Hot Deformation Behavior and Microstructural Characteristics of Ti–46Al–8Nb Alloy

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Abstract

Hot deformation behavior, microstructural evolution and flow softening mechanism were investigated in Ti–46Al–8Nb alloy via isothermal compression approach. The true stress–strain curves exhibited typical work hardening and flow softening, in which the dependence of the peak stress on temperature and strain rate was obtained by hyperbolic sine equation with Zener–Hollomon (Z) parameter, and the activation energy was calculated to be 446.9 kJ/mol. The microstructural analysis shows that the alternate dark and light deformed ribbons of Al-rich and Nb-rich regions appeared and were associated with local flow involving solute segregation. The Al segregation promoted flow softening mainly arising from the recrystallization of γ phase with low stacking fault energy. The coarse recrystallized γ and several massive γ phase were observed at grain boundaries. While in the case of Nb segregation, β /B2 phase harmonized bending of lamellae, combined with the growth of recrystallized γ grains and $\alpha + \beta + \gamma \rightarrow \alpha + \gamma$ transition under conditions of temperature and stress, leading to the breakdown of α_2/γ lamellar colony. During the hot compression process, gliding and dissociation of dislocations occurred in γ phase that acted as the main softening mechanism, leading to extensive γ twins and cross twins in α/γ lamellae and at grain boundaries. In general, homogeneous microstructure during the hot deformation process can be obtained in TiAl alloy with high Nb addition and low Al segregation. The deformation substructures intrinsically promote the formability of Ti–46Al–8Nb alloy.

Keywords Titanium aluminides · Deformation behavior · Activation energy · Dynamic recrystallization · Local flow softening

1 Introduction

TiAl-based alloys are promising for high-temperature structural applications because of their excellent low density, high-temperature strength, high specific modulus and creep resistance [1–4]. Recently, high-Nb TiAl alloys (Ti–45–46Al–5–9Nb–C, B, Y) were studied as high-

temperature TiAl alloys, for turbines of aircraft engines and gas-burning power-generation plants [5–7]. Unfortunately, the widespread application of TiAl–Nb alloys is limited because of low-temperature brittleness and poor workability [8–10]. This requires tuning of thermo-mechanical treatment to obtain fine microstructure and improved ductility [11, 12].

In the case of TiAl–Nb alloy, the high Nb addition may significantly change the TiAl phase diagram greatly, leading to multiphase structure and complex microstructure because of solute segregation, which can easily produce cracks, localized flow stress and microstructural inhomogeneity during the hot deformation process [13, 14]. The improved deformability depends on optimizing the deformation temperature, strain rate and strain–stress conditions.

Additionally, the high-Nb TiAl alloys studied previously [15, 16] had inhomogeneous microstructure and severe solute segregation (S -, β - and α -segregation), which may

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aggravate local flow during deformation, retained-bulk α_2/γ lamellae and partially recrystallized during the hot deformation process. The solute Al and Nb segregation has significant effect on stacking fault energy of TiAl alloy, which may lead to localized flow and inhomogeneous microstructure [17, 18]. Limited attention has been given to the deformation substructure of high-Nb TiAl alloy, which may promote distinct dynamic recrystallization (DRX). Therefore, it is significant to investigate the hot deformation and the origin of dislocations and twinning of Ti-46Al-8Nb alloy so that the DRX mechanism in TiAl as well as the deformation of lamellae can be clearly explained. On the other hand, the bcc B2 phase from the β -stabilizer is added to improve hot deformability, because it is softer than α and γ phases at elevated temperature. This has been studied for the deformation process in Ti-47Al-2Cr-0.2Mo, Ti-47Al-2Cr-1Nb and Ti-44Al-4Nb-1Mo-(B, Si), etc. [19–22]. But the contribution of B2 phase is not deeply understood depending on the origin, morphology and distribution in TiAl–Nb alloys.

In this study, the hot deformation behavior and microstructure of Ti-46Al-8Nb alloy were studied by isothermal compression process. First, the stress exponent (n) and hot deformation activation energy (Q) were obtained by inducing Zener–Hollomon (Z) parameter to understand the hot deformation of Ti-46Al-8Nb alloy, and then the microstructural evolution and flow softening mechanism with various Z values were analyzed. Finally, the particular interest was on the effect of the solute segregation, multiphase structure, and deformation substructure on the local flow behavior to control the microstructure and deformability of high-Nb TiAl alloys.

