Superplastic deformation mechanism of a γ -TiAl alloy with coarse and bimodal grain structure

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Abstract: The superplasticity of a Ti–42.5Al–8Nb–0.2W–0.2B–0.1Y (at.%) alloy with ~20 vol.% metastable β /B2 phase has been examined at temperature range of 850~1050 °C under an initial strain-rate of 10^{-4} s^{-1} . The results showed that though the proposed alloy has a coarse and bimodal grain structure which was unfavourable for fine-structure superplasticity, impressive elongation was obtained at 1000 °C (~380%) with a strain-rate sensitivity exponent value of about 0.46. The microstructural characterization showed that remarkable dynamic recrystallization occurred during superplastic tension, and the initial texture-free specimen gradually produced a $\langle 100 \rangle$ fiber texture whose the main component was cube texture. All these observations seem to be contradictory to the classical theory for fine-structure superplasticity (grain boundary sliding mechanism). Instead, it was suggested that the major deformation mechanism in this alloy was crystallographic slip accompanied by discontinuous dynamic recrystallization.

Keywords: *Titanium aluminides; Superplasticity; Recrystallization; Texture*

1. Introduction

In the case of fine-structure superplasticity (FSS), it is widely accepted that grain boundary sliding (GBS) is the predominant deformation mechanism, and thus some prerequisites are required to allow the readily onset of GBS, as summarized by a number of reviews [1-3]: (1) Equiaxed and fine grain structure is the major requirement; (2) High angle grain boundary

should be predominant due to its high sliding velocity; (3) Presence of second phase/particles impedes the excessive grain growth. Subsequently during the superplastic deformation, the materials generally show quasi-steady-state flow (under constant strain-rate) with high strain-rate sensitivity exponent (i.e., $m > 0.3$) [3,4], and the microstructure evolution also shows some unique features [2,5]: i) No essential change in grain shape/size; ii) Continuously reduction of the overall texture due to grain boundary sliding/grain rotation; iii) No massive dynamic recrystallization (DRX) occurs in the usual sense. Till date, considerable investigations have focused on the FSS of TiAl alloys with ($\alpha_2 + \gamma$) constituents and it was widely recognized that the predominant deformation mechanism is GBS [4,6,7]. Especially in our previous works [8], we found that the microstructure and texture evolution of the fine-structural Ti-43.5Al-8Nb-0.2W-0.2B alloy during superplastic deformation strongly supported the GBS mechanism.

As for the β /B2-rich TiAl alloys, the situation is somewhat different. Generally, the β /B2 phase can remarkably enhance the superplastic properties of TiAl alloys due to its soft nature at elevated temperature, and it was reported [9,10] that the β /B2-rich TiAl alloys can exhibit impressive superplasticity even in as-cast state. However, GBS is still thought to be the major superplastic deformation mechanism for these β /B2-rich alloys, probably because their microstructures are sufficiently fine and uniform to permit the predominance of GBS during deformation.

Given these situations, it may be interesting to examine the superplastic property of the TiAl alloys with coarse microstructure (unfavorable for FSS) but with considerably large amount of β /B2 phase (favorable for FSS), and (if superplasticity does exist) the deformation

mechanism may also change. For this purpose, in this paper we proposed a (β /B2+ γ)-TiAl alloy with coarse and bimodal grain structure to examine its superplasticity. In order to reveal the deformation mechanism, electron backscatter diffraction (EBSD) was applied to study the microstructure and texture evolution.

2. Experimental procedures

A TiAl ingot with a nominal composition of Ti-42.5Al-8Nb-0.2W-0.2B-0.1Y was prepared by vacuum arc re-melting (VAR), and the HIPing (hot isostatic pressing) was conducted at 1300 °C/140MPa for 4 h. Then a cylindrical rod was machined from the ingot and hot-forged at 1260 °C. The obtained microstructure of the forged pancake is shown in Fig. 1. One can observe that the alloy mainly consists of white β /B2 phase (~20 vol.%) and dark γ phase (~80 vol.%). A bimodal γ grain distribution can be clearly observed. The large γ grains with sizes of 8–30 μm display as island-like clusters, and the γ particles with much smaller grain size (1–8 μm) are distributed in the β /B2 matrix. The mean γ grain size is estimated (by area-fraction) to be about 13 μm .

Fig. 1 Microstructure of the as-forged TiAl pancake. The γ phase is in dark contrast and the β /B2 phase is in white.

Tensile specimens with a gauge dimension of 8×3×1.5 mm were machined from the forged-pancake, and the tensile direction (Tens.D) of the specimens was parallel to the radial

direction of the pancake while the transverse direction was parallel to the forging direction (FD). Tensile tests were performed at temperature range of 850~1050 °C with an initial strain-rate of 10^{-4} s^{-1} to examine the superplastic property. Moreover, strain-rate jump tests were performed at 1000 °C to evaluate the m value. With step sizes of 0.2 μm or 50 nm, the microstructure at various regions of the specimen (corresponding to different true strain) tested at 1000 °C/ 10^{-4} s^{-1} was characterized by using EBSD conducted on a JSM-6500F scanning electron microscope, and the texture was analyzed by JTEX software [11] using the EBSD data. Because the slight tetragonality of γ -TiAl lattice cannot be resolved by EBSD [12], we have ignored the order domain and treated the pseudo-cubic γ -TiAl lattice as normal f.c.c. lattice.

