THERMAL-MECHANICAL FATIGUE BEHAVIOUR OF MICROCAST-X INCONEL 718

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1 INTRODUCTION

Inconel 718 is a nickel based superalloy which is commonly used in aircraft engine applications, where severe mechanical and thermal stresses are exerted on components. Recently, a finer equi-axed version of cast Inconel 718 superalloy, called Microcast-X, has been developed by Howmet Corporation of USA. It is reported to have better properties than those of the conventionally cast, as well as wrought Inconel 718 [1]. Since the grain size of the material is significantly smaller than the conventionally cast Inconel 718, its weldability is also like to be better then that of conventionally cast Inconel 718. Therefore, a project was initiated to evaluate the weldability, microstructures and mechanical properties of Microcast-X Inconel 718 and determine the effect of welding on them, as well as to compare it with that of the wrought Inconel 718 superalloy. This communication is concerned with the thermomechanical fatigue aspects.

2 MATERIAL AND EXPERIMENTAL PROCEDURES

The material used in this study was produced by the Howmet Turbine Components Corporation using a proprietary investment casting technique, Microcast-X. Following casting, the material was hot isostatic pressed (HIP) at 1120°C for four hours under 103 MPa pressure. The 1.6 cm thick castings were 15cm x 20.5cm, excluding end risers. The chemical composition supplied by Howmet is given in Tables 1. Homogenization of Inconel 718 castings was performed to reduce chemical segregation and eliminate brittle, non-equilibrium Laves phase. Two different metallurgical conditions were introduced in the as-received alloy by doing solution heat treatment at 1050°C and 1150°C. Heat affected zone (HAZ) was simulated in the heat treated material by subjecting it to weld thermal cycle by Gleeble thermo-mechanical simulator. For comparison purposes, additional testing was performed on conventional wrought Inconel 718 conventionally heat treated.

In-phase thermo-mechanical fatigue tests were performed using the Gleeble simulator, with temperature cycled between 350°C and 650°C at a frequency of 0.1

Table 1. Chemical composition (wt%) of MX Inconel 718

Nickel	52.81	Carbon	0.04
Iron	18.88	Silicon	0.02
Chromium	18.85	Copper	0.01
Niobium	4.84	Zirconium	0.01
Molybdenum	3.04	Boron	0.004
Titanium	0.9	Phosphorus	0.003
Aluminum	0.48	Sulfur	0.001
Cobalt	0.1	Magnesium	0.001
Manganese	<0.1	Silver	<5 ppm

3 RESULTS

The as received Microcast-X alloy had a fine grained and non-dendritic structure, the equiaxed grain structure being consistent throughout the casting and the grain size was measured to be 87±5 μm. Typical grain sizes for conventionally cast 718 was in the range of 0.5 and 10 mm. The secondary phases present in the as-received MX 718 were Laves phase, NbC, δ phase, and γ″. The material also had 0.25 vol.% shrinkage porosity in it. In the solution treated condition Laves phase and δ needles were partially retained from the original cast structure, though the grain size remained unchanged from the as-received. As expected, the aging treatment resulted in the precipitation of γ″ and γ′ phases that were identified with the TEM and a representative TEM micrograph is shown in Fig 1.

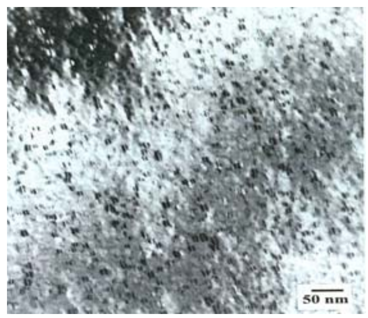

Fig. 1. TEM bright field image of aged MX718

The data from the thermal mechanical fatigue tests are shown in Fig. 2. Some interesting trends in TMF behavior are evident in the stress-life plot, there being a noticeable difference in TMF life between aged wrought and MX material. The wrought material required considerably more fatigue cycles to reach failure for a given loading. For example, at a maximum stress of 750 MPa the wrought material survived an entire order of magnitude of cycles longer than the MX material. The aged microstructure offered a considerable improvement over the non-aged and HAZ

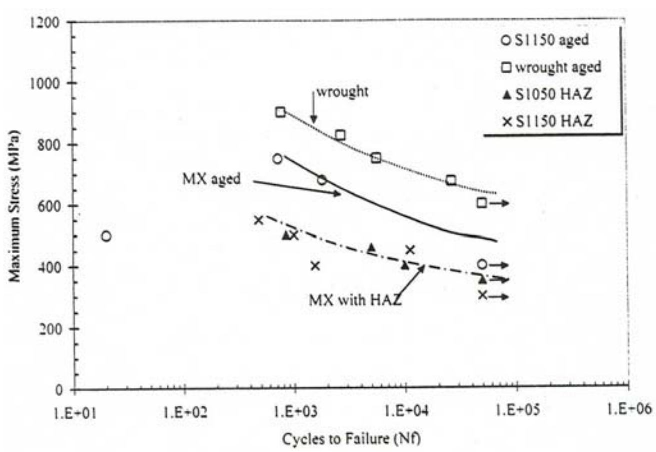

Figure 2. Thermal Mechanical Fatigue of MX and Wrought 718

simulated material, as would be expected. The aging treatment improved material strength and it could therefore withstand higher loads with less plastic strain. In addition, weld simulation resulted in partial liquation of the grain boundaries in HAZ and which reduced the resistance to intergranular crack propagation.

