DEVELOPMENT OF A DAMAGE TOLERANT MICROSTRUCTURE FOR INCONEL 718 TURBINE DISC MATERIAL

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SUMMARY

A new modified heat treatment has been developed for Inconel 718. This heat treatment leads to substantial improvements in elevated temperature crack propagation resistance with apparently limited loss in resistance to LCF crack initiation as compared to the conventional heat treatment for this alloy. This is a result of tailoring the microstructure to obtain the optimum combination of grain size, grain boundary structure and matrix precipitate morphology.

In the modified heat treatment, the material is solution treated at 1032°C/1h (below the grain coarsening temperature) then furnace cooled to 843°C and held for 4h to produce profusely serrated grain boundaries by precipitating along the boundaries orthorhombic δ- Ni_3Nb needles, hereafter referred to as the Ni_3Nb phase. After this, the material is partially solution treated at 926°C/1h, to dissolve the coarse intragranular γ'' precipitates (also of the basic composition Ni_3Nb but having a body centred tetragonal structure) that previously formed during furnace cooling from 1032°C to 843°C/4h. Finally the material is subjected to the conventional double aging heat treatment.

Relative to the conventional heat treatment, the new heat treatment reduces the FCGRs and CCGR of Inconel 718 by a factor of 2 and 5 respectively at 650°C. The new heat treatment does not alter the LCF life as a function of total strain relative to the conventionally heat treated material at 650°C. This is a significant result if damage tolerance concepts are used in turbine disc design.

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Introduction

Inconel 718 is used as disc alloy in a number of gas turbine engines. Disc forgings of Inconel 718 are generally solution treated (ST) at 955°C/1h then direct aged by cooling from this temperature to 718°C and held at 718°C for 8h, further cooled to 621°C and held at 621°C for 8h and finally air cooled (AC). This heat treatment has proved adequate from a safe life point of view, where disc life limits are established statistically on the basis of the number of low cycle fatigue (LCF) cycles required to form a detectable crack (~ 0.8 mm) in 1 in 1000 components. However, for those cases where damage tolerance design requirements must be satisfied, the conventional heat treatment may not be adequate since the crack growth rates may be too high to obtain a practical safe inspection interval (SII). In damage tolerance design, the fracture critical locations of discs are assumed to contain defects of a size corresponding to the detection limit of the nondestructive inspection (NDI) technique used to inspect the components. These defects are then assumed to act as propagating cracks and their rates of propagation are established on the basis of fracture mechanics principles, using experimental crack growth rate data.(1) The time or number of cycles to grow these inherent cracks to a predetermined dysfunction size are then used to establish a SII on the basis of which discs are repeatedly returned to service until a crack is eventually detected, Figure 1. If the crack propagation rates are excessively high, the SII may prove too short to be economically viable.

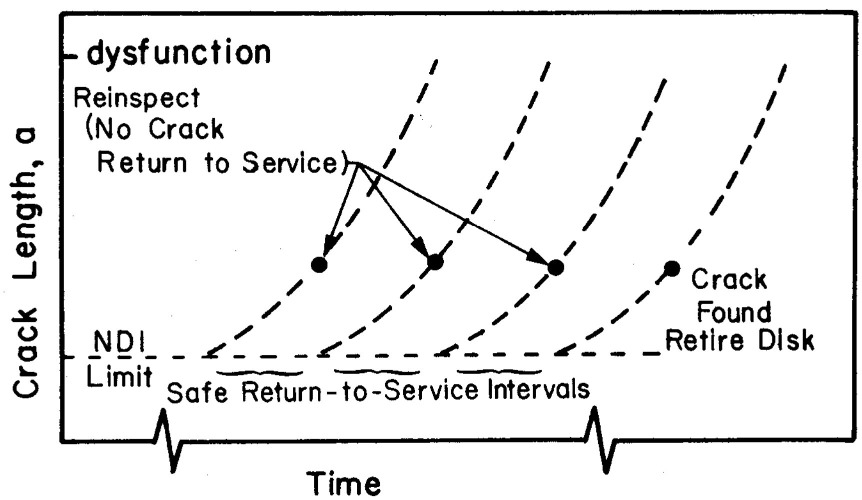

Fig. 1. Schematic representation of the damage tolerance based life prediction methodology.

Table 1. Chemical Compositions of two Inconel 718 heats in wt. %.

