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Microstructural evolution of a forged TiAl based alloy during heat treatment at subtransus temperature①

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Abstract: Microstructural evolution of a forged TiAl based alloy during heat treatment from 1 180 ℃ to 1 300 ℃ was investigated. The grain sizes of the alpha phases as well as the sizes and the volume fractions of the gamma phases were evaluated as a function of heat treatment temperature and time. When the alloys are isothermally heat treated at subtransus temperatures, the sizes of gamma phases(D_γ) increase slightly with heat treatment temperature and time and those of alpha phases(D_α) and the volume fractions of gamma phases(φ_γ) vary significantly with holding time in the early stages of heat treatments, but after heat treatments for 2 h, φ_γ reveal little variations with holding times and D_α approach limits, which can be described by $D_{\alpha 0} = 0.65\ D_\gamma / \varphi_\gamma$. Besides, it has been found that the alpha phases in the specimens heat treated at 1 260 ℃ and 1 300 ℃ contain lamellar structures, at low temperatures, however, appear featureless.

Key words: TiAl base alloy; heat treatment; microstructure **Document code: A**

1 INTRODUCTION

The TiAl-based alloy has been known to be one of the most promising candidates for high temperature structural applications[1]. The properties of the alloys strongly depend on their microstructures[2]. The TiAl based alloys with a fully lamellar microstructures generally display high fracture toughness and creep resistance, but poor ductility. Inversely, materials with a duplex microstructures have good ductility, but poor fracture toughness and creep resistance[3,4]. At the same time, thermomechanical processings, such as hot extrusion and canned hot forging, and heat treatments are used to control the microstructural development in these alloys[5,6]. Isothermal heat treatments at temperatures above the alpha transition produce fully lamellar microstructures and lead to rapid grain growth of the alpha phase in the TiAl based alloys, whose kinetics can be described by the Hall-Petch expression[7]. When the deformed alloys are heat treated at subtransus temperatures, duplex microstructures can be obtained. Because of the coexistence of the alpha and gamma phases, the microstructural evolution is very complex. Up to now, little work has been carried out to understand the effect of heat treatment temperature and time on the sizes of alpha (D_α), gamma (D_γ) phases, and the volume fraction of gamma phase (φ_γ)[8]. But these studies are very important to adjust the microstructures and mechanical properties of TiAl alloys with duplex microstructures[9]. To meet these objectives, microstructural evolution in a forged TiAl based alloy during heat treatment was investigated in the present work.

2 EXPERIMENTAL

The TiAl based alloy for test with a nominal chemical composition (mole fraction, %) of Ti-48Al-2Cr was melted in a vacuum consumable electrode furnace. The remelting technology was used to reduce the composition segregation. After homogenization at 1 050 ℃ for 72 h and hot isostatic pressing at 1 250 ℃ for 4 h under argon pressure of 150 MPa, canned hot forging reported in Ref.[6] was adopted to deform the TiAl based alloy by 80 %. A series of experiments were conducted by heat treating the forged specimens for different duration ranging from 10 min to 8 h at 1 180, 1 220, 1 260 and 1 300 ℃ followed by air cooling. Nephot-Ⅱ optical microscope was used to observe the microstructures. Semiquantitative methods, such as the point counting and linear intercept methods, were employed to determine the microstructural parameters such as D_α, D_γ and φ_γ.

3 RESULTS

Fig.1 shows the microstructures of the forged TiAl based alloy. A uniform deformed microstructure is observed in the specimen.

Fig.2 displays the microstructures of TiAl heat treated for 2 h at 1 180, 1 220, 1 260 and 1 300 ℃, respectively. It is clear that the microstructures

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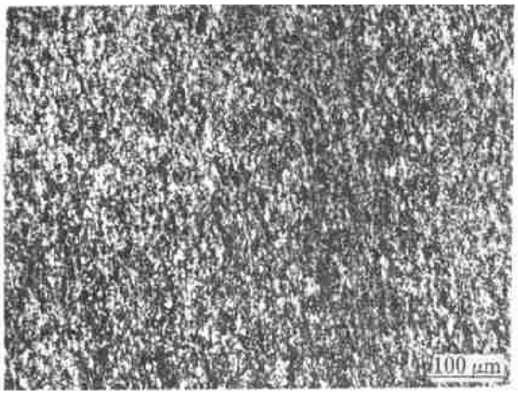

