ASSESSMENT OF DAMAGE ACCUMULATION AND PROPERTY REGENERATION BY HOT ISOSTATIC PRESSING AND HEAT TREATMENT OF LABORATORY-TESTED AND SERVICE EXPOSED IN738LC

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Summary

Damage accumulation during creep testing of IN738LC has been quantitatively characterized. The effects of hot isostatic pressing (HIP) on specimens interrupted after various strains have been considered in relation to both the healing of damage and the subsequent mechanical performance. HIP shows little advantage over re-application of the conventional heat treatment in regenerating the creep performance. Examination of service retired rotor blades from an industrial gas turbine show little evidence of creep damage and considerable recovery of creep strength is obtained by heat treatment without pressure.

Introduction

Hot Isostatic Pressing (HIP) is now a well established stage in the manufacture of certain cast and powder superalloy components. The HIP process leads to improved structural integrity, manifested as increased or more consistent mechanical properties, through accelerated consolidation of powders and elimination of porosity in castings. HIP is also being increasingly advocated as a method of rejuvenating service exposed parts in order to extend their operating lives; however, there have been differing views of the effectiveness of HIP for this purpose (1,2,3).

The principal advantages of HIP to material rejuvenation are claimed to be the healing of cavitation and/or cracking damage that is associated with high temperature fracture. It is an implicit assumption of this approach that fracture is significantly influenced by the evolution of such damage at relatively early stages in the creep life and, consequently, that the long tertiary creep regimes exhibited by nickel-base superalloys are a direct consequence of this damage development. However, there is little direct microstructural evidence of how such damage accumulates and, indeed, recent work indicates that tertiary creep in superalloys is due to changing intrinsic deformation kinetics rather than to the introduction of defects such as cavities or cracks (4,5).

The aim of the present study was to evaluate the effectiveness of HIP and heat treatment in removing creep damage and in extending the material life. Firstly, the accumulation of damage during long term creep tests is quantitatively characterised. The effects of rejuvenation procedures, applied at various fractions of the nominal life of laboratory creep specimens, on both the material microstructure and the cumulative performance of the alloy, are then assessed and related to the performance of material cut from service-exposed industrial gas turbine blades. Emphasis has been placed on relatively long term testing (up to $\sim 10^4$h at 850°C) of the cast nickel-base superalloy IN738LC in order to represent the service conditions of industrial turbine blades.

Experimental Procedures

The main experimental programme was carried out on IN738LC that was investment cast in the form of specimen blanks to tight specifications in order to minimise the extent of casting porosity. Other cast batches were also examined to assess the effects of higher levels of microporosity. All material was given the standard commercial heat treatment of 2h/1120°C/AC+ 24h/845°C/AC before testing.

Constant load creep tests were performed at 850°C on specimens with 50.8 mm gauge length and 7.6 mm gauge diameter. Strain was measured by capacitance transducers on extensometers attached to circumferential ridges defining the gauge length. Sets of tests at 170 and 250 MPa, which gave uninterrupted lives of $\sim 10^4$ and $\sim 10^3$h, were stopped after various elapsed times. Some specimens were slit longitudinally, polished to 1 µm grade diamond and examined by optical microscopy for evidence of damage. Other partially creep tested specimens were HIPped for 2h at 1180°C in an argon atmosphere at 170 MPa followed by reapplication of the conventional heat treatment. The gauge diameters of one set of HIPped specimens were reduced by 0.25 mm to correct minor distortions prior to continuing the creep tests with the original stresses; a second set was retested without this surface skimming.