Stock Type	C	Si	Cr	Ni	Mo	Nb + Ta	Ti	Al	Fe
Bar	0.03	1.12	18.3	53.2	3.0	5.17	1.0	0.42	Bal
Plate	0.05	0.21	17.9	53.1	3.06	5.11	0.96	0.47	Bal

This paper describes work aimed at developing a heat treatment for Inconel 718 which reduces elevated temperature creep and fatigue crack growth rates (CCGR and FCGR) without substantial sacrifice in LCF crack initiation life relative to conventional microstructure.

A number of heat treatments have already been developed for improving the fracture toughness[2], notch rupture ductility[3], notch stress rupture properties [4,5] and CCGRs and FCGRs[6,7] of Inconel 718. However, none of these heat treatments are capable of producing a microstructure that improves CCGR and FCGR without sacrificing LCF life. It is also clear that microstructural variables such as the grain size, the grain boundary Ni_3Nb morphology and the size of γ'' precipitates have to be controlled for optimizing these properties.[7] The following strategy was therefore adopted for developing a damage tolerant microstructure for Inconel 718 discs. This included:

(i) selecting a solution treatment temperature for achieving full solutioning of the Ni_3Nb, γ' and γ'' precipitates without inducing excessive grain growth which can reduce LCF crack initiation life,

(ii) developing a post solutioning heat treatment sequence for reprecipitating the Ni_3Nb needles along the grain boundaries to form serrated grain boundaries which suppress grain boundary sliding and

(iii) developing an additional heat treatment sequence for precipitating the optimum amounts of γ' and γ'' precipitates to strengthen the grain interiors without altering the serrated grain boundary structures.

Experimental Materials and Methods

The commercially available hot rolled Inconel 718 was procured in the form of 22 mm diameter bars and 12.7 mm thick x 50.8 mm wide plates. The bar and the plate stock were from two different heats and their chemical compositions are given in Table 1.

Heat Treatments and Microscopy

The solution treatments were carried out at 1032°C, 1050°C, 1066°C, 1080°C and 1093°C for 1 to 16h. The grain boundary Ni_3Nb precipitation kinetics were monitored over a range of aging temperatures (818°C to 917°C) for starting solution treatment conditions of 1032°C/1h, 1066°C/1h and 1080°C/1h. In this case, the specimens were solution treated, furnace cooled (4 to 7°C/min) and direct aged for 1 to 6h. Transmission electron microscopy, using replica and thin foil techniques, was carried out on a selected number of solution-treated and direct aged and direct aged plus partial solution treated specimens.

A series of heat treatments was also conducted on a selected number of specimens that were direct aged in the Ni_3Nb precipitation range in order to solution the overaged γ' and γ'' precipitated during direct aging. These heat treatments were carried out at 917°C, 926°C, 955°C and 975°C for 1 to 10h.

Low Cycle Fatigue Testing

ASTM E606 axial fatigue specimens were used for conducting fully reversed, constant amplitude LCF tests in a closed loop electrohydraulic testing system under total axial strain control using a triangular wave form and a constant strain rate of 0.002/s. The test section of each specimen was polished manually in the axial direction with successively finer grit emery papers (grade 320, 400, 600) to remove circumferencial machining marks. The specimens were tested over a strain range ($\Delta\epsilon$) of 0.65 to 2% at 650°C in a laboratory air environment.

An x-y recorder was used to obtain cyclic total axial strain versus load plots, which were converted to engineering stress-strain hysteresis loops using the specimen cross sectional area. Specimen failure, N_f, was defined by the number of cycles for a 5% drop in the steady state tensile stress value.

Creep and Fatigue Crack Growth Rate (CCGR and FCGR) Testing

All CCGR and FCGR tests were conducted in an electrohydraulic testing system. Initially, a tapered double cantilever beam (DCB) fracture mechanics specimen, having a constant K region over 31.75 mm,[7] was used for CCGR and FCGR testing at a stress intensity factor and range (K and ΔK respectively) of 45 MPa $\sqrt{m}$ in laboratory air environment at 650°C. An R-value of 0.1 and a frequency of 0.1 Hz were selected for FCGR testing using a sine wave form. Fracture surfaces were studied by scanning electron microscopy (SEM) to determine the crack growth rates.[7]

Statistically significant FCGR data bases were further generated, using standard 50.8 mm wide and 12.7 mm thick compact tension (CT) specimens conforming to ASTM E647 specifications, at 650°C in a laboratory air environment. The CT specimens were precracked at room temperature and the FCGR tests were conducted at an R-value of 0.1 and a frequency of 1 Hz using a sawtooth wave form. A direct current potential drop (DC-PD) technique having an accuracy of 0.085 mm and a precision of 0.025 mm was used to monitor the crack lengths at 650°C. The FCGR data was generated over a ΔK range of 24 to 80 MPa $\sqrt{m}$.

