Flow softening and dynamic recrystallization behavior of BT9 titanium alloy: A study using process map development

E. Ghasemi a, A. Zarei-Hanzaki a,⁎, E. Farabi a, K. Tesari b, A. Jäger b, M. Rezaee a

⁎ The Complex Laboratory of Hot Deformation and Thermomechanical Processing of High Performance Engineering Materials, School of Metallurgy and Materials Engineering, College of Engineering, University of Tehran, Tehran, Iran

A R T I C L E   I N F O

Article history:
Received 31 July 2016
Received in revised form 26 October 2016
Accepted 31 October 2016
Available online 2 November 2016

Keywords:
Titanium alloys
Hot compression
Processing map
Dynamic recrystallization
Electron backscatter diffraction

A B S T R A C T

In the present study, uniaxial compression tests were employed between 1000 and 1100 °C under the strain rates of 10⁻³, 10⁻² and 10⁻¹ s⁻¹ to investigate the hot deformation behavior of BT9 titanium alloy. Work hardening behavior interpretation and analytical investigations including calculation of deformation activation energy, and developing process maps were used to establish a numerical correlation between microstructural evolution and the flow behavior of the alloy. The results showed that dynamic recrystallization takes place at severe condition (T = 1000 °C and ̇ε = 0.1 s⁻¹), while dynamic recovery is the major microstructural mechanism at other condition. According to dynamic material model and Prasad’s instability criterion, the maximum power dissipation of 52% and 46% occur at 1000 °C/0.1 s⁻¹ and 1150 °C/0.001 s⁻¹, respectively. Electron backscattered diffraction images and high resolution optical images also revealed that continuous dynamic recrystallization is the governing mechanism at these deformation conditions resulting in a significant grain refinement. Considering the calculated deformation activation energies, power efficiency domains and the microstructural observations, 1000 °C/0.1 s⁻¹ was determined as the optimum deformation condition.

© 2016 Published by Elsevier B.V.

1. Introduction

The dual phase titanium alloys possess vital roles in aerospace industries, due to their superior properties such as high specific strength, fracture toughness and ductility. However, there is a significant debate about their ability to be processed into the complex shapes [1]. Moreover, it is believed that the formation of undesired microstructures during solidification in most of Ti-based alloys might result in low room temperature ductility and a large scatter in practical properties [2]. In order to overcome these shortcomings, optimized cycles of hot working and heat treatment has been employed [3].

As is well established the outcomes of applying any hot working cycle on Ti alloys are deeply related to their initial microstructures, the morphology of alpha phase and the processing parameters such as the amount of strain, deformation temperature and strain rate (encapsulated in Zener-Hollomon parameter) [4,5]. The dual phase titanium alloys, in particular BT9 one (which is known as TC11 in Chinese classification system), are conventionally hot worked (e.g., forging) at higher temperatures in two phase region (30–50 °C below its (α+β) to β transformation temperature, i.e. transus temperature). This hot working routine could generate equiaxed microstructure consisting of alpha grains in the transformed β matrix [6,7]. Although the equiaxed microstructure possesses high ductility and thermal stability, the related higher temperature mechanical properties (e.g. thermomechanical fatigue properties and the resistance to crack growth and propagation) are yet insufficient. To overcome these shortcomings, the β-forging process was developed for Ti-alloys. In comparison to the conventional forging process, the parts fabricated by β-forging process comprise of lamellar microstructures (lamellar alpha grains) in the transformed β matrices. The latter microstructure possesses promising properties including good high-temperature creep resistance along with impact and fracture toughness [8]. Moreover, it is worth noting that the formation of any desired microstructure is highly dictated by the operating dynamic restoration mechanisms during β-forging process [9,10]. In this regard, the understanding of the restoration mechanisms during β titanium hot working, which has been a controversial debate over the years [11,12], is highly necessitated.
One of the most important material parameters, which is believed to have a significant role on materials’ dynamic softening behavior during hot deformation, is the material’s stacking fault energy (SFE) [13]. BT9 alloy is a high SFE material due to its high Al content and the related BCC structure of β matrix in single phase region [14]. Accordingly, there is a general perception that β is prone to dynamic recovery (DRV) due to its BCC structure and high SFE value. However, the occurrence of dynamic recrystallization during hot deformation in the β phase region has been reported for TC11 alloy [7,15–17]. Jackson et al. [18] illustrated that the dominant restoration mechanism of Ti-10V-2Fe-3Al changes from DRX (dynamic recrystallization) to DRV by increasing the temperature. CHEN et al. [12] studied the hot deformation behavior of TC11 alloy in β phase field. Their results showed different dynamic recrystallization mechanisms at various deformation temperatures (i.e. geometrical or discontinuous dynamic recrystallization). Likewise, in some other researches the formation of recrystallized grains at β grain boundaries was attributed to the discontinuous DRX, although the growth of recrystallized grain was also reported [5,19–21]. Besides, it is believed that DRX is limited in a narrow processing window inside the β single-phase region [22]. However, only a few studies have paid attention to the effects of Zener-Hollomon parameter (Z) on the microstructural development during hot deformation of titanium alloys.

