Hot Deformation Behavior of Ti–6Al–4V Alloy in β Phase Field and Low Strain Rate

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Abstract Hot deformation of Ti–6Al–4V alloy in β phase field is studied by means of compression testing at different temperatures and strains. Microstructural observations indicate that dynamic recrystallization of β grains occurs in the designated experimental conditions. Dynamic recrystallization was analyzed on the basis of the JMAK kinetics equation and strain hardening rate versus stress curves. The apparent activation energy for hot deformation was calculated as 168.5 kJ/mol. The variation of dynamically recrystallized grain size with deformation temperature and strain was studied and an empirical equation was derived.

Keywords Hot deformation · Ti–6Al–4V alloy · Kinetics · Dynamic recrystallization

Introduction

Ti–6Al–4V alloy is recognized as a α + β titanium alloy, in which the low-temperature α phase with hexagonal close-packed (hcp) structure shows high strength and oxidation resistance. However, the high-temperature β phase with body-centered cubic (bcc) structure exhibits better ductility and formability. In fact, a suitable balance between Al (α -phase stabilizer) and V (β -phase stabilizer) elements produces interesting properties such as high specific strength, excellent fracture toughness, and acceptable heat and corrosion resistance [1, 2]. These properties make this type of alloy a desirable choice in the aerospace industry [3]. In spite of broad application of Ti–6Al–4V, its formability is

rather poor and the alloy is difficult to form into complex shapes [4]. The β transition temperature of the applied alloy is approximately 995 °C.

Considering the high ductility and high workability of the β phase, β -processing (mechanical working above β -transus), especially hot forging and extrusion [5, 6], plays a significant role in preparing defect-free and homogenous industrial products or semi-products.

Recent studies on equiaxed [7] and lamellar [8] β phase field Ti–6Al–4V microstructures reveal that dynamic recrystallization (DRX) occurs during hot deformation at strain rate range of 10^{-3} – 10^{-1} s $^{-1}$. DRX is now recognized as one of the softening mechanisms in hot deformation of crystalline materials [9]. This mechanism plays an important role in high-temperature processing due to generation of fine grains that lead to improvement of mechanical properties. Therefore, it can be applied to control the microstructure and the properties of alloys in industrial processing [10].

Generally speaking, the DRX mechanism is described as nucleation and growth of new grains during plastic deformation. Thus, the initiation of DRX occurs during sufficiently high temperature when a critical strain (ϵ_c) is attained [11]. Grain boundaries have been considered as preferred sites of nucleation due to creation of the critical conditions in these regions [12]. Small dislocation-free nuclei can grow by migration of high-angle boundaries into surrounding materials that contain dislocations [13].

Various models including Johnson–Mehl–Avrami–Kolmogorov (JMAK) model [14], Sandström and Lagneborg model [15], Monte Carlo computerized model [16], Cellular automata (CA) model [17], and mechanical approach model [18] have been considered for evaluation and prediction of the DRX behavior. Although various studies were focused on hot deformation behavior of

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Ti–6Al–4V alloy in β phase field, only limited efforts have been made to propose kinetics models that are based on microstructure evolution during DRX. Ding and Guo [19] and Zhao et al. [20] developed models for simulation of microstructural evolution and flow stress behavior of β phase field based on the fundamental metallurgical principles of DRX and the CA method.

In general, β phase processing leads to a finer structure in Ti–6Al–4V due to the action of the dynamic recrystallization mechanism. This increases the transformation rate of β to α during continuous cooling from the β phase field. Furthermore, a smaller β grain size shifts the TTT diagram to shorter times and decreases the martensite start (M_s) and martensite finish (M_f) temperatures. Therefore, the formation of martensite phase occurs at higher cooling rates and the hardenability of the alloy decreases. Moreover, it should be noted that β phase processing can also slightly affect the α phase morphology, owing to the displacement of the TTT diagram [21].

