An investigation into the warm deformation behavior of Ti–6Al–1.5Cr–2.5Mo–0.5Fe–0.3Si alloy

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A B S T R A C T

The microstructural evolution and the flow behavior of Ti–6Al–1.5Cr–2.5Mo–0.5Fe–0.3Si alloy were investigated in this research. The flow behavior of the alloy at temperatures in the range of 100–600 °C was studied using warm compression testing under the strain rate of 0.001–0.1 s⁻¹. The results indicate that the formability of the alloy is significantly increased at temperatures higher than 400 °C due to the activation of pyramidal slip systems. Moreover, optical observations confirm the occurrence of flow localization and adiabatic shear banding within the microstructure due to the adiabatic heating phenomenon at lower temperatures. The adiabatic heating is also led the experimental material to exhibit a negative strain rate sensitivity behavior. The tensile deformation behavior of this alloy was also studied at temperatures in the range of 100–400 °C through warm tension testing. The results show that the alpha phase grains are elongated along the tensile direction at lower temperatures but would be globularized at higher temperatures.

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1. Introduction

It is well-known that titanium alloys hold a wide range of applications in various industries. Among their different types, the α+β (two phase) alloys are extensively used in aerospace, compressor disks and blades of gas turbines owing to their high toughness and strength as well as good resistance to corrosion at high service temperatures [1]. To this end, their corresponding plasticity behavior at such service conditions has always been an important issue for materials engineers. Therefore, many researches have been conducted to study the deformation behavior of titanium alloys at high temperatures up to date [2–4].

Various mechanisms are activated during deformation of titanium alloys at high temperatures. Among them, phase transformation from alpha+beta microstructure to single beta phase [5], adiabatic shear banding [6–8], activation of new slip or twinning systems [9], fragmentation and globularization of alpha phase [10,11] and restoration mechanisms such as dynamic recovery (DRV) and/or dynamic recrystallization (DRX) are the most common ones [12,13]. The occurrence of dynamic recrystallization during hot deformation in the β phase region has been reported for α+β Ti–6Al–4V [14], α+β Ti-6246 [15], near-α IMI834 [16,17], α+β Ti–6Al–7Nb [18] and near-α Ti–1100 [19]. Jackson et al. [20] illustrated that the dominant restoration mechanism of Ti–10V–2Fe–3Al changes from DRX to DRV by increasing the temperature. During the hot working in the α+β region however, the globularization of α lamellae has been detected in several investigations [10,11,21,22]. Seshacharyulu et al. [23] studied the hot deformation behavior of Ti–6Al–4V in both the dual phase (α+β) and single phase (β) regions. Their results showed that at temperatures between 850 and 950 °C under strain rates ranging from 0.001 to 0.1 s⁻¹, globularization of the lamellar structure was occurred. At temperatures lower than 850 °C under strain rates below 0.1 s⁻¹, cracking at the prior β grain boundaries might take place under mixed mode conditions. For strain rates higher than 1 s⁻¹ and temperatures lower than 950 °C, the material exhibits a wide range of flow instabilities. Recent investigations have reported the occurrence of DRX during hot working of titanium alloys within the dual phase region too [24–26].

Ti-6Al–1.5Cr–2.5Mo–0.5Fe–0.3Si is one of the most common alpha-beta alloys in gas turbine blade industries (known as BT3-1). However, little attention has been paid regarding the medium-to-high temperature deformation behavior of this alloy. Specially, most of the performed studies aimed to investigate the deformation behavior of BT3-1 at temperatures higher than 800 °C where the primary processing of the alloy is conducted [27,28]. Unfortunately, the BT3-1 behavior at temperatures below 600 °C, which is considered as servicing temperatures, has not been

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discussed so far. It should be noted that the behavior of titanium alloys at temperatures higher than 800 °C totally differs from the medium (warm) temperatures regime (below 600 °C) in many aspects; for example alpha beta phase transformation would no more occur and deformation conditions are more severe. Therefore, the previous studies, which concentrate on the hot deformation conditions, are less helpful for the warm temperature regime.