The fracture surfaces of TMF specimens were examined with a SEM and were found to be similar for all the MX 718 specimens. Prior to fracture, numerous surface porosity had developed which were far greater in number than those present in the untested material. These porosities acted as the crack initiation sites. However, the material did not contain porosity levels high

Figure 3. Oxidation and separation at grain boundary triplepoint during testing of a HAZ simulated specimen loaded with a maximum stress of 350 MPa.

enough to explain all of the initiation sites. Crack initiation was also observed at sties of intergranular oxidation at open surfaces and along grain boundaries, as shown in Fig 3. Evidence of grain boundary sliding and active creep damage was also observed, and an example is shown in Figure 4. It is suggested that many of the cracks that developed at the specimen surface are evidence of creep-fatigue interaction.

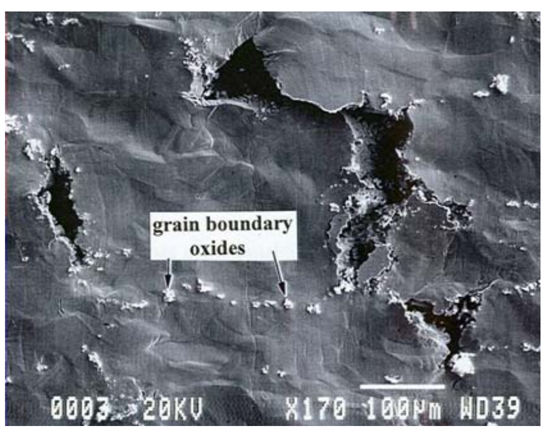

Figure 4. Oxidation and crack initiation along grain boundaries during testing of a HAZ simulated S1050 specimen loaded with a maximum stress of 460 MPa

Crack growth in the specimens tested was observed to be intergranular from initiation up to fast fracture. An example of intergranular crack is shown in Figure 5a. A similar preference for intergranular crack propagation has been observed during in-phase thermal mechanical fatigue of alloy 718. Fatigue damage is believed to occur only on the loading cycle and in-phase testing enables tensile creep components to damage the grain boundaries and promote intergranular growth. The test data did not produce any discernable differences between the S1050 and S1150 solution treatments.

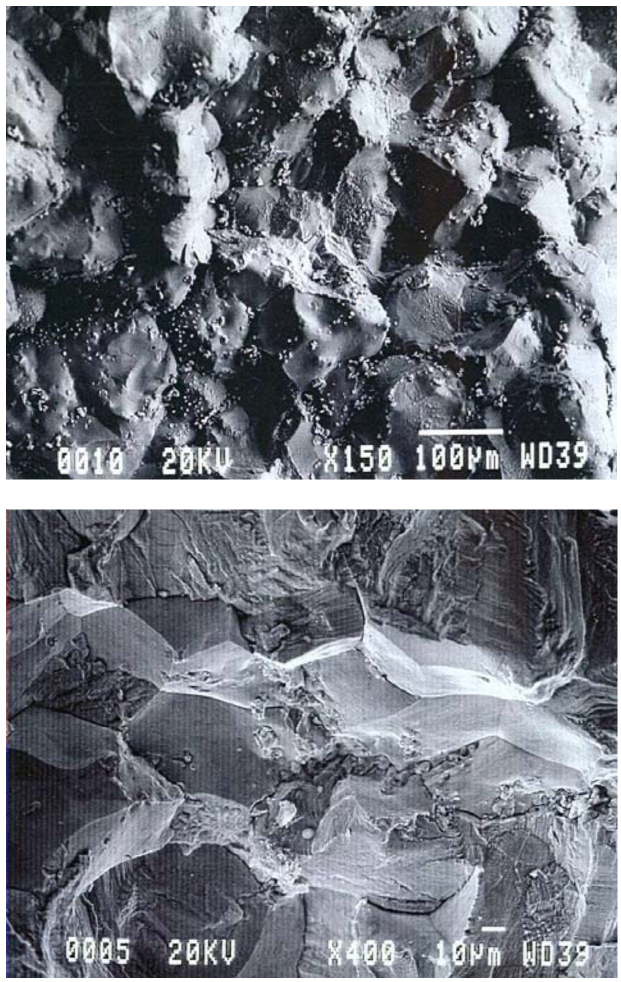

Figure 5(a & b). Shows the intergranular crack growth observed in aged and HAZ simulated microstructures. Oxidation of small particles, presumably (Ti,NB,Mo)C, located on grain boundary surfaces is evident.