Considering the hot deformation potential on improving the properties of titanium alloys and the lack of detailed understanding of the involved restoration mechanisms, the present work was planned to precisely track the dynamic softening behavior of BT9 Ti-alloy during hot deformation. This was supported by the proper microstructural examination. Moreover, a dynamic materials model (DMM) was developed to predict the safe and unsafe regions (windows) of BT9 experimental alloy during hot deformation. This was interrelated with the developed microstructures for further confirmation of dominant restoration mechanism. Consequently, a kinetic model for DRX including the relationship between Z parameter and β grain size with the fractional recrystallization was developed and discussed.

2. Experimental method

The experimental material was received in extruded condition as bars with 20 mm diameter, the composition of which listed in Table 1. The β transus temperature of the alloy is 1008 °C [17]. The initial microstructure of the as-received material is shown in Fig. 1. As is observed, the initial material shows a typical dual phase microstructure with a combination of the alpha phase (consisting equiaxed and lamellar morphologies) and the transformed beta phase as the matrix. The hot compression testing technique was utilized to thermomechanically process the experimental material. The corresponding cylindrical specimens were prepared by electro-discharge machining based on the ASTM-E209 standard (12 mm in height and 8 mm in diameter) [23]. The hot compression tests were carried out using a GOTECH AL-7000 universal testing machine coupled with a programmable resistance furnace at different temperatures in the range of 1000–1150 °C (with 50 °C intervals) under the strain rates of 0.001, 0.01 and 0.1 s^-1 up to the true (logarithmic) strain of 0.7 (Fig. 2). The specimens were initially heated up to the deformation temperature and soaked isothermally for 5 min prior to the straining. The specimens were then immediately quenched in water to preserve the microstructure. To reduce the die friction and ensure a uniform deformation, two thin pieces of mica were affixed to the flat dies. In a parallel way, an isothermal soaking treatment (5 min) was carried out at any test temperature (without any straining); this was followed by quenching the specimen to conserve the initial microstructure prior to any hot compression test temperature.

For microstructural examinations, each specimen was sectioned along the longitudinal direction. The standard grinding and polishing technique was employed for microstructural investigations. The optical micrographs were taken near the mid-diameter position by ML-7000 Microscope. Several areas were examined to quantitatively analyze the revealed microstructures (grain size, recrystallization volume fraction, aspect ratio). The newly formed grains were identified by classifying the grain aspect ratio and size. In general, a thermomechanically processed grain structure with an aspect ratio less than 2.5 and smaller grain size in comparison to the non-processed materials can be considered as a new recrystallized grain. Accordingly, the volume fraction of the recrystallized

### Table 1

<table>
<thead>
<tr>
<th>Element</th>
<th>Al</th>
<th>Mo</th>
<th>Zr</th>
<th>Si</th>
<th>Fe</th>
<th>V</th>
<th>Cu</th>
<th>Cr</th>
<th>Nb</th>
<th>Ti</th>
<th>%wt.</th>
</tr>
</thead>
<tbody>
<tr>
<td></td>
<td>6.3</td>
<td>3.5</td>
<td>1.65</td>
<td>0.2</td>
<td>0.03</td>
<td>0.03</td>
<td>0.02</td>
<td>0.008</td>
<td>0.005</td>
<td>Balanced</td>
<td></td>
</tr>
</tbody>
</table>
3. Results and Discussion