Results and Discussion

Solution Treatment Selection

Figure 2 shows a plot of average grain size versus solutioning temperature for a range of solution treatment times. It is noted that excessive grain coarsening commences between 1040° and 1050°C and this is because primary NbC precipitates (solvus 1040-1093°C) begin to dissolve in this temperature range. The γ'' and Ni_3Nb solvus temperatures for Inconel 718 are 900°C and 982 to 1037°C respectively. A solutioning temperature of 1032°C will prove optimum in achieving full solutioning of the Ni_3Nb, γ' and γ'' precipitates without inducing excessive grain growth. A solution treatment time of 1h is considered adequate for homogenizing the microstructure. Longer solution treatment times could increase the grain size through Ostwald ripening of NbC precipitates even if the solution treatment temperature is kept below the NbC solvus temperature, Figure 2.

Selection of a Direct Aging Treatment to Form Serrated Grain Boundaries

In Inconel 718, the precipitation of the grain boundary Ni_3Nb needles may under certain optimum conditions create a serrated grain boundary structure.[7] In this case, the serrations arise from the cellular precipitation of Ni_3Nb following the Tu-Turnbull mechanism instead of the motion of heterogeneous γ' precipitates (a mechanism suggested to produce serrations in high γ' volume fraction alloys).[8] Serrations due to the latter mechanism exhibit well-rounded peaks and valleys whereas serrations arising from Ni_3Nb needles are angular in nature. It is therefore important to establish the kinetics of Ni_3Nb precipitation in Inconel 718.

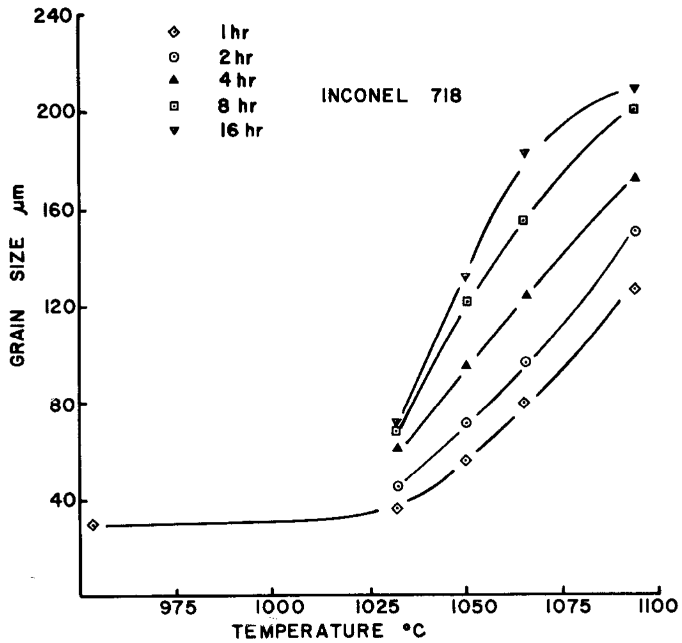

Fig. 2. Grain coarsening behaviour of Inconel 718

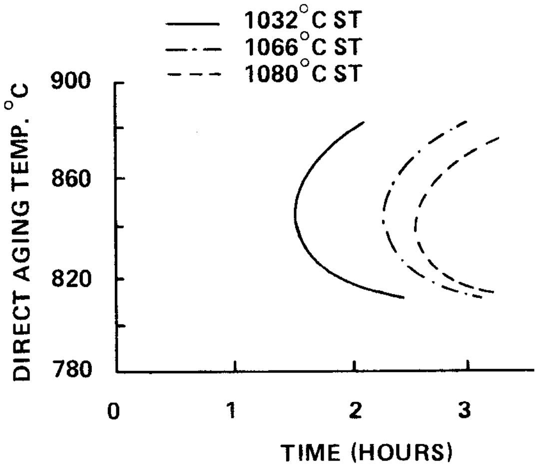

Fig. 3. Inconel 718 grain boundary Ni_3Nb time-temperature-precipitation (TTP) curves for samples solution treated (ST) at 1032°C, 1066°C and 1080°C.

