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Micro-tensile strength of a welded turbine disc superalloy

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ABSTRACT

A micro-tensile testing system coupled with focussed ion beam (FIB) machining was used to characterise the micro-mechanical properties of the weld from a turbine disc alloy. The strength variations between the weld and the base alloy are rationalised via the microstructure obtained.

1. Introduction

The development of micro-scale experiments has been initiated by the need to evaluate the mechanical behaviour of small volumes of materials and also by a desire to determine how the mechanical properties of a material change when external dimensions are greatly reduced [1–5]. Micro-tensile testing has been effectively used to characterise the mechanical properties of thin films, e.g. of MEMS.

The methods often used in preparing micro-tensile specimens include material deposition on substrate (additive process) [6–8], deep reactive ion etching (subtractive process) [9] and electro-discharge machining [10]. All these methods have been successfully used, but cannot be adopted for applications where a particular site of interest within a large volume of material is needed to be characterised. Focussed ion beam (FIB) is an invaluable tool for fabricating micron-sized structures, due to its ability to deposit Pt or W in controlled shapes, the availability of submicron (20–50 nm) ion beams, 3-D stages, and fully automated control [11].

Aside from the high accuracy in local positioning through direct visual control, in situ micro-tensile testing also offers insight into the real time material deformation process.

Micro-tensile tests have been carried out to characterise the local properties within fusion welds of HY-100 steel [10,12] and of two dissimilar stainless steels [13]. The strength at the centre of the weld was found to be considerably higher than that away from the centre. However, the dimensions of these microsamples ($300 \times 200 \times 200\ \mu m^3$) are larger than many welds produced by solid state techniques. For instance, inertia friction welding often produces much narrower welds with a bond line zone about 50 μm wide compared with the 1300 μm of typical fusion welds. Solid state joining is increasingly used in the aeroengine industry for multicomponent turbine engines. Detailed material characterisation within the narrow weld zone, where microstructural variation from the parent material has occurred, is essential to understand the overall mechanical performance of welded joints. The hardness of an inertia friction welded (IFW) RR1000 superalloy has been reported to be higher than that of the parent material [14]. As a result the true yield stress of the weld zone cannot be measured by standard tensile testing procedures [15].

This paper reports in-situ micro-tensile deformation of IFW RR1000, focussing on the yielding.

2. Materials and methods

RR1000 is a recently developed (Rolls Royce) nickel base superalloy processed via powder metallurgy with a nominal chemical composition of (wt%) 15.0 Cr, 18.5 Co, 5.0 Mo, 3.0 Al, 3.6 Ti, 2.0 Ta, 0.5 Hf, 0.015 B, 0.06 Zr, 0.027 C and balance nickel.

A thin cross-section slice was extracted from an inertia welded tube in RR1000 by electro-discharge machining (EDM). It was ground and thinned before a wedge shaped sample, which contained both the parent and the weld regions, was cut out (Fig. 1). The wedge shaped sample was further ground and polished to a thickness of about 100 μm. Micro-tensile samples with a 13 μm gauge length and 2 μm by 3 μm cross section were prepared from the wedge shape sample using a Quanta 3D FEG SEM FIB with a Ga^+ ion source operated at 30 kV.

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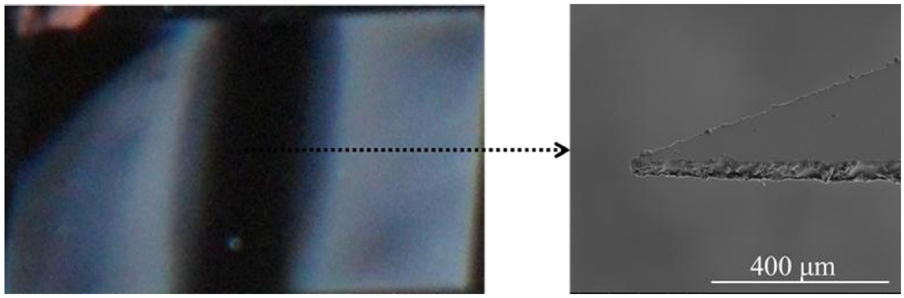

Fig. 1. A wedge shaped specimen (taken from the dark/etched (weld region) of the sample (mounted in conductive bakelite)) used in FIB for micro-tensile sample preparation.

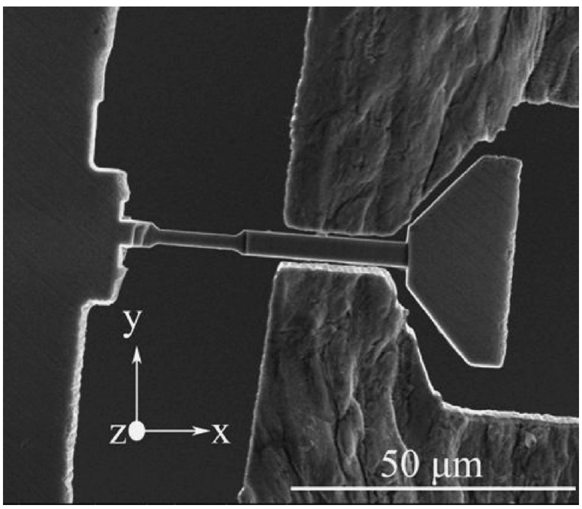

Fig. 2. SEM picture of a micro-tensile sample ($13 \times 2 \times 3\ \mu m^3$) prepared from the base alloy (Parent 1) and fitted into the sample grip before loading.

