Materials and Design 88 (2015) 32–40

Contents lists available at ScienceDirect

Materials and Design

journal homepage: www.elsevier.com/locate/jmad

Effect of the alpha grain size on the deformation behavior during isothermal compression of Ti–6Al–4V alloy

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ARTICLE INFO

Article history:
Received 24 June 2015
Received in revised form 8 August 2015
Accepted 25 August 2015
Available online 2 September 2015

Keywords:
Ti–6Al–4V alloy
Flow stress
Apparent activation energy for deformation
Processing maps
Microstructure

ABSTRACT

The effects of alpha grain size on the flow stress, the apparent activation energy for deformation (Q) and the processing maps of Ti–6Al–4V with an equiaxed microstructure are thoroughly investigated using isothermal compression tests, and detailed explanation is given based on the microstructure observation and quantitative analysis. The shapes of flow curves are dependent on the microstructure characteristic of the alloy before deformation and during the deformation process. The flow stress increases with increasing equiaxed alpha phase, but decreases with increasing alpha grain size. The Q-values for d_{r1} and d_{r2} are smaller than those for d_{r3} and d_{r4}, respectively, which is possibly attributed to that Ti–6Al–4V alloy for d_{r1} and d_{r2} which exhibits a very strong grain-boundary sliding (GBS) mode besides dominant dislocation glide/climb mechanism. The local efficiency maxima and unstable regions in processing maps change with the alpha grain size, which implies that proper hot-working domains should be modified in different grain size range so as to meet the precision forging process.

1. Introduction

Ti–6Al–4V alloy, as a type of two-phase titanium alloy, has been widely used in aerospace applications due to its high strength-to-weight ratio, attractive mechanical and corrosion resistant properties. Recently, a number of efforts have focused on constitutive modeling and the analysis of plastic flow to improve the hot workability of Ti–6Al–4V alloy [1–8]. Bai et al. [9] proposed a set of mechanism-based unified elastic-viscoplastic constitutive equations to model the mechanisms of plastic flow softening for the two-phase titanium alloy Ti–6Al–4V in hot forming conditions. Momeni and Abbasi [10] investigated the effect of hot working on flow behavior of Ti–6Al–4V alloy in single phase and two phase regions.

Although these reports have revealed the flow behavior and physical mechanisms which were beneficial to design the processing parameters, the effect of alpha grain size on the deformation behavior of Ti–6Al–4V alloy during isothermal compression is still needed. The Hall–Petch equation has been applied to describe the dependence of yield stress on grain size at room temperature for different metals and alloy systems. The equation shows that the yield stress is inversely proportional to the grain size, thus:

$$\sigma_y = \sigma_0 + k_y d^{-1/2} \tag{1}$$

where σ_y is the yield strength (MPa), σ_0 is the start stress for dislocation movement or the resistance of the lattice for dislocation motion (MPa), k_y is the strengthening coefficient (MPa·mm$^{1/2}$), and d is the average grain size (μm).

It seems to be a satisfactory description of the dependence of yield stress on grain size when a somewhat limited range of grain size [11]. However, the Hall–Petch behavior should be considered not a universal law. For example, Semiatin and Bieler [12] investigated the effect of alpha platelet thickness on the plastic flow of Ti–6Al–4V with a transformed microstructure, and pointed out that the experimental results at lower strain rates (0.001 and 0.01 s^{-1}) did not exhibit a distinct Hall–Petch behavior at 815 and 900 °C. Therefore, the effect of grain size on the flow stress was very complex and its influence varied with processing parameters such as deformation temperature, strain, and strain rate for different materials. Nowadays, a number of studies were conducted to recognize the alpha grain size effect on flow stress for many materials such as pure aluminum [13], 7075 aluminum alloy [14], AZ91 magnesium alloy [15], 47Zr–45Ti–5Al–3V alloy [16], stainless steel [17]. The objective of present research is to analyze the effect of alpha grain size on the flow stress of equiaxed Ti–6Al–4V alloy at different processing parameters, and detailed explanation is given with the help of the microstructure observation and quantitative analysis. Moreover, the Q-values for different alpha grain sizes are calculated on the basis of flow stress–strain curves. The processing maps of Ti–6Al–4V with different alpha grain sizes are established, in which the optimal processing parameters and unstable domains are provided.

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http://dx.doi.org/10.1016/j.matdes.2015.08.130
0264-1275/

Table 1
Chemical composition of the as-received Ti–6Al–4V alloy (mass fraction in %).

Al	V	Fe	Si	C	N	O	H	Ti
6.05	4.32	0.22	<0.04	0.024	0.006	0.16	0.003	Bal.

2. Materials and procedures

2.1. Materials

A piece of bar stock Ti–6Al–4V with a 30.0 mm diameter was used in this study. The chemical composition of this alloy was shown in Table 1. The microstructure of as-received Ti–6Al–4V alloy was composed of fine equiaxed primary α phase and a small amount of intergranular β. The β transus temperature for this alloy was determined to be 985 °C via a technique involving heat treatment followed by optical metallography.

In order to investigate the effect of alpha grain size on the deformation behavior of Ti–6Al–4V alloy, sections of the Ti–6Al–4Vbar stock were heated to produce four equiaxed microstructures with different alpha grain sizes. According to normal grain growth kinetics of titanium alloys, the static grain growth as an atomic diffuse process affected by thermal effect was related to grain boundary mobility. Thus, it was given by the following expression [18]:

$$D^n - D_0^n = Kt\exp\left(-\frac{Q_{pd}}{RT}\right) \quad (2)$$

where Q_{pd} is the activation energy for boundary diffusion ($kJ \cdot mol^{-1}$), D and D_0 denote the initial and final grain sizes (μm), t is the annealing time (s), R is the gas constant ($8.3145\ J \cdot mol^{-1} \cdot K^{-1}$), T is the absolute temperature (K), n is the grain growth exponent, and K represents the material constant.