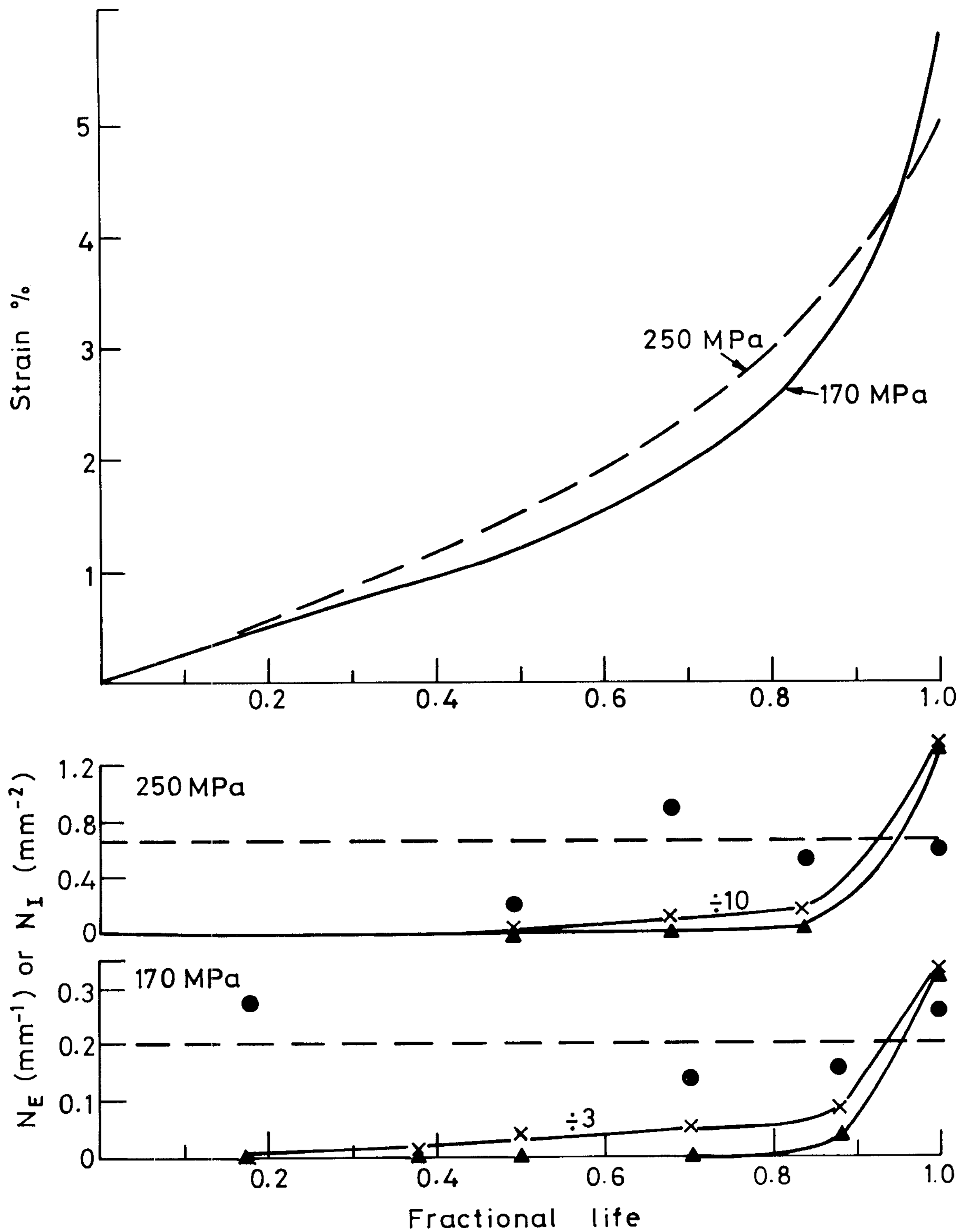

Fig. 1. a) Creep curves for IN738LC at 850°C with stresses of 170 and 250 MPa in terms of fractional life. Rupture lives were 8524 and 1038h respectively.

b) Creep damage measured on specimens stopped after various fractional lives

● – surface cracks

▲ – internal cracks (> 30 μm)

X – internal cavities and cracks (≳ 1 μm).

Industrial turbine blades of IN738LC that were retired after 23,000h of service at 800°C because of impact damage were examined metallographically for evidence of creep damage. Specimens prepared from spark machined rods cut from the blades were creep tested at 850°C and 250 MPa both with and without regenerative heat treatment.

