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High-temperature crack growth in a Ni-base superalloy during sustained load

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ABSTRACT

The high-temperature sustained load crack growth behaviour of a Ni-base superalloy was investigated using a combination of mechanical testing in controlled atmosphere, fractographical and microanalytical investigations, and finite element modelling. The results show that the local crack front geometry is uneven on two scales – jaggedness on the scale of 100 μm was observed in all specimens, whereas mm-scale waviness could occasionally be observed. The jaggedness can be explained by a percolation-type crack growth along weaker grain boundaries, whereas the large-scale waviness is presumably due to larger regions of the material having specific grain texture with high crack growth resistance. The uneven crack front is shown to potentially have considerable effects on the loading conditions at the crack tip, whereas ligaments of un-cracked material in the crack wake are deemed to have less effect on the crack tip loading due to their low area fraction. The ligaments fail intergranularly in the wake as the crack grows in the present case, as opposed to by creep fracture as previously proposed. Finally, the plastically deformed regions about the crack and crack tip are shown not to exhibit any elevated oxygen levels, implying that the damage in these regions is purely mechanical.

1. Introduction

Ni-base superalloys are well documented to exhibit time-dependent intergranular crack growth when subjected to sustained mechanical load at high temperatures in oxygen-rich environments [1–8]. Several extensive summaries on this topic have recently been compiled [9–11]. The most prominent mechanisms currently proposed are strain-assisted grain boundary oxidation (SAGBO) [12] and dynamic embrittlement (DE) [4]. The concept of DE is based on the reduction in cohesive strength of grain boundaries, resulting from stress-driven diffusion of an embrittling species into the boundary. The theoretical treatment of DE is today rather well developed [9], and the phenomenon is often invoked as an explanation to elemental embrittlement caused by both internal and external species. SAGBO, on the other hand, is a result of the formation of an intergranular oxide that cracks under the applied load. Some recent results have indicated that SAGBO, occurring on micron- or sub-micron scale, may indeed be a viable contributor to the environmental cracking of Ni-base superalloys [13,14], leading to the development of an emerging theoretical model [15]. In this paper, the term sustained load crack growth (SLCG) is used as an unbiased expression.

Time-dependent cracking generally occurs by (quasi-)brittle intergranular propagation. Bika et al. [16] reported that the crack front during SLCG in a sulphur-embrittled steel was jagged, with ligaments of un-cracked material in the wake of the main crack front. This was also later observed in a Ni-base superalloy in the work of Pfaendtner et al. [4] where it was proposed that the varying resistance to cracking due to grain boundary structure was the reason. The more susceptible grain boundaries will crack, leading to a locally propagating crack front between the more resistant boundaries. These resistant boundaries are then fully surrounded by the crack during the advancement of the crack front, and are left as ligaments in the wake. The ligaments and the lagging parts of the crack front were proposed to carry much of the applied load, while undergoing power-law creep due to the high stresses and temperatures [4]. Thus, the susceptible boundaries at the leading part of the crack will crack more slowly as the local stress intensities are lowered by the ligaments.

In accordance with this idea, it has been shown that grain boundary engineering by means of thermo-mechanical processing could be used to decrease the SCLG rate through the introduction of an increased fraction of resistant grain boundaries [17,18]. The slower crack growth was accompanied by an increase in the area fraction of ligaments in the crack wake. This effect was mainly

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attributed to the occurrence of special Σ3 boundaries [19,20], which have been shown to provide much more resistance to crack growth than random high-angle boundaries or other special boundaries (Σ19 and Σ31) [19]. The Σ3 boundaries were, however, not fully immune to cracking, but required significantly higher stress intensity factors to initiate the propagation, and therefore the total crack growth rate was lower. The effects of local crack front geometry and remaining ligaments have also recently become important from a life prediction perspective, as models are being developed based on the existence of a "damaged zone" where such features are implicitly controlling the crack growth rate [21–23].

In another aspect, although SLCG is largely considered as brittle, deformation ahead of the crack and in the crack flank has been reported [7,24]. Studies on the correlation between deformation and oxygen/oxidation for the considered conditions are scarce, although previous work on the interplay between slip bands and grain boundaries during SLCG is available [25]. However, such correlations have been shown to exist for stress corrosion cracking (SCC) in austenitic alloys [26,27] using modern analysis methods. Recent findings on SLCG [14,24] by the authors show that there are strong generic similarities between SCC and SLCG, although SCC concerns lower temperatures, far longer exposure times and water-bearing environments. The above clearly condone studies of oxygen in relation to plastic deformation and the "damaged zone" concept also for SLCG.

