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MECHANISMS OF IMPROVING CREEP RUPTURE LIVES BY RE-HEAT-TREATMENTS

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INTRODUCTION

During creep of metals and alloys, fracture generally occurs by nucleation, growth and link-up of grain boundary cavities, forming intercrystalline cracks. A method of delaying fracture to improve creep lives has evolved from experiments with gold [1]. Samples were crept into the tertiary stage at 668K, when cavities were clearly detectable on grain boundaries. Annealing at 1273K for short periods moved the boundaries away from the cavities while longer annealing times sintered out the cavities. In both cases the original creep properties were fully restored by the annealing treatments. It was therefore proposed that improved lives of commercial creep-resistant alloys could be achieved by heat-treatment procedures aimed at eliminating the cavities [2, 3]. However, when a material is heat-treated after creep, several changes can occur:

i) The dislocation structure may be altered, recovering some degree of primary creep on reloading.

ii) Grain growth or recrystallization may isolate within grains any cavities developing on the original boundaries. Cavities so isolated will not continue to grow during further creep.

iii) Cavities may change shape by surface diffusion or may sinter out. Sintering may be inhibited by gas pressure within the cavity or contamination of the void surface by oxide formation.

iv) The particle dispersion may be changed by the heat-treatment procedure in alloys containing two or more phases.

The present analysis aims to rationalize the wide range of observations on removal of creep damage by re-heat-treatment in terms of the factors causing the onset of tertiary creep.

IMPROVEMENTS IN CREEP LIFE OF SUPERALLOYS BY HEAT-TREATMENT

With Nimonic 80A, which consists essentially of a 15 vol.% dispersion of $Ni_3(Al,Ti)$ or γ' precipitates and occasional carbide particles in a nickel - 20% chromium matrix, cavities develop during creep at 1023K [3]. For tests interrupted early in the tertiary stages, complete restoration of the creep properties was achieved by annealing at the creep temperature or at the ageing temperature (1093K). Indeed the creep life could be increased by a factor of four by repetitive creep/heat-treatment cycles,

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i.e. a sample was crept at 1023K for a time, t_a, late in the secondary stage, annealed at 1093K and the test continued for a further period, t_a, and so on. Although the cavities appeared to be removed by the annealing treatment, failure eventually resulted from loss of creep and fracture resistance due to extensive overageing [3].

With Nimonic 105 (having $\sim$ 40 vol.% γ') sintering treatments at or just above the creep temperature did not restore the creep properties and more complex heat-treatment procedures were found necessary to obtain improved lives [4]. Using repetitive creep/heat-treatment cycles, increased creep lives at 1123K were obtained by employing a heat-treatment schedule which dissolved the γ' phase (leaving the carbides out of solution to prevent grain growth [4]) followed by ageing to redevelop the initial particle dispersion. Creep lives could, again, not be extended indefinitely suggesting that some permanent damage was still being accumulated [4].

The ineffectiveness of simple annealing treatments was confirmed by Nimonic 115 [5]. Thus, for more complex alloys, prolonged creep lives were only obtained in repetitive creep/heat-treatment cycles by using more sophisticated re-heat-treatments, the detailed procedure depending on the solvus temperature of the γ' and carbide phases [4 - 6].

FACTORS CAUSING THE ONSET OF TERTIARY CREEP

The first indication of eventual fracture is usually the acceleration in creep rate at the onset of the tertiary stage of creep. This acceleration is, in general due to

a) the development of grain boundary cavities and cracks to a size sufficient to affect the deformation processes [7, 8] or

b) microstructural instability, such as grain growth or recrystallization in single-phase materials [9] or changes in particle dispersion during creep of two-phase alloys [10, 11].

A method of distinguishing between these causes of tertiary creep has been suggested [10, 12] depending on the measurement of the time to the onset of tertiary creep, t_t, and the time to fracture, t_f. For many single-phase materials, when tertiary creep begins as a result of cavity development (in the absence of grain growth and recrystallization), both t_t and t_f are inversely proportional to the steady creep rate [12] so that the ratio t_f/t_t is a constant ($\sim$1.5) over wide ranges of stress and temperature. When tertiary creep is associated with grain growth [9] or overageing [10], considerably larger t_f/t_t ratios are recorded. The relationships between t_t and t_f for the materials considered in this analysis are shown in Figure 1. Whilst $t_f/t_t = 1.5$ for gold [1] and Nimonic 80A [3], large values of this ratio are found for Nimonic 105 [4] and Nimonic 115 [5].

