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Effects of microstructures on fatigue crack initiation and short crack propagation at room temperature in an advanced disc superalloy

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ABSTRACT

Fatigue crack initiation and early short crack propagation behaviour in two microstructural variants of a recently developed Low Solvus, High Refractory (LSHR) disc superalloy at room temperature has been investigated by three-point bending with replication procedure. The results shows that fine gained (FG) LSHR possesses higher fatigue life due to its better crack initiation resistance, limited crack coalescence and comparable Stage I crack propagation resistance to the coarse grained (CG) LSHR, although its resistance to Stage II crack propagation is inferior. Twin boundary (TB) cracking in the relatively large grains dominates the crack initiation process along with occasional crack initiation due to slip band cracking. Activation of the primary slip systems parallel to the TB at matrix and twin and high resolved shear stress associated with high Schmid factor (SF) are required for TB crack initiation. Cracks preferentially propagate along slip bands associated with high SF slip systems after initiation. But cracks also propagate along slip bands associated with slip systems with lower SF if the inclination angle between the slip band ahead of the crack tip and the crack segment of the crack tip is small enough to enable a steady transition (or non-deflected growth) of cracks across the grain boundary.

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1. Introduction

Aeroengine turbine discs normally operate at elevated temperatures under dynamic loads in an aggressive service environment over significant periods of time. This requires disc materials to possess high strength at elevated temperatures, good fatigue and creep performance under these service conditions, and excellent oxidation and corrosion resistance. Powder metallurgy (PM) Ni-based superalloys have been widely used for aeroengine turbine disc application due to their exceptional combined mechanical properties at elevated temperatures in combination with good oxidation/corrosion resistance [1–3]. Currently however, most of the disc alloys are designed for use at operating temperatures lower than 700 °C, beyond which severe environmental attack accelerated fatigue/creep failure may happen and thereby significantly reduces the service lifetime of the disc alloys [4].

In order to increase the operating temperature of disc alloys beyond 700 °C, to enhance fuel efficiency, produce higher thrust-to-weight ratio and reduce green-house gas emission, significant efforts have been made over past decades to develop more advanced disc alloys with superior tensile strength and fatigue/creep performance at elevated temperatures by optimising alloy composition and controlling the material processing and heat treatment. It is well known that high temperature tensile strength and creep performance can be effectively improved by increasing the amount of γ' forming elements (such as Al and Ti) and refractory elements (such as W and Mo) to enhance precipitation strengthening and solid solution strengthening effects [2]. However, addition of these elements usually results in a high solvus temperature of γ', which makes the supersolvus and dual microstructure heat treatment challenging due to the significant grain growth and increased propensity of quench cracking caused by the greater residual thermal stress [5]. In order to neutralise the increase in γ' solvus temperature, relatively high Co content is used in more recent disc superalloys as it enables lowering of the γ' solvus temperature and introduce an additional strengthening effect by mechanical twinning during deformation, especially at temperatures above 650 °C [6,7]. In fact, the Co content in these more recent disc alloys, such as RR1000, Rene' 104 and LSHR alloy, is about 5–7% higher compared with relatively older disc alloys, such as Udimet 720 and Waspaloy [3,6,8].

The LSHR alloy was recently developed for aeroengine turbine disc application by NASA. It combines the low solvus of Rene' 104 brought about by the high Co content and the high refractory element content of Alloy 10 [9]. Based on existing research carried out at NASA, it has been found that the LSHR alloy possesses exceptional high temperature tensile strength and creep resistance [6]. It is also claimed that the LSHR alloy has good processing

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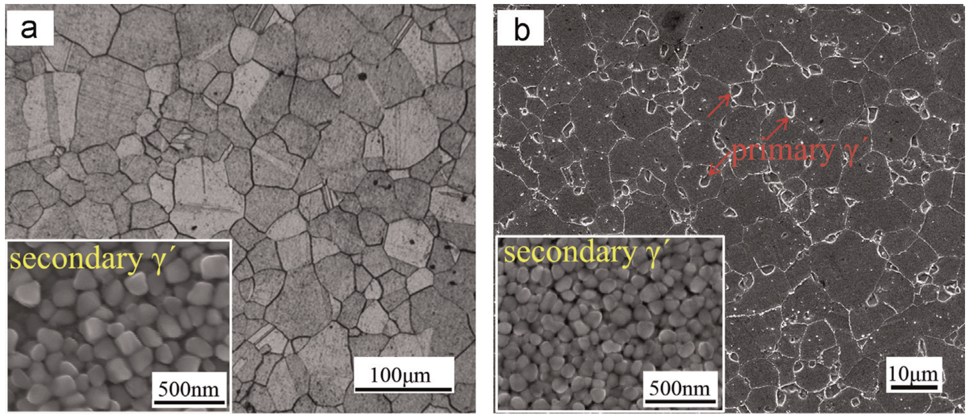

Fig. 1. Microstructures of (a) CG and (b) FG LSHR alloys obtained by supersolvus and subsolvus heat treatments respectively.

versatility due to the low γ' solvus temperature, which makes it possible to produce a dual microstructure turbine disc with optimised creep and fatigue performance at various disc locations by differentiated heat treatments [6,9,10].