Results

A series of creep tests with stresses of 170 and 250 MPa at 850°C were interrupted after various strains and examined for evidence of fracture damage. The full creep curves are shown in Fig. 1a. Two main categories of damage were apparent and these are described below.

a) Surface cracking invariably followed grain boundaries and there was evidence of environmental interaction (oxide, precipitate denuded zones). There was less surface cracking at the lower stresses, but in both cases these cracks formed at an early stage during creep, were relatively uniformly distributed along the specimen and did not develop rapidly.

b) Internal damage was also located predominantly at grain boundaries but was heterogeneously distributed and increased dramatically with elapsed strain. These features showed no evidence of environmental effects and varied in size from sub-micrometer cavities to quite long cracks formed by the linking of several cavities.

The development of both types of damage during creep is shown in Fig. 1b. The external damage, represented by the number of cracks per unit specimen edge length N_E, although readily visible, forms early during creep but, grows to about one grain depth then remains static. However, the visible internal damage N_I (number of cavities > 1 μm diameter per unit area of specimen) is at very low levels until late in the nominal creep life but then increases dramatically in the final ~ 10% of life. Typical microstructures indicating internal damage after various elapsed creep strain are shown in Figs. 2a-c. Removal of the external cracking by machining 0.25 mm from the gauge diameter after ~ 25% of the nominal life had no significant effect on the total rupture life.

HIP of creep exposed specimens removed most of the internal damage (Fig. 2d) but had little influence on the external cracking. The extent of internal damage, resolved by optical microscopy, after fracture of the specimens that had been interrupted for HIP was considerably less than in those tested conventionally (Fig. 2e). The damage accumulation during such composite creep tests is indicated in Fig. 3b.

Extensive examination by optical microscopy of the service exposed turbine blades showed little evidence of creep damage. A few isolated cavities (Fig. 2f) were identified but there were too few to give a significant value of N_I. However, large areas of the blades showed evidence of casting porosity.

A typical set of composite creep curves for specimens interrupted after 2.5% strain for HIP, both with and without surface skimming prior to retesting, and the corresponding measurements of internal damage are shown in Figs. 3a and b respectively. A full set of data is given in Table 1. In every case the specimens where 0.25 mm of the gauge diameter had been removed after HIP prior to retesting had significantly longer cumulative lives than those that were retested as HIPped. These data are shown graphically as plots of the cumulative life as a function of the percentage of nominal elapsed life prior to HIP in Fig. 4. The dashed line maps the ideal behaviour if the original properties were fully restored. HIP prior to testing, either in specimen or cast blank forms, had no significant influence on creep performance.