3.1. Stress-strain behavior

The typical true stress—true strain curves of BT9 titanium alloy obtained at different temperatures and strain rates are depicted in Fig. 3. As is observed the flow behavior is substantially influenced by changing the temperature and strain rate due to the significant sensitivity of the experimental material to the variation of temperature and strain rate. In this regard it is believed that the substantial increase in the average kinetic energy of atoms and the swift decrease in the critical slip shear stress may facilitate the dislocations movement [25]. Besides, the reduction of dislocation density at higher temperatures through restoration mechanisms can offset the work hardening thereby reducing the flow stress. Moreover, the probable dislocations density drop at lower strain rates diminishes their interactions thereby resulting in a lower impact of work hardening on the level of flow stress. Knowing all above, a more precise observation of flow stress variations may reveal three distinguished behavior (classified in Table 2 and highlighted in Fig. 3). The first category exhibits a peak stress followed by a dynamic softening regime down to the steady state behavior; this is apparent at 1000 °C under all strain rates and under the strain rate of 0.1 s⁻¹ at all testing temperatures (class I). This flow softening phenomenon may well indicate the occurrence of DRX in traditional qualitative point of view. On the other hand, the specimens which were deformed at 1050, 1100 and 1150 °C under the strain rate of 0.001 s⁻¹ show a flow leveling off trend, which can be considered as an indication of DRV; this flow behavior is identified as the second category (class II). It should be noted that, detecting the onset of DRX through calculations on flow stress curves can be a difficult task. However, some of the materials don’t show recognizable peaks [26]. In order to determine the onset of DRX precisely, the plotting of the work hardening rate (\(q\)) against true stress (\(s\)) is helpful. In this regard, the critical stress can be calculated by using the strain hardening rate versus true stress curve. In this study, the method proposed by Jonas [27] is used. First, work hardening rate (\(q\)) against true stress (\(s\)) was plotted by deriving true stress versus true strain. The inflection point is detected by fitting a third order polynomial to the \(q-s\) curves up to the peak point as follows:

\[
\theta = A\sigma^3 + B\sigma^2 + C\sigma + D
\]

where A, B, C, and D are constants for a given set of deformation conditions. The second derivative of this equation with respect to \(\sigma\) can be expressed as:

\[
d^2\theta/d\sigma^2 = 6A\sigma + 2B
\]

At critical stress for initiation of DRX, the second derivative becomes zero. Therefore,

\[
1000^\circ C & 0.001-0.1 & DRX \\
T > 1000 & 0.001 & DRV \\
T > 1000 & 0.01 & DRX & DRV with yield drop
\]

Table 2: The identified classes of various flow behaviors; DRX stands for dynamic recrystallization and DRV is the acronym for dynamic recovery.
Accordingly, the strain hardening behavior of BT9 alloy in class I and class II conditions are studied (Fig. 4). As is seen in Fig. 4a, the appearance of inflections is served as indication of DRX in class I conditions. However, the typical variations of strain rate hardening with stress in the class II conditions (Fig. 4b) are associated with structural dynamic recovery [26]. The beneficial effects of the recrystallization process on the grain refinement and suppression of the various deformation defects in semi-finished and finished products has been frequently reported in previous works [7,13,28-30]. In this regard, the obtained results in the β phase region have been compared with similar researches in the literature in terms of activated restoration processes, as are indicated in Table 3 [31-35].

The last category of flow stress curves, which happens at temperatures higher than 1000 °C under strain rate of 0.01 s⁻¹ (class III), is characterized by a discontinuous (sudden) drop in the flow stress, occurring in the initial stages of hot deformation, followed by a steady-state or flow softening behavior at higher strains. This type of hot deformation behavior is identified as yield drop phenomenon, which has been also found in some near alpha and alpha + beta titanium alloys, including Ti-15-3 [36], Ti-13V-11Cr-3Al [37], Ti-5.6Al-4.8Sn-2.0Zr alloy [38] and Ti-6242 [39]. The corresponding dynamic theory in this matter [40] represents the discontinuous yield phenomenon as simultaneous movement of mobile dislocations generated from grain boundaries. By further straining, the generated dislocations spread inwards and outwards the grain interior. This would eventually lead to higher applied stress to release the pinned dislocations. This results in a sudden stress drop and higher density of mobile dislocations.