The main purpose of this study is to evaluate DRX behavior in Ti–6Al–4V alloy regarding the flow curves, microstructural observations, and JMAK kinetics model. Another important aspect of this research is to establish a general relationship between the deformation process parameters (i.e., strain, strain rate, and temperature) and the dynamically recrystallized grain size.

Experimental Procedures

The starting material was hot forged commercial grade Ti–6Al–4V, the chemical composition of which is listed in Table 1. Cylindrical specimens of 12 mm height with 8 mm diameter were machined from the as-received stock for hot deformation testing. A groove was cut into the top and bottom surfaces of the cylinders to trap lubricant and assist in reducing friction. It is noted that a combination of glass powder, graphite, and mica sheet was used at both ends of the specimens as lubricant. In order to measure the actual temperature and any adiabatic temperature rise during testing, a small hole was drilled at mid-height of each specimen for inserting a thermocouple. Isothermal hot compression testing was performed in a computer-controlled servo hydraulic testing machine. The platens and specimen were surrounded by a resistance heating split

furnace, which was interfaced with a digital programmer and controller.

Prior to the deformation, each specimen was heated to the deformation temperature and held at this temperature for 5 min to eliminate the thermal gradient. Hot compression experiments were conducted at a constant true strain rate of 0.001 s^{-1} and different temperatures of 1000, 1050, and 1100 °C. For each temperature, the test was stopped at a predefined strain (five final true strains in the range of 0–0.7) and then the specimen quenched in water to capture the final microstructure.

The load–deflection data obtained in the compression experiments was converted to true stress–true strain data using the standard method. In order to plot the flow curves, data obtained were corrected for friction by applying the following equation [22]:

$$\sigma = \frac{Pc^2}{2(\exp(c) - c - 1)}, \quad c = \frac{2\mu r}{h}, \quad (1)$$

where σ is the friction corrected flow stress, P is the uncorrected flow stress, r and h are the instantaneous radius and height of specimen, respectively, and μ is the friction coefficient which is determined according to the amount of barreling for each specimen [23, 24].

In an earlier study on the flow behavior of Ti–6Al–4V alloy with lamellar starting microstructure, it was shown that the flow softening due to the deformation heating is negligible in compression testing at high temperatures and low strain rate of 0.001 s^{-1} [25].

The deformed specimens were sectioned at mid-plane, parallel to the compression axis. The microstructure of samples was characterized using light microscopy. Samples were mechanically polished through 1200 grit wet SiC emery paper, followed by fine polishing with alumina. The as-polished specimens were etched with Kroll's reagent (2% HF + 4% HNO₃) and light micrographs were recorded at various different magnifications. The area fraction of recrystallized grains was determined by a combination of automated classification of different grains based on their size and manual designation of those recrystallized grains not identified automatically.

Results and Discussion

Microstructural Observations

It is essential to describe a criterion for the detection of prior recrystallized microstructure from prior un-recrystallized β grains. For this purpose, the nature of dynamic recrystallization, i.e., the formation of dynamic recrystallized colonies, has been considered in the present research. In fact, it has been shown that as the recrystallized grains are

Table 1 Chemical composition of the starting material used in this investigation

Al	V	Fe	O ₂	C	N ₂	H ₂	Ti
6.05	3.98	0.2	0.18	0.02	0.01	0.0051	Balance

relatively small compared to the original grain size, the new nuclei preferentially form at sites adjacent to recrystallized grains, so that “colonies” of recrystallized grains spread out from grain boundaries as recrystallization proceeds [9].

It is interesting to note that the specimens were quenched in water to retain the high temperature deformation structures. Actually, the prior β grains are transformed into martensite structure (α'), whereas their boundaries are retained under this condition [26]. However, this phase transformation makes the detection of dynamic recrystallized grains very difficult [8] and can result in lower quality light micrograph images of the as-quenched microstructure.