2 Experimental Procedure

The intermetallic Ti-46Al-8Nb (at.%) alloy with an actual composition of Ti-46.3Al-7.6Nb (at.%) measuring approximately $\Phi 110$ mm $\times$ 400 mm was supplied in the form of cast cylindrical ingot, which was fabricated by the induction skull melting (ISM) under argon atmosphere. Cylindrical compressive samples with a diameter of 8 mm and a height of 15 mm were cut with an electric discharge machining perpendicular to the radial direction of the ingot in the center zone. Simulating isothermal compression tests were conducted on an MMS-300 thermo-simulator machine in an argon atmosphere in the temperature range of 1100–1250 °C and strain rates of 0.001–1 s^{−1}. The samples were heated at a rate of 5 °C/s and homogenized for 5 min at these test temperatures. All the samples were deformed to 0.6 true strain during compression and then water quenched to room temperature to study the microstructural evolution during deformation. To reduce

the friction, graphite papers were used between the samples and the machine grips. The deformed specimens were sectioned parallel to the compression axis for microstructural analysis.

The microstructure was characterized by scanning electron microscopy (SEM) in backscattered electron (BSE) mode, transmission electron microscopy (TEM) and electron backscattered diffraction (EBSD) techniques. The specimens for SEM observations were polished using standard metallographic procedure, while specimens for TEM were prepared by ion beam thinning. Electro-polishing was carried out in a solution consisting of 5 vol% perchloric acid + 35 vol% butanol + 60 vol% methanol at − 20 °C and 25 V to prepare EBSD samples. The EBSD data at intervals of 0.2 μ m were collected and analyzed by the OIM software.

3 Results and Discussion

3.1 Microstructure

Figure 1 shows the microstructure of as-cast Ti-46Al-8Nb alloy. The multiphase microstructure and solute segregation can be observed by SEM and TEMs. The lamellar colonies with alternating γ and α_2 phases were obtained from the $\alpha \rightarrow \gamma + \alpha_2$ transition with 0.2–0.8 μ m spacing, as shown in Fig. 1b. Additionally, γ phases at the grain boundaries and B2 phase with Nb enrichment in the core of the lamellar colonies were observed by TEM, as shown in Fig. 1c, d. The net-like Al segregation with black contrast and Nb segregation with white contrast exist in Ti-46Al-8Nb alloy, which mainly resulted from the peritectic reaction and low diffusivity of Nb [23]. The multiphase structure and high solute segregation in Ti-46Al-8Nb alloy may give rise to local flow and complex recrystallization process during hot deformation.

3.2 Flow Stress–Strain

The hot deformation process of TiAl alloys is sensitive to temperature, strain rate and extent of deformation, during which dislocations slide, climb and twins are nucleated, controlled by the thermal activation process. Then Zener–Hollomon (Z) parameter and thermal activation coefficients are introduced to establish the dependency of flow behavior on hot-deformed conditions in Ti-46Al-8Nb alloy.

The true stress–strain curves obtained from isothermal compression tests of the as-cast Ti-46Al-8Nb alloys are shown in Fig. 2. The flow stress rises rapidly with strain during the early stage of dislocation multiplication and exhibits a peak at a strain. The curve is followed by decrease of flow stress to a steady state with increasing

Fig. 1 Macrostructure **a** and microstructure of α_2/γ lamellae **b**, γ phase at lamellar boundaries **c** and B2 phase in lamellar colony **d** in the as-cast Ti-46Al-8Nb alloy

strain, showing dynamic recrystallization characteristic. With the increase in temperature and the decrease of strain rate, the flow stress decreases. At temperature of 1250 °C and strain rates of 0.01 and 0.001 s⁻¹, the flow stress is not sensitive to deformation conditions and was almost steady.