IMPROVEMENTS IN CREEP RUPTURE LIFE BY CAVITY SINTERING

For gold and Nimonic 80A, it appears that tertiary creep commences as a result of void development. With gold, no grain growth at recrystallization occurred during creep; cavities and cracks being readily detectable at the onset of tertiary creep. Also, with Nimonic 80A, when specimens were crept into the secondary stage, annealing at either the creep or

ageing temperatures delayed rather than promoted the start of tertiary creep on continuing the test [2], demonstrating that tertiary is associated with cavity development rather than overageing. Thus when cavity development causes tertiary (i.e. when $t_f/t_t = 1.5$) periodic removal of cavities extends the creep life. However, for this type of material, the extent to which the creep life can be improved by repetitive creep/heat-treatment programmes is limited since,

a) sintering treatments will not successfully remove cavities from late tertiary when interlinkage of cavities and cracks to the surface causes oxidation of the cavity surfaces, or when cavities are stabilized by gas pressure and

b) with materials such as Nimonic 80A, annealing at the creep or ageing temperature to eliminate cavities promotes overageing, causing gradual loss of creep and fracture resistance [10].

IMPROVEMENTS IN CREEP RUPTURE LIFE BY RESTORATION OF MICROSTRUCTURES

With Nimonic 105 and 115, heat-treatment schedules which dissolve and reprecipitate the γ' or carbides were needed to give increased lives in repetitive programmes [4, 5]. Metallographic examination of crept specimens before and after heat-treatment indicated that complete cavity sintering did not occur and the cavity incidence increased throughout repetitive tests giving a far higher incidence at fracture compared with that in uninterrupted tests. These observations demonstrate that improved lives in repetitive programmes with these alloys cannot be explained in terms of cavity sintering [4, 5]. An alternative interpretation can be advanced on the basis that when $t_f/t_t > 1.5$ (Figure 1), tertiary is caused by microstructural instability. For many complex alloys, cavities and cracks are only detectable just prior to fracture [13] indicating that rapid cavity and crack growth occurs late in the tertiary stage as increases in particle size and spacing take place. The improved lives obtained in repetitive programmes with Nimonic 105 and 115 [4, 5] can then be accounted for by the heat-treatment schedules maintaining, as far as possible, the original particle distribution. Thus, cavities grow at the rate expected for the initial structure rather than at the enhanced rates associated with overageing. In this way (although overageing can be prevented) periodic heat-treatments need not result in indefinite rupture lives since failure can occur by the gradual development of cavities.

Clearly, complete void sintering is not achieved even at the high temperature involved with the heat-treatment schedules used for Nimonic 105 and 115. Yet cavities can be removed simply by annealing at the creep temperature with gold and Nimonic 80A. This difference in sintering behaviour appears to be a consequence of gas stabilization [3, 14]. Oxygen does not dissolve in gold and the oxidation resistance of Nimonic 80A is extremely good. Conversely, the relatively poorer oxidation resistance of Nimonic 105 and 115, together with the higher creep temperatures used with these high strength alloys, may produce gas filled cavities which cannot sinter out.

CONCLUSIONS

Improved creep lives obtained by repetitive creep/heat-treatment programmes can be accounted for in terms of two types of material behaviour

a) When tertiary creep begins as a result of cavity development, (i.e. with materials having t_f/t_t ratios of ~ 1.5), improved lives can be obtained if the cavities can be sintered out periodically, the sintering behaviour depending on gas or oxide stabilization.

b) When tertiary creep begins as a result of microstructural instability, such as overageing (i.e. when $t_f/t_t > 1.5$), improved lives can be obtained, even without sintering out the cavities, by periodic heat-treatment schedules designed to minimize cavity growth rates by maintaining the original particle dispersion.

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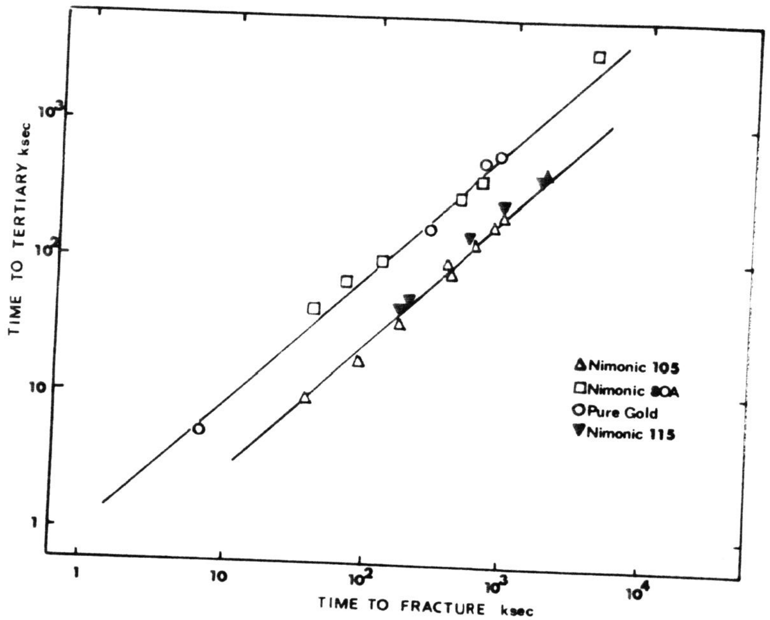

Figure 1 The Relationship Between the Time to the Onset of Tertiary Creep and the Time to Fracture for Pure Gold and Nimonic 80A, 105 and 115