A typical microstructure of the as-received hot forged bar is shown in Fig. 1. The initial microstructure consists of the kinked or separated lamellar α colonies in coarse prior β grains of approximately 125 μm . The kinking or break-up of the lamellae is a result of induced hot forging in the $\alpha+\beta$ phase field.

Figure 2 demonstrates the microstructural evolution of Ti-6Al-4V during hot deformation at 1050 $^{\circ}\text{C}$ and a strain rate of 0.001 s^{-1} and various strains. Colonies of dynamically recrystallized grains are observed at prior β grain boundaries, some of which are indicated by black arrows. In addition, the induced compressive stress causes extension of the unrecrystallized grains toward the direction of maximum shear stress, which is approximately 45° with respect to the compression axis. The micrographs also confirm that the volume fraction of DRX grains gradually increases with strain.

Micrographs of the specimens deformed up to strain of $\varepsilon = 0.5$ at a strain rate of 0.001 s^{-1} indicate that the increase of the deformation temperature from 1000 $^{\circ}\text{C}$ (Fig. 3a) to 1100 $^{\circ}\text{C}$ (Fig. 3b) leads to an increase in the volume fraction of DRX grains. This could be attributed to the increase in the diffusion rate at the higher temperature,

which accelerates both the grain nucleation rate and the grain boundary mobility.

It is interesting to note that the volume fraction of meta-dynamically recrystallized grains is negligible in all the deformed specimens. In fact, no evidence of new grains with smooth boundaries is observed and the new grains are almost dynamically recrystallized grains with irregular boundaries. This can be attributed to the very low deformation strain rate imposed in the present investigation.

DRX Analysis by θ – σ Curves

Concerning the critical condition for the initiation of DRX, the critical strain (ε_c), critical stress (σ_c), and saturation stress (σ_s is the steady-state stress for dynamic recovery in the absence of DRX mechanism) are important variables in this hot deformation study. It is impossible to calculate ε_c and σ_c when peak strain (ε_p) and peak stress (σ_p) are not detectable in the flow curve. In such a case, the curve of strain hardening rate ($\theta = d\sigma/d\varepsilon$) versus flow stress (σ) is used to determine ε_c , σ_c , and σ_s .

The θ – σ and S – σ curves for hot deformation of Ti-6Al-4V at 1000, 1050, and 1100 $^{\circ}\text{C}$ are shown in Fig. 4, where $S = -d\theta/d\sigma$. It is clear that θ nonlinearly decreases with increasing the flow stress until reaching to a zero value at σ_p . In earlier studies [10, 11, 18], it has been shown that the presence of a peak stress in a constant strain rate flow curve leads to an inflection point in the θ – σ curve. Concerning irreversible thermodynamics concepts, this point corresponds to the initiation of DRX. The inflection in θ – σ curve can mathematically be stated as

$$\frac{\partial^2 \theta}{\partial \sigma^2} = 0 \quad \text{or} \quad \frac{\partial S}{\partial \sigma} = 0. \quad (2)$$

Fig. 1 Microstructure of as-received material at different magnifications

Based on Eq (2), the coordinates of the minimum point in S – σ curve determines S_c and σ_c values.

Considering θ – σ curve, it can be assumed that in the absence of dynamic recrystallization, dynamic recovery takes place and θ follows a linear decline with the slope of $(-S_c)$. In such a case, σ_s can be considered as the stress value corresponding to $\theta = 0$, as shown in Fig. 4. The calculated values of ε_c , σ_c , ε_p , σ_p , and σ_s for hot deformation of Ti–6Al–4V at

different temperatures are listed in Table 2, in which all the calculated quantities decreased with temperature.

Figure 5 shows the variation of the abovementioned critical and peak stress and strain with temperature. It is evident that all the calculated values decrease with increasing temperature.