4 DISCUSSION

41.Crack Initiation

Crack initiation was observed to be intergranular for all TMF specimens. It is believed that the high upper temperature of 650°C combined with the low strain rates favored this mechanism over crack initiation at persistent slip bands on the sample surface caused by heterogeneous slip [3]. Other investigators have reported that intergranular initiation is predominant at higher temperatures for nickel-based alloys [4,5]. The two main reasons provided for this is the greater contribution of creep and environmental damage at high temperature. Creep assisted initiation occurs when neighboring grains move with respect to each other and separation occurs at stress concentrations along the grain boundary such as triple points. Environmental damage occurs by oxidation of grain boundaries accessible from the specimen surface. These regions become brittle and in some cases the strengthening precipitates are depleted. In TMF the thermal cycle produces strains between the oxidation products and the base metal because differences in thermal expansion [3]. Both the oxidation of grain boundaries on the sample surface and evidence of creep cavitation at grain boundaries was observed for the TMF specimens. It is proposed that both mechanisms were active in this investigation.

In the results section, a third initiation mechanism was suggested as the predominant initiation site for fatigue cracks. This mechanism was by cracking at the stress concentrations provided by intergranular porosity intersecting the specimen gauge surfaces. The MX 718 material used for this study was quantified as having shrinkage porosity content of 0.25%. This is extremely low, but several instances of the gauge surface intersecting with porosity were noted and fatigue cracking was observed to initiate from the defects.

4.2 Crack Propagation

Crack growth in the specimens tested was observed to be intergranular from initiation up to fast fracture. A similar preference for intergranular propagation was reported by Cook et.al. [2] for in-phase thermal mechanical fatigue of alloy 718. Out-of-phase testing between the same temperature has been reported to result in transgranular crack growth [2]. Fatigue damage is believed to occur only on the loading cycle and in-phase testing enables tensile creep components to damage the grain boundaries and promote intergranular growth.

The test data did not produce any discernable differences between the S1050 and S1150 solution treatments. This may be because the test group size was small, but it is still possible that the grain size difference between the two materials being an important parameter, could have been responsible for it. Maiya and Majumdar [7] reported a decrease in the fatigue life of 304 SS with increasing grain size attributed to wedge cracks forming at the grain boundaries more readily.

In addition, the entire fracture surface of TMF specimens was strongly dominated by intergranular growth, whereas in a previous examination of isothermal fatigue crack growth specimens the fracture surfaces were not so dominantly intergranular with some areas displaying faceted structures or striated growth [6]. Such observations should be expected given the slower strain rates of the TMF cycle.

5 CONCLUSIONS

1) Crack initiation in TMF was attributed mainly to intergranular shrinkage porosity, even though the level of porosity was about 0.25%
2) The path of the crack was mainly intergranular from initiation at porosities and sometimes from grain boundary cavitation, which formed due to creep-fatigue interactions
3) The Microcast-X alloy had lower resistance to thermal fatigue than the wrought alloy due to the presence of shrinkage porosities in the cast alloy and a less resistant microstructure to creep-fatigue damage.

4) As expected the presence of a synthetic HAZ also reduced the TMF behaviour of the Microcast-X alloy relative to the as heat treated material.

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7 REFERENCES

1. Bouse, G.K., R.A. Dunham and J. Lane, "Mechanical Properties of Fine-Grained Microcast-X Alloy 718 Investment Castings for SSME Gas Turbines, and Airframe Components", *Proc. Of Conference Superalloy 718, 625, 706 and Various Derivatives*, June 15-18, 1997, TMS, Warrendale, U.S.A., pp. 459-468.
2. Cook, T.S. J.H. Laflen, R.H. Van Stone, and P.K. Wright, "TMF Experience With Gas Turbine Engine Materials", *Thermal Mechanical Fatigue of Aircraft Engine Materials*, AGARD-CP-569, 1995: 2.1-2.15.
3. James, L.A., and W.J. Mills, "Fatigue Crack Propagation Behavior of Defective Weldments", *Int. J. Pres. Ves. & Piping,* Vol. 9, 1981: 367-383.
4. Spera, D.A. "What is Thermal Fatigue?", *Thermal Fatigue of Materials and Components,* ASTM STP 612, 1976: 3-9.
5. Boismier, D.A. and H. Sehitoglu, "Thermo-Mechanical Fatigue of Mar-M247: Part 1- Experiments", *Journal of Engineering Materials and Technology*, Vol. 112, 1990: 68-79.
6. Zagula, J. N.L. Richards and M.C. Chaturvedi, "Fatigue Crack Growth Behaviour of Microcast-X Inconel 718", *Proceedings of 4th Creep Fatigue Interaction Conference*, Kalpakkam, India, Indian Institute of Metallurgy, 2003, I43-52.
7. Maiya, P. and S. Majumdar, *Metallurgical Transactions A,* Vol. 8A, 1977: 1651.