In the present work, the microstructural evolution and the flow behavior of BT3-1 alloy at warm deformation region are investigated. For this purpose, the stress–strain curves, microstructural observations and fracture surface investigations obtained from tension and compression tests are employed to investigate the deformation behavior of the experimental alloy at servicing conditions (i.e. at temperatures in the range of 100–600 °C under quasi static strain rates).

2. Experimental method

The material used in this research is a BT3-1 titanium alloy with the chemical composition listed in Table 1. The β transus temperature of the alloy is 965–970 °C [29]. The initial microstructure of the as-hot rolled material is shown in Fig. 1. As is seen, the initial alloy possesses a typical dual phase microstructure, which consists of a combination of primary equiaxed alpha grains, lamellar alpha phase and matrix transformed beta phase.

In order to study the warm deformation behavior of BT3-1 alloy, the warm compression and warm tension testing methods were utilized. The related specimens were machined from the as-received material according to ASTM E-209 and ASTM E-21 standards (Fig. 2a and b). The warm compression tests were carried out at temperatures in the range of 100–600 °C under the strain rates of 0.001, 0.01 and 0.1 s⁻¹. The warm tension tests were performed at temperatures in the range of 100–400 °C under the initial strain rate of 0.001 s⁻¹. All specimens were homogenized for 5 min at testing temperature before deformation. Right after straining, the specimens were quenched in water to preserve the hot-deformed structures for further microstructural investigations.

In order to study the microstructural evolution, the specimens were sectioned along the longitudinal direction. The cut surfaces were prepared for optical microscopy examination using standard grinding and polishing techniques and were etched using Kroll’s reagent (5% HF; 15% HNO₃; 80% distilled H₂O). The optical observations were accomplished using ML-7000 Microscope. CLEMEX image analysis software as a quantitative metallurgy tool was employed to analyze the microstructural features such as phase fraction, morphology of phases and fraction of globularized area. The fracture surfaces of tensile specimens were investigated using Environmental-Scanning Electron Microscope (E-SEM), Model Cam Scan MV 2300.

3. Results and discussion

3.1. Hot compression flow behavior

The variations of compressive true stress – true (logarithmic) strain of BT3-1 alloy at different temperatures ranging from 100 to 600 °C are depicted in Fig. 3a–c. It is observed that the curves exhibit the strain hardening behavior at examined temperatures of 100–400 °C (Region I) for all strain rates. In this temperature regime, the failure of specimens has started earlier at logarithmic strain of about ~0.35 and led to the complete fracture at logarithmic strain of ~0.4.

At higher temperatures (between 500 and 600 °C), the flow behavior is completely changed, i.e. the stress increases up to a peak followed by a softening (Region II). Moreover, these specimens were easily deformed up to logarithmic strains of 0.7 without failure. This variation in the flow behavior of the alloy can be due to the activation of a new deformation mechanism, which will be discussed in detail using microstructural images in the subsequent sections.

3.2. Hot compression microstructures

The optical microstructures of the experimental alloy, obtained from the hot compression tests at temperatures of 200 °C and 600 °C are depicted in Fig. 4. Present phases are specified and marked with colored arrows in Fig. 4 (red arrow: primary alpha phase, yellow arrow: lamellar alpha phase and green arrow: β transformed). The microstructures of 200 °C and 600 °C were selected as a representative for Region I (100–400 °C) and Region II (500–600 °C) behaviors, respectively.

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Table 1
Chemical composition of BT3-1 experimental alloy.

<table>
<thead>
<tr>
<th>Element</th>
<th>Al</th>
<th>Mo</th>
<th>Cr</th>
<th>Fe</th>
<th>Si</th>
<th>Sn</th>
<th>Zr</th>
<th>Nb</th>
<th>V</th>
<th>C</th>
<th>Ti</th>
</tr>
</thead>
<tbody>
<tr>
<td>%wt</td>
<td>6.3</td>
<td>2.8</td>
<td>1.3</td>
<td>0.4</td>
<td>0.16</td>
<td>0.01</td>
<td>&lt;0.005</td>
<td>0.002</td>
<td>0.008</td>
<td>&lt;0.005</td>
<td>Balanced</td>
</tr>
</tbody>
</table>
Fig. 3. True stress–true strain curves of the experimental BT3-1 alloy at different temperatures under the strain rates of (a) 0.001, (b) 0.01, and (c) 0.1 s⁻¹.