Furthermore, parameter Z was introduced to illustrate the dependence of flow stress on strain rate and temperature during hot deformation. It has been demonstrated that hot deformation is a process controlled by thermal activation which is frequently described by Eqs. (1) and (2). These equations are suitable for low stress regime ($\alpha\sigma < 0.8$) and high stress regime ($\alpha\sigma > 1.2$), respectively,

$$Z = \exp(Q/(RT)) = A'\sigma_p^{n'}, \quad (1)$$

$$Z = \exp(Q/(RT)) = A''\exp(\beta\sigma_p), \quad (2)$$

while these equations can be combined into Eq. (3), which follows a hyperbolic sine relationship and can be applied over a wider range of stress ($0.3 < \alpha\sigma < 2.8$),

$$Z = \dot{\epsilon}\exp(Q/(RT)) = A[\sinh(\alpha\sigma_p)]^n, \quad (3)$$

where Z is Zener–Hollomon parameter, Q is the activation energy at high-temperature compression, R is the universal

gas constant, n and n' are stress exponents, and A, A', A'', α and β are constants, of which $\alpha = \beta/n'$.

First, according to the true stress–true strain curves (Fig. 2) and converting Eqs. (1) and (2) into natural logarithmic form, the relationships between $\ln\sigma_p - \ln(\dot{\epsilon})$ and $\sigma_p - \ln(\dot{\epsilon})$ can be obtained, as shown in Fig. 3a, b. The measured constants of n' and β were determined to be 5.43 and 0.013, respectively, by linear fitting, and α was calculated to be $2.43 \times 10^{-3} \text{ MPa}^{-1}$.

Assuming that the hot compression temperature T is a constant, $\ln(\dot{\epsilon})$ and $\ln[\sinh(\alpha\sigma_p)]$ will have a linear relationship such that the stress exponent n for the alloy can be calculated by the following equation:

$$n = \frac{\partial(\ln \dot{\epsilon})}{\partial \ln[\sinh(\alpha\sigma_p)]}. \quad (4)$$

The relationship between $\ln[\sinh(\alpha\sigma_p)] - \ln(\dot{\epsilon})$ is shown in Fig. 3c, and n was estimated to be 4.04. The n values obtained at different deformation temperatures were 4.79, 4.02, 3.90 and 3.44, respectively, in agreement with other alloys that lower n values are usually observed at higher temperatures. The n value above 1150 °C is not applicable

Fig. 2 True stress–strain curves at different strain rates obtained from the isothermal compression tests of Ti-46Al-8Nb alloy at: **a** 1100 °C, **b** 1150 °C, **c** 1200 °C, **d** 1250 °C

for dislocation-climb-controlled deformed alloys, $4.2 < n < 5.9$, which implied that another thermally controlled process played an important role at these compressive temperatures. The calculated values of $\alpha\sigma$ are in the range of 0.25–1.89, which suggest that only hyperbolic sine relationship is suitable for calculation in this alloy.

The relationship, $\ln[\sinh(\alpha\sigma_p)] - (1/T)$, is shown in Fig. 3d, and if the strain rate $\dot{\epsilon}$ is fixed, activation energy (Q) is expressed by:

$$Q = -nR \cdot \frac{\partial \ln[\sinh(\alpha\sigma_p)]}{\partial (1/T)}. \quad (5)$$

According to the above calculation results, Q was 446.91 kJ/mol.

The relationship between Z parameter and peak stress in Ti-46Al-8Nb alloys is as follows:

$$\sigma_p = 411.286 \sinh^{-1}(3.107 \times 10^{-15} \cdot Z)^{0.254}. \quad (6)$$

As depicted in Fig. 4, the data obtained between 1100 and 1250 °C fit well with the hyperbolic sine function, and Z value increases with increasing peak stress.

The comparison of peak stress and activation energy values for different TiAl alloys is listed in Table 1. The

Q value in our work is higher than that of Ti-45Al-7Nb-0.3W, Ti-45Al-7Nb-0.4W and Ti-47Al-1V alloys, but smaller than that of Ti-43.9Al-4.3Nb-(Mo, B, Si) alloy. It is noted that Q values increase with increasing Nb content and lower Al content. This is consistent with previous studies [24–28]. Meanwhile, the peak stress of 255.2 MPa in our work is obviously higher than that of other TiAl alloys at 1200 °C and strain rate of 0.01 s⁻¹, which is most likely to be related to high Nb content. In high Nb-containing alloys, solute Nb atoms would increase the elastic distortion energy, affecting dislocation thermal diffusion and increasing the activation energy [29]. The reasons may directly determine the characteristics of true stress–true strain curves in our work. Furthermore, the higher deformation activation energy and flow stress suggest that the alloy is difficult to be deformed, because it needs more external energy to activate the dislocations and other substructures. Consequently, the microstructural evolution at various Z values and identification of hot deformation mechanisms are significant and are addressed below.

