EFFECT OF THERMAL-MECHANICAL TREATMENT ON THE FATIGUE CRACK PROPAGATION BEHAVIOR OF NEWLY DEVELOPED ALLVAC® 718PLUS™ ALLOY

Xingbo Liu¹, Jing Xu¹, Nate Deem¹, Keh-Minn Chang¹, Ever Barbero¹, Wei-Di Cao², Richard L. Kennedy², Tadeu Carneiro³

¹West Virginia University, Morgantown, WV 26506-6106, USA

²Allvac, an ATI Technologies Company, Monroe, NC 28111-5030, USA

³Companhia Brasileira de Metalurgia e Mineração, São Paulo - SP 04552-902, Brazil

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Abstract

The newly developed Allvac® 718Plus™ alloy has shown mechanical properties superior to alloy 718 and comparable to Waspaloy at the temperature up to 704°C. Previous results showed that fatigue crack propagation (FCP) resistance without holding time has no significant difference between three alloys with 718Plus being the best and 718 the lowest. During the hold-time FCP tests, 718Plus shows comparable results to those of Waspaloy and better than Alloy 718. In this paper, the effect of various thermal-mechanical treatments, including direct aging, pre-treatment, and long-term exposure, on the hold-time fatigue crack propagation behavior of 718Plus alloy were investigated and the results are summarized as follows:

(1) The hold-time fatigue crack propagation rates (FCPRs) of DA sample are almost the same as that of the alloy after conventional solution plus age heat treatment, while the FCP of DA sample is slower than conventional 718Plus alloy under the 650°C, 3S loading condition; (2) the long-term exposure tests show^o that the alloy's hold-time fatigue cracking resistance is improved after exposed at 760°C for 350 hours; (3) the fine-grain alloy shows better hold-time FCP resistance than that of coarse-grain alloy, which is attributed to the delta-phase effect; and (4) it is indicated that after pre-treatment at 857°C for up to 24 hours, the alloy's fatigue cracking resistance is improved because of delta-phase at grain boundaries.

Introduction

In recent years, a new Ni-base superalloy, Allvac 718Plus, has been developed to meet the objectives of increasing the temperature capability 55°C higher than that of alloy 718 and with comparable processing characteristics as alloy 718 [1]. The nominal chemical composition of the alloy is listed in Table I, as compared with that of alloy 718 and Waspaloy:

Table I: Chemical Composition of Allvac 718Plus in Comparison with Alloy 718 and Waspaloy

Alloy	C	Ni	Cr	Mo	W	Co	Fe	Nb	Ti	Al
718Plus	0.020	Bal.	18	3.0	1.0	9.0	10	5.4	0.7	1.4
718	0.025	Bal.	18	3.0	--	--	18	5.4	1.0	0.5
Waspaloy	0.035	Bal.	19.5	4.2	--	13.0	--	--	3.0	1.3

Extensive studies demonstrated that this alloy has shown superior tensile and stress rupture properties to alloy 718 and comparable properties to Waspaloy at the temperature up to 704°C [2, 3]. However, relatively speaking, the data on fatigue crack propagation (FCP) resistance of this alloy are still insufficient. Large civil fixed wing aircraft, and many military aircraft are designed using Damage Tolerance (DT) approach such as in accordance with Civil Aviation Authority (CAA) and Federal Aviation Authority (FAA) regulatory requirements. The DT design philosophy states that a component or structure enters service with pre-existing flaws or cracks. Linear Elastic Fracture Mechanics (LEFM) theory is used to predict the rate of crack growth. The crack growth rate must be shown to produce cracks of sufficient size that they can be detected by periodic inspection before they reach a dangerous size for the safety of the component or structure [4]. Therefore, it is critical to study the FCP behaviors of Ni-base superalloys.

Figure 1: Fatigue crack propagation behaviors of 718PLUSTM alloy at 650°C, as compared to that of Alloy 718 and WASPALOY

FCPRs of 718Plus alloy at conventional solution plus age condition were previously evaluated for test frequency of 0.33Hz and with 100 seconds hold at maximum load, as compared to that of Alloy 718 and WASPALOY[5]. Figure 1(a) shows that alloys 718Plus, 718 and Waspaloy have similar fatigue crack growth rates under 3 seconds triangle loading at 650°C with 718Plus being slightly better. As indicated in Figure 1(b), WASPALOY shows the best resistance to fatigue crack growth under hold time fatigue condition while the resistance of 718Plus is better than that of Alloy 718.

The objective of this paper is to investigate the effect of various thermal-mechanical treatments, including direct aging, pre- solution treatment, and long-term exposure, on the hold-time fatigue crack propagation behavior of 718Plus alloy.

Experiments

Materials

The 718Plus alloy employed in this investigation was provided by ATI Allvac . (Monroe, NC). The alloys melted by VIM+VAR (1.5 ton) and forged to 200 mm round billet, and then various thermal-mechanical treatments were undertaken. The samples used in this study are summarized in Table II.