Fig. 5 exhibits the dependence of peak stress in the three classes on the deformation temperatures and strain rates. At any temperature, an increase in the strain rate results in higher peak stress. At a given temperature, the peak stress is increased by increasing the strain rate; this indicates that the restoration processes have more time to be proceeding. Furthermore, there is a noticeable decrease in peak stress by temperature increment from 1000 °C to 1050 °C. This may be due to the complete dissolution of α phase (harder phase) in the β matrix (softer phase). This is well in accordance with the previous researches regarding the peak points of Ti-based alloys [41].

3.2. Analysis on kinetic aspects of deformation

The relation between processing parameters, namely deformation temperature and strain rate, on the flow stress can be expressed following kinetic equations [42,43]:

\[
\dot{e} = A \sinh(\alpha \sigma)^n \exp \left( \frac{Q}{RT} \right)
\]  

\[
\sigma = C e^m
\]

Fig. 4. The plots of work hardening rate (θ) against flow stress for experimental BT9 alloy associated with: (a) class I, and (b) class II types of flow behavior.
Table 3
The main macro-plastic deformation mechanisms of various titanium alloys during deformation under different condition.

<table>
<thead>
<tr>
<th>No.</th>
<th>Alloy</th>
<th>Deformation parameters</th>
<th>Macro-deformation mechanisms in the β region</th>
</tr>
</thead>
<tbody>
<tr>
<td>1</td>
<td>Zhu et al. [31]</td>
<td>ε &lt; 1 s⁻¹</td>
<td>DRX</td>
</tr>
<tr>
<td>2</td>
<td>TC21</td>
<td>ε &gt; 1 s⁻¹</td>
<td>DRV</td>
</tr>
<tr>
<td>3</td>
<td>Balasubrahmanyan et al. [32]</td>
<td>ε = 0.001–0.1 s⁻¹</td>
<td>DRX</td>
</tr>
<tr>
<td>4</td>
<td>Ding et al. [33]</td>
<td>ε = 1</td>
<td>30% DRX</td>
</tr>
<tr>
<td>5</td>
<td>Han et al. [34]</td>
<td>ε &lt; 1 s⁻¹</td>
<td>42% DRX</td>
</tr>
<tr>
<td>5</td>
<td>Seshacharyulu et al. [35]</td>
<td>ε &gt; 1 s⁻¹</td>
<td>48% DRX</td>
</tr>
<tr>
<td>6</td>
<td>Eli grade Ti-6Al-4V</td>
<td>ε &lt; 0.1 s⁻¹</td>
<td>Instable deformation</td>
</tr>
<tr>
<td>6</td>
<td>Present study</td>
<td>High Z</td>
<td>Large grained super-plasticity in β phase</td>
</tr>
<tr>
<td></td>
<td></td>
<td></td>
<td>DRX (80%)</td>
</tr>
</tbody>
</table>

lnε = \(-\frac{Q(ε, T)}{RT}\) + \(n(ε, T)\ln(\sinh(α(ε, T)σ))\) + \(\ln A(ε, T)\)  \hspace{1cm} (10)

\[Q(ε, T) = Rn(ε, T)\left[\frac{\ln(\sinh(α(ε, T)σ))}{α(ε, T)}\right]_τ\]  \hspace{1cm} (11)

In this regard, the variation of Q value with temperature and strain rate is plotted through a 3D activation energy map for the logarithmic strains of 0.1, 0.3, 0.5 and 0.7 in Fig. 6. At a glance, the Q (ε, T) varies between 152 and 266 kJ mol⁻¹ over the deformation conditions. As is observed, the uppermost Q value is associated with the temperature/strain rate of 1000 °C/0.1 s⁻¹, where the material is deformed up to the strain of 0.1; the minimum Q value is obtained at 1150 °C/0.001 s⁻¹ and the logarithmic strain of 0.7. Moreover, under the strain rate of 0.1 s⁻¹, in comparison to the intermediate strain rates the Q-value is relatively higher. This incremental trend of Q with increasing the strain rate can be related to the rapid changes in the state of dislocations configuration (i.e., higher rate of dislocation generation, denser dislocation tangles, denser deformation cell walls, finer substructures, and so on). These may well end to the stronger strain hardening effects thereby increasing the energy needed for additional deformation.
3.2.1. Variation of Q-value for class I flow curves