In this work we study the high-temperature SLCG in a Ni-base superalloy in order to highlight some additional aspects of environmentally assisted cracking. The investigation was performed on the Ni-base superalloy Allvac 718Plus, which was heat treated to remove grain boundary precipitates in order to emphasise the environmental effects. It will be shown here that there is no observable oxygen diffusion ahead of the crack tip, or in the plastically deformed regions around the crack. The ligaments observed in the crack wake have a minor effect on the crack tip loading in the present case due to their low number. They subsequently fracture by brittle oxygen-induced failure as the stresses in the ligaments increase during crack growth, rather than through creep as has been previously proposed. On the other hand, geometrical variations along the crack front resulting from the crack growth process will be shown to potentially have a large influence on the local loading situation at different parts of the crack. This has not been considered in previous studies of sustained load or dwell-fatigue crack growth, and it further adds to the complexity of the dynamic environmental embrittlement phenomenon.

2. Experimental

2.1. Materials and heat treatment

The material used in the investigation was the Ni-base superalloy Allvac 718Plus (wrought bar). The nominal composition of the alloy is given in Table 1. In order to simplify the oxygen interaction at the grain boundaries, the alloy was heat treated to produce grain boundaries free of δ or η phase precipitates. Such a condition has been found to be more susceptible to dwell-time crack growth [8]. To achieve this the following heat treatment was

Table 1
Nominal composition of the Allvac 718Plus alloy (wt%).

Element	Ni	Cr	Fe	Co	Nb	Mo	Al	W	Ti	C	P	B
wt%	Bal.	18	9.5	9.1	5.4	2.7	1.45	1.0	0.75	0.02	0.006	0.005

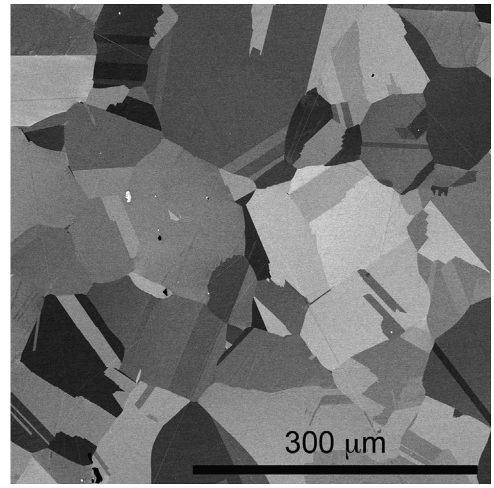

Fig. 1. SEM image of the investigated microstructure of Allvac 718Plus. The grain boundaries are free of precipitates. Extensive twinning from heat treatment is clearly visible, as are some Nb and Ti carbides (bright and dark, respectively).

performed: 1040 °C/1 h (WQ)+788 °C/8 h (FC)+704 °C/8 h (FC), where WQ and FC refer to the water quenching and the furnace cooling, respectively. The first step dissolves the grain boundary phases along with existing hardening precipitates, whereas the subsequent steps correspond to the standard ageing treatment. Fig. 1 shows the as heat-treated microstructure, with no grain boundary phases but with annealing twins and occasional Nb and Ti primary carbides. The hardness was measured to 429 HV10kg, which is in line with what is expected of Allvac 718Plus in the age-hardened condition.

2.2. Mechanical testing

Single-edge notched (SEN) mechanical testing specimens were machined to the geometry shown in Fig. 2(a). A rectangular specimen waist cross-section of 8 × 3 mm was used where a 0.2 mm deep through-thickness notch was produced on one of the 3 mm sides using electric discharge machining (EDM). Mechanical testing was performed using a MTS model 661 hydraulic rig with a custom designed environmental chamber, as shown in Fig. 2(b). The chamber allowed a controlled environment to be maintained while performing direct current potential drop (DCPD) monitoring of crack growth and induction heating of the specimen.

The specimens were pre-cracked in two stages, first at room temperature (RT) and then at 700 °C, as shown schematically in Fig. 2(c). The pre-crack load sequence was designed such that the size of the plastic zone at the crack tip would have minimal influence on the high-temperature sustained load cycle. Thus, pre-crack loads (10 Hz triangular waveform with a load ratio R=0.05) were reduced incrementally to yield a final pre-crack plastic zone of similar size as for the start of the sustained load testing. Using this approach, pre-cracks were first grown at RT to an approximate length of 2.5 mm (step 1 in Fig. 2(c)) and then additionally around 0.5 mm at 700 °C (step 3).

Following pre-cracking two different load sequences, referred to as Type 1 and Type 2, were applied. For both test types a single 600 s sustained load was applied at 700 °C using a load of 3 kN (125 Mpa nominal stress) in a controlled environment of $^{16}O_2$ and $^{18}O_2$ at 1:1 volume ratio (step 4). Upon termination of the sustained load, the environment was immediately changed from the oxygen mix to forced convection cooling in Ar while at the same time turning off the induction heating and releasing the mechanical load to zero (step 5). The sample temperature was registered < 500 °C within 15 s and approximations by finite