A number of studies have documented the general kinetics of Ni_3Nb precipitation as a function of solution treatment and aging temperatures.(6) However, there is little or no information available for direct aging conditions where the material is slow cooled from the solutioning to the intermediate aging temperature and held at the aging temperature to produce grain boundary Ni_3Nb needles and serrations. For solution treatment temperatures of 1032°, 1066° and 1080°C, the time-temperature-precipitation (TTP) curves for a range of direct aging conditions are presented in Figure 3. It is somewhat surprising to note that lower solution temperatures enhance Ni_3Nb precipitation upon direct aging. Similar results have also been reported by other workers.(4)

It has been suggested that a lower solution treatment temperature produces a smaller grain size and a larger grain boundary area which accelerates the nucleation kinetics of Ni_3Nb needles.(4) Mechanistically, however, nucleation rate is sensitive to the solution treatment temperature only when quenching to the aging temperature is rapid and the equilibrium mole fraction of solute adsorbed at the grain boundaries varies markedly from one solution treatment condition to another.(9) Neither of these conditions are entirely satisfied in the present experiments because cooling to direct aging temperature is slow and the solution treatment temperatures lie close to or above the Ni_3Nb solvus thus minimizing the grain boundary solute concentration differences within the temperature range studied. It is possible that the differences in the Ni_3Nb TTP-curves in Figure 3 are instead related to the differences in the growth kinetics of Ni_3Nb during direct aging. From 1080°C, the cooling rate to the direct aging temperatures might be relatively faster than from 1032°C resulting in somewhat higher point defect densities in the 1080°C solution treated specimens. A higher defect density would precipitate a larger volume fraction of heterogeneous γ'' during direct aging thus relieving the matrix Nb super-saturation for grain boundary Ni_3Nb precipitation and vice versa, Figure 4. Therefore, the differences in the defect density and Nb supersaturation from one solution temperature to another may be responsible for delaying the growth of Ni_3Nb precipitates with increasing solutioning temperature during direct aging.

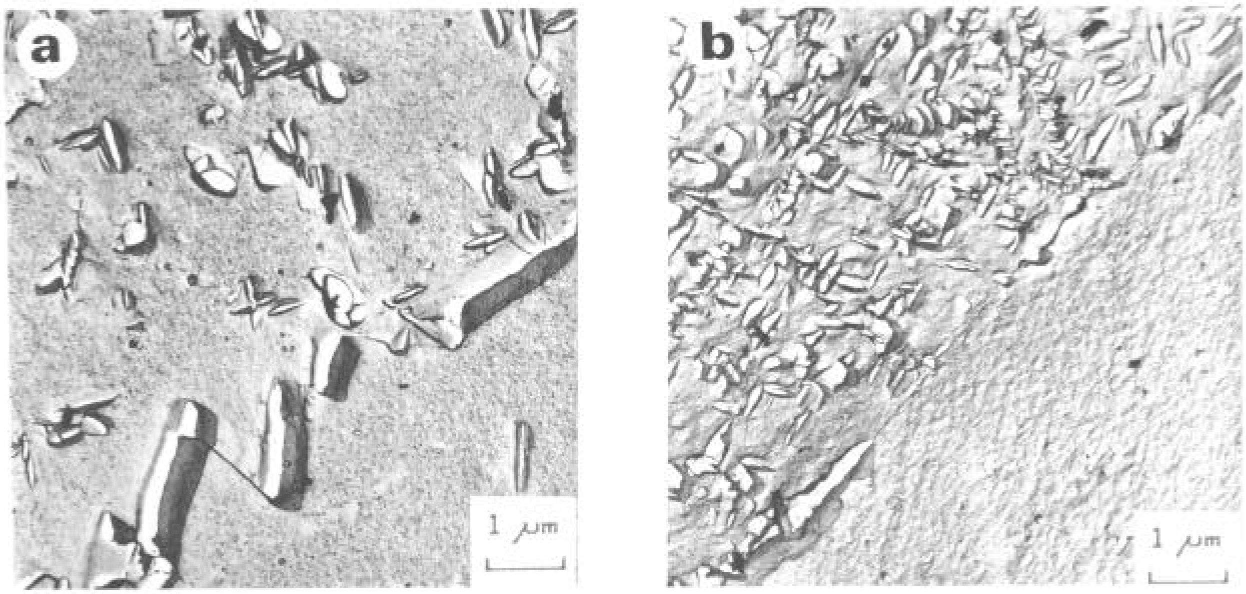

Fig. 4. Coarse γ'' precipitates formed during furnace cooling from the solution treatment temperature to the direct aging temperature of 843°C.(a) ST 1066°C (b) ST 1080°C.