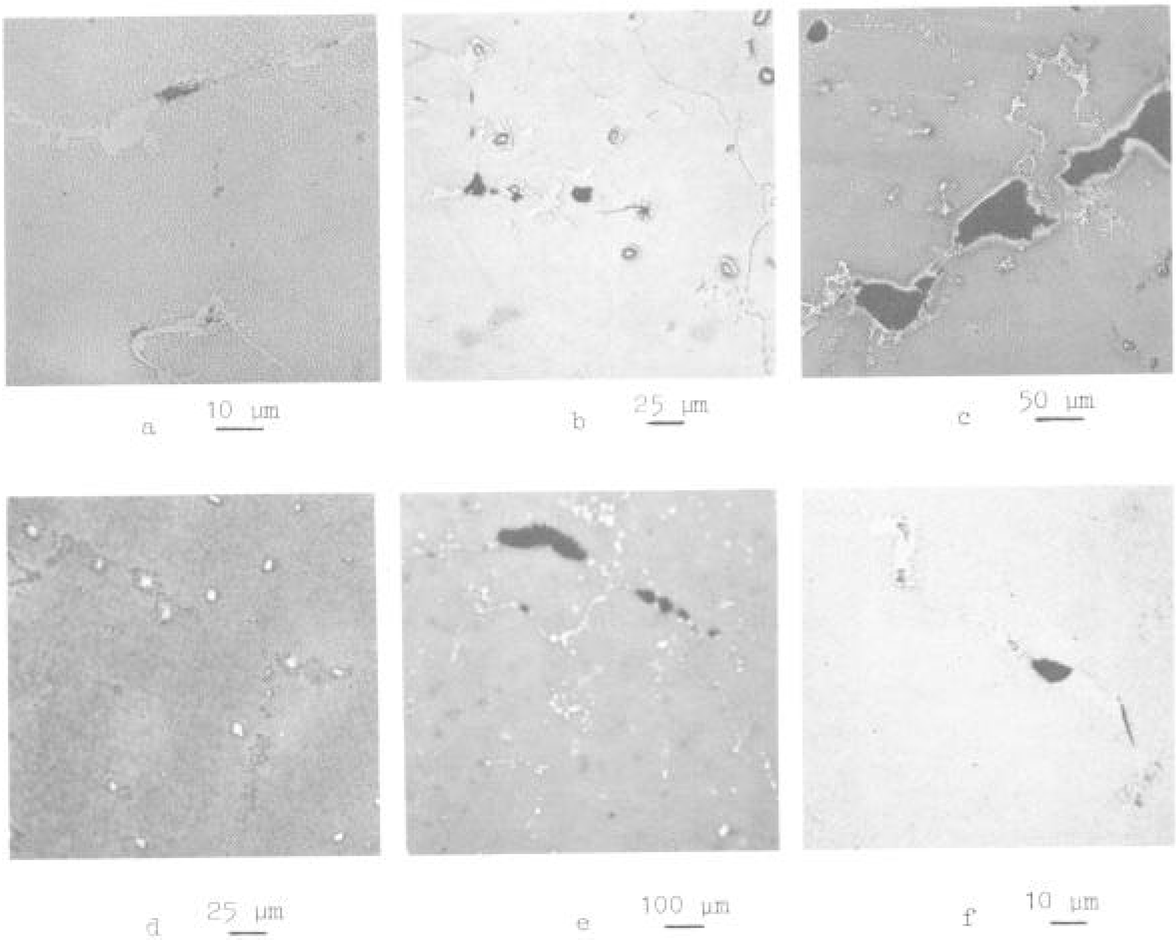

Fig. 2. Microstructures of IN738LC after various creep deformations, heat treatments and service exposures
(a) - (c) Examples of damage after various creep strains at 850°C/ 170 MPa
(a) 0.8% (b) 1.9% (c) 5.5% (fracture) (d) 3% + HIP
(e) 2.6% + HIP + 4% (fracture) (f) industrial blade after 20,000h at ~ 800°C.

The creep performance of the service exposed material was remarkably consistent showing little variation with position in the blade. Creep data for the original blade material are not available but the lives of 280 - 400h represent about 40% of those of the low porosity castings described above and of other commercial sources of the alloy; the level of cavitation observed would indicate that at least 70% of the original life should remain. Previous tests have shown that high casting porosity leads to low rupture lives (6) so that the intrinsic creep behaviour of the original blade material will almost certainly have been inferior to that measured in the laboratory assessment of low porosity castings. Reapplication of the conventional heat treatment to the service exposed blade material, either in bar or specimen form, gave the improvement in rupture life indicated in Fig. 5 which largely amounts to full restoration of the original properties, in view of the casting porosity. HIP processing is currently being carried out. However, since this will influence both casting porosity and creep damage it will be difficult to separate the rejuvenation component from the effect of consolidation of the original cast microstructure.