In a recent study [11], it is confirmed that the ratios of DRX critical stress and strain to their corresponding peak

Fig. 2 Light micrographs of Ti–6Al–4V specimens deformed at 1050 °C/0.001 s^{−1} and strain of **a** $\varepsilon = 0.1$, **b** $\varepsilon = 0.2$, **c** $\varepsilon = 0.3$, **d** $\varepsilon = 0.4$, and **e** $\varepsilon = 0.5$, where the *black arrows* indicate dynamically recrystallized grains at prior β grain boundaries

Fig. 3 Light micrographs of Ti–6Al–4V specimens deformed at **a** 1000 °C/0.001 s^{−1} and strain of $\varepsilon = 0.5$, **b** 1100 °C/0.001 s^{−1} and strain of $\varepsilon = 0.5$, where the *black arrows* indicate dynamically recrystallized grains at prior β grain boundaries

values are constant. Concerning data presented in Table 2 and Fig. 5, the following empirical relationships can be derived:

$$\varepsilon_c \approx 0.4\varepsilon_p \quad (3)$$

$$\sigma_c \approx 0.95\sigma_p \quad (4)$$

Evaluation of DRX Kinetics

The JMAK equation is extensively applied to study the kinetics of DRX [20, 24]. This equation represents the variation of the volume fraction of dynamically recrystallized grains (X) with transformation time (t) under an isothermal condition

$$X(t) = 1 - \exp(-Kt^n), \quad (5)$$

where K and n are the kinetics constant and the JMAK exponent, respectively. In general, the constant K is stated as a function of temperature which depends on the rate of nucleation and growth

$$K(T) = K_0 \exp\left(-\frac{Q}{RT}\right), \quad (6)$$

where Q is the apparent activation energy, K_0 and R are the pre-exponential factors, and the gas constant, respectively. By taking natural logarithms of Eq (5), the relationship can be expressed as

$$\ln\{-\ln(1 - X)\} = \ln K + n. \quad (7)$$

Plotting $\ln\{\ln[1/(1 - X)]\}$ versus $\ln(t)$ results in a straight line with the intercept and slope equal to $\ln(K)$ and n , respectively.

Figure 6 shows the variation of $\ln\{\ln[1/(1 - X)]\}$ versus $\ln(t)$ at a strain rate of 0.001 s^{−1} for different temperatures. It is clear that a linear plot appropriately fits to the experimental data, from which K and n at different

temperatures are calculated. Utilizing Eq 5 and data presented in Fig. 6, variation of the volume fraction of dynamically recrystallized grains (X) with transformation time (t) at different temperatures was calculated and shown in Fig. 7.

The mean value of the calculated JMAK exponent (Fig. 6) is close to $n = 1$. It is generally accepted that the n value is an indication of nucleation mode and varies between 1 and 3 [24]. For isothermal diffusion-controlled transformation processes, when nucleation occurs heterogeneously at grain boundary edges and early site saturation is assumed, the product phase grows essentially with cylindrical symmetry around the grain edges and an exponent of 1 is obtained [27].

Based on Eq 6, the JMAK model can be used to determine the apparent activation energy for hot deformation. As depicted in Fig. 8, plotting $\ln(K)$ versus the reciprocal of the absolute temperature ($1/T$) leads to a straight line with a slope is equal to $(-Q/R)$. The apparent activation energy for hot deformation of Ti–6Al–4V in β phase field is estimated to be about 168.5 kJ/mol, which is comparable with the activation energy calculated by Prasad et al. [28]. The calculated apparent activation energy in this study is comparable to that calculated for the self-diffusion in β -Ti (153 kJ/mol) [8]. This confirms that the DRX mechanism in β phase field is a diffusion-controlled phenomenon.

