The elevated-temperature fatigue behavior of boron-modified Ti–6Al–4V(wt.%) castings

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This work investigated the effect of nominal boron additions of 0.1 and 1.0 wt.% on the elevated-temperature (455 °C) fatigue deformation behavior of Ti–6Al–4V(wt.%) castings for maximum applied stresses between 250 and 450 MPa (R = 0.1 and 5 Hz). Boron additions resulted in a dramatic refinement of the as-cast grain size, and larger boron additions resulted in larger titanium-boride (TiB) phase volume percents. The boron-containing alloys exhibited longer average fatigue lives than those for Ti–6Al–4V, which was suggested to be related to the reduced as-cast grain size and the addition of strong and stiff TiB phase. The Ti–6Al–4V–0.1B alloy exhibited the longest average fatigue lives. The TiB phase cracked during the fatigue experiments and this resulted in a decreasing Young’s modulus with increased cycle number. Each alloy exhibited α-phase cracking and environmentally assisted surface edge cracking.

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1. Introduction

Conventional titanium (Ti) alloys containing small additions of boron (B) have been gaining considerable interest in recent years due to their attractive mechanical properties including high room temperature (RT) specific strength and Young’s modulus (E) along with reasonable elongation-to-failure (εf) [1–8]. Table 1 lists the RT tensile properties of as-cast Ti–6Al–4V–xB(wt.%)\(^1\) (0 ≤ x ≤ 1) alloys [7,8]. The significant increase in strength and E in Ti–B alloys arise from the strong and stiff titanium-boride (TiB) phase (E~371–482 GPa) that precipitates in situ during solidification [2,5]. TiB reinforcement in Ti is advantageous over other reinforcements as there exists a crystallographic arrangement between Ti and TiB resulting in a clean interface without significant reactivity [9]. Due to the similarity of the thermal expansion coefficients (8.2 × 10\(^{-6}\) C\(^{-1}\) for Ti and 6.2 × 10\(^{-6}\) C\(^{-1}\) for TiB), processing of such alloys is less challenging than for other ceramic-reinforced alloys. In addition, recent work has shown that the addition of trace amounts (~0.1 wt.% of B to Ti–6Al–4V(wt.%)) decreases the as-cast grain size by approximately an order of magnitude [3]. This drastic reduction in the as-cast grain size leads to significant benefits including increasing yield strength while reducing or avoiding time spent on expensive and energy-intensive thermomechanical processing. Small B additions do not result in significant density changes, thus the higher strength and stiffness of Ti–B alloys provides important increases in specific strength and stiffness.

Although Ti–B alloys are currently used for various ambient-temperature commercial applications (e.g. exhaust valves of automotive engines) [1], understanding the mechanisms of deformation and fracture is necessary to consider these materials for fracture-critical applications (e.g. aerospace) especially under fatigue loading conditions. To date no studies of the elevated-temperature fatigue behavior of cast Ti–6Al–4V–xB alloys have been performed. This study focused on the deformation evolution and mechanisms in Ti–6Al–4V–xB during elevated-temperature (455 °C) fatigue. The main objectives were to access the effects of B additions on the fatigue behavior of Ti–6Al–4V and to evaluate and understand the fatigue deformation behavior of Ti–6Al–4V–xB alloys as a function of stress level and B concentration.

2. Experimental

2.1. Materials and processing

Table 2 lists the compositions of the alloys, which were measured through inductively coupled plasma optical emission spectroscopy and inert gas fluorescence. Castings of 70 mm diameter and 500 mm length were produced for each composition at Flowserve Corporation, Dayton, OH via induction skull melting. All castings were hot isostatically pressed at 900 °C and 100 MPa for 2 h. No post-processing heat treatments were performed prior to fatigue testing.

The as-processed materials were prepared for imaging using conventional metallography techniques and scanning electron microscopy (SEM) was used to examine the as-cast grain size, the
α- and β-lath widths, the phase distributions and their morphologies. Phase volume percents and lath widths were determined using ImageJ image analysis software on several backscatter electron (BSE) SEM photomicrographs acquired using either a CamScan4FE Field Emission SEM or a Cambridge Stereoscan 250 Mk 2 SEM. The lath widths were measured for over 300 laths for each of the α and β phases, for each alloy composition, via the line-intercept method.

