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Yield point phenomena in TIMETAL 125 beta Ti alloy

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ABSTRACT

The yield point phenomena in TIMETAL 125 beta-Ti alloy was studied using hot compression tests in the temperature range of 680–880 °C and at strain rates of 0.001–1 s^{-1}. The yield drop increased with decreasing temperature or increasing strain rate. Indeed, the yield point was sharper at high strain rates and low temperatures. The Mo equivalent was taken as a compositional index to determine the influence of chemical composition on the yield point phenomenon. There was a direct relationship between the yield drop and Mo equivalent at different deformation conditions. The total thermal–mechanical energy (Q) for yielding was determined using the power-law constitutive equation. A second-order polynomial equation was used to establish a relationship between Q and the deformation temperature. The contributions of mechanical and thermal energies (Q_M and Q_T) to Q were determined based on a semi-empirical power-law equation. The developed model showed that there is a direct relationship between the amount of mechanical energy and the yield drop. It was also found that the ratio of thermal to the total energy (Q_T/Q) could increase from 0.2 to more than 0.8 by increasing temperature and decreasing strain rate.

1. Introduction

Beta titanium alloys are widely used in high-tech applications due to their low density, good strength at low temperatures, and hot corrosion resistance [1]. Different heat or thermomechanical treatments may be adopted to impart them their unique mechanical properties [2–5]. A discontinuous yielding as a sharp stress increase to a peak and a following abrupt decrease is the major phenomenon in the studies on the mechanical properties of Ti alloys. This behavior termed as the yield point phenomenon (YPP) has been the topic of many investigations. YPP has been reported in most of β-Ti alloys such as Ti–8Mn and Ti–15Mo [6], Ti–6.8Mo–4.5Fe–1.5Al [7], Ti–10V–4.5Fe–1.5Al [8], and Ti–15V–3Al–3Cr–3Sn [9]. It has been also reported in ultra fine-grained pure Ti as a small discontinuous yielding during tensile tests [10]. The major reason often proposed to account for YPP is that the initial dislocations in the material are not able to convey the early stages of yielding. This is often due to pinning the pre-existing dislocations by the solute atoms present in the alloy [11,12]. In a material in which the primary dislocations are immobilized, stress abruptly increases to activate new sources of dislocations. In the multiplication of dislocations grain boundaries has been introduced as the protruding dislocation sources in addition to the typical ones [13]. The internal plastic strain rate increases as the dislocation density rises. When the internal plastic strain rate precedes the outer one applied by the testing apparatus, stress degrades to bring them into equivalence [14]. In such materials, stress increases up to a critical point (the upper yield point, σ_U) to run the dislocation sources and when too many dislocations start to move, the flow stress suddenly descends to the lower yield point (σ_L). The plastic deformation in such cases often occurs locally in so-called "Luder bands" which expand in the workpiece during straining [15].

There are some papers dealing with modeling YPP and its underlying phenomena. Most of the applied models have followed a plasticity approach to describe σ_U, σ_L, Luder bands formation and the start of continuous deformation [16–18]. The plasticity approaches as well as phenomenological ones have been used to model the flow curves of beta Ti alloys under hot working conditions [18,19]. The current models do not elucidate how stress and temperature take part to the dislocation multiplication process. Indeed, the participation ratio of stress or temperature to the total energy of yielding has not been quantified or related to the flow behavior. In addition, the influence of alloying elements on the extent of YPP has been neglected so far. Hence, current work has aimed at shedding light on the mentioned deficiencies in the better understanding of the yield point phenomena in a beta Ti alloy.

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2. Experimental procedure

The ingot of TiMETAL-125 (Ti–6V–6Mo–6Fe–3Al) with the chemical composition presented in Table 1 was cast using a vacuum arc remelting (VAR) furnace and then homogenized at 1100 °C.

The ingot was then hot rolled at 1100 °C, quenched in water and then heat treated to get ensured of a homogenous single-phase beta grain structure.

The cylindrical specimens with 8 mm diameter and 12 mm height were machined from the hot rolled heat treated strip according to the ASTM E209 standard. The hot compression tests were carried out in argon inert atmosphere at temperatures in range of 680–880 °C, with interval of 50 °C, and strain rates of 0.001 s^{-1}, 0.01 s^{-1}, 0.1 s^{-1}, and 1 s^{-1} by an Instron 8502 servo-hydrolic universal testing machine. Graphite powder was applied on the contacting surfaces of the samples and anvils to reduce friction and therefore samples barreling. At the final step, the load–stroke data obtained from the hot compression machine were converted to true stress–true strain curves using standard equations.

Table 1
Chemical composition of TIMETAL-125 used in this investigation.

Element	V	Fe	Al	Si	Mo	Ti
Content (Wt%)	5.62	4.60	3.07	0.18	4.01	Bal.