Table II Tested Samples

Heat Treatment	CH		CH+LE		DA	PST+CH
Grain Size	8-10	4-6	8-10	4-6	8-10	4-6
Sample No.	A1	B1	A2	B2	A3	B3

Note: CH – Conventional Heat treatment: 955°C/1hour/AC + 787°C/2hour/FC

→650°C/8hour/AC

LE – Long-term Exposure: 760°C/350hour

DA – Direct Aging

PST- Pre-solution treatment: a special heat treatment to precipitate delta phase before CH

Fatigue Crack Propagation Test

FCP tests were performed by employing the single-edge-notched (SEN) specimens, see Figure 2. The specimen was pre-cracked up to 2.54mm with lower stress intensity triangle cycles at room temperature, and then was heated to higher temperature for the fatigue crack growth tests. The tests were carried out at high temperatures and with different loading cycles (3 seconds triangle wave, trapezoid wave with 3 seconds + 100 seconds loading at maximum stress). The tests were conducted under constant load control. The R ratio (K_{min}/K_{max}) was always set to 0.1.

Fractographical analyses were conducted by means of SEM

Figure 2: Dimensions of the Single-Edge Notched Specimens

The increment of crack length during fatigue and sustained crack growth tests were monitored continuously by using a dc potential drop technique. While a constant dc current was passed through the specimen, the crack length was monitored by a pair of potential probes mounted on the front edge of the specimen across the pre-machined notch. The measured dc potential drop at any crack length was normalized and converted into the corresponding crack length by a single analytical relation, namely Johnson's equation:

$$A = \frac{2W}{\pi} \times \cos^{-1} \left[\frac{\cosh(\pi Y / 2W)}{\cosh \{ (U / U_0) \cosh^{-1} [\cosh(\pi Y / 2W) / \cos(\pi A_0 / 2W)] \}} \right] \quad (1)$$

where A and A_0 represent the actual and initial crack lengths, U and U_0 are updated and initial measured potential drops, and W , Y represent the specimen width and one half of the potential probe span respectively.

Many different mathematical expressions for the stress intensity of a SEN specimen are available, and we use Tada's empirical equation for its accuracy over a wide range of crack length.

$$K = \frac{P}{\sqrt{W}} \frac{\sqrt{2 \tan \theta}}{\cos \theta} \left[0.752 + 2.02 \left(\frac{A}{W} \right) + 0.37 (1 - \sin \theta)^3 \right] \quad (2)$$

where P represents the applied load, θ equals to $(\pi/2) \times (Y/W)$. W , Y and A have the same meaning as in Eq.(1). The accuracy of this equation has been confirmed to be better than 0.5% providing that no bending moment is applied.

Experimental Results

FCP of 718Plus with Various Grain Sizes

FCPRs of 718Plus alloy with various grain sizes were evaluated for test frequency of 0.33Hz and with 100 seconds hold at maximum load. It is indicated by Figure 3 that the FCPRs of groups B1 and A1 are almost the same, while FCPRs of B1 are slightly lower than that of group A1 for the tests without hold-time at 650°C and 704°C. It is also shown in Figure 3 that on the contrary to the results of the tests without hold-time, for the tests with 100 seconds hold-time, the FCPs of fine-grained A1 are slower than that of B1 at 650°C.

As shown in Figure 4 (a), examination of the fatigue fracture surfaces by SEM revealed transgranular crack propagation with striations for group A1 at room temperature. The fracture mode of A1 at 650°C is the mixture of intergranular and transgranular modes. When the temperature is up to 704°C, intergranular is predominant mode

and the whole surface covered by a layer of oxide film, indicating that extensive oxidation happens during the crack growth. Figure 5 shows the fracture surfaces of group B1 after 3-second FCP tests at room temperature, 650°C and 704°C. It is evident that the fracture mode of group B1 is intergranular with well-defined fatigue striation at the temperatures up to 650°C. It is also interesting to find the deformation pattern of twins. As seen in Figure 5 (c), the slip bands in matrix could not pass through the twin, they stop at twin boundary. At the same time, the fracture surface of the twin shows clear river-like cleavage pattern, which indicates there is another

Figure 3: Fatigue Crack Propagation Rates of A1 & B1

different slip system operating than the ones in matrix. It also can be found that the slip bands in matrix lead to several micro-crack in the twin. Figure 5(d) indicates that the fracture mode of B1 at 704°C is full intergranular.

(a) A1- Room Temperature

(b) A1 – 650°C

(c) A1 – Benchmark between RT and 650°C

(d) A1 – 704°C

Figure 4: SEM fracture surface micrograph of A1 after FCG tests with the frequency of 0.33Hz at Room Temperature, 650°C and 704°C.

FCP of DA alloy

Figure 6 shows the fatigue crack propagation of DA 718Plus, as compared to the alloy after conventional heat treatment. It is indicated that the DA sample has slower fatigue crack growth rate than that of the alloy after conventional heat treatment under the 650°C, 3S condition. Under the 3+100S hold-time fatigue condition, the DA and conventional alloys have almost the same FCP rates.

(a) B1 – Room Temperature

(b) 650°C

(c) B1 – Room Temperature

(d) B1 – 704°C

Figure 5. SEM fracture surface micrograph of B1 after FCG tests with the frequency of 0.33Hz at Room Temperature, 650°C and 704°C.