Fig. 4. Microstructure of the warm compressed BT3-1 alloy at (a) 200 °C, 0.001 s⁻¹, (b) 200 °C, 0.1 s⁻¹, (c) 600 °C, 0.001 s⁻¹, and (d) 600 °C, 0.1 s⁻¹. (For interpretation of the references to color in this figure legend, the reader is referred to the web version of this article.)
The occurrence of ASB arises from the adiabatic heating phenomenon, which is common in low thermal conductivity alloys such as Titanium. As a result of adiabatic heating, the temperature at a local part (highly strained areas) of the specimen would increase because of the deformation heat converted from high amount of plastic work and a low thermal conductivity of the alloy [30]. If the decrease of strength due to heating is more than that of the combined strain and strain rate hardening, the material exhibits flow instability [31], such as adiabatic shear band and flow localization. It is reported that the occurrence of ASBs can result to the failure of the materials since they may result to cavitation and crack initiation [32]. Fig. 5 illustrates the fractured specimen after being tested at 200 °C under strain rate of 0.1 s⁻¹. It is seen that the orientation of crack toward the compression axis is about 30°, which is very similar to the orientation of ASB toward compression axis in Fig. 4b.

The adiabatic heating also plays an important role in the flow behavior of the experimental alloy. To investigate this role, the temperature rise due to adiabatic heating was calculated through \( \Delta T = \frac{0.95 \times p \int_0^\varepsilon \sigma \, d\varepsilon}{\rho C_p} \) formula. In this formula, \( \rho \) is density of the titanium alloy in g/cm³, \( C_p \) is the specific heat, and \( \varepsilon \) is the strain. The \( \eta \) constant is dependent on strain rate value and is equal to 0, 0.25 and 0.5 for the strain rate of 0.001, 0.01 and 0.1 s⁻¹, respectively [30]. The temperature rise within the specimen due to the adiabatic heating phenomenon is presented in Table 2. As it can be seen, there is a greater adiabatic heating effect at relatively higher strain rates, which is attributed to the higher heat generation rate, and limited heat dissipation capacity at high deformation rates. In fact, at higher strain rates, the actual temperature at which the alloy is being tested would be higher and work hardening rate of the specimen would be lower, accordingly. The latter leads to negative values of strain rate sensitivity factor (m) as is seen in Table 2.

As is seen in Fig. 4c and d, some new microstructural features are emerged at 600 °C. For instance, the grain boundary serration in alpha phase is appeared which are marked by arrows in Fig. 4c and d. The serration of grain boundaries along with the appearance of peak stress in flow curves can be an indication for dynamic recrystallization in titanium alloys which were reported in several papers previously [24,33,34]. Yoo et al. [9] expressed that in titanium as a HCP metal, only \(<a>\) basal slip system is activated at lower temperatures. However, pyramidal \(<a+c>\) slip systems would be activated at higher temperatures. The activation of pyramidal slip systems increases the deformation inhomogeneity, which may lead to the occurrence of DRX and the appearance of serration along grain boundaries. The significance of a pyramidal
slip system is more emphasized by the fact that it can provide five independent deformation mode. At the temperature range of 100–400 °C, the alloy lacks from the sufficient formability due to the limited slip systems. While at 500 and 600 °C, the addition of pyramidal slip fulfills the Von Mises criterion for a general homogeneous plastic deformation and improves the formability of the material.

A possible source mechanism for non-basal \((\text{c+a})\) slip dislocations is proposed based on the formation of an attractive junction between glissile \((\text{a})\) and sessile \((\text{c})\) dislocations from the prism plane into a pyramidal plane. The driving force for the junction formation, which comes from the long-range elastic interaction between \((\text{c})\) and \((\text{a})\) dislocations, is relatively large in most HCP metals. The cross-slip process is energetically unfavorable in Ti from a viewpoint of the change in anisotropic elastic line tension. However, it becomes favorable at elevated temperatures above 300 °C for pure Ti and higher temperatures for Aluminum-containing-titanium alloys [9]. This difference is attributed to the stacking fault lowering effect of Aluminum atoms.

