Materials Science & Engineering A 679 (2017) 455–464

Contents lists available at ScienceDirect

Materials Science & Engineering A

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

Tensile behavior of Ti-22Al-24Nb-0.5Mo in the range 25–650 °C

Hongze Zhao[a,b], Bin Lu[a,*], Min Tong[a], Rui Yang[a]

[a] Institute of Metal Research, Chinese Academy of Sciences, Shenyang 110016, China
[b] School of Materials Science and Engineering, University of Science and Technology of China, Shenyang 110016, China

ARTICLE INFO

Keywords:
Ti_2AlNb-based alloy
Microstructures
Tensile behavior
Fracture mechanisms

ABSTRACT

The tensile and fracture behavior of an orthorhombic alloy Ti-22Al-24Nb-0.5Mo (at%) with three types of microstructure were investigated in the temperature range between 25 °C and 650 °C. The O+bcc microstructure (basketweave lath type) possesses an elongation maximum at 500 °C and its drop at 650 °C has been attributed to dynamic strain aging as suggested by previous investigators. The α_2+bcc microstructure has the highest room temperature elongation (ε_f > 18%) and the cause of the elongation drop at 650 °C is suggested to be the fine O phase precipitates from the bcc phase. An optimal balance between strength and plasticity was achieved in the α_2+O+bcc microstructure.

1. Introduction

Ti_2AlNb-based alloys have exhibited potential for elevated-temperature structural applications since the 1990s. Different from commercial titanium alloys the Ti_2AlNb-based alloys feature an orthorhombic (O) phase (*Cmcm* system based on Ti_2AlNb), apart from a hexagonal close-packed α_2 phase ($D0_{19}$ structure based on Ti_3Al), and a body-centered cubic β phase (disordered structure, the allotrope of titanium) or B2 phase (ordered structure) [1]. It has also been recognized the presence of β/B2 phase is essential in imparting ductility and toughness to this alloy system [2]. The O phase is known to increase creep resistance and strength for these Ti_2AlNb-based alloys [3,4].

Most development and understanding of Ti_2AlNb-based alloys to date have concentrated on creep and tensile behavior [1,2,4–19]. The plastic behavior of the Ti-Al-Nb system alloys with basketweave lath structure was examined by Banerjee et al. [5] who found that a significant increase of the total elongation with increasing temperature was followed by a ductility minimum around 650 °C. Gogia et al. [2] and Nandy and Banerjee [6] reported that the ductility minimum around 650 °C occurred as a consequence of dynamic strain aging effects observed over this temperature range in the Ti-27Al-(18–24) Nb (at%) alloys. Rowe et al. [7] found that the ductility minimum around 650 °C for the Ti-22Al-25Nb alloy existed when it was tested in vacuum, but disappeared in the Ti-22Al-27Nb alloy tested in air. This work suggested that the cause of the ductility minimum was not environmental effects. Li et al. [15] investigated crack initiation and propagation in deformed microstructure of Ti_2AlNb alloys and indicated that microstructure had important effect on the tensile fracture mechanism of the alloys.

The purpose of the present research was to compare the tensile behavior of a typical Ti_2AlNb-based alloy with different microstructures in the temperature range 25–650 °C. The main objective was to clarify the relationships between microstructure and the mechanisms of fracture. Special emphasis is placed on understanding the cause of the ductility minimum around 650 °C.

2. Experimental

Cast ingots with the nominal composition of Ti-22Al-24Nb-0.5Mo (at%) were prepared by consumable vacuum arc remelting. The ingots were hot forged and then hot rolled into bars with 40 mm diameter. Chemical analysis and gas analysis of the ingot composition were conducted and the results are listed in Table 1. It can be seen that the actual composition is close to the nominal composition and the gas impurity contents were low.

The rolled bars were cut into small cylinders 10.5 mm in diameter and 60 mm in height by electro-discharge machining (EDM) and then heat treated. The various heat treatment schemes and obtained microstructures were shown in Table 2. The small cylinders were machined into M10 tensile samples with 5 mm diameter and 25 mm gage length. The samples had been tested at room temperature on an electronic universal testing machine (AG-100 kG) and at elevated temperature on an AG-X250KN machine. The samples were classified into three groups (H1, H2, H3) according to heat treatment, and all three groups of samples were tested at room temperature (25 °C), 500 °C and 650 °C. In addition H1 samples were tested at 300 °C, and H2 samples tested at 300 °C and 700 °C. At room temperature, the

* Corresponding author.
E-mail address: blu@imr.ac.cn (B. Lu).

http://dx.doi.org/10.1016/j.msea.2016.10.047
Received 7 June 2016; Received in revised form 11 October 2016; Accepted 14 October 2016
Available online 17 October 2016

Table 1
Composition of the studied alloy obtained by chemical analysis.

	Al (at%)	Nb (at%)	Mo (at%)	O (ppm)	N (ppm)	H (ppm)
Nominal	22	24	0.5	–	–	–
Actual	20.8	23.9	0.51	540	80	10

Table 2
Heat treatment schemes and corresponding microstructures for Ti-22Al-24Nb-0.5Mo alloy.

Scheme	Solution treatment	Aging treatment	Microstructure
H1	980 °C/2 h AC	–	α_2+bcc
H2	980 °C/2 h AC	780 °C/24 h AC	α_2+O+bcc
H3	1030 °C/1 h AC	780 °C/24 h AC	O+bcc

Note: AC denotes air cooling.

crosshead speed was 0.45 mm/min before yield, while it was 2.4 mm/min after yield. At elevated temperature, the loading rate was 200 MPa/min before yield and after yield the crosshead speed was 2.5 mm/min.