3. Results and discussion

3.1. Effects of processing variables and chemical composition

The true stress–strain curves of beta Ti alloys have been investigated in different references. In this research YPP has been investigated through hot compression of TIMETAL125 and the results are supported by other similar results in the literature. The flow curves calculated from the load-stroke data at various temperatures and strain rates are presented in Fig. 1.

The YPP is observed as a discontinuous sharp yield point which is evident at high strain rates of 0.1 and 1 s^{-1}. At low strain rates of 0.001 and 0.01 s^{-1}, YPP is very weak at low temperatures, e.g. 680 °C, and vanishes at higher temperatures. The results show that the yield drop ($\sigma_U - \sigma_L$) (YD) depends on the deformation temperature and strain rate. Figs. 2 and 3 exhibit how YD depends on temperature and strain rate, respectively. As temperature rises or strain rate decreases, YD declines. Indeed, at temperatures beyond 780 °C and low strain rates of 0.001 and 0.01 s^{-1} YD is nearly zero and the flow curves are characterized by continuous yielding. However, by decreasing temperature and increasing strain rate the yield point becomes sharper and YD rises to 90–100 MPa, e.g. at 680 °C-1 s^{-1}.

The absence of YPP at high temperatures can be easily signified if the dependence of dislocations mobility to temperature is taken into account. With increasing the mobility of dislocations at high temperatures [20], as is in parallel to faster diffusion, they can be unleashed from the solutes pinning. Therefore, by increasing the density of free dislocations available to take part in yielding prior to deformation, σ_U decreases and approaches to σ_L.

The influence of strain rate can be analyzed in view of its influence on σ_U and the rate of dislocation production at the early stages of yielding. As the primary dislocations are pinned by the solute atoms, increasing strain rate gives rise to increasing the deformation resistance of the material up to σ_U. This is a natural reaction to increase the velocity of existent dislocations and propel the new sources to produce new dislocations. As stress approaches σ_U, the internal strain rate increases and becomes tantamount to the external one applied by the machine, just at the upper yield point. Beyond the upper yield point, stress degrades to lessen the internal strain rate and equalize it with the external strain rate at the lower yield point.

Previous investigations have agreed that there should be a relationship between the concentrations of solute atoms and the extent of YPP in beta Ti alloys [21]. Through an investigation on Ti–13V–11Cr–3A, Abbasi et al. [12] showed that the apparent activation energy to reach the upper yield point is comparable to that required for the diffusion of V into Ti. However, Table 2 shows that the activation energies for the diffusion of common alloying elements in beta Ti are nearly the same and a contribution of all solute atoms should be taken into consideration.

In order to study the influence of chemical composition on YPP and YD, some previous publications in the literature were probed. All the references observed YPP in different beta Ti alloys, which were hot compressed at various temperatures and strain rates. Fig. 4 incorporates all information in the literature with the results of current research. It is implied that YD can be related to the concentration of alloying elements incorporated in the Mo equivalent (Mo_{eq}) defined by the following equation [29]:

$$Mo_{eq} = Mo + \frac{Ta}{5} + \frac{Nb}{3.6} + \frac{W}{2.5} + \frac{V}{1.5} + 1.25\,(Cr + Ni) + 1.7\,(Mn + Co) + 2.5Fe \tag{1}$$

As shown, although YD clearly increases as Mo_{eq} rises, developing a strong relationship is hardly achieved because some other parameters such as deformation condition and grain size should be considered [21]. It seems that the variation in the curve slopes is more due to neglecting the influence of grain size than deformation condition.

3.2. Contribution of thermal and mechanical energies to yielding

As aforementioned, in a material with pinned dislocations, yielding occurs when dislocations are multiplied to some extent. The temperature and stress energies should cooperate to produce the required number of dislocations for yielding. The upper yield stress (σ_U) in Fig. 1 is an index for the dependence of yielding phenomena to temperature and strain rate. The total thermal–mechanical energy required to yield the material can be measured by a suitable constitutive description of σ_U.

The variation of σ_U with temperature and strain rate can be defined by a power-law constitutive equation as follows [30]:

$$\dot{\varepsilon}\exp\left(\frac{Q}{RT}\right) = A\sigma_U^n \tag{2}$$

where, A and n are the material constants, Q is the total thermal–mechanical activation energy for the yield, and R is the universal gas constant. The value of Q can be determined by drawing the variations of σ_U with temperature and strain rate as in Fig. 5. According to Eq. (2) the slopes indicated in Fig. 5(a) give the values of n at different temperatures. In similar, the slopes in $\ln \sigma_U - 1/T$ (Fig. 5(b)) are equal to Q/nR, from which Q can be determined for different deformation temperatures as presented in Fig. 6. The variation of Q with temperature can be correlated with the previous results which showed higher YD at low temperatures. This is easily signified because Q reflects the resistance to yielding and therefore increases at low temperatures where the material more resists against yielding. The polynomial trend of Q–T curve provides the possibility of calculating the value of activation energy at any temperature in the range of 680–880 °C.

There are two sources, temperature and deformation, which