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Tensile and fracture behavior of boron and carbon modified Ti-15-3 alloys in aged conditions

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ABSTRACT

This work illustrates the effect of boron and carbon addition on the mechanical behavior of a beta Ti alloy, Ti–15V–3Cr–3Al–3Sn (Ti-15-3), in differently aged conditions. The alloys were prepared by consumable vacuum arc melting followed by forging and hot rolling. These were subsequently solution treated and aged at different temperatures above 500 °C for 8 h. Standard tensile and plane strain fracture toughness tests were carried out to understand the mechanical behavior of the alloys and its correlation with the microstructural features characterized by scanning and transmission electron microscopy. Both the boron- and the carbon-containing alloys exhibit improved strength with comparable elongation to failure values as compared to the base Ti-15-3 alloy. The presence of TiB and TiC precipitates in a matrix of fine α with β results in lower fracture toughness (K_{IC}) in the boron- and carbon-containing alloys as compared to the base alloy. However, at higher aging temperatures K_{IC} improves due to more tortuous crack path because of the presence of coarse α-phase. An empirical relationship has been proposed correlating K_{IC} with the volume fraction, size and interspacing of α in these alloys.

1. Introduction

Ti–15V–3Cr–3Al–3Sn is one of the important metastable β-Ti alloys, which is used for several airframe structural applications such as: environmental control system duct, engine bracket, sheet, plate, aircraft springs etc. [1,2]. The properties of Ti-15-3 alloy significantly depend on the thermo-mechanical processing conditions. Heat treatment usually involves solution treatment followed by aging. Several researchers have shown improvement in tensile properties of this alloy by modifying the microstructure employing different heat treatments and thermo-mechanical processing [3,4]. Recently, alternate routes are being tried to enhance the mechanical properties of different β-Ti alloys by incorporating hard non-deformable particles by the additions of boron and carbon [5–7]. The presence of boride and carbide particles in Ti-15-3 alloy plays an additional role in modifying the α–β structure [8–10] over heat treatment and thermo-mechanical processing. An understanding of the evolution of microstructures and consequently their effects on the resultant properties of the boron and carbon modified alloys appear essential for better exploitation of these alloys in service. Specifically, properties such as fracture toughness, which is an important design parameter for β-Ti alloys, need careful attention because these alloys are used for different structure critical applications in aerospace industry.

Studies related to tensile and fracture properties and their dependence on the structure of boron and carbon modified Ti-15-3 alloy after heat treatment or thermo-mechanical processing are limited. Though some attempts have been made earlier to study the tensile behavior of modified Ti-15-3 alloys, research on the fracture behavior of in-situ modified β-Ti alloy with boron and carbon is almost non-existent. As the variation in the microstructure and its constituents considerably affects fracture toughness of Ti alloys, generating information on fracture toughness and associated understanding of the fracture behavior of the base and the modified Ti-15-3 alloys are considered important. Establishing correlation of fracture properties with the corresponding microstructures would be some requisite supplement.

The effects of boron and carbon additions on tensile and fracture properties of a β Ti alloy, Ti-15-3, have been studied in detail and are reported here. Additionally, the effect of heat treatment on the properties of thermo-mechanically processed Ti-15-3 alloys modified by boron and carbon additions has been investigated in order to achieve optimization of strength, fracture toughness and

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heat treatment schedule.

2. Experimental

Ti–15V–3Cr–3Sn–3Al alloy and two of its modifications with 0.2 wt% B and 0.2 wt% C were prepared by consumable vacuum arc melting in the form of ingots of 20 kg. The preparations of the alloys and their chemical compositions are already reported elsewhere [11]. Ingots were hot forged in the β-phase field to a height reduction of 50% using an open die 1 t pneumatic power hammer (MASSEY, Germany). Before forging, the ingots were soaked at 1050 °C for 1 h in a furnace. The forged ingots were next processed to a thickness of 15 mm using a two-high rolling mill (DEMAG, Germany). The alloys were soaked at 950 °C for 2 h in a furnace before hot rolling. During the repeated forging and rolling steps, ingots were annealed intermittently to maintain the processing temperature. The wrought alloys were first solution treated above β-transus temperature at 900 °C for 1 h followed by water quenching to retain the metastable β-phase. The solution treated alloys were subsequently aged at different temperatures for the duration of 8 h to develop two-phase equilibrium α–β structures avoiding the presence of any metastable phase. The heat treatment details and the sample coding are given in Table 1.

Specimens for microstructural examinations were prepared using standard metallographic techniques. A Quanta 400 (FEI) scanning electron microscope was used for the examination of the microstructures and the fracture surfaces of the failed specimens. A Tecnai G^2 20T TEM was used for investigations of the foils, which were prepared by electro-polishing at −50 °C using a twin jet electro polisher (FISCHIONE Instruments, Model 110). A solution of 5% H_2SO_4 in methanol was used as the electrolyte.