Studying low cycle fatigue (LCF) performance at elevated temperatures indicates that the typical fatigue-initiation sites and the overall fatigue life of LSHR alloy are related to the microstructures, test temperatures and strain ranges employed [6,10]. Specifically, cracks mainly initiate from internal inclusions and/or pores and occasionally initiate from crystallographic facets in FG LSHR at elevated temperatures (427 °C and 704 °C). This has been observed when the applied total strain is less than 0.8% (which is usually associated with longer fatigue life). Whereas cracks predominantly initiate from crystallographic facets of larger grains in the CG LSHR variant, which usually produces a shorter fatigue life at similar moderate strain ranges [10]. When higher strain ranges were applied by conducting LCF tests at higher temperature (704 °C), an oxidation assisted crack initiation process was observed to come into effect, usually forming intergranular cracks, especially when a long dwell time was applied at the peak time during LCF tests [6]. The short crack propagation behaviour after initiation which is sensitive to the local microstructure adjacent to the crack tips was not investigated in these studies, and the intrinsic crack initiation process (in the absence of environmental attack) that is linked to the cyclic deformation processes at lower temperatures in the LSHR alloy has also not received much attention. It is generally accepted that fatigue crack initiation and short crack growth processes are important to optimise as they contribute to the majority of fatigue life of a turbine disc during service. This is due to the high overall component stresses which result in a relatively small extent of fatigue crack propagation prior to fast fracture and thereby limit the fatigue life dependency to the short crack growth regime [11–15]. As a result, a systematic assessment of crack initiation and subsequent short crack growth behaviour is necessary at both lower and elevated temperatures in order to evaluate the intrinsic (without environmental attack) and extrinsic (with environmental attack) fatigue crack processes in appropriate microstructural variants of the LSHR alloy. This is expected to provide a better understanding of the fatigue crack initiation and propagation processes and to contribute to the ongoing development of optimised disc alloys. It should also be noted that the bore region of a turbine disc will experience lower temperatures in service and optimisation of turbine alloys requires good fatigue resistance at both low and high temperatures. In this paper, crack initiation and subsequent short crack propagation behaviour in LSHR alloy at room temperature was investigated, and the effects of grain size, grain orientation and primary γ' precipitate distribution have been studied and are discussed. A companion study on crack initiation and short crack propagation in LSHR alloy at elevated temperatures will be presented in another paper.

2. Materials and experimental procedures

2.1. Materials

The LSHR alloy used in this study was provided by NASA. Composition (in wt%) of the LSHR alloy is 12.5Cr, 20.7Co, 2.7Mo, 3.5Ti, 3.5Al, 0.03C, 0.03B, 4.3W, 0.05Zr, 1.6Ta, 1.5Nb, Ni bal. Specimens used for the short crack tests were extracted from a turbine disc which was fabricated by canning atomised LSHR alloy powder followed by hot isostatically pressing, extruding and isothermally forging. The extracted specimens were supersolvus heat treated at 1171 °C and subsolvus heat treated at 1135 °C to yield CG and FG microstructures respectively, followed by the same dual aging heat treatments [4]. The obtained microstructures are shown in Fig. 1, and the measured grain size and γ' size are shown in Table 1. The details of microstructure evaluation can be found in our previous publication [4].

2.2. Experimental procedures

Fatigue tests were conducted on polished U-notch CG and FG LSHR specimens under three-point bend loading on an Instron 8501 hydraulic testing machine at room temperature with a 20 Hz sine waveform and a load ratio of 0.1. The applied load was calculated to produce a maximum nominal elastic stress of 1020 MPa at the notch root using simple elastic beam theory for the uncracked ligament. The dimension of the U-notch specimen and the position of the loading rollers are shown in Fig. 2. The notch has a depth of 1.25 mm with a curvature radius of 2 mm. This notch type was chosen to provide an elastic stress concentration of around 2, i.e. representative of that seen in the fir tree root fixings used to secure blades to turbine discs. The notch was ground and then polished using dental felts by 1 μm diamond polishing paste before testing. Some of the tests were interrupted at certain intervals to make a replica of the notch root surface with a silicone compound (provided by Struers Ltd) to monitor crack evolution. Some

Table 1
Statistical data on size of grain, primary γ' and secondary γ' in LSHR alloy.

Materials	Grain size (μm)	Primary γ' (μm)	Secondary γ' (nm)
CG LSHR	38.38 ± 18.07	N/A	153 ± 29
FG LSHR	8.14 ± 2.77	1.74 ± 0.48	89 ± 15

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