Fig. 3 Relationship between peak stress and deformation conditions for the Ti-46Al-8Nb alloy: **a** $\ln \sigma_p - \ln(\dot{\epsilon})$, **b** $\sigma_p - \ln(\dot{\epsilon})$, **c** $\ln[\sinh(\alpha\sigma_p)] - \ln(\dot{\epsilon})$, **d** $1000/T - \ln[\sinh(\alpha\sigma_p)]$

Fig. 4 Relationship between peak stress and Z parameter

3.3 Hot Deformation Microstructure

Figure 5 shows hot deformation microstructural evolution with decreasing $\ln Z$ values at different temperatures and strain rates for Ti-46Al-8Nb alloy. It may be noted that inhomogeneous microstructure was observed, with alternating deformed ribbons of Al-rich (dark region) at grain

boundaries and Nb-rich (light region) at the center of lamellae.

Figure 5a, b shows a typical ribbon structure with elongated α_2/γ lamellae and fine γ grains at the grain boundaries (dark Al-segregated region). At true strain of 0.6 at lower temperature and higher strain rate, the lamellar colonies showed a “Z”-shape kink or elongated morphology along the direction of normal to the stress in Fig. 5a. Dynamic recrystallization (DRX) could only be observed in Al-rich region. With decreasing $\ln Z$ value, the α_2/γ interfaces became normal to the compression stress and were further decomposed to some extent. Localized DRX can be observed in Al-segregated region and Nb-segregated region, as shown in Fig. 5b, c. Moreover, the α_2/γ lamellae colonies and solute segregation disappeared gradually with decreasing $\ln Z$. With further decrease in $\ln Z$ in Fig. 5d, e, the elongated α_2/γ lamellae decomposed into multiple structures with equiaxed α_2/γ lamellae, γ - and α -phase. The locally deformed Nb segregation ribbon gradually disappeared and was occupied by DRX grains, while the Al-segregated region was composed of coarse DRX grains and massive γ phase, as shown in Fig. 5d, e. Finally, the coarse

Table 1 Activation energy and peak stress of TiAl alloys during high-temperature compression deformation [11, 24–26]

Composition (at.%)	Initial microstructure	T (°C)	$\dot{\epsilon}$ (s ⁻¹)	σ_p (MPa)	Q (kJ/mol)
Ti–45Al–7Nb–0.3W	Near γ , HIP	1050–1200	1×10^0 – 1×10^{-3}	140	414
Ti–47Al–1V	Nearly lamellar (IM)	1000–1200	1×10^0 – 1×10^{-3}	125	404
Ti–43.9Al–4.3Nb–(Mo, B)	Nearly lamellar, HIP	1100–1250	1×10^0 – 1×10^{-2}	60	540.3
Ti–45Al–7Nb–0.4W	Nearly lamellar (PM)	1000–1200	1×10^{-1} – 1×10^{-3}	160	420
Ti–46Al–8Nb [this work]	Nearly lamellar (IM)	1100–1250	1×10^0 – 1×10^{-3}	255	446.9

The peak stress was obtained at 1200 °C and 0.01 s⁻¹

Fig. 5 Microstructures of hot-deformed Ti–46Al–8Nb alloys at different $\ln Z$ values: **a** $\ln Z = 36.85$ (1100 °C/0.1 s⁻¹), **b** $\ln Z = 33.17$ (1150 °C/0.01 s⁻¹), **c** $\ln Z = 31.89$ (1200 °C/0.01 s⁻¹), **d** $\ln Z = 30.69$ (1250 °C/0.01 s⁻¹) and the insertion shows the detailed view of the white box, **e** $\ln Z = 29.58$ (1200 °C/0.001 s⁻¹), **f** $\ln Z = 28.39$ (1250 °C/0.001 s⁻¹)

DRX grains and massive γ phase were observed with lowest $\ln Z$ value in hot-deformed Ti–46Al–8Nb alloy, as shown in Fig. 5f.