2.2. Fatigue experiments

Dogbone-shaped specimens were machined via electrical discharge machining (EDM). Recast layers, which formed during EDM, were removed through silicon carbide paper grinding to a 600-grit finish. Open-air, tension-tension fatigue experiments were performed on a horizontal servohydraulic frame previously described by Hartman and Russ [10]. The test temperature was 455 °C. The test temperature was 455 °C and the applied maximum stresses ranged between 250 and 450 MPa with a stress ratio of 0.1. Strain was measured using an alumina-rod high-temperature extensometer, with a 12 mm gage length, attached to the gage section of the specimen. The strain-life behavior was documented throughout the experiments and both the Young’s modulus and hysteresis behavior were tracked at certain cycles during the experiments. In such cases, 20 stress–strain data points were acquired throughout the desired cycle. The specimens were locally heated using a Barber Coleman’s temperature controller and two sets of heating banks, located approximately 5 mm above and below the sample, each containing four evenly spaced quartz lamps. Specimen temperatures were monitored by four chromel–alumel type-K thermocouples located within the specimen’s gage section. Targeted test temperatures were maintained within ±5 °C. The test specimens were soaked at 455 °C for at least 20 min prior to applying load in order to minimize the thermal stresses. Runout was determined to be greater than 1,000,000 cycles and the test frequency was 5 Hz. For the fractured samples, the furnace was turned off immediately upon fracture to prevent oxidation of the exposed surfaces. A minimum of three experiments were performed for each alloy at the following maximum applied stress levels: 350, 400, and 450 MPa. Maximum stresses of 250 and 450 MPa were also evaluated. The fractured surfaces and gage sections of the specimens were analyzed using ImageJ software on several backscatter electron (BSE) SEM photomicrographs acquired using either a CamScan4FE Field Emission SEM or a Cambridge Stereoscan 250 Mk 2 SEM. The lath widths were measured for over 300 laths for each of the α and β phases, for each alloy composition, via the line-intercept method.

Table 1

<table>
<thead>
<tr>
<th>Alloy</th>
<th>YS (MPa)</th>
<th>UTS (MPa)</th>
<th>ε (×10&lt;sup&gt;-6&lt;/sup&gt;)</th>
<th>E (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ti–6Al–4V–0.05B [7]</td>
<td>879</td>
<td>944</td>
<td>8.5</td>
<td>110</td>
</tr>
<tr>
<td>Ti–6Al–4V–0.1B [7]</td>
<td>900</td>
<td>966</td>
<td>7.3</td>
<td>116</td>
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<tr>
<td>Ti–6Al–4V–0.4B [7]</td>
<td>913</td>
<td>973</td>
<td>5.2</td>
<td>126</td>
</tr>
<tr>
<td>Ti–6Al–4V–1B [8]</td>
<td>975</td>
<td>1045</td>
<td>2.5</td>
<td>146</td>
</tr>
</tbody>
</table>

3. Results

3.1. Microstructure

Table 3 lists the average volume percents of the TiB and β phases for each alloy along with the average α- and β-lath widths. Photomicrographs of the microstructures for each alloy are illustrated in Fig. 1. The addition of 0.1B to Ti–6Al–4V refined the as-cast grain size (referred to as the prior-β grain size) from 1700 to 200 μm. The photomicrographs in Fig. 1 show that higher B concentrations lead to higher TiB phase (darkest phase) volume percents. The volume percents of TiB in the Ti–6Al–4V–0.1B and Ti–6Al–4V–1B alloys were 0.6 and 5.8, respectively. The volume percents of the β-phase remained relatively constant (13 < V<sub>β</sub> < 16) for each alloy. The 0.1B addition reduced the thickness of the grain boundary α-phase from approximately 10 μm to approximately 2 μm, which is expected to improve the workability of this alloy system and increase the ε<sub>f</sub> [6].

The TiB phase formed a necklace structure around the prior-β grain boundaries in the Ti–6Al–4V–0.1B alloy, see Fig. 1b, which helped restrict grain growth. The average α-lath width varied slightly with B concentration and was always between 3.6 and 4.2 μm.

3.2. Fatigue behavior

The fatigue S–N curves obtained for the as-cast alloys are shown in Fig. 2. For the maximum stress levels of 350 MPa, 400 MPa, and 450 MPa, the three Ti–6Al–4V–0.1B specimens tested each exhibited longer lives than the specimens of the other alloys. Comparing the Ti–6Al–4V–1B and Ti–6Al–4V alloys, there was overlap in the N<sub>f</sub> ranges exhibited at a given maximum stress level, though the Ti–6Al–4V–1B alloy exhibited greater average N<sub>f</sub> values. It is noted that the S–N behavior for the Ti–6Al–4V alloy was quite similar to that for a cast-then-forged Ti–6Al–4V alloy, with an α-lath width size between 1 and 2 μm, tested at 450 °C [11].