Tensile tests were performed at room temperature using cylindrical specimens of 5 mm gage diameter and 25 mm gage length. These tests were carried out in a screw driven Instron machine (model: 5500R) at a nominal strain rate of $6.66 \times 10^{-4}\ s^{-1}$. Three tensile tests were conducted on specimens in each heat treated condition and the average values of the tensile properties obtained from these tests are reported.

Estimations of plane strain fracture toughness of the materials were carried out using a 100 kN hydraulic Universal testing machine (Dartec, UK with upgraded controller from BISS, Model No. BI-46-402) using 1/2 CT specimens prepared as per ASTM E 399-12 standard. Notches were initially machined to 0.35 times the width (W) of the specimen followed by fatigue pre-cracking up to 0.45–0.55 W. The load (P) and the crack opening displacement (COD) data obtained from the tests were recorded to estimate the critical load for crack initiation (P_Q) using the standard 5% secant method. The conditional fracture toughness (K_Q) values were estimated following ASTM E399-12 standard. The major validity criteria for plane strain fracture toughness (ASTM E399-12, 2013) were examined prior to assigning K_Q to be K_{IC}. The criteria that have been examined are: (i) the ratio P_{max}/P_Q should not exceed 1.1, (ii) specimen thickness and the ligament size (W–a) should be greater than the values of $2.5(K_Q/\sigma_{YS})^2$, and (iii) the ratio of the deviation of the load–COD curve (trace) from the tangent drawn from the initial linear part (the elastic loading line) at $0.8P_Q$ and P_Q should be less than 0.25 for all the specimens. The K_Q values of all the specimens were found to satisfy the above-mentioned validity criteria for K_{IC} estimation except the ones estimated for SA600 specimens. The 1/2 CT specimens of the base alloy aged at 600 °C appear to be in the plane stress fracture toughness regime. Fracture toughness of SA600 specimens has been considered as K_Q in further discussions.

Table 1
Heat-treatment details and sample codification.

Sample	Heat-treatment	Aging temp. (X °C)	Specimen code
Ti-15-3	Solution treatment (ST): 900 °C/1 h/ water quenched + Aged: @X °C/8 h/ air cooled	500	SA500
		535	SA535
		550	SA550
		600	SA600
Ti-15-3-0.2B		500	BSA500
		535	BSA535
		550	BSA550
		600	BSA600
Ti-15-3-0.2C		500	CSA500
		535	CSA535
		550	CSA550
		600	CSA600

*where X is aging temperature

3. Results

3.1. Microstructure

The SEM micrographs of the boron- and carbon-containing alloys aged at 500 °C and 600 °C are shown in Fig. 1(a–d). Two-phase α–β structures are observed in all the alloys along with boride and carbide particles in the boron- and the carbon-containing alloys, respectively. Carbide particles are more uniformly distributed than the boride particles, which are mostly in the form of clusters and are located near or at the grain boundaries. Coarsening of the α-phase is observed in all the alloys due to an increase in aging temperatures from 500 °C to 600 °C (Fig. 1). Size of the α phase is finer in the boron- and carbon-containing alloys compared to the base alloy due to the presence of the boride and the carbide particles [7,8,11]; the refinement of the matrix microstructure is more pronounced in the carbon-containing alloy. The width and the interspacing of the α-phase precipitates, measured (considering at least 200 aligned as well as randomly oriented α-laths) using a number of SEM and TEM micrographs for the boron- and carbon-containing alloys in different aging conditions, are shown in Table 2. Similar measurements for the base alloy, as reported elsewhere [12], are also shown in Table 2 for comparative analysis. A significant change in the size of α-phase is observed in the boron- and carbon-containing alloys with respect to that of the base alloy.

The theoretical estimates of the volume fractions of α phases, obtained using Thermo-Calc® software, for the boron- and the carbon-containing alloys at different temperatures are shown in Fig. 2; the estimations for the base alloy are also included in this figure for comparison [12]. The volume fraction of α gradually decreases with increase in aging temperature from 500 °C to 600 °C in the boron- and carbon-containing alloys like that in the base alloy. The percentage volume fraction of α at any particular aging temperature is almost same in all the three alloys, as considerable change in volume fraction of α phase is not expected due to addition of only 0.2% boron or 0.2% carbon in the alloys, subjected to similar heat treatment.

The results of X-ray diffraction for the boron- and the carbon-containing alloys in solution treated and in solution treated plus aged conditions are shown in Fig. 3. The presence of α-phase can be seen from the XRD traces (Fig. 3) in the aged specimens, unlike that in the solution treated specimens of the investigated alloys. Peaks corresponding to orthorhombic TiB and face-centered cubic TiC phases are also observed in the boron- and the carbon-containing alloys, respectively. The details of the XRD pattern of the boron-containing alloy near the (110) β Ti in Fig. 3a has been