidification parameters affect the microstructure of the alloys, and also influence theirs mechanical behaviors [14]. Therefore, the effects of solidification parameters on the microstructure and mechanical property have been investigated widely for different alloys, such as Al-based alloys [15–17], Pb-based alloys [18,19], Zn–Cu alloys [20], and Sn–Ag–Cu alloys [21,22]. These researches focused on eutectic or peritectic alloys with a low melting point. For binary alloy with a high melting point, however, the effects of solidification parameters on the microstructure and mechanical property have rarely been reported [11,23,24]. Hence, present work is focused on the TiAl-based alloy with a high melting point.

There is a linear relationship between microhardness and yield stress [11,25], which promises the mechanical properties of directionally solidified TiAl ingots to be predicted from the values of HV. Similar linear dependence of the yield stress on hardness was also observed in wrought TiAl alloy with fully lamellar structures [26]. It appears that the microhardness analysis offers a relatively simple way to predict the mechanical properties of the materials [27]. Therefore, the values of Vickers microhardness are useful for quality control of directionally solidified TiAl alloys.

In this paper, the aim of present work was to experimentally investigate the effect of growth rate on the microstructure and microhardness in directionally solidified Ti–49Al (at.%) alloy, and compared the results with the previous similar experimental

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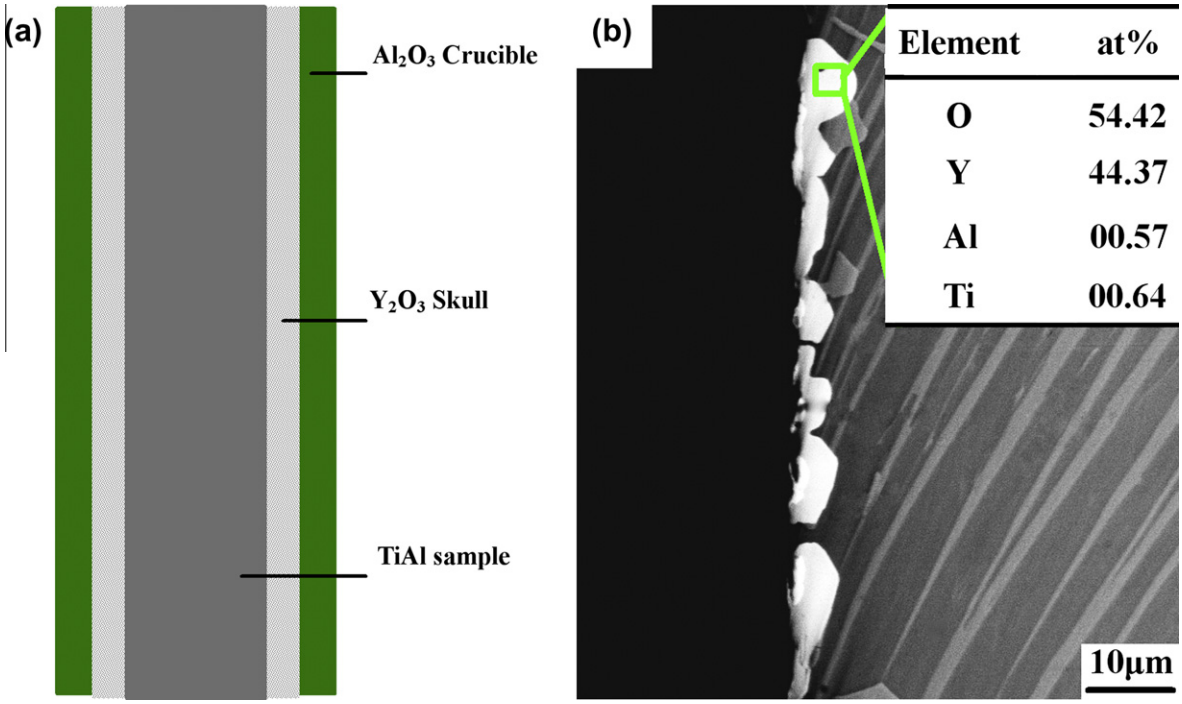

Fig. 1. (a) The schematic diagram of the specimen with a skull-aided technology; (b) microstructure of the surface of the specimen after directional solidification. The inside table in (b) is the EDX results of the white phase on the surface.

results for TiAl-based alloys. The results will provide a reference for the microstructure control of TiAl-based alloys by directional solidification.

