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EFFECT OF SO₂ AND WATER VAPOUR ON THE LOW-CYCLE FATIGUE PROPERTIES OF NICKEL-BASE SUPERALLOYS AT ELEVATED TEMPERATURE

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Abstract:

In this study the effect of SO₂ + water vapour on strain controlled low cycle fatigue resistance of three different nickel based superalloys has been studied at 450°C and 550°C. A negative effect was found on both the crack initiation and crack propagation process. The effect increases with increasing temperature and is likely to be influenced by both the chemical composition and the grain size of the material. In general the negative effect decreases with decreasing strain range even if this means that the total exposure time increases. This is explained by the importance of the protective oxide scale on the specimen surface, which is more likely to crack when the strain range increases. When the oxide scale cracks, preferably at the grain boundaries, oxidation can proceed into the material, causing preferable crack initiation sites and reduced fatigue resistance.

Keywords: Nickel based superalloys, Low cycle fatigue, Environmental effect, Sulphur, Water vapour

1 Introduction

Today, the applicability of fine grain polycrystalline nickel base superalloys in gas turbine applications is often limited by their susceptibility to fast intergranular cracking during fatigue in combination with extended hold times at high temperatures and high tensile stresses.

However, the time dependent intergranular cracking of nickel based superalloys, under both sustained and cyclic loads, is generally believed to be due to environmental interactions at the crack tip [1-3]. The main damaging species that is always present is oxygen, but other more aggressive species most likely also have a significant impact on the service life of nickel base components [4,5]. For instance, the typical environment for gas turbine components could contain varying amounts of moisture, sulphur (in industrial environments) and chlorine (in coastal environments). Furthermore, in future gas turbine designs, one must be prepared for an increased variety of fuels. Alternative gas sources, such as syngas or biogas produced from landfill gas or digester gas contain significant amount of H_2S , which on combustion will transform to SO_2 and water vapour. Thus, the future need of conversion from fossil-based fuels will lead to more chemically aggressive environments for the materials and, consequently, a better understanding of the environmental influence on service lives of critical components is needed.

In many cases, the environments that increase crack growth rates during fatigue loading do not produce significant general attack in an unstressed material and thus conventional high temperature corrosion tests may not be useful for predicting the behaviour of components in service [6,7].

The negative influence of oxygen exposure during LCF was illustrated by Woodford et al. during the 70'ties when the response of IN706 to testing at a constant plastic strain range and

frequency in air and vacuum was investigated [8]. For testing in air a pronounced minimum in life were found at temperatures around 700°C while very little effect on life in vacuum tests over the entire temperature range were noticed. Another conclusion from this study is that the environmental influence may extend to a temperature as low as 400°C even without introducing hold times.

Compared to lab experiments the service conditions of typical gas turbine components involve lower cyclic frequencies and much longer hold times under load and elevated temperature. The impact of frequency and environment is illustrated by the results of Chang et al. [9] for Inconel 718. In vacuum and in air at 400°C the alloy show pure cycle dependent crack propagation at all frequencies, but with increasing temperature and decreased frequency the crack propagation rate in air becomes more and more time dependent. The importance of the environment can also be seen during sustained load crack propagation as shown for Inconel 718 in the study Floreen et al. [10] where the time dependent crack growth rate at 650°C was ~100 times faster in air than in helium. The processes that typically contribute to time dependent crack propagation are thermally activated and therefore become increasingly important with increasing temperature and when the time-dependency increases the mode of fracture is observed to changes from transcrystalline to intercrystalline.

Much of the early work assumed that the time dependence originated from the nature of deformation at the crack tip and the term creep fatigue was applied but the dominant role of the environment is now well established [1,2] and two dominating theories to explain this behaviour can be found: stress accelerated grain boundary oxidation (SAGBO) and dynamic embrittlement (DE). The SAGBO process involves oxidation of grain boundaries ahead of the crack tip and subsequent cracking of the oxide, exposing new surfaces to the oxygen. The DE

theory on the other hand advocates embrittlement of the grain boundary by oxygen diffusion, separation of the embrittled boundaries and subsequent oxidation of the fresh surfaces. DE requires oxygen diffusion over very short distances.