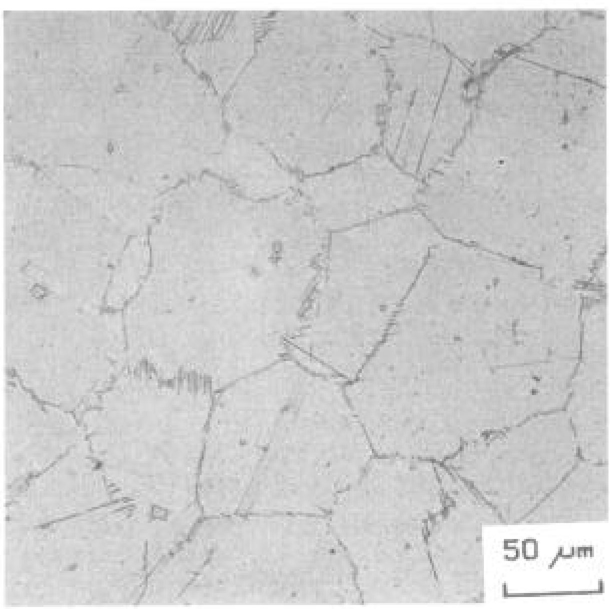

Fig. 5. Serrated grain boundary morphology in specimens solution treated at 1032°C/1h $\xrightarrow{F.C.}$ 843°C/4h/AC+926°C/1hr $\xrightarrow{F.C.}$ 718°C/8h $\xrightarrow{F.C.}$ 621°C/8h/AC.

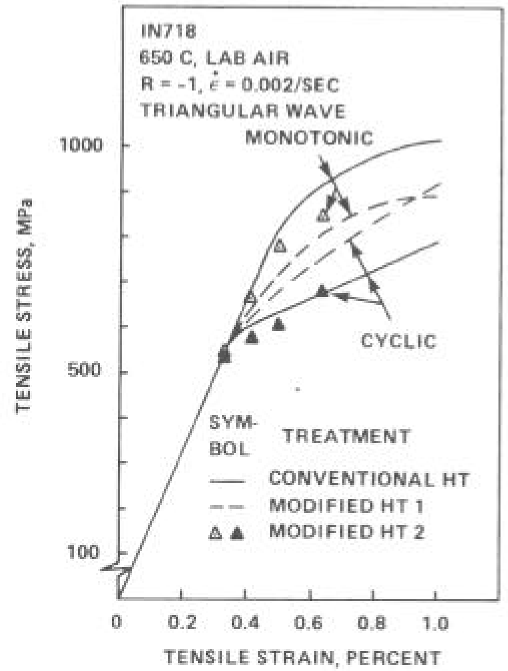

Fig. 6. Monotonic and cyclic stress-strain data for specimens subjected to conventional and modified heat treatments.

The nose of the Ni_3Nb TTP-curves lies in the vicinity of 843°C in all cases, Figure 3. Direct aging at 818°C formed short Ni_3Nb needles with limited serrations at the grain boundaries whereas direct aging at 917°C led to Laves phase precipitation. A direct aging treatment of 843°C/4h produced profuse serrations, Figure 5, whereas longer aging times led to intragranular Ni_3Nb precipitation. Therefore, a direct aging treatment of 843°C/4h was selected for inducing a serrated grain boundary structure.

