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Behavior and modeling of flow softening and ductile damage evolution in hot forming of TA15 alloy sheets

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ABSTRACT

Flow softening and ductile damage behavior of TA15 titanium alloy with initial bimodal microstructure were studied using uniaxial hot tensile tests at different temperatures (750–850 °C) and strain rates (0.001–0.1 s^{-1}). SEM examination of deformed specimens shows that deformation mainly occurs in β and secondary α phase (α_s). Globularization of α_s is also observed. According to the SEM observations of the cracked specimens, the mechanism of ductile damage is attributed to the breakdown of the compatibility requirements at the α/β interface. Based on the experimental results, a set of mechanism-based unified elastic-viscoplastic constitutive equations have been formulated to model the flow behavior and the damage evolution of TA15 alloy in hot forming conditions. Dislocation density, ductile damage evolution, deformation heat, phase transformation and globularization of the α_s have been modeled. The model constants have been determined by using a Genetic Algorithm (GA)-based optimization method. Furthermore, the proposed constitutive equations were evaluated in terms of correlation coefficient (*R*), average absolute relative error (*AARE*), and root mean square error (*RMSE*). The results indicate that the calibrated predictions, including flow stress, volume fraction of each phase, and fracture strain, are in good agreement with experimental results.

1. Introduction

TA15 (Ti–6Al–2Zr–1Mo–1 V) alloy, a typical near α titanium alloy, has been widely used in manufacturing structure parts in aerospace industry due to its excellent performance in high-temperature strength, specific strength, thermal stability, weldability, low growth rate of fatigue crack and strong corrosion resistance [1–3].

There are different microstructures in TA15, such as fully lamellar, bi-modal and fully equiaxed microstructure. The bi-modal microstructure consists of equiaxed α phase and lamellar β-transformed microstructure. This kind of microstructure makes the alloy possess a good balance of strength and ductility [4]. Consequently, the bi-modal microstructure is usually required for many applications. As a consequence, it is of great importance to understand the deformation mechanism, flow behavior and failure behavior of TA15 alloy with bi-modal structure.

Up to now, much work has been done on flow softening behavior and constitutive modeling for the hot forming of titanium TA15 alloy. Fan et al. [1] developed a set of constitutive equations coupling microstructure evolution on the basis of hot compression tests on TA15 alloy with initial equiaxed microstructure. The loss of Hall-Pech strengthening effect was modeled to predict the flow softening. Sun et al. [3] proposed a set of mechanically based equations for modeling the evolution of dislocation density, recrystallization and grain size. The model was used to predict microstructure evolution during isothermal compressive-deformation process of TA15 alloy with the initial microstructure consisting of primary α phase (α_p) in β-transformed matrix. Gao et al. [5] developed a set of physically based constitutive model. The model was successfully used to predict the flow softening and globularization of TA15 alloy with initial lamellar microstructure. Bai et al. [6] put forward a set of mechanism-based unified elastic-viscoplastic constitutive equations, which was used to model the dominant mechanisms of flow softening for the Ti–6Al–4 V alloy with equiaxed microstructure in hot forming conditions. All the above constitutive models were based on compression tests and did not consider the ductile damage effects on flow behavior.

Ductile failure can be seen as the ultimate stage of internal degradation, result of the progressive nucleation, growth and coalescence of voids and micro-cracks that accompany large plastic strain [7]. Much work has been devoted to the understanding of cavity behavior of titanium alloy. Semiatin et al. [8,9] and Nicolaou et al. [10–13] studied the cavity behavior of Ti–6Al–4 V with different microstructures under various deformation conditions. Ko et al. [14] investigated cavity growth behavior during the superplastic deformation of submicrometer-grained titanium alloy. Dong et al. [15] studied the cavity nucleation behavior of TA15 alloy with initial lamellar microstructure using compression test. Modeling of damage is necessary for the prediction of

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formability and failure in metals under various loading conditions. Kumar et al. [16] studied the evolution of damage in near α titanium alloy IMI834 with bimodal microstructure and developed a continuum damage mechanics model, based on Lemaitre's concept of equivalent stress. Lin et al. [17] proposed a set of damage constitutive equations to model the flow behavior and damage evolution of Ti–6Al–4 V under superplastic forming conditions. Limited work has been conducted on modeling of flow softening and damage during hot tensile testing of TA15 alloy, which is necessary to simulate the hot forming process of sheet metal parts.

The aim of this study is to introduce a reliable method for modeling the flow softening and damage behavior of TA15 alloy with initial bimodal microstructure. Firstly, the as-received TA15 sheets were annealed in vacuum in order to obtain bimodal microstructure. Secondly, isothermal tensile tests and metallography examinations were employed to investigate the flow softening and damage behavior. Thirdly, based on the above experimental results, a set of constitutive equations has been formulated to model the flow and damage behavior of TA15 sheets. And then the model was calibrated using experimental results by a Genetic Algorithm (GA)-based optimization method. Finally, the accuracy of the predicted results was evaluated by statistical analysis.