It has also been shown that water vapour content can promote time dependent cracking in Inconel 718 independent of the oxygen. Valerio et al. [11] found that the crack propagation rate is increased by two orders of magnitude in a mixture of water vapour and argon (without any O_2) compared to dry argon. They identified both oxygen and hydrogen as embrittling species and found a higher concentration of niobium at the grain boundaries. Also sulphur and chlorine might lead to more rapid embrittlement and intergranular cracking at elevated temperatures as demonstrated by Woodford et al. [8]. When exposing pure nickel (Ni 270) at 800°C in chlorine and sulphur, the tensile ductility as a function of test temperature decreased significantly compared to exposure in air. Whereas there was no measurable oxygen embrittlement for this exposure, the entire specimen cross sections were embrittled and showed intergranular fracture at room temperature for both the chlorine and sulphur penetration. Grain boundary fracture following sulphur exposure indicated decohesion at smooth interfaces with no evidence of chemical reactions. This appearance is quite different from the intergranular fracture shown for air exposure. The lowest temperature of sulphur embrittlement, 450°C, is well below the melting temperature of any known nickel sulphide. It is therefore believed that the embrittlement is due to elemental diffusion of sulphur down the grain boundaries. Other chalcogens (Se, Te) and the halogens (F, Cl, Br, I) as well as metal vapours (Pb, Bi) may undoubtedly also lead to gas phase embrittlement [2].

Even if the negative effect of Niobium on the crack growth rates in air has been illustrated by Gao et al. [12] systematic studies ranking different Nickel-based alloys are rarely reported.

Furthermore, the mechanisms discussed above mainly concerns crack propagation and very few studies are unfortunately devoted to the environmental effects on the crack initiation process. The goal of the current project is therefore to give a better understanding of the influence of environmental impact on nickel base superalloys during low cycle fatigue (LCF) in combination with extended hold times at elevated temperatures and high tensile stresses. The mechanisms and sensitivity for embrittlement and intergranular cracking due to SO_2 + water vapour will be studied for the currently widely used alloys, Nimonic 901 and Inconel 718, as well as the future candidate alloy 718Plus.

2 Experimental procedures

The materials studied in the present work are materials used in, or intended for, gas turbine disc applications. The materials represent 3 different generations of Nickel-based disc alloys typically seen in small and medium size gas turbines. Nimonic 901 represents the first generation and is today mainly replaced by Inconel 718 while Alloy 718plus is a rather recent alloy developed by ATI Allvac to be a lower cost alloy capable of use at higher temperatures than Inconel 718. The materials and their chemical composition are listed in Table 1. All materials tested originate from real disc forgings supplied by Aubert & Duval and as such they have a processing and a microstructure representative for disc applications. Even if the forgings are not exactly the same for all three materials, the size of the forgings are comparable, 200-300 kg. The heat treatment for each alloy is summarized in Table 2.

Table 1 Nominal composition in wt% of the investigated wrought Ni-base Superalloys

Alloy	Ni	Cr	Co	Mo	W	Nb	Al	Ti	Fe	Mn	Si	C	B
Nimonic 901 (UNS N09901)	78.5	12.5		5.75			0.35	2.9				0.05	
Inconel 718 (UNS N07718)	52.5	19		3		5.4	0.5	0.9	18.5	0.2	0.2	0.04	
Alloy 718Plus (UNS N07818)	51.5	18	9	2.7	1	5.45	1.45	0.7	10			0.025	0.004