The above results indicated the local flow softening with alternating Al-rich ribbon and Nb-rich ribbon in the hot-deformed Ti–46Al–8Nb alloy. The solute segregations resulted in severe microstructural inhomogeneity, and the coarse DRX structure with massive γ phase was obtained at high temperature and low strain rate. This inhomogeneity is different from the previous studies on hot-deformed microstructure of Ti–44Al–8Nb alloy by Zhou et al. [28] and Ti–43.9Al–4.3Nb–0.9Mo–0.1B–0.4Si alloy by Li et al. [27], which had the multiphase microstructure of γ , α , B2 equiaxed crystals and retained α_2/γ lamellae. A detailed evolution of microstructure in the hot-deformed Ti–46Al–8Nb alloy is presented below.

3.4 Flow Localization and Deformation Mechanism

The local flow softening mentioned above can be closely related to high solute segregation in Ti–46Al–8Nb alloy.

Figure 6 shows the evolution of hot-deformed microstructure in Al-rich region at the grain boundary with different $\ln Z$ values. The dark elongated ribbon containing a number of DRX grains with the size of less than 2 μm at high $\ln Z$ value was observed, as shown in Fig. 6a. Furthermore, the DRX grains in Al-segregated region grew gradually with the decrease in $\ln Z$, and γ grains in the Al-segregated region grew rapidly from 1 to 40 μm , as shown in Fig. 6a–d. Finally, grown DRX grains and massive γ phase were present in the Al-segregated region at the grain boundary. The above results can be attributed to the fact that γ phase always recrystallized first at the grain

Fig. 6 Microstructures of Al-segregated region in the hot-deformed Ti-48Al-8Nb alloy with the different Z values: **a** $\ln Z = 37.77$ ($1150\text{ }^{\circ}\text{C}/1\text{ s}^{-1}$) and the inset shows a detailed view of the white box, **b** $\ln Z = 35.47$ ($1150\text{ }^{\circ}\text{C}/0.1\text{ s}^{-1}$), **c** $\ln Z = 31.89$ ($1200\text{ }^{\circ}\text{C}/0.01\text{ s}^{-1}$), **d** $\ln Z = 29.58$ ($1200\text{ }^{\circ}\text{C}/0.001\text{ s}^{-1}$)

boundary. The fine DRX grains would accelerate the crystal boundary migration and grain rotation, which produces stress concentration, leading to local softening at the grain boundary. The softening can be further facilitated by high Al-enrichment, which promotes $L(\alpha_2/\gamma) \rightarrow \gamma$ transition under temperature and deformation stress, as well as Nb solution, which decreases the stacking fault energy of γ phase. Finally, completely DRX grains are obtained first in Al-segregated region of the hot-deformed Ti-46Al-8Nb alloy.

The detailed substructure and deformation mechanism were investigated in the γ phase in Al-segregated region by TEM as shown in Fig. 7. The dislocation movement and pileup in γ grains can be observed in Fig. 7a. It is known that deformation of $L1_0$ structure γ -TiAl always occurs on $\{111\}$ planes by glide of ordinary dislocations with Burgers vector $\mathbf{b} = 1/2 \langle 110 \rangle$ and super dislocations with Burgers vector $\mathbf{b} = \langle 101 \rangle$ and $1/2 \langle 112 \rangle$, respectively [30]. These processes occur when there is sufficient thermal activation to allow long-range diffusion of dislocations. The comparatively low stacking fault energy of γ phase with high Nb in solution makes the dislocations difficult to cross slip or climb, which promotes dislocation pileup and sub-grain formation in γ phase.

In addition, extensive twinning with different thickness was observed in γ phase. Figure 7b, c shows mechanical twins or thin intersecting twins with fixed angle of about $70^{\circ}22'$. These twins were identified to be twinning systems of type $\mathbf{b} = 1/6 \langle 11\bar{2} \rangle$ that propagate on oblique $\{111\}$ planes [31], and the different morphologies are viewed from different $[111]$ and $[\bar{1}10]$ zone axes, respectively. Although the mode of twin nucleation in Ti-46Al-8Nb alloy is not clarified at this time, we speculate that it is related to the very low stack fault energy of the alloy. The stacking faults easily extend and partial dislocations increase rapidly under the deformation conditions, which would activate twinning. The deformation of the twins and the interaction between the twins and dislocations promote nucleation of DRX γ grains, as shown in Fig. 7d.