2. Material and experimental details

2.1. Material details

As aforementioned, the material used in the current study is cold-rolled TA15 alloy sheet with the chemical composition listed in Table 1. The β-transus temperature is 990 °C.

2.2. Heat treatment

In order to obtain bimodal microstructure, the material was subjected to dual-annealing in vacuum. The specimens used for vacuum annealing were sealed in quartz glass tubes with the vacuum degree of 10^{-5} MPa, as shown in Fig. 1(a). The annealing diagram is shown in Fig. 1(b) with the following two steps. (1) Heating up to 950 °C with 20 °C s^{-1}, holding for 2 h, followed by air cooling and (2) re-heating up to 600 °C with 20 °C s^{-1}, holding for 2 h and final air cooling. Then cool them in air.

The original microstructure of as-received material is equiaxed microstructure, shown in Fig. 2(a). After heat treatment, a typical bimodal microstructure containing equiaxed α-phase and acicular α-phase surrounded by the transformed β-matrix (Fig. 2(b)) is obtained.

2.3. Isothermal hot tensile test

Specimens used in the hot tensile test were previously machined by wire-electrode cutting. Fig. 3(a) shows the dimensions of the specimen. In order to abate the influence of rough cutting surface on the tensile tests, cutting-profiles of the specimens were polished using 800# emery papers.

The isothermal tensile tests were conducted with a computer-controlled universal testing machine (DDL50) having a heating device. The deformation temperatures were 750 °C, 800 °C, and 850 °C. And the strain rates were 0.001 s^{-1}, 0.01 s^{-1} and 0.1 s^{-1}. For a uniform temperature distribution, all specimens were held for 3 min at corresponding deformation temperature. First, the specimens were stretched to fracture at different temperatures and strain rates. Then, hot tensile tests at different deformation extents (30% and 60% of the fracture strains) were carried out to obtain the global evolution of the microstructure. The actual strain is determined using the measured displacement. The deformed specimens (Fig. 3(b)) were cooled immediately in air after deformation. Load–displacement curves were automatically recorded during the isothermal tensile tests.

Table 1
Chemical composition of as-received TA15 sheets (wt.%).

Elements	Al	Zr	Mo	V	O	H	N	Ti
wt.%	6.67	1.97	1.18	1.41	0.075	0.0052	0.0087	bal.

2.4. Metallographic examination

After tensile tests, cross-sections of the specimens were subjected to the standard mechanical grinding and polishing routine (ASTM: ASTM E3-11), and were etched using Kroll solution ($HF:HNO_3:H_2O = 1:3:7$). Microstructures of the specimens were observed using JSM-6510A scanning electron microscope (SEM). The β-phase fractions were measured using the SEM images of microstructures by a quantitative metallographic image analysis system (Image-pro plus). And the globularization fractions of α_s were estimated using SEM images by Image-pro plus considering α phase with the aspect ratio (length/width) lower than 2 as a globular. The fractographs of the cracked specimens were also analyzed using SEM.

3. Experimental results and discussions

3.1. Calculation and correction of stress–strain curves

In order to eliminate the influence of necking on the true stress calculation, correction should be conducted. For a round tensile specimen, the well-known Bridgman equation (Eq. (1)) is usually used to correct the true stress [18]. Zhang et al. [19] proved that Bridgman's correction can still be used for necking correction of the true average stress obtained from rectangular cross-section specimens.

$$\sigma_c = \left[\left(1+\frac{2r}{a}\right)\ln\left(1+\frac{a}{2r}\right)\right]^{-1}\cdot\sigma \tag{1}$$

where a is the current radius of the neck and r is the radius of curvature of the neck surface in the longitudinal plane at the minimum section. σ is stress before correction, and σ_c is the corrected stress. The correction starts once the maximum load has passed.

The difficulty in the application of the Bridgman correction equation is the determination of the neck geometry parameter, the value of a/r. It can be calculated according to the empirical formula proposed by Le Roy et al. [20]:

$$\frac{a}{r} = 1.1(\varepsilon - \varepsilon_{P_{\max}}) \tag{2}$$

where $\varepsilon_{P_{\max}}$ is the strain when stress arrives the maximum value, and ε is the true strain.

Fig. 4 shows the corrected true stress–strain curves. Like most of the former studies [21–23], the flow stress decreases with increasing deformation temperature and decreasing strain rates. For all the flow curves, the flow stress reached a peak value at a low strain and then decreases with increasing strain. The strain corresponding to the peak stress ranges from 0.05 to 0.15. It increases slightly with increasing strain rate and decreasing deformation temperature. This behavior is the same as the compression test results of TA15 alloy with equiaxed structure reported by Fan et al. [22]. With the increase of strain, the specimens are eventually pulled to rupture. The strain to failure ε_f is affected significantly by the deformation temperature and strain rate. At the same deformation temperature, ε_f reduces sharply with increasing strain rate. At the strain rate of 0.1 s^{-1} or 0.01 s^{-1}, ε_f rises up evidently with rising deformation temperature. However, when the strain rate is 0.001 s^{-1}, the ε_f dose not grow up monotonically with growing temperature. And the ε_f is nearly equal for the cases of 800 °C and 850 °C.