Table 2 Heat treatment applied to each material prior to testing

Alloy	Heat treatment
Nimonic 901	Solution heat treatment at 1090°C for 3 hours, first ageing at 775°C for 4 hours, second ageing at 720°C for 24 hours
Inconel 718	Solution heat treatment at 970°C for 3.5 hours, first ageing at 720°C for 8 hours, second ageing at 620°C for 8 hours
Alloy 718Plus	Close die forging at 1058°C, Presolution at 870°C, Solution heat treatment at 954°C, first ageing at 788°C for 8 hours, second ageing at 704°C for 8 hours

The grain size is significantly larger ($\sim 100\mu\text{m}$) in Nimonic 901 compared to Inconel 718 ($\sim 10\mu\text{m}$). Alloy 718Plus in the condition tested in this study has a grain size somewhere in between the other two alloys ($\sim 30\mu\text{m}$). Sometimes disc forgings can display anisotropic material behaviour. Therefore all test specimens are orientated in the tangential direction relative to the disc forging for all materials.

The LCF test equipment has been specially modified to allow for testing in controlled environments. The base is an Instron 1380 load frame equipped with a 30 kN load cell and a SFL SF892A split furnace. Normal tests in air are run with both an internal and an external

extensometer, whereas tests in controlled atmosphere are performed with only the external extensometer. The external extensometer measures the relative displacement of the two pull rods just outside the furnace. A test series in controlled atmosphere was therefore always accompanied by a corresponding test series in air that was used to calibrate the extensometers. A schematic drawing of the setup for testing in controlled environments is given in Figure 1. To make testing in controlled environments possible, the internal extensometers are removed and three quartz glass tubes are inserted. The outermost tube is a snug fit against the furnace and sealed by cooled stainless steel collars at both ends. The top collar also contains a stainless steel bellow. During testing this tube is filled with argon under a slight overpressure. The two innermost tubes are sealed by glass-fibre padding at both ends and the aggressive gas is led into the innermost tube through a small diameter quartz tube. Between the inner and the middle tube a chamber is formed and the gas outlet is placed here. Since both the argon and the aggressive gas are introduced under slight overpressure, the excess gas seeps into this chamber through the glass-fibre padding and it is then led out of the test setup through a quartz tube.

The experimental set-up thus makes it possible to conduct strain-controlled mechanical testing in a well-controlled environment. At least 4 tests with different strain ranges were conducted for each combination of material, temperature and environment. The strain range was varied in order to achieve life times between 100 and 6000 cycles to failure, the latter number is a typical requirement for turbine discs. Most specimens were allowed to rupture and the numbers of cycles to final failure (N_f) were noted for each specimen. The strain rate used was 6%/min and a dwell time of 5 min at the maximum strain was applied in all tests. The strain ratio was $R = \epsilon_{\min} / \epsilon_{\max} = 0$. A stable hysteresis loop taken at approximately midlife was also evaluated for each test in order to determine the inelastic strain range as well as the

stress range. For all tests, no or very limited creep relaxation was noted during the dwell time, the intention of the dwell time is however to expose the material to the combined effect of mechanical loads and a corrosive environment.

Tests were conducted in both in air and in SO₂ + water vapour. The later environment was achieved by mixing;

- Gas with 25% SO₂ and 75% N₂
- Humidified air (RH = 70% at 90°C)

leading to a resulting SO₂ content in the test environment of 3% and a dew point temperature of approximately 80°C. The effect of this environment on the LCF life was tested for two different temperatures; 450°C and 550°C. During testing at elevated temperature the added water vapour correspond to a relative humidity of approximately 0.05% at 450°C and 0.02% at 550°C.

In order to study the deformation and fracture appearance the tested specimens were analysed using a Jeol JSM-6610LV scanning electron microscope. The main fracture surfaces were analysed “as tested” while secondary cracks were analysed in cross sections perpendicular to the longitudinal axis. The cross section samples were prepared by grinding and mechanical polishing. All elemental maps were produced by EDS measurements.

