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An alternative method to separate and analyse the microtextures and microstructures of primary alpha grains and transformed beta grains in near-α titanium alloy Timetal 834

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Abstract

An alternative technique to that recently put forward by Germain et al. [Materials Characterization 54 (2005) 216–222] to separate the orientations of primary alpha (α_p) grains from those of transformed beta (α_s) grains in the near-α titanium alloy Timetal 834 is presented. This new method involves correlating orientation image maps (OIM) obtained through electron back-scattered diffraction (EBSD) with optical images of the same area. By using optical microscopy and an appropriate etch, strong contrast between α_p and α_s is obtained enabling regions within an OIM containing the α_p and α_s to be determined using relatively straightforward image analysis. Results are presented for both high-resolution microstructure analysis and texture level EBSD datasets for material subjected to a simulated industrial thermomechanical forging process. A comparison is also made between the texture datasets for both α_p and α_s obtained using this method and that obtained using grain size distribution acquired directly from the EBSD dataset. In this case very little difference was found between the separated textures, suggesting that in the first instance a morphological/grain size approach direct from the EBSD dataset is sufficient for observing any trends in texture evolution of both α_p and α_s.

Keywords: Timetal 834; Bimodal microstructure; EBSD; Microtexture; Texture; Image analysis

1. Introduction

Timetal 834 (nominal composition Ti–5.8Al–4Sn–3.5Zr–0.7Nb–0.5Mo–0.35Si–0.06C) is a near-α titanium alloy possessing an excellent combination of tensile strength, creep and low cycle fatigue resistance up to 600 °C [1]. Under dynamically loaded conditions it has the highest temperature capability of any titanium alloy and is almost exclusively used in structure critical components such as jet engine compressor sections. Critical to the success of the alloy is its capability of being thermomechanically processed and heat-treated high in the α+β phase field. Thus, when processed to achieve an optimum combination of creep and fatigue resistance, a bimodal microstructure of approximately 10–30% primary α (α_p) grains within a fine transformed beta (α_s) matrix is produced, an example of which is shown in Fig. 1a [2]. However, the alloy can be susceptible to premature in-service failure when subjected to relatively long periods of constant stress during its loading cycle, so-called low cycle dwell fatigue (LCDF) [3]. Bache et al. [4] propose that

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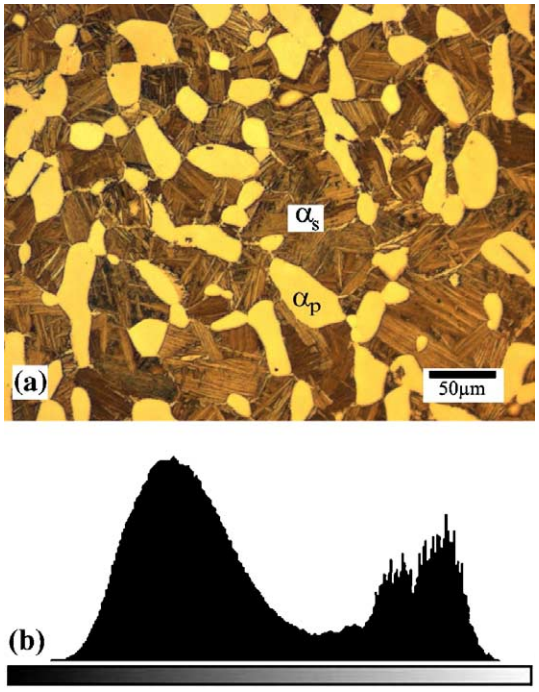

Fig. 1. (a) Optical image of Timetal 834 containing 18% primary α (α_p) in a matrix of transformed β (α_s) and (b) the gray level histogram of (a) showing two well-defined peaks corresponding to α_p and α_s.

under such circumstances crack initiation is dominated by the development of cleavage facets in the α_p closely aligned to the basal (0001) plane. This would suggest that crystallographic texture within this alloy, particularly that of the α_p, plays a significant role on dwell fatigue behaviour and, therefore, service life.

The current work is part of a larger investigation on the effect of thermomechanical processing parameters on the texture evolution of the individual phases within Timetal 834. Historically, determination of these textures has been complicated as any global textures obtained using bulk X-ray analysis will be a function of the volume fraction of each transformation product, α_p and α_s. Moreover, since α and α have the same α-hcp crystal structure and typically there is only 10–30% α_p, any α_p texture will be difficult to quantify as the global texture will be dominated by the α_s.

Consequently, a key issue in elucidating the effects of texture in alloys such as Timetal 834 is the separation of the α_p and α_s textures. The most obvious technique to perform this task would appear to be electron backscattered diffraction (EBSD). Most importantly EBSD enables the construction of an orientation image map (OIM) as the spatial locations of individual orientations are obtained [5]. Consequently, to automatically distinguish the orientations of α_p from those of α_s, all one would have to do is find a characteristic feature within the OIM that discriminates between α_p and α_s. However, Germain et al. [6] found that the OIM and its associated EBSD dataset by itself did not contain enough information to categorically distinguish the α_p and α_s orientations. They instead proposed a technique correlating the OIM with a corresponding back-scattered electron (BSE) image, for which high chemical contrast is observed between α_p and α_s. Once a good match is achieved between the OIM and BSE images, each pixel in the OIM is defined as either α_p and α_s based on the corresponding grey level intensity in the BSE image. This resulted in less than 6% of pixels being wrongly assigned, suggesting that a technique had been identified with a high reliability of describing the true global α_p and α_s textures. This paper documents a similar method that correlates the EBSD data within the OIM with an optical image of the same area processed using relatively simple image analysis. By using an optical image greater contrast can be achieved than that in a BSE image making image correlation much more straightforward and, in turn, more reliable.

2. Description of the method

The procedure for obtaining separate orientation datasets for both α_p and α_s can be split into three steps: 1. Acquisition of orientation data from EBSD to generate the OIM and placement of map location markers, 2. Location of the mapped region using the optical microscope and subsequent image analysis, and 3. Correlation of the OIM with the optical image and separation of the orientation datasets for each microstructural component.

At this stage, it is important to note that the crucial issue in the successful application of the technique is the location of the orientation mapped region in the optical microscope. This can be achieved relatively simply by taking advantage of an artefact of the specimen-electron beam interaction. By etching the specimen immediately after EBSD mapping a large unetched area equivalent to that scanned during OIM acquisition is revealed, an example of which is shown in Fig. 2. This we believe is due to carbon contamination, a well-known phenomenon in electron microscopy [7], and we shall refer to this area as the ‘blank’.

The appearance of the blank is directly related to the step size and the time taken per data point acquisition. When a small step size is used, <0.5 μm, with a long data acquisition time, >0.5 s, the blank is very strong and easily recognisable. However, when a larger step size is used with increased data acquisition rates, typical