3. Results and discussion

3.1 Tensile property

Fig. 2(a) shows the flow curves of the alloy at various temperatures under an initial strain-rate of 10^{-4} s^{-1} . It can be observed that the elongation is increased with temperature, and shows the maximum value (~380%) at 1000 °C without visible necking. In addition, the m value at superplastic conditions (temperatures between 950~1050 °C and strain-rates around 10^{-4} s^{-1}) was estimated to be ~0.46 which is a typical value for FSS, as shown in Fig. 2(b). That is, though the microstructure is relatively coarse, superplastic phenomenon is quite evident.

3.2 Microstructure and texture evolution

The microstructure and texture at different regions of the specimen tested at 1000 °C are shown in Fig. 3, one can note that the microstructure in the grip region (where $\epsilon \approx 0$) is slightly changed in comparison with the initial state. However, at the deformed region, the large γ

grains gradually break into smaller grains due to dynamic recrystallization (DRX), hence the grain size distribution tends to be homogenous and its bimodal nature is significantly weakened (Fig.3 (e,h)). Evidence for DRX is confirmed by the great number of DRX nuclei found by EBSD with small step size (50 nm). As shown in Fig. 4, grain boundary bulging frequently appears, and most of them were twin-related to the parent grains (TB), while the opposite nuclei boundaries with high angles migrate towards the other grains. This twinning-related DRX nucleation mechanism is quite critical in some alloys such as copper [13], and is regarded as a typical mark of discontinuous DRX (DDRX). Because DRX requires stored strain energy as driving force, it is readily deduced that the crystallographic slip is highly activated during superplastic deformation of the present alloy. Apparently, the intensive occurrence of DDRX is inconsistent with the general features of FSS (mentioned in section 1).

Fig. 2 Flow curves of the alloy at various temperatures; (c) Stress-strain rate correlations at 1000 °C.

To further reveal the deformation mechanisms, textures of γ major phase have been investigated. As shown in Fig. 3(c,f,i), from the (100) pole figure one can note that the initially texture-free alloy produces a weak $\langle 100 \rangle$ fiber texture with a prominent cube texture during superplastic deformation, and its intensity is gradually increased. One should note that this is totally inconsistent with the general features of FSS which is characterized by

continuous weakening of the texture, as mentioned in section 1. Moreover, it should be pointed out that the predominance of cube texture is a mark of full-recrystallization in f.c.c. alloys as well as in TiAl alloys [14]. The formation of the present cube texture may be rationalized by the following way. It is well-known that $\beta/\text{B2}$ phase is much softer than γ phase at elevated temperatures, and thus $\beta/\text{B2}$ phase contributes for a considerable part of the deformation. Then the strain accumulation in γ phase is quite slow so that the DRX can restore the deformed grains in time. Consequently the recrystallization texture can replace the deformation texture.

Fig. 3 Microstructures and textures at various regions (corresponding to different local strain) of the specimen tested at 1000 °C. (a,d,g) IPF maps for γ major phase; (b,e,h) γ grain size histograms where the y-axis is in grain area; (c,f,i) Pole figures of γ major phase.

From the above observations, we can conclude that the microstructure and texture evolution of the present alloy are quite inconsistent with classical FSS theory. Indeed, the deformation mechanism of the present alloy may not be GBS but rather seems to be a slip-DDRX mechanism. Actually, the slip-DRX mechanism was firstly proposed by Johnson, Packer and their co-authors [15,16] when studied the superplasticity of hot-rolled Zn-Al eutectic alloys. They found that the initially round specimen became elliptical after an elongation of about

100 %, and they attributed this shape change to crystallographic slip. Although this slip-DRX model has not been widely accepted and subject to some criticisms [17], according to the present study we thought that it should be re-considered in certain cases especially for the materials with unfavorable microstructure for FSS. However, more investigations are necessary to rationalize the origination of the large m value (the major requirement for large elongations) of the present alloy.

Fig. 4 γ grain boundary map shows the dynamically nucleated grains. The IPF map shows the detailed orientation information of the γ nuclei in the red rectangular region.

5. Conclusion

In this study, we reported an unusual superplastic phenomenon in a TiAl alloy with coarse and bimodal grain structure. The results of this investigation suggested that crystallographic slip and discontinuous DRX were important or even predominant processes operating during deformation, rather than grain boundary sliding which is widely recognized as the major mechanism for fine-structure superplasticity. This evidence is associated with the intense occurrence of DRX and development of cube texture during superplastic deformation of the present TiAl alloy.

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