3 Results

3.1 Mechanical testing

3.1.1 Nimonic 901

The results from the strain controlled LCF testing can be found in Figure 2, where Figure 2(a) shows the total strain range versus cycles to failure while Figure 2(b) shows the corresponding plastic strain range for each test as a function of cycles to failure. At 450°C, very little effect of the environment can be seen and this holds for the total strain range versus cycles to failure as well as for the plastic strain range versus cycles to failure. However, when LCF testing is performed at 550°C in an environment containing SO₂ + water vapour, a significant decrease in LCF life compared to testing in air is found. Furthermore it should be noticed that the environmental impact is more significant for high strain ranges but as the strain range decreases the difference compared to tests performed in air also decreases. This observation holds for both the total strain range versus cycles to failure in Figure 2(a) and the plastic strain range as a function of cycles to failure in Figure 2(b). When the data for plastic strain range versus cycles to failure for all tests are compared, see Figure 2(b), one interesting observation is that the data from all tests at 450°C and the tests in air at 550°C all fall on approximately the same trend line, which indicates that the main failure mechanism is the same in all these tests. The tests in SO₂ + water vapour at 550°C on the other hand show lower LCF life for equivalent plastic strain ranges, which indicates that another failure mechanism is active under these conditions.

3.1.2 Inconel 718

The mechanical strain range versus the number of cycles to failure for all tests on Inconel 718 are plotted in Figure 3(a). One can see a reduction in LCF life with increasing temperature for

tests in both air and in SO₂ + water vapour. Furthermore, a significantly lower LCF life for tests in SO₂ + water vapour compared to tests in air can also be noticed.

In Figure 3(b) the plastic strain range versus the number of cycles to failure are plotted for all tests on Inconel 718. One can see that the plastic strain range at both 450°C and 550°C is significantly reduced in an environment containing SO₂ + water vapour compared to tests in air. For the tests performed in air one can also see a tendency for lower resistance to plastic deformation at 550°C compared to tests at 450°C.

3.1.3 Alloy 718Plus

The mechanical strain ranges versus the number of cycles to failure for all tests on alloy 718Plus are plotted in Figure 4(a). As for Inconel 718, one can see a lower LCF life for tests in SO₂ + water vapour compared to tests in air. This holds for both temperatures but is less pronounced as the number of cycles to failure increases.

In Figure 4(b) the plastic strain range versus the number of cycles to failure are plotted for all tests. One can see that the plastic strain range at both 450°C and 550°C is reduced in an environment containing SO₂ + water vapour compared to tests in air. The reduction at 450°C is however lower compared to that exhibited by Inconel 718. Unlike Inconel 718 the test performed in air at 550°C seems to resist the same amount of plastic strain range as the test performed at 450°C.

3.2 Scanning electron microscopy

3.2.1 Nimonic 901

At 450°C, independent on the environment, mainly transcrystalline cracks have been observed, as illustrated in Figure 5. However, when the temperature is increased to 550°C, the

test performed in SO₂ + water vapour shows a different behaviour with pronounced intercrystalline fracture appearance as illustrated in Figure 6.

3.2.2 Inconel 718

The fatigue fracture appearance in air at 450°C is more or less completely transcrystalline as seen on the fracture surface in Figure 7(a). However, at 550°C there is a tendency to also have partly intercrystalline fatigue fracture surfaces even in air, as seen in Figure 7(b). Figure 8 shows a cross-section of an Inconel 718 sample tested in air at 550°C. The secondary crack has a partly intercrystalline crack path and no significant amount of oxidation product can be seen in the crack, see e.g. the crack tip in Figure 8(b).

The fracture surface appearance when Inconel 718 is tested in SO₂ + water vapour is illustrated in Figure 9a, where one can observe alternately darker and brighter areas. In higher magnification it can be seen that the darker areas of the propagation zone have been worn against the opposing fracture surface, see Figure 9(b). The fracture surfaces are oxidized to a varying degree and it is therefore not possible to observe any distinct grain contours. However, there is a pattern of rounded topographic features of the same size as the grains indicating that intercrystalline cracking might have occurred.