To obtain four different alpha grain sizes, the heat treatments were utilized through the following procedures: (1) heating at 930 °C for 2 h, 14 h, 36 h, and 60 h, respectively, and then furnace-cooling to room temperature (FC); and (2) heating at 600 °C for 2 h, and then air-cooling to room temperature (AC). Fig. 1 shows the resulting microstructures consisting of different alpha grain sizes: (a) $d_{r1} = 9.7$ μm; (b) $d_{r2} = 12.7$ μm; (c) $d_{r3} = 15.2$ μm; and (d) $d_{r4} = 17.3$ μm.

Additional heat treatments were given to simulate the subsequent preheat used prior to isothermal compression. Samples with four different alpha grain sizes were heated at 860 °C, 890 °C, 920 °C and 960 °C, and soaked for 5 min. Typical optical micrographs of Ti–6Al–4V microstructures with different alpha grain sizes during heating prior to isothermal deformation are given in Fig. 2. The corresponding microstructure variables, including the grain size and volume fraction of equiaxed alpha are listed in Table 2. These measurements suggest a competition between the alpha grain growth and the α → β phase transformation during heat treatment.

2.2. Isothermal compression tests

Cylindrical compression specimens with 8.0 mm diameter × 12.0 mm height were machined from the heat-treated bars; the specimen dimension was from the Chinese standard GB/T 7314-2005. Isothermal compressions were carried out a Gleeble-3500 simulator at deformation temperatures of 860 °C, 890 °C, 920 °C and 960 °C, strain rates of $0.01\ s^{-1}$, $0.1\ s^{-1}$, $1.0\ s^{-1}$ and $10.0\ s^{-1}$. Specimens were heated to test temperature, soaked for 5 min, and then upset under constant strain rate conditions to a height reduction of 60%. Flow stress–strain curves were recorded automatically during isothermal compression. After isothermal compression, the specimens were cooled in air to room temperature so as to be consistent with the actual forging process of Ti–6Al–4V alloy. To observe microstructure evolution, the specimens were axially sectioned, electropolished and chemically etched in a solution of 30 ml HF, 10 ml HNO_3, 70 ml H_2O_2 and 200 ml H_2O and checked on a Leica DMI3000M optical microscope. The microstructures were observed in the central region of each specimen. The grain size and

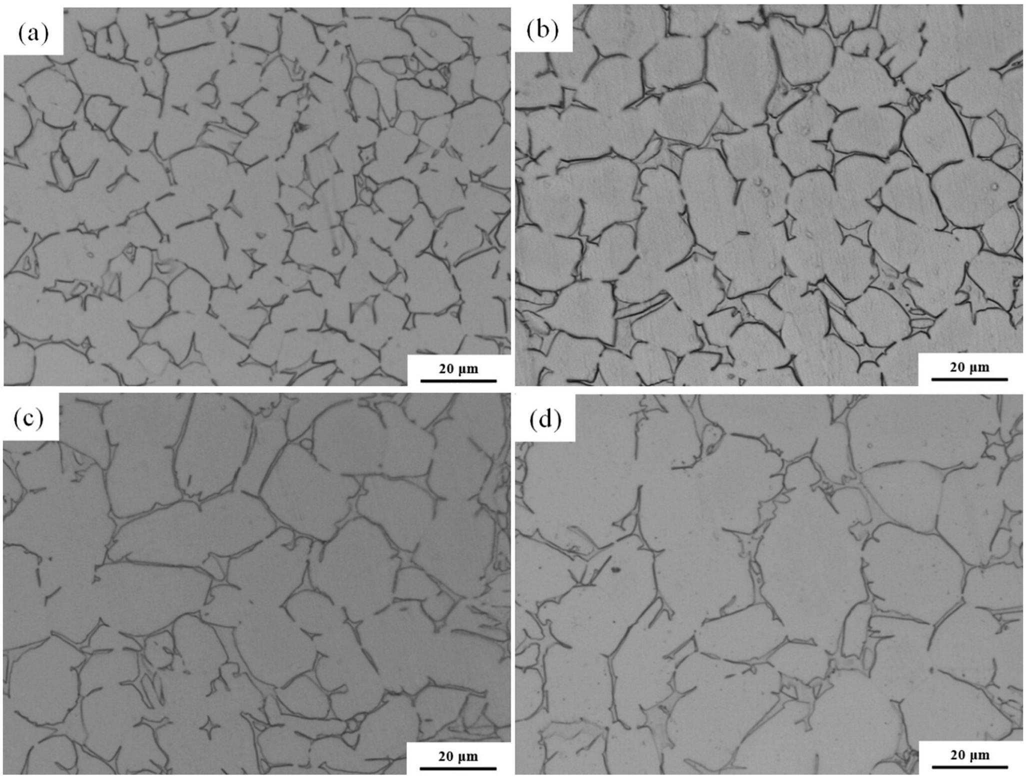

Fig. 1. Optical micrographs of equiaxed Ti–6Al–4V microstructures at different holding times: (a) 930 °C/2 h, FC + 600 °C/2 h, AC; (b) 930 °C/2 h, FC + 600 °C/14 h, AC; (c) 930 °C/36 h, FC + 600 °C/2 h, AC; and (d) 930 °C/2 h, FC + 600 °C/60 h, AC.