Selection of Aging Treatments for Precipitating Optimum Amounts of γ' and γ''

All direct aging treatments that produce serrated grain boundaries also induce coarsening of matrix γ' and γ'' precipitates, Figure 4, which may be harmful to LCF life. Therefore, it is necessary to introduce a partial solution treatment in order to dissolve the coarsened matrix γ' and γ'' precipitates and reprecipitate them in the optimum morphology without dissolving the grain boundary Ni_3Nb precipitates and straightening the serrated grain boundaries. After direct aging, partial solutioning at 917°C did not completely dissolve the coarse γ'' precipitates whereas at 955°C it led to excessive growth of Ni_3Nb needles while at 975°C it started to dissolve the Ni_3Nb needles. Excessive Ni_3Nb needle growth should be avoided because it removes Nb (the main γ'' forming constituent) from the grains thus decreasing the matrix strength. A partial solution treatment of 926°C/1h was thus deemed adequate for solutioning the coarse γ' and γ'' precipitates formed during direct aging. Following the suggestion of other workers,[4,7] it was further decided to reprecipitate γ' and γ'' through a standard double aging treatment, i.e. 718°C/8h $\xrightarrow{\text{F.C.}}$ 621°C/8h/AC.

Two modified heat treatment schedules were selected to assess which would provide the best balance of mechanical properties relative to the conventional heat treatment, Table 2. A typical microstructure of the modified heat treated Inconel 718 is shown in Figure 5.

Mechanical properties

The modified heat treatments decrease the Inconel 718 yield strength relative to the conventional heat treatment, Figure 6 and Table 2. These trends are not unexpected because heavy grain boundary Ni_3Nb precipitation in modified heat treated materials removes some Nb (the element responsible for γ'' precipitation) from the matrix which leads to a decrease in the matrix strength.

Under LCF conditions, all materials hardened initially and then softened but cyclic softening was more pronounced in conventionally heat treated specimens. Upon plotting LCF life as a function of plastic strain range, Figure 7(a), it is evident that both modified heat treatments reduced the Inconel 718 LCF life by 40 to 50%. The superior LCF life (in terms of plastic strain range) of the conventionally heat treated specimens can be attributed to their finer grain sizes, Table 3, which promote homogeneous deformation and retard crack nucleation by reducing stress concentrations. Upon plotting the LCF data as a function of total strain range, Figure 7(b), all data fall within experimental scatter. In terms of disc LCF life, the transition fatigue life (N_t), i.e. LCF life where elastic and plastic strains are equal, is an important parameter.[10] For LCF life greater than N_t elastic strain predominates whereas for LCF life lower than N_t plastic strain predominates. The N_t values for the conventional and modified heat treated materials were the order of 300 and 15 cycles respectively, Table 3. Typically turbine discs are designed to have a safe life of 10,000 cycles which is considerably greater than these N_t values. It is therefore likely that the total strain LCF data would be used to predict LCF lives of Inconel 718 turbine discs. It can thus be concluded that relative to the conventional heat treatment the modified heat treatments will not alter the LCF life of Inconel 718 discs at 650°C.

Table 2. Grain size and yield strength data for Inconel 718

Heat Treatment	Heat Treatment Schedule	Grain size μm	0.02% Proof Stress MPa
Conventional	955°C/h $\xrightarrow{F.C.}$ 718°C/8h $\xrightarrow{F.C.}$ 621°C/ 8h/AC	20-40	860
Modified H.T.1	1032°C/1h $\xrightarrow{F.C.}$ 843°C/4h+926°C/1h $\xrightarrow{F.C.}$ 718°C/8h $\xrightarrow{F.C.}$ 621°C/8h/AC	35-80	680
Modified H.T. 2	1032°C/1h $\xrightarrow{F.C.}$ 843°C/4h + 926°C/1h/AC + 718°C/8h $\xrightarrow{F.C.}$ 621°C/8h/AC	35-80	790

Table 3. LCF and constant K CCGR and Constant ΔK FCGR data at 650°C

Heat Treatment	Transition strain range in LCF	LCF life at transition strain in cycles	CCGR is mm/h at 45 MPa $\sqrt{m}$	FCGR in mm/cycle x 10^{-3} 45 MP $\sqrt{m}$
Conventional	0.83	300	2.67	3.0
Mod. H.T. 1	1.40	15	0.52	1.7
Mod. H.T. 2	1.20	15	0.58	1.8