In the cross section in Figure 10 it can be seen that the secondary cracks of samples tested in SO₂ + water vapour at 550°C are filled with corrosion products and the fracture surfaces have an oxide layer. It is interesting to see that corrosion is completely localized to the cracks while there is only a very thin oxide layer on the surface of the test specimen in between the secondary cracks. The longer secondary cracks, which are several hundred microns in length, have a meandering propagation path, which seems to have followed grain boundaries. However, the heavy corrosion of the cracks makes it difficult to determine the precise

propagation path of the original crack and its relation to surrounding grains. In Figure 11 one can see a very short secondary crack which is also very much affected by oxidation. In this study, this type of oxidation pit typically seems to be located within grain boundaries.

EDS mapping performed on a sample tested in SO_2 + water vapour at 550°C can be seen in Figure 12. The EDS analysis did not indicate any significant oxygen penetration ahead of the crack tip. However, the crack penetrating along the grain boundary is clearly filled with oxides, which seems to have a layered structure. Close to the main material the oxide is rich in Nb and sulphur while the centre has higher concentrations of Fe and Ni. The Ni also seems to diffuse preferably outward towards the surface where it forms a nodular like corrosion product.

3.2.3 Alloy 718Plus

In general, the fracture behaviour of alloy 718Plus seems to be more similar to that seen in Inconel 718 than in Nimonic 901. In Figure 13 one can see a “finger” like appearance on the fracture surface with worn and unworn areas, similar to what was seen for Inconel 718. The fracture surfaces on specimens tested at 550°C are heavily oxidized but for material tested at 450°C clear signs of crack branching can be seen and the topographic features are approximately of the same size as the grains, which indicate intercrystalline cracking.

4 Discussion

From the strain controlled experiments performed in this study it is clear that the presence of SO_2 + water vapour has a negative effect on the low cycle fatigue resistance of wrought nickel based superalloys. The environmental effect is typically manifested by lower resistance to cyclic plastic deformation. This can be seen for Nimonic 901 at 550°C in Figure 2(b), for

Inconel 718 at both 450°C and 550°C in Figure 3(b) and for alloy 718Plus at both 450°C and 550°C in Figure 4(b). In good agreement with the lowered resistance to plastic deformation there is also a change in fracture appearance from transcrystalline crack propagation to intercrystalline crack propagation in the presence of SO₂ + water vapour. Furthermore it is found that the intergranular cracking often is irregular on a macroscopic level since the crack front penetrates like fingers into the material, leaving unbroken ligaments behind, see Figure 9 and Figure 13. When these ligaments finally crack they are subjected to more plastic deformation with the result that these areas are later on rubbed against each other as the cracks propagate further into the material. Similar observations have been done for Inconel 718 tested in air at higher temperatures and longer dwell times, see reference [1].

At elevated temperatures alloys like Inconel 718 are known to be prone to intercrystalline cracking and two dominating theories to explain this behaviour are found: stress accelerated grain boundary oxidation (SAGBO) and dynamic embrittlement (DE) [2]. The SAGBO process involves oxidation of grain boundaries ahead of the crack tip and subsequent cracking of the oxide, exposing new surfaces to the oxygen. The DE theory on the other hand advocates embrittling of the grain boundary by oxygen diffusion, separation of the embrittled boundaries and subsequent oxidation of the fresh surfaces. DE requires oxygen diffusion over very short distances.

During low cycle fatigue, the crack propagation mechanisms listed above only applies to a small part of the total testing time since the main part of the time is spent on crack initiation (or at least propagation of very small cracks, i.e. cracks significantly shorter than the typical grain size of the material). The question is then if there is an environmental effect also on the crack initiation mechanism. Clearly some of the secondary cracks seen in Figure 10 and

Figure 11 are very short and very much affected by oxidation and should probably initially be seen as oxidation pits. Typically, these pits seem to be located in grain boundaries and as they crack they must be important for the crack initiation process. The negative effect of these pits is not just as stress concentrations, it also enhances the transportation of embrittling elements down along the grain boundaries. Crack initiation mechanisms influenced by oxidation can thus be enhanced by the presence of water vapour and SO_2 .