Fig. 6 shows that the extent of Q-value for class I flow behavior varies in the range of 200–266 KJ mol$^{-1}$ which is somewhat higher than that of self-diffusion barrier energy for β titanium alloy (~153 kJ mol$^{-1}$) [22]. According to the reported Q-values for hot deformation of dual phase titanium alloys in single phase region, the occurrence of DRX in the experimental material appears to be achievable in these conditions. Gao et al. [49] calculated the Q-value to be 281 KJ mol$^{-1}$ in conditions similar to the present class I behavior. Moreover, Zhang et al. [50] showed that in β phase regions the apparent activation energy at lower temperatures (similar to class I conditions) is about 218.26 KJ mol$^{-1}$. However, in single-phase metals, it has been frequently observed that the activation energy was appreciably higher where the dynamic recrystallization was the rate controlling mechanism as opposed to the dynamic recovery due to the dislocation climb and cross-slip processes [51,52].

3.2.2. Variation of Q-value for class II flow curves

At higher temperatures, due to the thermally activated nature of dislocation movement the energy barrier for the deformation can be significantly reduced. Therefore, the activation energy for further deformation would follow a decreasing trend, (Fig. 6). The calculations of the deformation activation energy at higher temperatures ends to the values lower than 170 KJ mol$^{-1}$. This is close to what is called the self-diffusion activation energy of BCC phase in titanium alloys (~153 kJ mol$^{-1}$) and provides the fact that a diffusion controlled mechanism is about to happen during deformation. Moreover, the reported Q-values for titanium-based alloys is in agreement with the present proposed value and being regarded as the activation energy for the dynamic recovery (DRV) in the literature [22]. Therefore, by referring to the flow curves and the calculated Q-values, it is postulated that the dynamic recovery is the governing restoration mechanism at higher temperatures.

3.2.3. Variation of Q-value with true strain

It is believed that under hot tensile and compression deformation can activate a large number of slip systems, resulting in an increased value of the activation energy before reaching to the peak stress value [53,54]. As is seen in Fig. 6, the activation energy is decreased by increasing the strain beyond the peak strain at all deformation temperatures, but reaches to a constant value at logarithmic strains higher than 0.5. As was aforementioned, this sudden change in Q-value variation behavior can be justified through the occurrence of energy consuming mechanisms such as DRV and DRX. This can provide the much needed work softening for further processing the material. In other words, DRX and DRV possess a positive contribution on the deformation by consuming the stored energy and making the workpiece easier to deform to the higher strains [55]. Therefore, it is easy to see the decrease in the activation energy above the peak strain [56].

3.3. Microstructural evolution during hot deformation

Beside the flow stress behavior, the microstructural investigation on the deformed specimens should be conducted for further clarifying the dominant restoration mechanisms. For titanium alloys, the microstructural evolution during hot deformation is strongly affected by the processing parameters (temperature and strain rate). The effects of strain rate and deformation temperature are simultaneously expressed by the Zener–Hollomon parameter (Z) (Eq. (12)).

\[ Z = \dot{\varepsilon} \exp\left(\frac{Q}{RT}\right) \]  

(12)

Similar to the previous sections the Zener–Hollomon parameter map depicted in Fig. 7 has been divided into the flow stress sub-classes and the variations of Z value with strain rate and temperatures were interrelated with the developed microstructures.

3.3.1. Evolution of microstructure in the class I flow behavior

Fig. 8 shows the evolution of microstructure during hot compression of BT9 experimental alloy under class I condition. As is clearly seen the initial coarse grain material (Fig. 2) has been replaced by newly recrystallized grains. However, the DRX process tends to occur in the materials with low SFE [57] while the β phase in BT9 alloy possesses high SFE values [14]. The DRX process in the materials with high SFE depends strongly on the developed cell structure through DRV process and would occur at larger strains specially by various severe plastic deformation (SPD) methods [30]. For the beta phase in experimental material, which is considered as a high SFE material, due to vigorous dislocations climb and cross-slip, which are highly necessitated for dislocation annihilation and rearrangement during DRV, the accumulated strain energy would not be sufficient to commence the DRX process. Moreover, the latter is supported by thermally activated processes, such as diffusion and substructure formation, therefore tends to take place at lower strain rates and higher temperatures. In fact the DRX nucleus would form only where the dislocation generation and accommodation becomes extensive enough in the microstructure [13]. In order to justify the critical condition for the onset of DRX in the mentioned temperature and strain rate, Varshni model has been employed as is follows [58]:

\[ \rho_c = \left( \frac{20 \gamma_c \varepsilon}{3b^2 \Omega \mu^2 M_{BM}} \right)^3 \]  