3. Results

3.1. Microstructure

The microstructure of as-rolled Ti-22Al-24Nb-0.5Mo alloy is shown in Fig. 1(a) and (b). In these images, a small amount of equiaxed α_2 phase is distributed at or near grain boundaries. A large number of fine O phase particles were formed in the process of cooling after rolling.

The microstructure after H1 heat treatment is depicted in Fig. 1(c), (d), where the bcc phase is the matrix and the dark particles are the α_2 phase. It is called α_2+bcc microstructure. The fine O phase in Fig. 1(a) disappeared in the process of solution treatment. Equiaxed α_2 phase formed mainly at the prior β grain boundaries and acicular α_2 phase (seen more clearly in Fig. 1(d)) precipitated mainly from the bcc phase matrix.

The microstructure after H2 heat treatment is shown in Fig. 1(e), (f), which is called α_2+O+bcc microstructure. It can be seen that the morphology and distribution of the α_2 phase are similar that after H1 heat treatment. In the process of aging treatment, a large number of fine O phase laths were precipitated from prior β grains. Some O phase particles formed at the rim of α_2 phase (such as the elongated O phase particle at the rim of the α_2 particle indicated in Fig. 1(f)) and are called rim O phase.

The microstructure after H3 heat treatment is shown in Fig. 1(g), (h) and is called O+bcc microstructure. It was found that some O phase particles precipitated continuously at the prior B2 phase grain boundaries and frequently these O phase precipitates were aligned. Roughly there were two size levels of the O phase laths (fine and coarse) and in some cases the O phase laths extended to greater than 10 μm long without impediment by neighboring O laths.

3.2. Mechanical properties

Representative true stress-strain curves for H1, H2, and H3 heat treated samples at room temperature are shown in Fig. 2. Here the true strain was calculated from the extensometer reading. The samples with α_2+bcc microstructure were broken with necking, but those with α_2+O +bcc or O+bcc microstructure were broken without necking at room temperature. It can be seen that the sample with α_2+bcc microstructure exhibited no work hardening (Fig. 2(a)), while the samples with α_2+O +bcc or O+bcc microstructure showed a little work hardening effect (Fig. 2(b), (c)). Data of yield strength and elongation from H1, H2, and H3 heat treated samples are presented in Fig. 3. The H1 heat treated samples with α_2+bcc microstructure exhibited the largest elongation value of 18% along with the lowest yield strength of 888 MPa at room temperature (Fig. 3(a)). For H2 heat treated samples, from room temperature to 650 °C, the yield strength dropped by approximately 25% and the elongation increased by approximately 50% (Fig. 3(b)). For samples with O+bcc microstructure, the yield strength decrease was nearly linear from room temperature to 650 °C, but the elongation appeared to have a maximum around 500 °C (20.8%), after which it dropped by nearly 40% at 650 °C (Fig. 3(c)).

3.3. Fractography

The fractographs of tensile samples with α_2+bcc microstructure are shown in Fig. 4. Ductile dimples were evident throughout the fracture surfaces of samples tested at room temperature (Fig. 4(b)). In addition, the room temperature tested sample exhibited a cup-cone-type fracture representative of ductile metals. However, when tested at 650 °C the elongation decreased by approximately 70% as compared to room temperature (Fig. 3(a)). The fracture surface of Fig. 4(c) showed that the fracture mechanism was intergranular at the edge of the round tensile sample, but there were some big cracks at the center of the sample and there were some dimples near the cracks (Fig. 4(d)).

The fractographs of tensile samples with α_2+O+bcc microstructure are shown in Fig. 5. At room temperature the fracture mechanism is quasi-cleavage. At low magnification, the fractograph exhibited river shape patterns (Fig. 5(a)), but at high magnification there were a few dimples and slip bands associated with the bcc phase (Fig. 5(b)). The fracture surface at 500 °C showed that failure initiation and propagation in the center of the samples occurred transgranularly in a relatively smooth manner (Fig. 5(c)). At high magnification, it can be seen that the fractograph exhibited a number of equiaxed dimples (Fig. 5(d)). The fractographs after tensile test at 650 °C were similar to those tested at 500 °C, but there were more and larger equiaxed dimples compared to the case at 500 °C.

Fig. 6 shows the fractographs of tensile specimens with O+bcc microstructure. At room temperature apparently intergranular fracture occurred (Fig. 6(a)). Slip bands were present at some facets (Fig. 6(b)), while at other facets there were only very smooth surfaces and no slip bands could be seen. At 500 °C, the fracture surface indicated that cracks had propagated perpendicularly to the loading direction and the samples exhibited transgranular fracture (Fig. 6(c)). High magnification image revealed a number of dimples (Fig. 5(d)). At 650 °C, interestingly, a mixture of intergranular and transgranular fracture was observed (Fig. 6(e)). Macroscopic examination of the failed tensile bars after testing at 650 °C showed that crack initiation was by intergranular facture at the edge of the samples and then rapid transgranular fracture occurred across the rest of the cross-section. High magnification image (Fig. 6(f)) showed fewer dimples than samples tested at 500 °C.

Metallographic sectioning of tensile samples with α_2+bcc microstructure revealed that the α_2 phase had undergone significant deformation in the necked region at room temperature (Fig. 7(a)). From Fig. 7(b), it can be seen that some primary α_2 phase particles had cracked after significant deformation and the secondary α_2 phase particles were oriented parallel to the tensile axis (the vertical direction). During testing at 650 °C the α_2 phase had hardly deformed (Fig. 7(c)) and the secondary α_2 phase particles were not oriented parallel to the tensile axis (Fig. 7(d)).

The samples with α_2+O+bcc microstructure exhibited a few cracks that were approximately normal to the tensile axis and near the fracture surface the α_2 phase had undergone a small amount of deformation at room temperature (Fig. 8(a), (b)). During test at