The amount of oxide seen in the cracks of Inconel 718 tested in water vapour + SO_2 is, as stated above, very high, see Figure 10. This can be compared to the tests in air at the same temperature where very little oxide can be found along the crack and only over a very short distance ahead of the crack tip, see Figure 8. If only the surface of the specimens are considered instead, the surface corrosion in air and in SO_2 + water vapour are rather similar. The surface scale therefore seems to remain adherent and protective on the main part of the surface. However, due to the strain introduced during LCF cycling, the surface oxide scale will probably rupture locally (preferably at the grain boundaries) and intergranular microcracks will initiate which promotes inward diffusion of oxygen and sulphur and outward diffusion of mainly nickel, see Figure 12. The crack penetrating along the grain boundary is filled with oxides, which seems to have a layered structure. Close to the main material the oxide is rich in Nb and sulphur while the centre has higher concentrations of Fe and Ni. The Ni also seems to diffuse preferably towards the surface and forms a nodular like corrosion product. In a study by Browning and Henry [13] on the influence of water vapour and different oxygen partial pressures it was found that slow crack growth rates in Alloy 718 were associated with the formation of a Cr-rich crack tip oxide scale while fast crack growth rates were associated with the formation of a Ni- and Fe-rich crack tip oxide scale. Thus, it is very likely that the corrosion process active in the current study has significantly increased the

crack propagation rates for tests performed in SO₂ + water vapour, especially at the higher test temperature.

The observations described above will obviously have implications for the prediction of service life of components made of nickel-base superalloys. Crack propagation under service like conditions are often considered as the combined effect of a pure cycle dependent mechanism and a time dependent mechanism, which is controlled by creep and/or oxidation processes, see e.g. reference [14]. Crack initiation on the other hand is typically considered to be influenced by the in-elastic strain range only and the in-elastic strain range will then include both cycle dependent plasticity and time dependent creep deformation.

However, the present study also show that the resistance to an in-elastic strain range decreases in an environment consisting of SO₂ + water vapour. Furthermore, a higher environmental impact is seen for the tests with higher strain ranges compared to those with lower strain ranges, see Figure 3 to Figure 5. Thus, it is clear that the environmental effect on LCF life is not only time dependent, since the general trend for all materials investigated in this study is that the environmental effect decreases with increasing number of cycles to failure even if this also means that the total exposure is longer, as schematically illustrated in Figure 14. One possible explanation to this behaviour is that the strain, which causes cracking of the oxide scale at the grain boundaries on the specimen surface, is important. A schematic illustration of the failure mechanism can be found in Figure 15. Without any externally applied load the material is able to create a protective oxide scale on the specimen surface. However, when the external load is high enough the protective scale on the surface will break, preferably at the grain boundaries, and oxidation can proceed down the grain boundaries. This mechanism is

related to the materials resistance to fatigue crack initiation in an SO₂ + water vapour environment.

Of the materials investigated in the present study, Inconel 718 has by far the best low cycle fatigue resistance in air. On the other hand the environmental impact is stronger for Inconel 718 at 450°C compared to Nimonic 901. This indicates that Inconel 718 is more sensitive to SO₂ + water vapour than Nimonic 901. Despite the stronger environmental effect for Inconel 718, the LCF properties in SO₂ + water vapour is still better for Inconel 718 compared to Nimonic 901 due to the exceptionally good initial LCF resistance for Inconel 718 seen in air. The tests on 718Plus show a tendency that is very similar to that of Inconel 718.