(13)

where, \( \rho_c \) referring to critical dislocation density for onset of DRX, \( \gamma_c \) denotes the boundary energy and \( M_{BM} \) represents the boundary mobility, which could be calculated by Ref. [59]:

\[ M_{BM} = \frac{b \delta D_{GB} V_{in}}{b^2 RT} \]  

(14)

Where;

\[ D_{GB} = D_0 \exp\left( -\frac{Q_{diffu}}{RT} \right) \]  

(15)

Fig. 7. The variation of strain rate logarithm with temperature in hot compression tests of Ti–6.5Al–3.5Mo–1.5Zr–0.3Si experimental alloy. (The superimposed numbers on the contour lines are the lnZ-values; Z is Zener–Hollomon parameter).
The relation between strain rate, temperature and the dislocation density would suggest that in the case of titanium alloys as a high SFE material, lower temperatures and higher strain rates (high Z values indicated in Fig. 7) can facilitate the critical dislocation density for the occurrence of partial dynamic recrystallization. Also due to high stacking fault energy of $\beta$ phase and formation of sub-boundaries inside the pre-existing $\beta$ grains [60] (which are marked by arrows in Fig. 8), the mechanism of DRX is postulated to be continuous dynamic recrystallization (CDRX). It is worth mentioning that the confirmation of latter mechanism demands more advance characterization. Moreover, as is shown in Fig. 8 the degree of recrystallization is decreased by the increment of temperature. This illustrates the competitive behavior between DRV and DRX. In this regard, it is comprehended that increment of temperature may debilitate the critical dislocation density for DRX and might enhance the DRV condition [61].

### 3.3.2. Evolution of microstructure in the class II flow behavior

The evolution of microstructure under class II condition is illustrated in Fig. 9. As is seen, increasing the temperature and decreasing the strain rate (low Z values) would result in less recrystallized grains volume fraction, which is a clear identification of dynamic recovery becoming the dominant restoration process in this hot working condition. In addition, due to the longer time under lower strain rate, the grain growth in limited recrystallized grains has led to a sharp increase in the grain size. Original $\beta$ grains are elongated perpendicular to the compression axis with serrated grain boundaries (Fig. 9). The latter is related to the preferred dynamic recovery at the prior grain boundaries. As is believed by many researchers [62–64] these $\beta$ grains contain many low angle boundaries. Moreover, in these conditions, the equiaxed $\beta$ grains are bounded by high-angle boundaries including subgrains inside. This indicates that dynamic restoration occurs mainly by DRV at this deformation condition.

### 3.3.3. Quantitative analysis on developed microstructures

Fig. 10 shows the effects of strain rate and deformation temperature on the fraction of recrystallization. As is seen, the fraction of recrystallization is decreased by increasing the temperature and decreasing the strain rate. Interestingly, the maximum value of recrystallization volume fraction (which is about 80%) is associated with the highest strain rate and lowest temperature, i.e. 1000°C/0.1 s$^{-1}$. On the other hand, the corresponding minimum value (14%) is observed at highest temperature and lowest strain rate (1150°C/0.01 s$^{-1}$).
A quantitative analysis on the relation between microstructural features (average size of newly formed grains and recrystallization fraction) and processing parameter (Z-value) is depicted in Fig. 11. As is observed, the size logarithm of freshly formed grains is inversely proportional to the Zener–Hollomon parameter (Z) and the logarithmic fractional recrystallization value shows a linear relationship with Z; this can be approximated on the grounds of experimental data as:

\[
\begin{align*}
    d_{\text{DRX}} &= kZ^{-N} \\
    X_{\text{DRX}} &= k'Z^N
\end{align*}
\]

(16)

where K and N are material constants, and can be extracted by taking logarithm from both side of Eq. (16). Therefore, with respect to Fig. 11 the grain size of recrystallized β phase and the fractional recrystallization can be expressed as:

\[
\begin{align*}
    d_{\text{DRX}} &= 1106Z^{-0.235} \\
    X_{\text{DRX}} &= 0.158Z^{0.343}
\end{align*}
\]

(17)
The average absolute error is only 5.23% and 6.52% for the predictions. These indicate that an accurate estimation of grain size and fractional recrystallization in Ti–6.5Al–3.5Mo–1.5Zr–0.3Si alloy during the hot deformation in single phase region can be extracted by Eq. (17). According to Fig. 11a, the average grain size at higher Z values (higher strain rates and lower temperatures) is less than that of lower ones.

