The Effect of Metallurgical Instabilities on

The Behavior of IN 718

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Abstract

The effects of temperature, time and deformation on the mechanical properties and microstructural stability of IN 718 are considered. It is shown that τ " overaging, the formation of δ and oxidation are the primary factors that limit the high temperature use of this alloy. Recent studies in which stability issues are addressed are reviewed and it is pointed out that the stability may be significantly improved by chemical manipulation.

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Introduction

Alloy In 718 is and has been widely used as a disk material in modern jet engines. In this application it experiences temperatures up to 650 C, high stresses and strains in both the bore and dovetail regions and aggressive environments that include. The alloy generally possesses good creep, low cycle fatigue (LCF), and fatigue crack propagation (FCP) properties. The physical metallurgy of this system has been discussed extensively in this conference proceedings; the principal phases present in IN 718 are:

- 1. An FCC (γ) matrix
- 2. A strengthening precipitate $\gamma\,^{\shortparallel}$ (DO $_{22}) which is coherent with the matrix.$
- 3. The $\gamma^{\,\prime}$ precipitate (L $_{|2}\!)$ which is found in conventional Ni-base alloys.
- 4. The δ phase which like γ " has an ideal composition of Ni₃Nb but which has an orthorhombic crystal structure. δ phase occurs in either a blocky grain boundary constituent or as Widmanstatten plates in the matrix.
- 5. MC carbides, which may be present in the matrix and along grain boundaries.
- 6. M.C which may form as a boundary film during service.
- 7. Minor phases.

In the normally heat treated condition, the alloy consists of γ " precipitates in the γ matrix with modest amounts of the other phases mentioned above. During service, the structure of the alloy can change with some of the phases mentioned above becoming more prominent and influencing the mechanical properties and performance of components. In addition, structural changes such as twinning and oxidation also take place. The effect of these structural changes on the performance of IN 718 is considered in the body of this report.

The Effect of Long-term Exposure on Creep and Tensile Properties

Since In 718 is used in applications where exposure at elevated temperatures for extended times is required, the effect of long-term exposure has been recently studied [1]. vacuum-melted heats of IN 718 were prepared, heat-treated and subsequently exposed (no stress) for times of 1000 and 10,000 h at 650C and for 10,000 h at 575C. The microstructures had uniform grain structures (ASTM 8-11) and discrete grain boundary δ precipitates. While the grain size remained relatively stable, the δ did coarsen preferentially on grain boundaries for the material that was exposed at 650C for 10,000h. In addition, the γ " coarsened considerably at both temperatures for the 10,000 h exposure and there was some precipitation of γ' at 650C, consistent with what would be expected based on TTT diagrams. After exposure, tensile (RT & 650C) and creep properties (650C & 705C) were determined. It was found that the modest exposure at 575C actually increased the strength and ductility while the exposure at 650C for 10,000 h resulted in a decreased ductility. The 650C exposure for 1000 h had relatively little effect on the microstructure or on the tensile properties. In creep, the material that was exposed at 575C and tested at 705C at a stress of 415 MPa exhibited a longer life than the unexposed material, showing that the effect of exposure could be beneficial to the

engineering properties of IN 718. However, the material exposed at 650C showed a significant decrease in the life when tested under the same conditions, the decrease being most severe for the material that was exposed for 10,000 h. The major effect of the thermal exposure was reported to be weakening of the grain boundaries through the precipitation of δ . Since δ is rich in Nb, it changes the local grain boundary chemistry and creates conditions under which γ' forms in the grain boundary area. Since the γ' is apparently a less efficacious strengthener than the γ'' , a weaker grain boundary with lower creep lives and tensile ductilities will result.

Use of Phase Transformations to Improve Properties

The mechanical properties of IN 718 depend, to a significant degree, upon the morphology and size of the grains and the morphology and volume fraction of the $\gamma^{\prime\prime}$. These can be controlled through a knowledge of the time-temperaturetransformation behavior for the phases of interest which essentially represents the stability of the alloy. For example, it has been shown [1] that the volume fraction of r" is very sensitive to the cooling rate from the annealing temperature with the "equilibrium" volume fraction being reached only after 100 h for water quenching and air cooling. (While classical Ni base alloys generally very rapidly attain a constant volume fraction of γ' , instances have been reported for very rapidly cooled material in which considerable times were required to attain "equilibrium" concentrations [2].) The implications of this are obvious inasmuch as the properties in service will change. This effect was seen for the alloys that were subsequently maintained at 575C for which the strength increased. The increase in strength is directly attributable to the increase in the amount of Y".