From the above results, the low-temperature brittleness of TiAl alloys is triggered by the small extent of operating slip systems in γ phase at low temperature, while more operative slip systems in γ -TiAl, comprising dislocations and super dislocations, are easily activated during high strain energy deformation at elevated temperature and consequently promote twinning and dynamic recrystallization.

Figure 8 shows detailed microstructure evolution and flow softening in the Nb-segregated region in the hot-

Fig. 7 TEM micrographs of deformed γ phase in Al-segregated region. **a** γ sub-grains induced by dislocation movement, **b** mechanical twins in γ grains, **c** dislocations pileup at the boundaries and intersection of the twins in γ grains, **d** DRX γ grains induced by $1/6\langle 112 \rangle\{111\}$ twins

deformed Ti-46Al-8Nb alloy. In contrast to the Al-rich region, the recrystallized grains were lacking, while a number of bending α_2/γ colonies can be observed, as shown in broken-line boxes of Fig. 8a, b, which shows typical shear deformation around the Nb-segregated region. The β -stabilizer Nb enrichment with B2-phase precipitation in the α_2/γ colonies can be observed, and the soft bcc-structure B2 phase with multiple-slip systems at high temperature may promote local flow around Nb-segregated region, leading to bending of α_2/γ lamellae. Simultaneously, the local shear strain around the Nb enrichment may promote α_2/γ colonies to DRX process of γ and α grains, as shown in Fig. 8b, which accelerates the decomposition of the α_2/γ lamellar structure.

Additionally, Fig. 8c, d shows detailed morphologies of bending α_2/γ colonies. The broad γ lath, thin α_2 lath and the DRX grains can be observed at the kinking point, which shows the breakdown of α_2/γ lamellae to equiaxed α and γ

grains, under the deformation stress and temperature. DRX grains in broad γ lath can be observed to be 2–6 μm , and the growth of recrystallized grains may destroy the lamellar structure, as shown in the dotted circle in Fig. 8c, d. From the above results, the bcc soft B2 phase from Nb segregation can easily induce local shear strain in lamellar colony at the deformation temperature, which promotes α_2/γ lamellar colonies to bend and break down. Therefore, appropriate addition of Nb in TiAl alloy may be beneficial to fine grains and homogeneous microstructure of the hot-deformed TiAl alloy.

The retained microstructure of bending lamellae is further illustrated by electron backscattered diffraction (EBSD), as shown in Fig. 9. The multiple morphologies including bending α_2/γ lamellae with 80.3° , recrystallized γ lath and equiaxed grains (γ and α_2 phases) at the bending points are observed in image quality (IQ) map, as shown in Fig. 9a. The γ -phase laths with low stacking fault energy

Fig. 8 Microstructures in Nb-rich region in hot-deformed Ti-46Al-8Nb alloy. **a** bending lamellar and elongated B2 phase, **b** deformation ribbon in Nb-enriched region and the inset shows detailed view of white box, **c**, **d** the bending lamellae and detailed microstructure

are first broken by the recrystallized grains with high-angle grain boundary (marked with black line) in Fig. 9b. Simultaneously, discontinuous and thin α_2 laths are observed, resulting from $\alpha \rightarrow \gamma$ transition and the growth of recrystallized γ grains. Furthermore, the equiaxed α and γ grains exist at the bending point mainly due to the growth of recrystallized γ phase and decomposition of α lamellae. Additionally, the grain orientation spread in Fig. 9c indicates that the bending lamellae has high local strain with high dislocation density and distortion energy (red color region), while the equiaxed grain region has low distortion because of stress release through phase transition and recrystallization. The lamellae with high distortion energy would be subsequently destroyed if the recrystallization is more sufficient. Simultaneously, the severe torsion during the deformation process results in the different lamellar orientations between the upper part and lower part of the selected region in Fig. 9d. Generally, the recrystallization and growth of γ phase accompany $\alpha \rightarrow \gamma$ transition and breaking of α_2/γ lamellar structure, especially bending

lamellae with high distortion energy, followed by the decomposition of α_2/γ lamellar to equiaxed γ and α grains. This also promotes flow softening process for the α_2/γ lamellar structure [32].

