Materials Science & Engineering A 708 (2017) 469–477

Contents lists available at ScienceDirect

Materials Science & Engineering A

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

Superplastic deformation behaviour and microstructure evolution of near-α Ti-Al-Mn alloy

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ARTICLE INFO

Keywords:
Superplasticity
Titanium alloys
Microsize grain structure
Dislocation structure

ABSTRACT

Superplastic deformation behaviour of conventional sheets of a near-α titanium alloy (Ti-2.5Al-1.8Mn) was studied by a step-by-step decrease of the strain rate and constant strain rate tests in a temperature range of 790–915 °C. The research found that superplastic deformation is possible in a temperature range of 815–890 °C and a constant strain rate range of 2×10^{-4} to $1 \times 10^{-3}\,s^{-1}$ with elongation above 300% and *m*-index above 0.4. Also, the research identified the optimum superplastic temperature range of 815–850 °C and constant strain rate of $4 \times 10^{-4}\,s^{-1}$ which provide a maximum elongation of 600–650%. Strain hardening is accelerated by dynamic grain growth at high temperatures of 865 and 890 °C. High dislocation activity is observed at superplastic flow in α-phase. Constitutive modelling of superplastic deformation behaviour is performed, and possible deformation mechanisms are discussed. It is suggested that grain boundary sliding between the α-grains is accommodated by a dislocation slip/creep mechanism.

1. Introduction

Application of titanium alloys in aerospace engineering is growing every year [1]. Titanium- based alloys are widely used for superplastic forming (SPF) of complex shape aerospace components [2]. SPF allows saving materials and improving product quality, in particular, they increase strength and reduce final weight of the components [1–3]. A very important advantage of the SPF method is high accuracy of die surface reproduction and a possibility to achieve very complex shape parts, which is essential for Ti-based alloys due to their poor formability at low temperatures [3]. One of the first studies on superplasticity of titanium was performed on the Ti-5Al-2.5Sn α-type alloy [4]. Subsequent studies found that excellent superplastic properties are observed in α + β type titanium alloys with high volume fraction of the β-phase [2]. The greater part of the literature on superplastic deformation of titanium alloys pay particular attention to α + β type alloys, especially Ti-6Al-4V [2,3,5–8]. At the same time, near-α titanium alloys are also very attractive for airspace applications [1]. Their advantages are as follows: light weight, superior fatigue and creep properties at elevated temperatures, adequate strength, toughness and weldability. As a result, near-α alloys are extensively used in jet engines. In recent times, studies on superplasticity of near-α alloys have considerably increased [9–18] due to rapidly growing applications of SPF of Ti alloys in airspace engineering. To perform SPF, one needs to understand the superplastic tensile behaviour and microstructure evolution during the deformation of near-α type alloys. Sun and Wan [13] studied the superplasticity of the TA15 alloy with an initial grain size of 2 μm. They showed that the alloy exhibited maximum elongation of 1074% at 900 °C and strain rate of $3.3 \times 10^{-4}\,s^{-1}$. Based on the analysis of the *m* values, the apparent activation energy and the microstructure evolution, the research demonstrated that the grain boundary sliding in the studied TA15 alloy is accommodated by the grain boundary diffusion at low strain rates and high temperatures; and also by the dislocation glide creep at high strain rates and low temperatures. Lin et al. [15] studied the microstructure and texture evolution of a near-α Ti–6.0Al–1.21Nb–9.04Zr–3.88Sn–1.59W–0.28Si alloy during tensile tests at 900 °C and initial strain rates of 3×10^{-2} and $3 \times 10^{-3}\,s^{-1}$. The authors reported that the principal deformation process was dynamic recrystallisation. It was also found that the dynamic recrystallisation process weakened the initial textures. Another near-α

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http://dx.doi.org/10.1016/j.msea.2017.10.017
Received 21 August 2017; Received in revised form 22 September 2017; Accepted 4 October 2017
Available online 05 October 2017
0921-5093/

type alloy, Ti-5Al-4Sn-2Zr-1Mo-0.25Si-1Nd, was tensile tested in a temperature range of 885–935 °C and a strain rate range of 8.3 × 10^{-4} to 1.33 × 10^{-2} s^{-1} [16]. It was shown that the dynamic recrystallisation occurs at both primary and steady stages of superplastic deformation and is responsible for the softening effect at a stable flow stage. It was concluded that the main deformation mechanism at the steady stage is grain boundary sliding accommodated by the grain rotation process. Similar effects of superplastic deformation in the near-α Ti-5.3Al-3.5Sn-3.0Zr-1.0Mo-0.4Ta-0.5Nb-0.25Si alloy were observed by Liu et al. [17] and in the (α + β) type alloy observed by Xu et al. [18].

Even though the superplastic deformation behaviour of some near-α titanium alloys has been recently studied, there is not much literature on the superplasticity of the Ti-2.5Al-1.8Mn near-α aerospace alloy. This paper provides a better understanding of the superplasticity phenomena required to perform SPF via a detailed study of the tensile deformation behaviour and the microstructure evolution of Ti-2.5Al-1.8Mn alloy.

1.1. Material and methods

Conventional sheets of Ti-2.5 wt%Al-1.8 wt%Mn alloy with a thickness of 1.55 mm were analysed. Thermo-calc (database TTTi3) was used to analyse the phase volume fractions and compositions which formed under equilibrium conditions at various temperatures. The as-annealed and as-deformed microstructures were studied after water quenching from annealing and deformation temperatures, accordingly. Annealing and deformation processes were performed in argon atmosphere to prevent oxidation of the surface. The samples for the microstructure studies were prepared by mechanical grinding on SiC papers and polishing on colloidal silicon suspension. Subsequent etching by $15\%HF + 15\%HNO_3 + 70\%C_2H_5O_8$ solution for 5 s was performed. Scanning electron microscopy (SEM) using a TESCANVega 3 fitted with EDS and EBSD techniques were applied for compositions, phases parameters, grain and subgrain size analyses. The EBSD analysis was performed with a step size of 0.4 µm and a scan area of 250 × 250 µm.