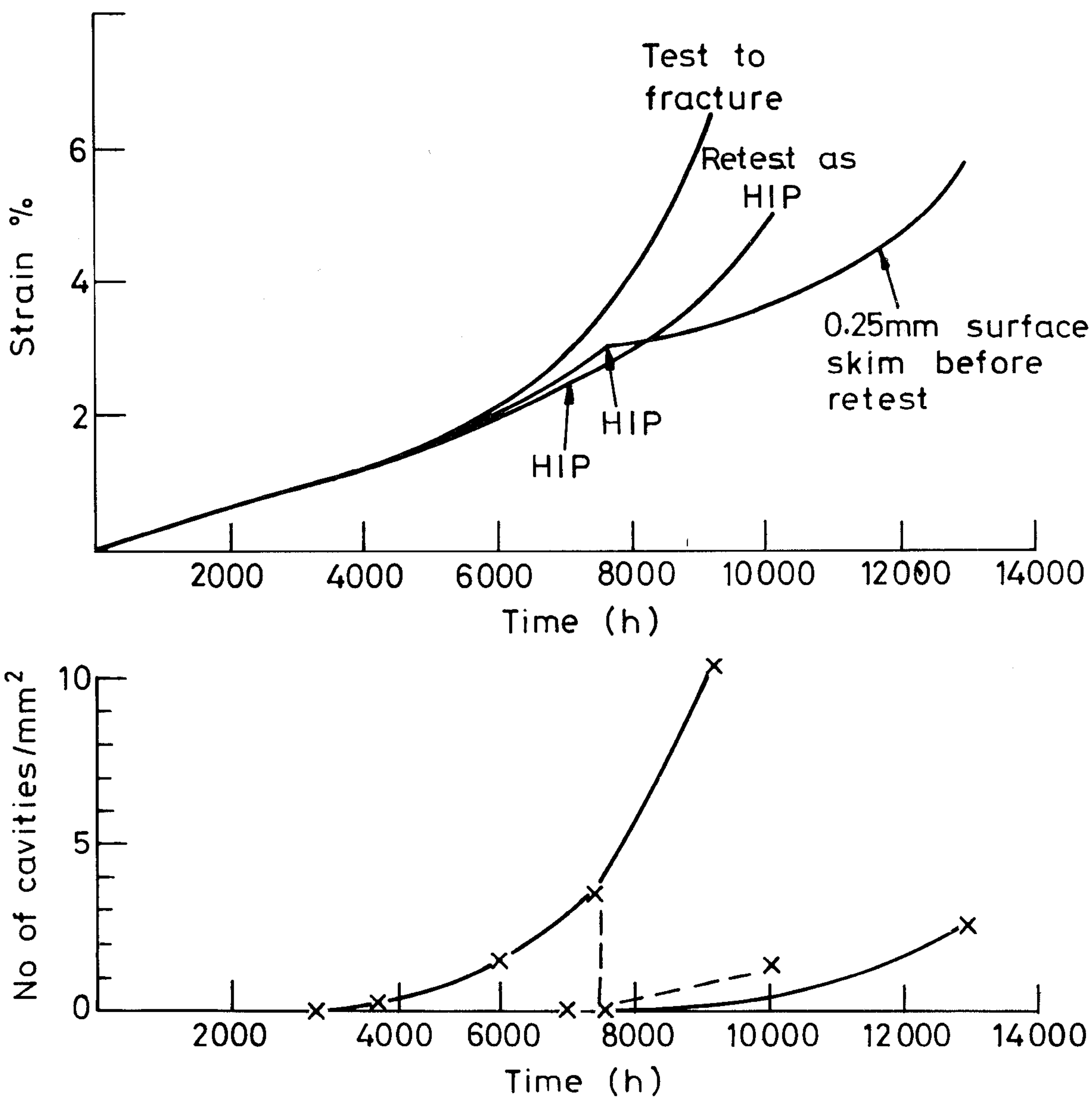

Fig. 3. (a) Composite creep curves for IN738LC tested at 850°C/ 170 MPa both with and without interruptions for HIP.
(b) Associated creep damage.

Discussion

The present results clearly show that although edge cracking is a very obvious feature of damage from the early stages of creep in IN738LC it does not, in general, cause fracture and, indeed, it becomes less prominent in low stress tests. Internal damage in the form of cavities and microcracks, on the other hand, is not apparent during the early stages of testing and only develops rapidly to cause failure in the final 10 - 15% of life. The

Table 1. Cumulative creep data for HIPped IN738LC tested at 170 MPa/850°C.