Fig.1 Optical microstructures of forged TiAl

depend on heat treatment temperatures. Fig.3 illustrates the temperature dependence of D_α, D_γ and φ_γ. From Fig.3, it is found that φ_γ decreases sharply with the increase of temperature, especially from 1 220 to 1 260 ℃. D_α increase gradually with temperature increasing from 1 180 ℃ to 1 220 ℃. When the temperatures are above 1 220 ℃, D_α increase sharply. The variation of D_γ with temperature is very weak. From Fig.2, it is also found that the relatively coarse alpha phase grains obtained at 1 260/1 300 ℃ contain lamellar structures. In contrast, the alpha phases obtained at 1 180/1 220 ℃ appear featureless.

Microstructures obtained from the specimens heat-treated at 1 260 ℃ are illustrated in Fig.4. D_α increases from 14 μm to 34 μm when the holding time increasing from 10 min to 2 h. Change of φ_γ from 38 % to 24 % is also observed. But the change of D_γ with the holding time is weak. Fig.5 illustrates the variations of D_α, D_γ and φ_γ with the holding time at 1 260 ℃. It can be seen that the changes of D_α, D_γ and φ_γ after about 2 h of heat-treatment are negligible. Therefore, the size of the alpha phase grains of TiAl heat treated for 2 h at 1 260 ℃ can be thought as the limit at this temperature. Similarly, D_α of the alloys heat treated for 2 h at various temperatures may also be treated as the limits $D_{\alpha 0}$ at those temperatures.

4 DISCUSSION

When the forged TiAl-based alloy is heat treated at subtransus temperatures, microstructural evolutions happen in the alloys so that the equilibriums between the alpha and gamma phases may be achieved. When they are heat treated at 1 260 ℃, 2 h is enough for the alloys to achieve the phase equilibrium. Therefore, increasing the holding time more than 2 h leads to little variation of φ_γ.

When the forged TiAl are heat treated at subtransus temperatures, D_α becomes time independent after about 2 h. This may result from the retardation of the alpha phase grain growth by gamma phase particles. The driving force for interface migration is mainly contributed to interfacial energy. The dragging force exerted by the second phase particles on a