Hysteresis plots of selected samples during defined cycles are illustrated in Fig. 3. The E tendency to decrease with increasing cycle number indicating that damage was occurring in the specimens. The damage was observed to be in the form of both TiB-phase cracking and α-phase cracking which will be shown later. An E value decrease of 14% was the maximum decrease for a Ti–6Al–4V–0.1B sample tested at ε<sub>f</sub> = 990,554. However, usually only an E decrease of 5% was exhibited for the samples prior to fracture. For all of the runout samples, an E value decrease of less than 5% was observed. No phase instability was observed on the post-fatigue gage sections even after more than two million cycles of fatigue exposure at 455 °C. In such cases the measured volume percents of the TiB, α, and β phases were similar both before and after the experiments.

For the B-containing alloys, TiB-phase cracking was evident within the deformed gage sections (see Fig. 4b and c). This behavior was similar to that for a Ti–6Al–4V/TiB composite containing 0.5 wt.%B [4]. The α-phase also exhibited a significant amount of cracking, as seen in Fig. 4a, while cracks were only rarely observed within the fine β-phase. For each alloy, cracks were observed along...
Fig. 1. Low- and high-magnification backscattered electron SEM images of the (a and b) Ti–6Al–4V, (c and d) Ti–6Al–4V–0.1B, and (e and f) Ti–6Al–4V–1B alloy microstructures. The TiB phase is black, the α-phase is gray, and the β-phase is white.

the edges of the samples and grew perpendicular to the loading direction (see Fig. 5). Thus the environment affected the fatigue performance and it is felt that the α-case on the surface, due to the oxidation, assisted this surface cracking. The final failure was likely initiated by one of these cracks which grew into the sample.

SEM photomicrographs of fracture surfaces for each alloy are illustrated in Figs. 6 and 7. Fracture initiation sites were typically located at surface locations (see Fig. 6a and b). In some cases, multiple surface crack initiation sites were evident. Fig. 6a and b illustrates the crack initiation, crack propagation, and overload regions for a Ti–6Al–4V specimen. The overload regions contained ductile dimples (see for example Figs. 6b and 7c). The crack growth regions of the sample consumed a large extent of the thickness of the samples (see Fig. 8). Fig. 8 illustrates the environmental susceptibility of the alloys as the colored region on the fracture surface indicated the crack growth region which oxidized during the experiment. The shiny metallic regions outside the colored region in Fig. 8 represent the overload regions/surfaces which were not oxidized. The fracture surface appearance of the Ti–6Al–4V–1B specimens (Fig. 7b) was somewhat different than that for the Ti–6Al–4V and Ti–6Al–4V–0.1B specimens as larger faceted regions were evident and fewer dimples were evident. The fracture surface of Ti–6Al–4V–0.1B indicated a more tortuous crack path than that for Ti–6Al–4V–1B. In some cases, the Ti–6Al–4V–1B fracture surfaces exhibited cleavage, indicating brittle fracture.

4. Discussion

4.1. Microstructure

The α-lath width is controlled by the thermodynamics during the β → α phase transformation. The TiB and β phases form in the eutectic reaction [6]. When the B concentration is low, such as in Ti–6Al–4V–0.1B, the TiB phase primarily forms a necklace structure around the prior-β grains (see Fig. 1b) as has been described
previously [3]. When the B concentration is higher, such as in Ti–6Al–4V–1B, the TiB phase forms not only at the prior-β grain boundaries, but also inside the grains (see Fig. 1c). The TiB phase facilitates the nucleation of more α laths [12]. Thus it is expected that larger TiB volume percents would result in a reduced α-lath width during the β → α phase transformation.

Sen et al. [7] studied the fatigue crack growth behavior of as-cast Ti–6Al–4V, Ti–6Al–4V–0.05B, Ti–6Al–4V–0.1B, and Ti–6Al–4V–0.4B cast alloys. The Ti–6Al–4V and Ti–6Al–4V–0.1B materials of their study were cut from the same ingot as those examined in the current study. They observed decreasing α-lath widths for the alloys containing greater B contents. The α-width measurement (6.8 ± 1.8 μm) for their as-cast Ti–6Al–4V alloy was significantly larger than that measured in the current study. The measured α-lath width of the Ti–6Al–4V–0.1B (4.2 ± 1.2 μm) alloy in their work was almost identical to that measured for the same alloy in the current study. Their measurement of the α-lath width for a Ti–6Al–4V–0.4B (3.5 ± 1.1 μm) alloy was very close to that found for the Ti–6Al–4V–1B alloy of the current study. The trend observed in both studies was that little decrease in the α-lath widths occurs for B contents greater than 0.1B. Overall, neither the α-lath nor β-lath widths changed dramatically for the alloys examined in the current study, although a slight decrease in the average α-lath width was observed for the Ti–6Al–4V–1B alloy compared with the Ti–6Al–0.1B alloy.