Spec. no.	Minimum* creep rate h^{-1}	Duration h	Ductility %		Comments
			Elong	RA	
H2A	1.5×10^{-6}	9194	6.4	10.00	test to fracture
H3A	0.65×10^{-6} 1.3×10^{-6}	1510 +7351=8861	0.36 +5.8=6.16	- 9.8	0.25mm removed retest to fracture
H6A	5.9×10^{-7} 4.2×10^{-6}	7104 +2912=10016	2.5 +4.4=6.9	2.5 +7.5=10	HIP, no machining retest to fracture
7DCH10D	2.4×10^{-6} 1.7×10^{-6}	7555 +5733=13288	2.6 +4.0=6.6	2.6 +8.9=11.5	HIP, 0.25mm removed retest to fracture
H7A	1.4×10^{-6} 4.4×10^{-6}	3651 +5173=8824	0.85 +3.1=3.95	0.8 +5.8=6.6	HIP, no machining retest to fracture
7DCH6D	2.9×10^{-6} 2.7×10^{-6}	3005 +7052=10057	0.80 +6.0=6.8	1.0 +11.0=12.0	HIP, 0.25mm removed retest to fracture
H8A	4.8×10^{-7} 3.2×10^{-6}	1652 +4836=6488	0.4 +3.2=3.6	- 4.6	HIP, no machining retest to fracture
7DCH5D	3.3×10^{-6} 2.5×10^{-6}	1505 +5756=7263	0.5 +7.4=7.9	0.6 +9.0=9.6	HIP, 0.25mm removed retest to fracture
H10A	2.4×10^{-6}	9300	5.5	8.8	HIP as test-piece test to fracture
H1E	2.2×10^{-6}	8485	5.7	7.0	HIP as cast-blank, manufacture-specialised test to fracture

* during each phase of testing

extent of grain boundary cavitation in the earlier stages is much too small to account for the long tertiary creep regime (Fig. 1a) in terms of increasing net stress through loss of section. Rather, the increasing creep rate is due to changing deformation kinetics, as discussed by Dyson and McLean (4) and Henderson and McLean (5).

HIP is very effective in eliminating cavitation damage and, together with reapplication of the conventional heat treatment, this is accompanied by some extension in creep life when the surface of the HIP/heat-treated gauge length is removed prior to retesting. However, the very large difference in density of internal damage at fracture observed by optical

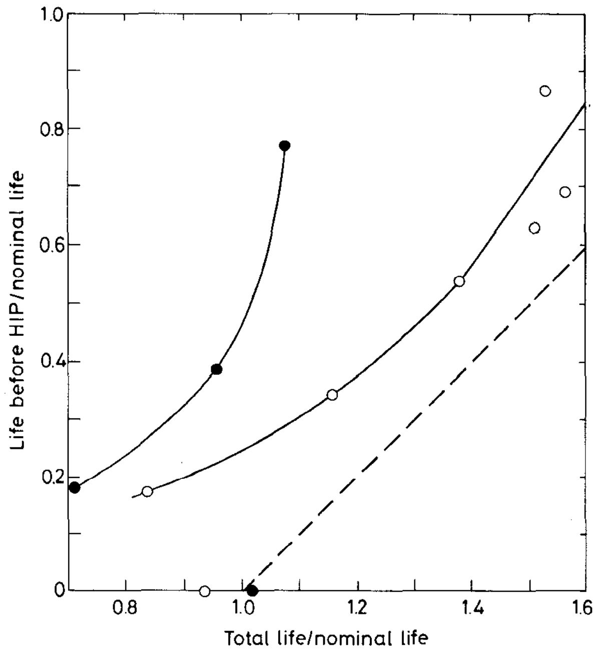

Fig. 4. Cumulative rupture life of IN738LC with intermediate HIP treatments after different elapsed fractional lives. (Tests at 850°C/ 170 MPa).