The lower environmental impact from water vapour + SO₂ on Nimonic 901 can have several possible explanations. Since the grain boundaries seem to play an important role also for the crack initiation process it is likely that a larger grain size as seen for Nimonic 901 can be beneficial. Furthermore, the alloying element Nb (which is rather similar in Inconel 718 and 718Plus) has been attributed to an increased time dependent crack growth (creep crack growth), see reference [2]. Also in the present study it is found that a Niobium-rich corrosion product is formed along the crack front in Inconel 718 which consolidates the negative effect of Niobium on LCF life in SO₂ + water vapour. Nimonic 901, which seems to be less susceptible to environmental fatigue at least at 450°C does not contain any Niobium. The positive effect of a large grain size and a low Niobium content found in this study is probably a general relationship that can be applied to other Nickel based alloys as well.

5 Summary and Conclusions

There is a negative effect of SO_2 + water vapour on the low cycle fatigue resistance of nickel based superalloys. The negative effect is found in both the crack initiation and the crack propagation process. This study shows that the negative effect increases with temperature and is more severe for alloys with small grains and high niobium contents. The environmental effect is manifested by a decreased resistance to cyclic plastic deformation and a transition from transcrystalline to intercrystalline fracture behaviour. In general this negative effect increases with the degree of plastic deformation while for lower mechanical strain amplitudes where the number of cycles and the total exposure time is higher the environmental impact is reduced. The surface corrosion in air and in SO_2 + water vapour are found to be rather similar in this study and it is therefore concluded that the surface scale can remain adherent and protective if the strains on the oxide scale are low. However, for LCF tests with higher strain ranges, the oxide scale will rupture preferably at the grain boundaries and intergranular microcracks will initiate which promotes inward diffusion of embrittling elements such as oxygen and sulphur, which reduces the fatigue resistance.

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Figure 1: Schematic drawing of the equipment setup for testing in controlled environments. The glass tubes and glass fibre padding seals can be seen around the specimen. The inlet gas flow tube ends up in the inner chamber next to the specimen and the outlet gas tube can be seen in the middle chamber. Placement of the thermocouples can also be seen on the specimen.

Figure 2: LCF results for Nimonic 901, (a) total strain range versus cycles to failure, (b) plastic strain range of the mid-life loop versus cycles to failure

Figure 3: LCF results for Inconel 718, (a) total strain range versus cycles to failure, (b) plastic strain range of the mid-life loop versus cycles to failure

Figure 4: LCF results for Alloy 718plus, (a) total strain range versus cycles to failure, (b) plastic strain range of the mid-life loop versus cycles to failure

Figure 5: Transcrystalline fracture appearance in Nimonic 901 tested in SO_2 + water vapour at 450°C , (a) longitudinal cross section, (b) fracture surface

Figure 6: Intercrystalline fracture appearance in Nimonic 901 tested in SO_2 + water vapour at 550°C, (a) longitudinal cross section, (b) fracture surface

Figure 7: Fracture appearance in Inconel 718 tested in air (a) Transcrystalline 450°C, (b) Partly intercrystalline at 550°C

Figure 8: Longitudinal cross section of a secondary crack in an Inconel 718 sample tested in air at 550°C. (a) shows the entire crack whereas (b) shows the crack tip.

Figure 9: Fracture surface of an Inconel 718 sample tested in SO_2 + water vapour at 550°C . The darker areas of the propagation zone are worn surfaces. (a) low magnification (b) higher magnification of the area indicated in Figure 9(a)

Figure 10: Longitudinal cross section with secondary cracks in Inconel 718 tested in SO₂ + water vapour at 550°C .

Figure 11: Inconel 718 tested at 550°C in SO_2 + water vapour showing cracking at an oxidation pit located in a grain boundary.

Figure 12: EDS-analysis of an Inconel 718 sample tested at 550°C in SO₂ + water vapour

Figure 13: Fracture surface of Alloy 718plus tested in SO_2 + water vapour at 450°C showing a finger like appearance with worn and unworn areas

Figure 14: Schematic illustration of the environmental impact on LCF life.

Figure 15: Schematic illustration of the failure mechanism during environmental assisted LCF failure