Figure 10 shows TEM micrographs describing the evolution of α_2/γ lamellar structure in hot-deformed Ti-46Al-8Nb alloy. Under deformation stress and temperature, extensive γ twins and dislocations are generated in lamellar colonies in Fig. 10a, which may be regarded as the prior condition of twinning and dislocation-induced dynamic recrystallization. Simultaneously, the recrystallized γ grain growth and $\alpha \rightarrow \gamma$ transition break the α_2/γ lamellae, as shown by dotted region in Fig. 10a, and then the multiple-phase structure retained α_2/γ lamellae, recrystallized α , γ grains is obtained in Fig. 10b, especially in the bending lamellar region. The breakdown of the α_2/γ lamellae contributes to stress relaxation, while the retained lamellae are accompanied by high dislocation density and distortion energy, as shown by the bright-field image in Fig. 10c. Moreover, the majority of α_2/γ lamellae can

Fig. 9 EBSD images of the evolution of bending lamellar colony of Ti-46Al-8Nb alloy. **a** Image quality (IQ) map, **b** phase map overlaid with high- and low-angle grain boundaries, **c** grain orientation spread (GOS) map, **d** inverse pole figure (IPF) map

decompose into equiaxed γ and α grains, and pileup of dislocations at the grain boundaries (Fig. 10d).

In the hot-deformed TiAl alloys, the retained α_2/γ lamellae with high resistance to deformation leads to flow anisotropy. Semiatin et al. [33] reported that the remnant lamellar colonies deteriorate the thickness gage of rolled sheets because of high flow stress. It is common that the inhomogeneous microstructure with coarse retained α_2/γ lamellae leads to wide scatter in mechanical properties in

forged, extruded and rolled TiAl alloys [34–36], which should be eliminated by improved hot deformation process and subsequent heat treatment. In the case of Ti-46Al-8Nb alloy with peritectic reaction, the retained α_2/γ lamellae can be aggravated by severe localized flow softening in the high Al-segregated region at the grain boundary, while the Nb-segregated region with soft B2 phase can promote the breakdown of kinked α_2/γ lamellar. Thus, homogeneous microstructure during the hot-deformed process can be

Fig. 10 Bright-field TEM images of the microstructures of hot-deformed Ti-46Al-8Nb alloy: **a** γ twins and growth of γ grains, **b** multiple-phase structure in the kink region, **c** retained lamellae, **d** recrystallized grains and sub-grain structure

obtained in TiAl alloys with high Nb content and low Al segregation.

4 Conclusions

The hot deformation behavior and microstructural characteristics of hot-deformed Ti-46Al-8Nb alloy were studied, and the following conclusions were obtained:

- (1) The true stress–strain curves of hot-deformed Ti-46Al-8Nb alloy exhibited typical work hardening and flow softening, and the peak stress was described by a Z parameter and the following equation: $\sigma_p = 411.286 \sinh^{-1}(3.107 \times 10^{-15} \cdot Z)^{0.254}$. The activation energy (Q) of the hot-deformed Ti-46Al-8Nb alloy was calculated to be 446.9 kJ/mol.
- (2) The inhomogeneous microstructure was observed in hot-deformed Ti-46Al-8Nb alloy, which was closely related to the localized flow in solute Al- and Nb-segregated regions. Al segregation promoted localized flow softening due to recrystallization of γ phase with low stacking fault energy, as well as the $L(\alpha_2/\gamma) \rightarrow \gamma$ transition under the deformation conditions. Nb segregation with soft B2 phase induced bending lamellar structure and breakdown of α_2/γ lamellar colony.
- (3) The γ -phase recrystallization and twinning, the kinked α_2/γ lamellar and $L(\alpha_2/\gamma) \rightarrow$ equiaxed $\alpha + \gamma$ transition, soft β phase and dislocation slip were the main softening mechanisms. Homogeneous microstructure during the hot-deformed process can be obtained in TiAl alloys containing high Nb and low Al segregation.

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