● - retest as HIP

0 - 0.25 mm skim

microscopy of HIPped and conventionally heat treated specimens casts doubt on a direct connection between the healing of cavities and extension of rupture life. Indeed Stevens and Flewitt (2) have previously shown that similar life extensions can be obtained in IN738 by re-heat-treatment without pressure.

Since the rate of deformation is an important factor both in the growth rate of cavities (eg. in constrained cavity growth (7)) and in the final instability leading to linkage of cavities, removal of cavitation will be of limited benefit unless the creep strength is also restored by HIPping. Where the damage is removed but the creep rate remains at a higher level than in the virgin material, the newly nucleated creep cavities may grow more rapidly than originally leading to accelerated fracture. However, examination of Fig. 3 and Table 1 shows that HIP/heat-treatment does reduce the creep rate to about the original level. Moreover, if the existing cavity density and subsequent nucleation were substantially reduced by HIPping, the post-HIP creep strain would be expected to be greater than the fracture strains of the un-HIPped specimens whereas the contrary occurs. It is more likely that the relatively low post-HIP fracture strains result from an increased density of sub-micrometer cavities that are not resolved by optical microscopy and that only cracks formed by the linking of cavities have been recorded.

The reasons for the poorer performance of the specimens that were retested in the HIP/heat-treated condition relative to those that had been surface skimmed were not obvious from metallographic examination. However, Woodford and co-workers (8,9) have shown that environmental effects during heat treatment can lead to significant reductions in both rupture life and ductility. In particular heat treatment of specimens of IN738 in air for 24h at 900°C led to a 25% reduction in rupture life; thus application of the commercial heat treatment directly to the specimen would be expected to give a similar deterioration in properties. A puzzling aspect of the present results is that the degradation does not occur when virgin alloy specimens