3.3. Hot tension tests

The warm tensile true stress–true strain curves of BT3-1 titanium alloy in the temperature range of 100–400 °C under the strain rate of 0.001 s\(^{-1}\) are depicted in Fig. 6. As is observed, the curves exhibit the normal strain hardening behavior up to the necking point with lower stress levels at higher temperatures. The variations of the both strength and total elongation values as a function of testing temperature under the strain rate of 0.001 s\(^{-1}\) are elucidated in Fig. 7. As is expected, with increasing temperature, the strength levels (Yield stress and ultimate strength) reduce obviously. But variation of total elongation values are negligible. The limited elongation to fracture of the experimented BT3-1 titanium alloy may be reasoned considering the insufficient slip systems operating in this low temperature range.

The fracture surface observations were consistent with the aforementioned formability trend. The fracture morphology of the tensile specimens, under strain rate of 0.001 s\(^{-1}\) are depicted in Fig. 8. Non-uniform distribution of shallow dimples can be detected in this figure. The formation of dimples with shallow depth and non-uniform distribution would be an indication of low ductility fracture in non-brittle materials. Fracture surface morphologies at different temperatures are similar, relatively. In fact, the size, distribution and depth of dimples are not affected by the temperature, since no significant change was occurred in deformation behavior of the alloy.

The optical microstructures of hot tensioned specimens at temperatures of 100–400 °C under the strain rate of 0.001 s\(^{-1}\) are depicted in Fig. 9. Present phases are identified and marked with colored arrows in Fig. 9 (red arrow: primary alpha phase, yellow arrow: lamellar alpha phase and green arrow: \(\beta\) transformed). As is seen, alpha grains are elongated along the applied tensile strains. Moreover, the size and the interspace of alpha layers are increased at higher temperatures. Meanwhile they are gradually transferring toward alpha islands at 400 °C.

Using CLEMEX Software, the average aspect ratio of alpha grains was calculated (Table 3). As is observed, the average aspect ratio of alpha grains decreases from 3.6 to 2.1 by raising the temperature. Globalarization process in dual phase titanium alloys is known to be responsible for the observed microstructural change and has been mentioned for various titanium alloys in previous papers. The globularization consists of two steps, as are
follows. The $\alpha/\alpha$ boundaries are firstly formed across the $\alpha$ lamellae either through recrystallization or shearing of the lamellae. Following to this step, $\beta$ phase penetrates in these boundaries and globularization takes place [35]. The details of this process are explained by Seshacharyulu et al. and are depicted in Fig. 10 schematically [23]. For the lower temperature regime, deformation is highly nonuniform, resulting in regions of intense localized shear in which the lamellar alpha colonies break up and begin to spheroidize after a tensile test. Since, globularization of $\alpha$ lamellae is a thermally activated process, increasing the temperature has facilitated the diffusional processes.

4. Conclusion

In this study, the microstructural evolution of BT3-1 titanium alloy during warm working in the temperature range of 100–600 °C was investigated. The following conclusions are drawn from the results of this investigation:

1. The warm compression tests indicated that the material would exhibit poor workability at temperatures lower than 400 °C due to limited slip systems. However, the activation of pyramidal systems at higher temperatures improved the formability of the experimental materials significantly.
2. The adiabatic heating phenomenon was identified to be responsible for negative values of strain rate sensitivity factor and the appearance of adiabatic shear bands within the microstructures.
3. Microstructural observations using warm tension tests confirmed the occurrence of dynamic globularization of alpha grains at temperatures in the range of 100–400 °C; this would be intensified at higher temperatures.

<table>
<thead>
<tr>
<th>Table 3</th>
<th>Microstructural features from warm tensioned samples.</th>
</tr>
</thead>
<tbody>
<tr>
<td>T (°C)</td>
<td>100 °C</td>
</tr>
<tr>
<td>Average aspect ratio of alpha grains</td>
<td>3.6</td>
</tr>
<tr>
<td>Fraction of globularization</td>
<td>20.1</td>
</tr>
</tbody>
</table>

Fig. 9. Under fracture surface micrographs of the warm tension specimens stretched at (a) 100 °C, (b) 200 °C, (c) 300 °C, and (d) 400 °C under the strain rate of 0.001 s⁻¹. (For interpretation of the references to color in this figure legend, the reader is referred to the web version of this article.)

Fig. 10. Schematic representation of globularization process.
References