Use has also been made of the γ " and δ phase transformations to produce microstructures that have superior FCP properties. For example, one study was carried out to characterize the effects of γ " and grain size [3]. Large grains were produced by annealing above the δ solvus (determined to be 1000C by metallographic techniques) at 1093C for 1 h and water quenching. In this case the δ was essentially totally dissolved and the grain boundaries were no longer pinned. The grain size obtained by this treatment was approximately 250 microns. To obtain fine grained material, annealing was carried out at 968C for 1 h with subsequent water quenching. The grain size for this treatment was about 20 microns. Ageing was carried out at 718C for 5h and for 200h with subsequent water quenching to obtain very fine and very coarse γ ".

When tested at 427C it was found that a spectacular decrease in the FCP rate was obtained for coarse-grained materials in the underaged condition. The reasons for the improvement were related to differences in the fundamental deformation mechanisms and not to "extrinsic" effects such as crack closure.

Another study [4] a heat-treatment was employed such that solution annealing was carried out below the grain coarsening temperature (1032C). The material was subsequently slowly cooled to 843C and held for 4h to produce serrated grain boundaries through precipitation of grain boundary δ . Evidently there was a dynamic interaction between the growing grains and the

precipitating δ such that the grains were locally pinned in some areas and free to grow in others. (A similar morphology has been produced in IN 792 through warm working and simultaneous precipitation of $\gamma'[5]$.) Following this the material was partially solution treated at 926C for 1h to eliminate any coarse γ " that had previously formed. The heat-treatment was concluded with a standard double age. While this modified heat treatment does decrease the tensile strength (presumably because of the Nb that is tied up in the form of δ as opposed to the strengthening precipitate γ "), there were significant improvements in the FCP properties when tested at 650C (approximately a factor of 2 decrease in the FCP rate). This improvement was attributed to the authors to be due to a significant reduction in the amount of grain boundary sliding that occurs for serrated grain boundaries [6] since at this temperature it is probable that there was a significant component of creep crack growth during the fatigue cycle. However, as discussed below, the boundaries also contained a significant amount of δ which could reduce the environmental sensitivity by providing traps for oxygen.

The Effects of Deformation and Environment

In a sense, environmental damage may also be considered to be a form of metallurgical instability since the basic structure of the alloy may be altered, either on an atomic scale where no new phases are formed but where properties are none-the-less altered as a result of different atomic species being present. The importance of the environment was pointed out by Kane and Floreen [7]. They compared the rate of creep crack growth in He and in air at 650C and found that in a He atmosphere the rate of crack growth was at least an order of magnitude lower than in air. In both air and He voids were observed to form and cracking was transgranular indicating that the same creep damage mechanism The authors drew the conclusion that the effect was operative. of the environment could be to accelerate the rate of cavity formation although the mechanism by which this could occur was not specified.

The effects of environment on FCP have also been considered and it is generally accepted that the FCP rate at elevated temperatures is very sensitive to environmental factors although the precise mechanism of environmental acceleration is not fully understood. It has been shown that raising the temperature from 25C to 550C results in a significant increase in the FCP rate especially at stress intensities below approximately 20 MPa(m) $^{1/2}$ At these stress intensities it was found that at both temperatures the deformation substructure consisted of intense bands in which twinning had taken place. As the test frequency decreased, the amount of intergranular cracking and twinning also increased so the effect of the formation of deformation twins was not to intrinsically damage the material but to accelerate environmental grain boundary damage. There is thus a negative synergistic effect between the deformation mode and the effects of environment. Such effects have been observed in classical Ni base alloys and have been used as the basis of a life prediction methodology [9]. The predominant effect of the environment on the FCP properties of IN 718 has been clearly documented in a study in which different microstructures (e.g. small grain, large grain, and necklace) were used [10]. In air at 650C, the microstructure exhibited a crack necklace growth

approximately 3 orders of magnitude lower than the fine grained microstructure. However, when the same microstructures were tested in vacuum at the same temperature, there was essentially no difference in the FCP behavior. Such results would tend to indicate that the main effect of different microstructures on FCP is to provide different degrees of susceptibility to environmental attack.