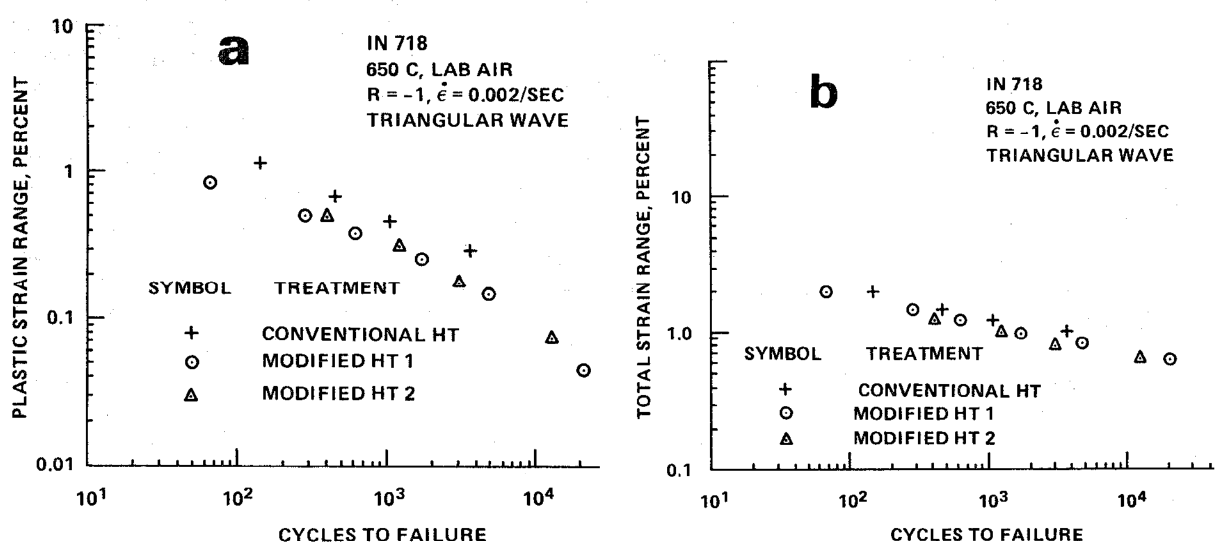

Fig. 7. LCF life as a function of (a) plastic strain range and (b) total strain range for Inconel 718 at 650°C.