Disc-shaped samples with a diameter of 3 mm and a major axis parallel to the deformation direction were used for the transmission electron microscopy (TEM). TEM studies were performed using the JEOL JEM–2000 EX microscope. The discs were electrochemically thinned by twin-jet polishing using Struers Tenupol as the A3 electrolyte at a temperature of (20 ± 1) °C and a voltage of 28 V.

Superplastic indicators and mechanical properties at room temperature were determined using a uniaxial tensile test on a Walter – Bay LFM100 test machine with a program service for the in-situ traverse motion. The samples with a gauge section size of 6.0 × 1.55 mm and a gauge length of 17 mm were cut parallel to the rolling direction. Step-by-step decreasing strain rate tests were done to evaluate the temperature, the strain rate ranges of superplasticity and the strain rate sensitivity *m*-index. The strain rate was decreased 1.5 times in each step in a strain rate range of 1 × 10^{-2} to 5 × 10^{-5} s^{-1}. The index *m* was determined as ln*s*-ln*e* slope in each step. The constant strain rate tests were performed in a temperature range from 790 to 915 °C and a strain rate range of 2 × 10^{-4} to 1 × 10^{-3} s^{-1} to determine the elongations to failure, stress values and the strain hardening coefficient *n*.

2. Experimental results

2.1. Microstructure before start of superplastic deformation

A two-phase structure of α (dark in Fig. 1) and β (bright in Fig. 1) phases is observed in the as-proceeded sample. The initial volume fraction of the β-phase is 6% as calculated by a linear intercept method.

Fig. 2a shows the polythermal section of the Ti-Al-Mn phase diagram (Thermo-calc model). The equilibrium polymorphic transformation temperature (β-transus) of the studied composition of Ti-2.5Al-1.8Mn is calculated as 906 °C.

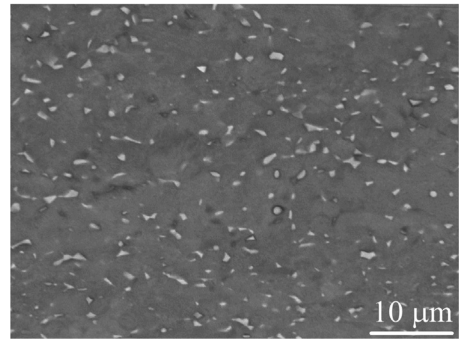

Fig. 1. Initial microstructure of the studied material.

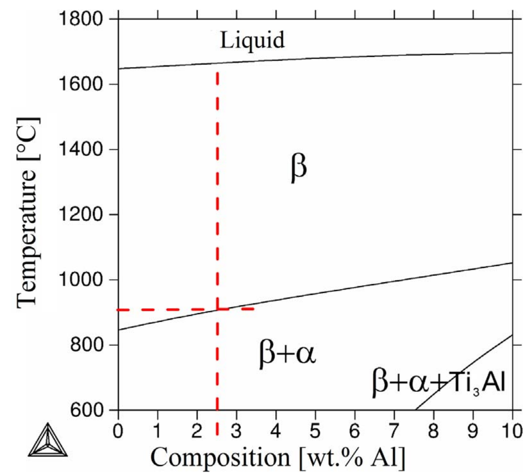

Fig. 2. Polythermic section of the Ti-Mn-Al phase diagram 1.8%Mn (Thermo-Calc, TTTi3 database).

The microstructure studies were performed after 30 min of annealing at the temperature range of 790–915 °C with a step of 25 °C and subsequent water quenching to analyse the grain size and the volume fraction of the α and β-phases before the start of superplastic deformation (Fig. 3). The martensitic plates of α′-phase are observed in the transformed β matrix as a result of the fast cooling at temperatures above 840 °C (Fig. 3d–f). Only coarse transformed β-grains with a size of (300–400) µm consisting of α′ martensite needles are formed after rapid cooling from a high temperature of 915 °C (Fig. 3f). Thus, the temperature of 915 °C belongs to the single β-phase field, which is in agreement with the ThermoCalc data (Fig. 2).

The β-phase volume fraction increases with increasing the temperature from 10% at 790 °C to 70% at 890 °C (Fig. 4a). Fine alpha (3.5 µm) and beta (1.5 µm) grains are observed after annealing at 790 and 815 °C (Fig. 4b). The grains of the α-phase slightly grow to 4.7 µm at a high temperature of 890 °C. The average β-grain size increases more than twice- from 1.5 to 3.5 µm- with increasing the temperature from 815 °C to 890 °C (Fig. 4b). The concentration of Al in both phases and Mn in the α-phase changes insignificantly in the studied temperature range (Fig. 4c). Mn concentration in the β-phase significantly decreases from 5.3 to 1.1 wt%, according to the EDS phase composition analysis and from 7.2 to 2.2 wt%, according to Thermocalc calculation, with increasing the temperature from 790 to 890 °C. It is notable that the equilibrium concentration values (dotted lines in Fig. 4c) and the SEM-EDS data (solid lines in Fig. 4c) are different, which means that the diffusion processes were incompete at annealing, and the alloy exhibits a non-equilibrium state even at high temperatures.