Dynamically Recrystallized Grain Size

As the grain size of metals and alloys directly influences their mechanical properties, the prediction of dynamically recrystallized grain size is an important tool. The average dynamically recrystallized grain size can be estimated by the following equation [29]:

Fig. 4 Plots of θ and S in terms of flow stress at different deformation temperatures. **a** 1000 °C, **b** 1050 °C, and **c** 1100 °C

Table 2 Calculated values of ε_c , σ_c (MPa), ε_p , σ_p (MPa), and σ_s (MPa)

	$T = 1000\text{ }^{\circ}\text{C}$	$T = 1050\text{ }^{\circ}\text{C}$	$T = 1100\text{ }^{\circ}\text{C}$
ε_c	0.094	0.087	0.081
σ_c	22.70	20.33	14.93
ε_p	0.228	0.226	0.204
σ_p	23.53	21.25	16.31
σ_s	24.68	22.39	17.70

Fig. 5 Critical and peak values versus the reciprocal of the absolute temperature. **a** Critical and peak strains and **b** critical and peak stresses

Fig. 6 Plots of $\ln\{\ln(1/(1-X))\}$ versus $\ln(t)$ at different temperatures

$$D_{\text{drx}} = \left(\frac{4S}{\pi N} \right)^{1/2}, \quad (8)$$

where D_{drx} , S , and N are the average diameter, the area and the number of the dynamically recrystallized grains, respectively. Based on Eq 8 and experimental data, the average dynamically recrystallized grain size of Ti–6Al–4V specimens deformed at different temperatures and

Fig. 7 Variation of volume fraction of dynamically recrystallized grains versus time at different temperatures

Fig. 8 Plot of $\ln(K)$ versus the reciprocal of the absolute deformation temperature

Table 3 Average dynamically recrystallized grain diameter (μm) at different temperatures (T) and strains (ϵ)

ϵ	$T = 1000^\circ\text{C}$	$T = 1050^\circ\text{C}$	$T = 1100^\circ\text{C}$
0.1	33.0	37.7	56.2
0.2	34.5	39.5	60.1
0.3	36.6	41.5	63.5
0.4	39.0	43.0	66.1
0.5	41.3	46.4	70.6

strains are calculated and listed in Table 3. It is evident that the dynamically recrystallized grains grow by increasing either the deformation strain or the deformation temperature.

It has been generally accepted that the dynamically recrystallized grain size is proportional to the process parameters, i.e., strain, strain rate, time, and temperature. A plot of $\ln(D_{\text{drx}})$ versus $\ln(t)$ at different deformation temperatures is shown in Fig. 9, where a linear fit appropriately correlates to the experimental data. It is interesting to note that the slopes of the linear curves are about identical; however, the intercepts of the curves are dependent on the

Fig. 9 Variation of dynamically recrystallized grain size versus deformation time

Fig. 10 Variation of the intercept of the lines in Fig. 9 (I) versus the reciprocal of the temperature

temperature, as shown in Fig. 10. Considering experimental data presented in Figs. 9 and 10, the following empirical equations can be derived:

$$\ln(D_{\text{drx}}) = 0.12 \ln(t) - \frac{9191.63}{T} + 10.11 \quad (9)$$

$$\ln(D_{\text{drx}}) = 0.12 \ln(\epsilon) - \frac{9191.63}{T} + 10.94 \quad (10)$$

Conclusions

Hot compression test was performed on Ti–6Al–4V alloy with lamellar starting microstructure in the temperature range of 1000–1100 °C (β phase field) at low constant strain rate of 0.001 s^{-1} . Based on the results obtained, the following conclusions can be drawn:

- Microstructural observation shows that DRX is the dominant restoration mechanism under the imposed deformation conditions and MDRX rarely occurred.

- Strain hardening rate versus stress curves ($\dot{\theta}$ – σ) were analyzed and critical and peak stress and strain were calculated. Empirical relations of $\varepsilon_c \approx 0.4\varepsilon_p$ and $\sigma_c \approx 0.95\sigma_p$ were established.
- The JMAK equation was effectively employed to predict the volume fraction of dynamically recrystallized grains in the Ti–6Al–4V alloy. JMAK exponent of 1 was obtained for all the experimental conditions which indicates dominant early site saturation and grain boundary edge nucleation.
- The dependence of dynamically recrystallized grain size on the deformation time and temperature was investigated and an empirical relationship was derived.

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