2. Experimental procedures

Master ingot with nominal composition of Ti–49Al (at.%) was prepared by using Ti (99.96%) and Al (99.99%) of commercial purity in a cold crucible induction furnace under argon atmosphere. The samples were machined to rods with 3 mm diameter and 100 mm in length from the ingot by a spark machining. The directional solidification experiments were performed in Bridgman-type furnace. The details of the furnace were described in Ref. [24]. The sample was placed into 99.99 pct pure alumina crucibles of 4/5.5 mm in diameter (inside/outside diameter) and 150 mm in length. Before directional solidification, the chamber of the furnace was evacuated down to 5×10^{-3} Pa, and then was back-filled with high purity argon to prevent the evaporation of aluminum. The specimen was heated to 1773 K during 4 h and thermal stabilized for 30 min, and then was directionally solidified with different growth rates range from 5 μm/s to 30 μm/s at a constant temperature gradient of 12.1 K/mm. After growing about 30 mm, the sample was quenched into the liquid Ga–In–Sn alloy to restore the solid–liquid interface.

The temperature gradient was measured by W/Re thermocouples that were placed near the outside surface of the alumina tubes, as was illustrated in Ref. [24]. The temperature gradient can be changed by changing the temperature of the sample. To keep the temperature gradient constant during directional solidification, the temperature of the cooler and the hotter part of the furnace were kept constant by an automatic temperature controlling system.

A skull-aided technology was used to avoid the reaction between the Al_2O_3 crucible and the melt of TiAl alloy during directional solidification. The fine powders of high-purity Y_2O_3 were filled in the interspace between the Al_2O_3 crucible and the specimen, as shown in Fig. 1. The powers of Y_2O_3 formed an integral coat which segregating the specimen from the crucible completely. After heating, the coat became a thin-integral skull and segregated the melt from Al_2O_3 crucible. Because of the high stability of Y_2O_3 to TiAl melts, the severe reaction between the crucible and the melt is replaced by the slight reaction between the skull and the melt.

After directional solidification, the longitudinal and transverse sections of the specimens were cut, polished and etched with a solution of 10 ml HF–10 ml HON_3–180 ml H_2O for further analysis. Both optical (OM) and scanning electron microscopy (SEM) were used to characterize the microstructure of the specimens. The phases were identified by a Rigaku D/max-RB X-ray diffractometer with monochromatic Cu-Kα radiation. SEM backscattered electron imagining (BSE) was used to help identify the phases present. The typical images of growth morphologies and microstructures of directionally solidified Ti–49Al (at.%) alloy were shown in Figs. 2 and 3. The typical BSE images of lamellar structures were shown in Fig. 4.

The dendritic spacings were measured from both the transverse sections and longitudinal sections of the specimens. The dendritic spacings measured from the transverse sections were used the triangle method, and measured from the longitudinal sections were used the linear intercept method, as described Refs. [15,22,28,29]. Because of the measurement of λ_L from transverse sections are more accurate than that from longitudinal sections [18,19,30], the values of λ_L were measured from the transverse sections of specimens on the BSE images according to the method described in Refs. [18,30]. The measured values are the average values and given in Table 1.

Microhardness measurements were made with a standardized Vickers measuring test device using 100 g load and a dwell time of 10 s on the surface of longitudinal and transverse section with polished and slightly etched surfaces. The microhardness is average of at least 15 measurements. The values of *HV* are also given in Table 1.