The microtesting system used in this study includes a piezo-electric drive to apply the load to the sample via a miniature load cell with a load capacity of 0.5 N and a resolution of 0.001 mN. The tensile system was under displacement control at a resolution of 40 nm. The applied load and elongation were measured, and used to plot a stress–strain curve, together with the recorded SEM images.

Electron back scatter diffraction (EBSD) was employed to determine the loading direction and the active slip systems, in order to evaluate the critical resolved shear stresses.

3. Results and discussion

Fig. 2 shows a micro-tensile specimen prepared from the parent RR1000. This specimen was tested by pulling to fracture. Fig. 3b is the stress vs. strain plot obtained from the tensile test. From both the in-situ observation of the formation of slip and the stress–strain results of this test, the yield strength of the alloy and the ultimate tensile strength were determined to be 619 MPa and 699 MPa, respectively.

The sample loading direction $[\bar{2}00]$ was determined using the EBSD Kikuchi pattern from the grain in the sample gauge length as illustrated in Fig. 3d. The slip plane was determined to be $(\bar{1}11)$ from the orientation relationship between the slip band on the sample and the Kikuchi pattern (plane trace analysis). Since slip direction is on slip plane, $(hkl)\,[uvw]=0$ was used to determine the corresponding slip direction $[\bar{1}\bar{1}0]$. Although more than one direction can satisfy this condition and since it is also difficult to determine the actual slip plane and direction by using only the EBSD Kikuchi pattern, the slip system that gives the highest Schmid's factor has been selected for the plastic deformation of the sample. The selected plane was validated by measuring the angle between the slip band and a plane on the Kikuchi pattern. Table 1 shows the schmid factors calculated for the twelve easy slip systems in the crystal structure. Critical resolved shear stress, τ_c calculated for the sample of about 253 MPa can be estimated via the Schmid equation.

Another tensile sample prepared from the parent material with loading direction $[\bar{5}30]$ was tested, and the sample was observed to start yielding at a stress level of 530 MPa by the activation of the $(\bar{1}11)$ $[\bar{1}0\bar{1}]$ slip system. A τ_c of 255 MPa can be derived, which agrees well with the value obtained for the previous sample.

Fig. 4 shows a micro-tensile sample prepared from the bond line of the weld with loading direction $[3\bar{6}\bar{1}]$. It was observed that the sample yielded at a stress level of 604 MPa (Fig. 4b and c) via the activation of two slip systems $(1\bar{1}\bar{1})[01\bar{1}]$ and $(1\bar{1}1)[011]$ (Fig. 4e) and a critical resolved shear stress of 300 MPa can be estimated.

Another sample prepared from the bond line zone with a $[13\bar{3}]$ loading direction was also tested. This sample yielded at 713 MPa on $(1\bar{1}1)[110]$ with a critical resolved shear stress of 306 MPa (Table 2). This consistency in the critical resolved shear stress measured indicates that micro-tensile testing is viable for testing local strength.

Another sample was prepared from the weld zone where constitutional liquation features were observed on the grain boundaries (Fig. 5a). EBSD shows that the grain boundary between grains A and B, and that between B and C, are typical high angle grain boundaries. It was observed that the sample started yielding in grain A, the slip bands then propagated into grains B and C, and the sample subsequently failed along the liquated grain boundary between B and C (Fig. 5h). This suggests that the high angle boundary, where a significant constitutional liquation product was observed, was weakened. Table 2 shows a summary of the tensile data from the experiment.

It is generally expected that the yield strength within the weld of a γ' strengthened nickel base alloy should be greater than in the base alloy. This is because the re-precipitation of a high volume fraction of tertiary γ' in the size range 10–40 nm at the weld increases the strength [14,16]. The size and distribution of γ' of the weld and the parent are compared in Fig. 6.

It is evident from Table 2 that the critical resolved shear stress in the weld is indeed higher than that in the base alloy. Although there is a slight uncertainty in the alignment of the length axis of the sample and of the sample holder in the *z* direction, efforts were made to ensure that the flanks of the sample holder and of the sample head are parallel and equally spaced in order to minimise out-of-plane loading of the specimens, as illustrated in Fig. 2. The measured yield strength in the micro-tensile samples is lower ($\sim$30%) than the expected yield strength in the bulk material. This could be a result of the size with crystallographic orientation of the micro-tensile sample [17–19].

Based on the premise that the deformation of γ' precipitation hardened nickel alloys at room temperature is via two coupled