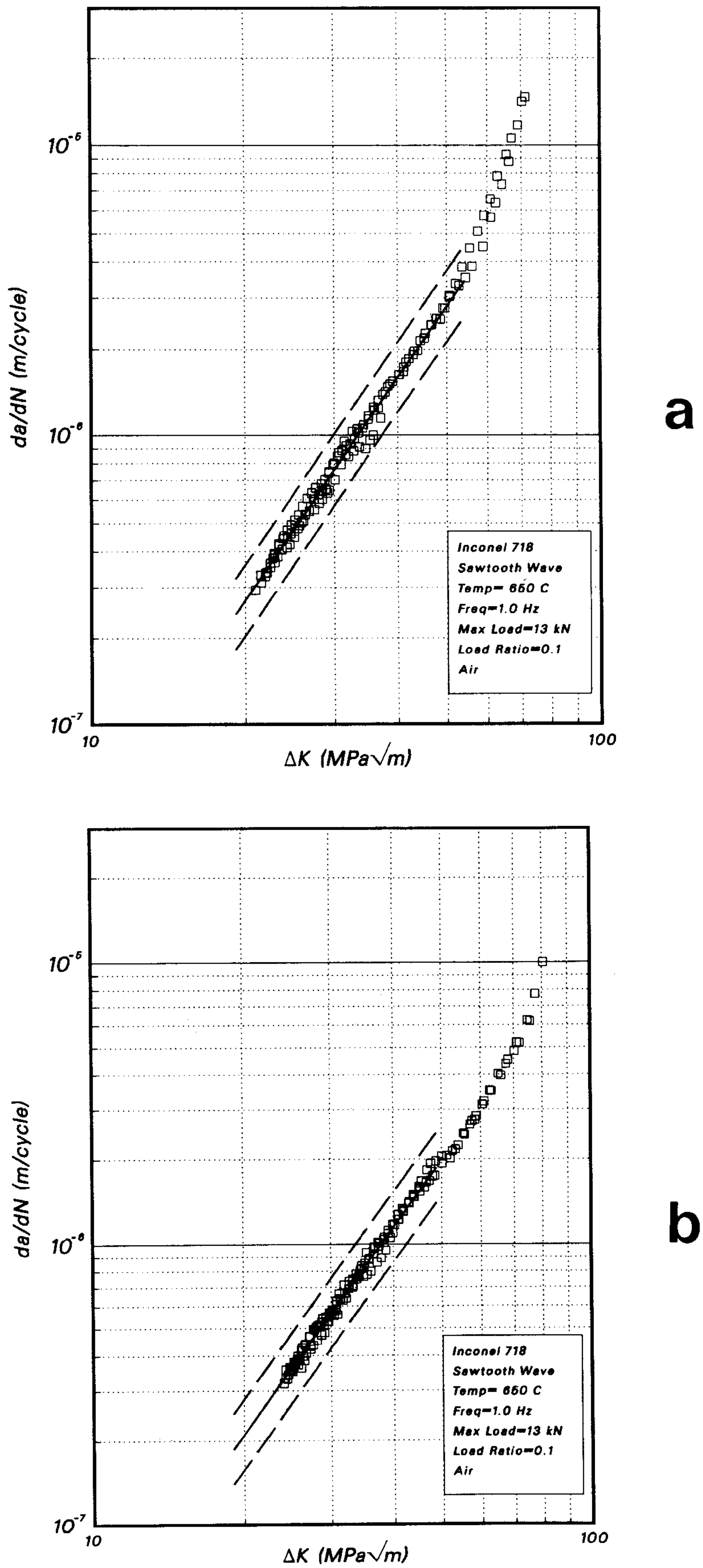

Fig. 8. FCGR data for Inconel 718 at 650°C, (a) conventional heat treatment and (b) modified heat treatment No. 2.

Relative to the conventional heat treatment both modified heat treatments reduced the FCGR and CCGR by a factor of 2 and 5 respectively at 45 MPa $\sqrt{m}$, Table 3. Since both modified heat treatments revealed similar CCGR and FCGR values at 45 MPa $\sqrt{m}$ under constant K and ΔK conditions, FCGR data on CT specimens was only generated for materials subjected to the conventional and the modified heat treatment No. 2. The CT specimen results revealed that relative to the conventional heat treatment the modified heat treatment improved the FCGR by a factor of 2 over a ΔK range of 30 to 80 MPa $\sqrt{m}$, Fig. 8. Furthermore, Figure 8 also indicates that the modified heat treatment might also improve the fracture toughness of Inconel 718. At 650°C, the superior crack growth resistance of modified heat treated materials could be primarily due to a coarser grain size and the presence of serrated grain boundaries which suppress grain boundary sliding as demonstrated in earlier studies.(7)

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References

1. A.K. Koul et al, Proc. AGARD-SMP Conf. on Damage Tolerance Concepts for Critical Engine Components, San Antonio, Texas, AGARD-CP-393 (1985), 23-1 to 23-22.
2. W.J. Mills, The Effect of Heat Treatment on the Room Temperature and Elevated Temperature Fracture Toughness Response of Alloy 718, Trans ASME, J. of Eng. Mat. and Tech., 102 (1980), 118-126.
3. E.L. Raymond, Effect of Grain Boundary Denudation of γ' on Notch-Rupture Ductility of Inconel Nickel-Chromium Alloys X-750 and 718, Trans. Met. Soc. AIME, 239 (1967), 1415-1422.
4. J.F. Muller and M.F. Donachie, The Effects of Solution and Intermediate Heat Treatments on the Notch Rupture Behaviour of Inconel 718, Met. Trans, 6A(1975), 2221-2277.
5. D.J. Wilson, Relationship of Mechanical Characteristics and Microstructural Features to the Time Dependent Edge Notch Sensitivity of Inconel 718 Sheet, Trans ASME, J. of Eng. Mat. and Tech., (1973), 112-123
6. H.F. Merrick, Effect of Heat Treatment on the Structure and Properties of Extruded P/M Alloy 718, Met. Trans. 7A(1976), 505-514.
7. R. Thamburaj et al., Proc. Int. Conf. on Creep, Tokyo, Japan, JSME/ASME, 1986, 275-282.
8. A.K. Koul and R. Thamburaj, "Serrated Grain Boundary Formation Potential of Ni-Base Superalloys and Its Implication," Met. Trans., 16A (1985), 17-26.
9. K.C. Russell and H.S. Aaronson, "Influence of Solution Annealing Temperature Upon Precipitate Nucleation Kinetics at Grain Boundaries," Scripta Metall., 10 (1976), 463-469.
10. T.S. Cook, "Stress-Strain Behaviour of Inconel 718 during Low Cycle Fatigue," J. Eng. Mat. and Tech., 104 (1982), 186-191.