3.4. Processing map for BT9 experimental alloy

For many years, the processing map based on dynamic materials model (DMM) has been considered as an accurate mean for interrelating the microstructure with the processing parameters [65–67]. It is a useful approach to predict the flow instability, adjust the processing parameters and regulate the microstructure during the plastic flow of metals and alloys. This model considers the work piece as a power dissipater which at a given $\varepsilon$: consists of two harmonizing parts: the G content and the J co-content, which are related to temperature rise and microstructure dissipation, respectively. Therefore, total power $P$ can be expressed as follows:

$$P = \sigma \dot{\varepsilon} = G + J = \int_0^\varepsilon \sigma d\dot{\varepsilon} + \int_0^\sigma \dot{\varepsilon} d\sigma$$

(18)

The following equation divides the power between G and J:

$$\frac{dJ}{dG} = \frac{\dot{\varepsilon} d\sigma}{d\sigma d\varepsilon} = \frac{d(\ln \sigma)}{d(\ln \dot{\varepsilon})} = m$$

(19)

For an ideal linear dissipater ($m = 1$), J reaches to its maximum, which is $J_{max} = P/2$. This results in a dimensionless parameter called the efficiency of power dissipation ($\eta$) of a non-linear dissipater and expressed as [68]:

$$\eta(T, \dot{\varepsilon}) = \frac{J}{J_{max}} = \frac{2m}{m + 1} |\dot{\varepsilon}|$$

(20)

The disparity of $\eta$ with temperature and strain rate establishes a power dissipation map exhibiting different domains, which directly corresponds to the specific microstructural mechanism. Usually, high efficiency domains are associated with definite microstructural mechanisms that have dissipated the power in the most proficient way during deformation. The stability of deformation is also described by DMM using an inequality proposed by Naghdi and Ziegler [69] in which the stability is a function of metallurgical changes. Therefore, instability criteria can be defined as a dimensionless parameter, $\xi(\dot{\varepsilon})$, at a constant temperature given by:

$$\xi(\dot{\varepsilon}) = \frac{d[\ln|m/m + 1|]}{d[\ln \dot{\varepsilon}]} + m > 0$$

(21)

Accordingly, the instability refers to negative $\xi(\dot{\varepsilon})$ values at a given deformation condition, where metallurgical evolutions may lead to instable plastic deformation (flow localization [21], shear banding [70], dynamic strain aging [71], mechanical twining and kinking or flow rotations [72,73]). The variation of $\xi(\dot{\varepsilon})$ is usually shown as contour lines in a temperature and strain rate space known as instability map. By superimposing the 2D contour map of efficiency and instability, the processing map can be obtained at a certain strain. Interestingly, the calculated instability criterion for the experimental material shows no negative values, expressing the satisfactory workability of material at the given conditions. Typical processing map of Ti-6.5Al-3.5Mo-1.5Zr-0.3Si for a true strain of 0.7, constructed at temperature ranging from 1000 to 1150 $^\circ$C and strain rate ranging from 0.001 to 0.1 s$^{-1}$, is shown in Fig. 12.

The contour numbers indicate the efficiency of power dissipation ($\eta$). Moreover, the optimum power dissipation efficiency represents the condition where the materials are proper for hot working since...
high extent of microstructural evolution could occur inside the workpiece. According to Prasad and Seshacharyulu [9], the optimum power dissipation efficiency of metallic materials is dependent on stacking fault energy (SFE); the previous studies indicate that low SFE alloys possess lower power dissipation efficiency. This is attributed to the close relation between SFE of the materials and their dynamic restoration behavior. According to the recent researches, the optimum power dissipation efficiency for the Ti alloys is reported to be in the range of 35–50% [9]. As is seen in Fig. 12, there are two domains holding optimum dissipation efficiency (with efficiency peak of 52% and 46% at 1000 °C and 1150 °C under strain rate of 0.1 and 0.001 s⁻¹, respectively, marked D₁ and D₂) for the present experimental conditions. For the materials falling into these stable regions, the related microstructural evolution is likely to be governed by DRX, superplastic deformation or DRV [9–11].