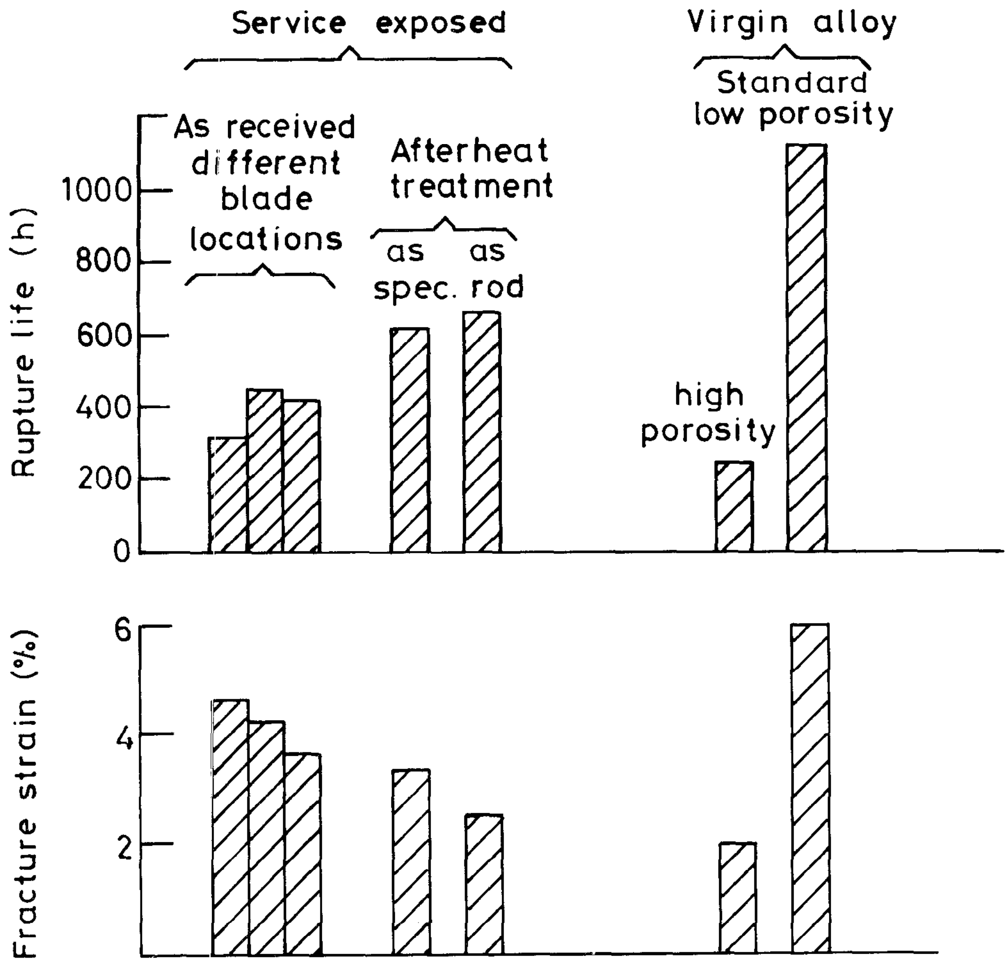

Fig. 5. Stress rupture life and ductility of IN738LC from a service exposed turbine blade and after re-heat-treatment. Data for virgin alloy are included for reference. (Tests at 850°C/250 MPa)

are HIP/heat-treated but requires some deformation to have occurred. Further work is required to establish whether any of the degradation occurs during the HIP cycle or if it is restricted to the heat treatment. Whatever the reasons, such environmental interactions will affect component performance after rejuvenation treatments and are likely to be particularly important for thin section parts.

The investigation of microstructure and properties of large service-retired turbine blades has confirmed that this component is subject to creep damage of the type observed in test specimens and that the creep properties can be recovered to a large degree by re-heat-treatment without recourse to HIP. Investigations of the effects of HIP are in progress and will be reported verbally; however, interpretation of these data must be ambiguous since HIP both removes creep damage and reduces the original level of casting porosity.

The present study only considers creep damage and shows that HIP gives no greater restoration of creep strength than has been previously reported for simple re-heat-treatment when this failure mode dominates. Other components, such as small aero-engine blades, are likely to be subject to considerable thermal fatigue damage where local elements of plastic strain, occuring during service cycles, can lead to the early formation of large numbers of cavities which Dyson et al (10) have shown to lead to substantially lower ductilities and rupture lives. The present work does not preclude the possible advantages of HIP in rejuvenating such components, but further work is required to establish the effects.

Conclusions

1. Creep fracture of IN738LC occurs by the development of internal grain boundary damage, largely in the final 10% of life, and not by the growth of surface cracks for the test conditions considered.

2. Creep damage observed by optical microscopy is effectively healed by hot isostatic pressing.

3. HIP only partially restores the original creep performance to a level previously obtained by re-heat-treatment without pressure.

4. Service exposed industrial turbine blades have limited creep damage; the original properties are restored by re-heat-treatment.

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References

1. M.H. Haafkens, p 931 in "High Temperature Alloys for Gas Turbines 1982", ed. by R. Brunetaud et al, Reidel Publ. Co., Dordrecht, Holland 1982.

2. R.A. Stevens and P.E.J. Flewitt, Proc. of ICSMA 5, vol 1, pp 439-441, Aachen, FDR August 1979.

3. J.P. Dennison, I.C. Elliott and B. Wilshire, p 671, in "Superalloys 1980", ed. by J.K. Tien et al, ASM, Metals Park, OH 1980.

4. B.F. Dyson and M. McLean, Acta Met. 31, 17 (1983).

5. P.J. Henderson and M. McLean, Acta Met. 31, 1203 (1983).

6. H.R. Tipler and M.S. Peck - Final report of programme UK 17 of Round 2 of COST50. NPL report no. DMA(A)33, June 1981, National Physical Laboratory, Teddington, UK.

7. B.F. Dyson - Canadian Metall. Qu. 18, 31 (1979).

8. D.A. Woodford and R.H. Bricknell, p 633 in "Superalloys 1980", ed. by J.K. Tien et al, ASM, Metals Park, OH 1980.

9. D.A. Woodford, Metall. Trans. 12A, 299 (1981).

10. B.F. Dyson, R.K. Varma and Y. Lindblom, unpublished work.