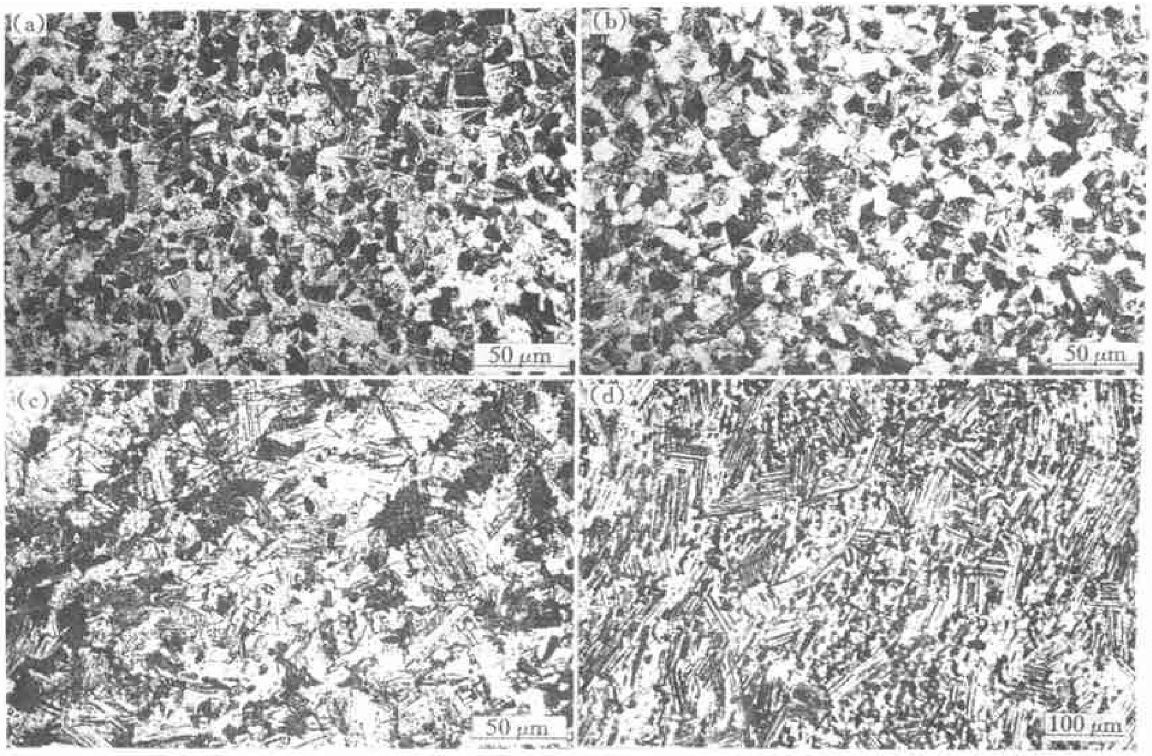

Fig.2 Microstructures of forged TiAl after heat-treatment at 1 180 ℃(a), 1 220 ℃(b), 1 260 ℃(c) and 1 300 ℃(d) 2 h followed by AC