The occurrence of DRX during hot deformation is associated with higher driving energy for the formation of subgrains and migration of grain boundaries. Therefore, the power dissipation efficiency would be accordingly higher for microstructural evolution during DRX process. Moreover, it is clearly observed that the regions with high dissipation efficiency (highlighted by the occurrence of DRX) are positioned at moderate-temperature and high-strain-rate processes.

In order to understand the accurate microscopic deformation mechanisms underlying the stable regions for BT9 experimental alloy, verifying the reliability of process parameters predicted by the processing map and clarifying the characteristics of microstructural evolution, the microstructures of experimental alloy deformed under the specific deformation conditions are characterized by EBSD images. Fig. 13 shows the orientation and phase maps of deformed and water quenched BT9 specimens. The martensite (α’) phase possesses a typical needle shape structure (Fig. 13a and c). The preferred crystal orientations of α’ has given rise to some distinctive prior β grains. Fig. 13b and d demonstrate the present phases in the microstructures corresponding to Fig. 13a and c, respectively. As is observed, almost all the β phase has been transformed to martensite structure (α’) after deformation and water quenching. However, for materials undergoing allotropic transformations such as many α+β titanium alloys, the high temperature phase is not stable at room temperature and cannot be retained after cooling. The prior β grain boundaries are the main sites for the nucleation of martensite laths. As Fig. 13 shows, the martensite laths formed at the vicinity of the prior β grain boundaries have different orientations. However, at some points due to special orientation of β planes [63] (indicated by the black box in Fig. 13a) the nucleated α laths would have close orientations. Moreover, different martensitic laths were identified through EBSD mapping. First, are laths formed at the early stage of the phase [74]. These are nucleated at the prior β boundaries and stretched across the β grain. And second, are secondary laths nucleated due to the

![Fig. 13.](a), (c) Orientation maps of experimental alloy deformed at 1000 °C/0.1 s⁻¹ and 1150 °C/0.001 s⁻¹, respectively, the martensitic (α’) structure after deformation and water quenching is seen; and (b), (d) represent the phase maps corresponding to (a) and (c), respectively.)
and microstructural evolution has been conducted. The stress-titanium alloy using work hardening behavior, processing maps

4. Conclusion

A detailed study on the hot deformation behavior of a novel BT9 titanium alloy using work hardening behavior, processing maps and microstructural evolution has been conducted. The stress-strain curves showed two distinct restoration behavior, one a peak followed by a gradual decrease to a steady state stress (DRX) and the other a gradual increase to a peak followed by a steady state stress (DRV). A further analysis using the activation energy and work hardening behavior, the material revealed that an increase in strain rate and a decrease in the deformation temperature would highly facilitate the dynamic recrystallization process. In addition, the electron backscattered diffraction (EBSD) analysis and high resolution optical images reveal that the continuous dynamic recrystallization (CDRX) plays a dominant role in the DRX region. The other deformation behavior was identified as dynamic recovery, which was dominant at other testing conditions. Moreover, the significant decrease in the recrystallization fraction at high temperatures and low strain rates (low 2 values) was contributed to the competitive nature of the two main restoration mechanisms and the increased share of the dynamic recovery during the hot deformation. Moreover, the plotted processing maps didn’t exhibit any instability domains. Two peaks in power dissipation of 52% and 46%, occurring at 1000 °C/0.1 s⁻¹ and 1150 °C/0.001 s⁻¹ was contributed to occurrence of DRX and DRV, respectively. This further provided facts for identification of the optimum processing condition which was at 1000 °C and strain rate of 0.1 s⁻¹.

Acknowledgment

Financial support offered by GACR GBP108/12/G043 is appreciated.

References


[37] G. Liu, S. Zhang, L. Chen, Hot deformation behavior of Ti-6.5 Al-3.5 Mo-1.5 Zr-0.3 Si alloy with acicular microstructure, J. Cent. South Univ. Technol. 18 (2011) 1163–1172.