### 4.2. Fatigue behavior

The improved average fatigue lives of the B-modified alloys compared with Ti–6Al–4V is believed to be due to the microstructural differences. Ti–6Al–4V exhibits superior high-cycle smooth bar fatigue lives when the slip length is small [13–16]. Reducing the grain size from 12 to 2 μm in equiaxed Ti–6Al–4V microstruc-

![Fig. 2. Maximum applied stress versus cycles to failure curves where runout specimens are marked with an arrow.](image)

![Fig. 3. Stress versus strain plots at the marked cycle numbers for (a) Ti–6Al–4V tested at a maximum stress of 450 MPa, (b) Ti–6Al–4V–0.1B tested at a maximum stress of 450 MPa, and (c) Ti–6Al–4V–1B tested at a maximum stress of 300 MPa.](image)
Fig. 4. SEM BSE photomicrographs of a (a) Ti–6Al–4V specimen which failed after 44,153 cycles at a maximum stress of 350 MPa. Cracking within the α-phase laths was exhibited, (b) Ti–6Al–4V–0.1B specimen which failed after 990,554 cycles at a maximum stress of 400 MPa. (c) Ti–6Al–4V–1B specimen which failed after 32,718 cycles at a maximum stress of 350 MPa. Cracking within the TiB phase was exhibited for the B-modified alloys in (b) and (c).

Fig. 5. SEM low-magnification BSE photomicrographs of the edges for a Ti–6Al–4V specimen which failed after 44,153 cycles at a maximum stress of 350 MPa.
Fig. 6. SEM secondary electron photomicrographs of the fracture surface for a Ti–6Al–4V specimen which failed after 28,848 cycles at a maximum stress of 350 MPa. The crack initiation sites are indicated with arrows. The overload region in (b) exhibited ductile dimples.

Fig. 7. SEM secondary electron photomicrographs of the fracture surface for a (a) Ti–6Al–4V–0.1B specimen which failed after 990,554 cycles at a maximum stress of 400 MPa, and for a (b and c) Ti–6Al–4V–1B specimen which failed after 32,718 cycles at a maximum stress of 350 MPa. The overload region in (c) exhibited ductile dimples.
Fig. 8. Low-magnification digital image of the oxidized crack growth region (dark colored) for a Ti–6Al–4V–1B specimen which fractured after 25,555 cycles for a maximum applied stress of 400 MPa. The ruler scale is in millimeters.

Since the strength and $E$ of the TiB phase are significantly higher than those for the $\alpha$ and $\beta$ phases, much of the load placed on the alloy is supported by the TiB phase. As a consequence of the load sharing by the TiB phase, stress relaxation in the $\alpha$ and $\beta$ phases can occur. The highly loaded TiB phase can break at locations where stresses are maximum and where flaws accumulated. Load sharing was also explained to be the reason for the increased creep resistance of B-modified Ti–6Al–4V cast alloys compared with Ti–6Al–4V castings [20].

5. Summary and conclusions

Cast Ti–6Al–4V, Ti–6Al–4V–0.1B, and Ti–6Al–4V–1B alloys were fatigue tested at 455 °C and maximum applied stresses between 250 and 450 MPa ($R = 0.1$). Boron additions dramatically decreased the as-cast grain size, however, the $\alpha$- and $\beta$-lath widths did not change significantly with respect to B content. The B-containing alloys exhibited longer average fatigue lives than those for Ti–6Al–4V, and the Ti–6Al–4V–0.1B alloy exhibited the longest average fatigue lives. This was explained to be a result of the higher $\varepsilon_f$ value of Ti–6Al–4V–0.1B compared with the other alloys. The enhanced fatigue resistance of the Ti–6Al–4V–1B alloy compared with Ti–6Al–4V was explained to be a consequence of the load-sharing mechanism by the strong and stiff TiB phase that precipitated as well as its presence as a barrier to dislocation motion. Thus, in spite of the observed TiB-phase cracking, small B additions resulted in improved fatigue resistance. Overall, this work suggests that Ti–6Al–4V–xB alloys may serve as adequate replacements for elevated-temperature fatigue-driven applications of Ti–6Al–4V.

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