As stated earlier, the precise mechanism of such attack and degradation is only poorly understood. It has been hypothesized that oxidation in IN 718 may occur in a two stage process [11]. In the first stage an oxide of the form a spinel (NiFeO) is produced which is not very protective. In the second stage the more protective Cr_2O_3 . Such a mechanism has been used to explain why FCP occurred below the threshold at 650C for short hold times (~30s) and also why there was saturation of this effect at longer hold times (~1000s) corresponding to the formation of the Cr-rich oxide.

The effects of δ on the properties may also be subtle. is known that deformation of IN 718 at elevated temperatures promotes the formation of δ platelets. Presumably the deformed regions in the γ " act as nuclei for δ . The local re-ordering of the γ " lattice along with the similarity of the δ and γ " compositions should provide for a rapid nucleation event as well as rapid growth since only short range diffusion would be required to convert γ into δ . The formation of "deformationinduced" δ will naturally reduce the strength since Nb is drawn from the principal strengthening phase, $\gamma^{\,\prime\prime}\,.$ The effect of δ is not entirely negative however, since at the grain boundaries it can provide some oxidation resistance, either through its intrinsic oxidation resistance or by providing oxygen traps through an increase in the boundary area. Such a mechanism would tend to reduce the effective diffusion rate and to require higher effective concentrations of oxygen in order to form deleterious The ready formation of δ by deformation processes at 650C and above is a major consideration in limiting its use to temperatures below 650C.

Alloying for Stability

From the preceding discussion it may be concluded that a reasonable goal of an alloy improvement program would be to decrease the amount of \hat{o} that forms during extended exposure and to increase the stability of the γ " and/or γ '. The results of such a program are described in a recent article [12] in which the effects of the (Al + Ti)/Nb and Al/Ti ratios were studied. It was shown that by increasing these ratios the length of the γ " was reduced by a factor of 2. The reason for this was attributed to the increased misfit between the precipitate phases The γ' acts as a heterogeneous nucleating and the γ matrix. agent for the γ ". In addition, the increased Al will increase the amount of γ' and since Nb can partition to the γ' , there will be less Nb available for the formation of γ ". An alternative explanation, in view of the results quoted in [1], might be that the rate at which γ " forms its equilibrium volume fraction is reduced significantly and the alloys are "effectively" stable to long term exposure. In addition to the more stable γ ", the tendency to form δ was greatly reduced. It was shown that after 1000h at 732C conventional IN 718 contained significant amounts

of δ whereas the modified compositions having showed only minimal amounts of δ . This was explained on the basis of a reduced amount of γ " and a correspondingly reduced area for nucleation of δ as a result of the smaller γ " particles. A study of the kinetics of δ formation in the modified alloys showed that the rate of δ formation was greatly reduced. The mechanical properties of such alloys showed significantly improved long term stress rupture and tensile properties although the fatigue properties were not quoted.

Concluding Remarks

Like all alloys used at high temperatures, IN 718 is metastable with respect to overaging of the primary strengthening precipitate. In addition, undesirable phases such as δ may form with exposure and the rate of formation of such phases may be greatly accelerated by deformation debris. The stability of $\gamma^{"}$ and reduction in the amount of δ can be obtained through appropriate chemical manipulation.

The fatigue and creep properties of IN 718 are greatly affected by the grain size and structure as well as by the form and distribution of the γ ". In general, excellent creep and FCP properties can be obtained from large grained materials.

IN 718 is particularly susceptible to environmental attack which may be in turn accelerated by the deformation mode. It has been hypothesized that much of the environmentally sensitive behavior may be understood in terms of the formation of a two stage oxidation process in which a relatively unprotective oxide forms initially, followed by the formation of a protective Crrich oxide.

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