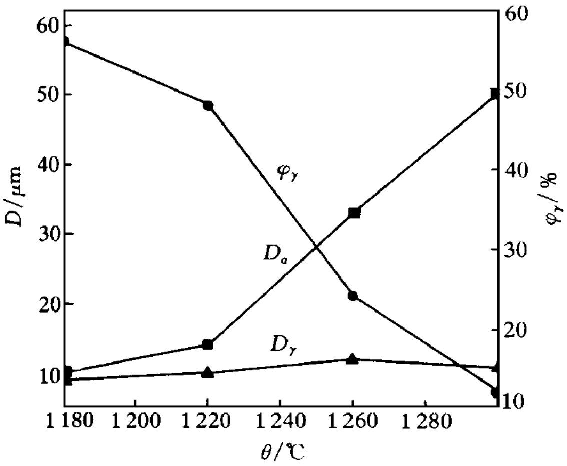

Fig.3 D_α, D_γ and φ_γ as a function of heat treatment temperature

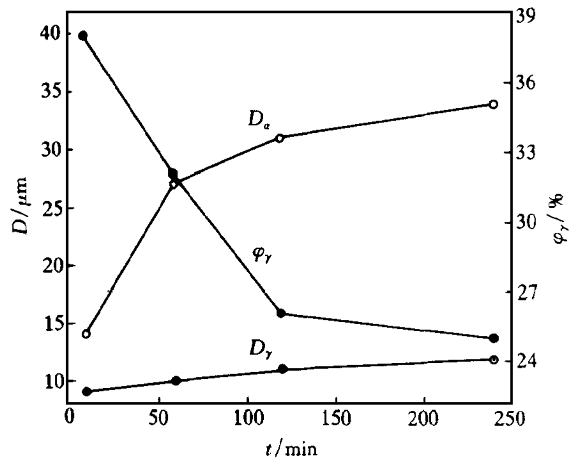

Fig.5 D_α, D_γ and φ_γ as a function of holding time

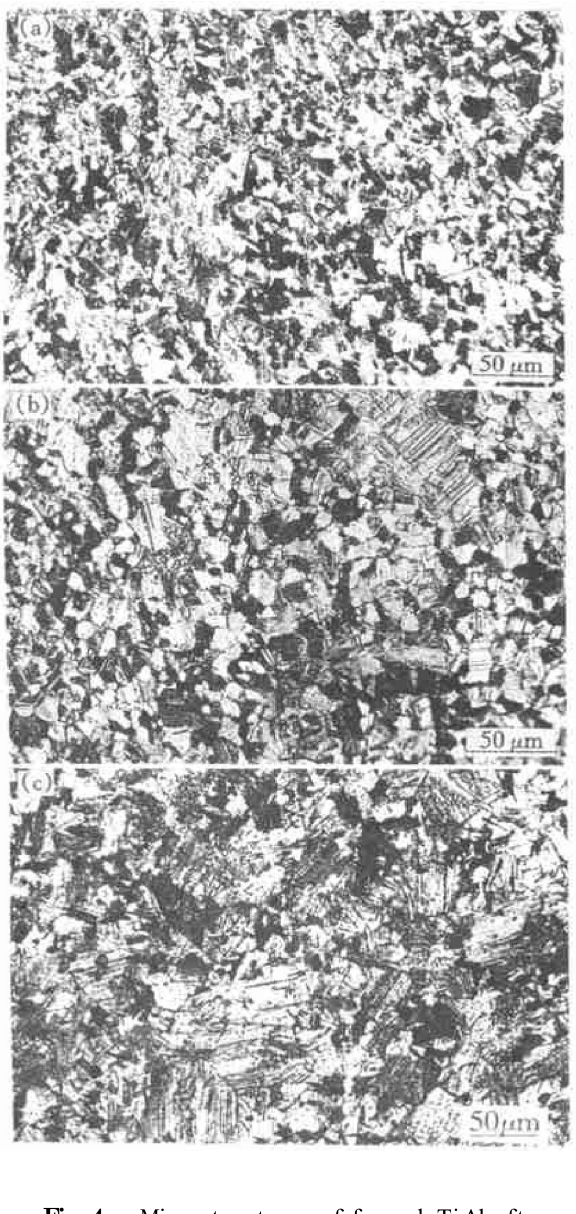

Fig.4 Microstructures of forged TiAl after heat treatments at 1 260 ℃ for 10 min(a), 1 h(b) and 4 h(c) followed by AC

migrating grain boundary is mainly determined by their volume fractions and sizes. The theoretical expression of the limit grain size at high volume fractions can be estimated from the following equation

$$D_{\alpha 0} = 4 D_\gamma / (3 \varphi_\gamma) \quad (1)$$

From Eqn.(1), it can be seen that the limit sizes of the alpha grains are in direct proportions to D_γ and in inverse proportions to φ_γ. But the statistical results of $D_{\alpha 0} \varphi_\gamma / D_\gamma$ of the experimental values are from 0.62 to 0.67. So, for this study the empirical expression can be described by

$$D_{\alpha 0} = 0.65 D_\gamma / \varphi_\gamma \quad (2)$$

From the comparison between Eqns.(1) and (2), it can be found that the experimental limit sizes of the alpha grains are much smaller than the theoretical values. This may result from the existence of a small amount of substitutional element Cr and traces of interstitial elements such as nitrogen and oxygen. It was reported that additions of substitutional elements such as Cr, Mo and Nb and interstitial elements such as N and O can retain the interface migration[10,11]. Zheng[10] discovered that the addition of Cr in TiAl leads to the formation of tiny particles of β_2 phase along the grain boundaries, thus hindering the interface migration. At the same time, Yun et al[12] reported that additions of a very small amount of O and N elements can reduce the interfacial energy and lower the driving force for the interface migration. Therefore the interfacial migrating rate in TiAl containing O and N elements is relatively slow. Under the common role of the second phase particles and the existence of Cr, N and O elements, relatively small microstructures can be obtained in the forged TiAl that are heat treated at subtransus temperatures.

The alpha phases in TiAl alloys heat treated at 1 260 ℃ and 1 300 ℃ contain lamellar structures. But the alpha phases in the specimens that are heat treated at 1 180 ℃ and 1 220 ℃ appear featureless. The different morphologies in the specimens can be explained on the basis of the aluminum content of the alpha. Jones et al[12] reported that the cooling rates for the formation of lamellar structures in the TiAl based alloy with different compositions depend on the aluminum content. It is relatively easy to form lamellar structures in high aluminum content alloys. In the low aluminum content alloys, the cooling rates for the formation of lamellar structures are relatively slow. When the forged TiAl based alloys are heat treated at high temperatures, the aluminum contents of the alpha phases are sufficiently high to decompose into $\alpha_2 + \gamma$ lamellae. In contrast, low temperature heat treatments lead to a high volume fraction of gamma phases and relatively low levels of aluminum contents in the alpha phases. Therefore, the alpha phases undergo ordering to α_2 during air cooling.

5 CONCLUSIONS

1) When the forged TiAl alloys are isothermally heat treated at subtransus temperatures, D_α and φ_γ vary significantly with holding time in the early stages of heat treatments, but after heat treatment for 2 h, φ_γ varies slightly with holding time and D_α approaches limits, which can be described by $D_{\alpha 0} = 0.65 D_\gamma / \varphi_\gamma$. In the mean time, D_γ reveals a little variations with temperature and time.

2) The alpha phases in the specimens heat treated at 1 260 ℃ and 1 300 ℃ contain lamellar structures, at low temperatures, however, appear featureless.

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