SEM fractographical analysis (Figure7) show that the alloy obtained elongated grain structure after DA treatment, as expected. The fracture mode of DA alloy after 650°C/3S test is a mixture of intergranular and transgranular modes, while the alloy after 650°C/3+100S test shows a predominantly intergranular failure.

The fatigue cracking resistance of DA superalloys are generally different from their conventional counterparts. As proposed by Lynch etc.[6], the possible differences in microstructure between DA 718 and conventional 718 which could be responsible for the cracking resistance include: (1) the volume fraction, morphology and distribution of carbides and delta phases; and (2) the grain size, shape, and morphology, and grain-boundary-misorientation distributions; rate of grain-boundary diffusion, extent of grain-boundary sliding, and degree of crack branching could be affected. Since most of the microstructural features are inter-related so that it is difficult to establish their relative importance for the alloys' cracking resistance.

Figure 6: FCP of DA 718Plus, as compared to the alloy after conventional heat treatment.

(a) A3, 650°C, 3S

(b) A3- 650°C, 3+100S

Figure 7: SEM fracture surface micrograph of A3 (DA) after FCP tests

Long-term Exposure Effect on FCP

The microstructure of most of Ni-base superalloys will be changed after long-term service at high temperatures. In general, there are three major changes in the alloys: (1) grain growth; (2) coarsening of strengthen phases; (3) the transformation of meta-stable phase to equilibrium phase, for instance, gamma double-prime (DO22) to delta-phase transformation. In accordance to the microstructural evolution, the mechanical properties will be changed. Typically, the alloys will lose part of their strength and ductility after long-term exposure.

In this investigation, alloy 718Plus thermal-stability was studied by exposing the alloy at 760°C for 350 hours. Figure 8 shows the alloy's FCP, as compared to the alloy after conventional heat treatment. It can be seen in the figure that long-term exposure does not change the fatigue cracking resistance under 650°C/3S condition. On the other hand, the hold-time fatigue cracking resistance of the alloy was improved by the long-term exposure, which means that this alloy has good long-term structural stability. The reason for this phenomenon needs to be further investigated.

Figure 8: FCP of the alloy after exposure at 760°C for 350 hours, as compared to the alloy after conventional heat treatment.

Discussions

Hold-Time FCP & Grain Size Effect

Fatigue crack propagation is generally attributed to the damage in front of the crack tip. In the cycle dependent "pure" fatigue, the material in front of the crack tip is damaged only by cyclic loading. However, if the test is conducted at higher temperature and there is hold-time at max load in fatigue cycle, the time-dependent behaviors must be taken into account. Damage by cyclic loading accounts for a very less amount of the total damage of materials. Considering the hold-time fatigue, during the hold-time at maximum loading, the material in front of the crack tip is damaged by the diffusion of oxygen and creep, and the resistance against cracking is lowered. During the next unloading and loading, the crack will pass through the damage zone and result in further crack growth. Then the cycle repeats again. This kind of crack growth is obviously time-dependent. The size of damage zone represents the resistance of materials against the crack growth. It should be pointed out that the existence of *Damage Zone* in several Ni-base superalloys has been confirmed by a specially designed testing method and the *Damage Zone* size can be used to evaluate the superalloys' resistance to hold-time FCP [7].

Creep and grain boundary oxygen diffusion/oxidation, are accepted to be the two primary reasons for time-dependent FCP of superalloys, and both are closely related to the grain size and grain boundary morphology of the alloys. It is generally agreed that coarse-grain alloys should have better resistance to time-dependent FCP than that of fine-grain alloys, although grain size has little effect on cycle-dependent FCP. For instance, Yuen's study [8] shows that by coarsening the grain size from 22 to 91 microns in alloy 718, near threshold crack growth rates were reduced and ΔK_{th} (threshold stress intensity range) values increased. Roughness-induced crack closure explains the influence of grain size. The coarse-grain material resulted in much rougher fracture

surfaces at near threshold growth rates. From grain boundary oxygen diffusion/oxidation point of view, increasing grain size can reduce the total area of grain boundaries, and therefore, reduce the time-dependent fatigue crack propagation rates of Ni-base superalloys

The results of this investigation show that on the contrary to “common sense”, fine-grain 718Plus alloy (B1) has better resistance to time-dependent FCP, which can not be explained by above mentioned mechanisms. Further investigation indicates that the forging route to achieve the coarse grain also reduces the grain boundary delta phase precipitation, and this caused the reduction of the alloy’s hold-time FCP resistance.

Delta-Phase Effect

Since both delta-phase and gamma double-prime contain Nb, in the early days delta-phase was considered “detrimental” phase in superalloys because the precipitation of delta phase may reduce the alloys’ strength by consuming Nb. However, recent investigations show that globular delta-phase distribution along the grain boundaries has retardation effect on grain boundary crack propagation at creep/fatigue interaction condition and the existence of reasonable amount of delta-phase can improve stress rupture ductility of alloy 718 [9]. In addition, delta-phase has been used to reduce the growth rate of grain size by pinning the grain boundaries. The grain size of delta-processed (DP) alloy 718 is below ASTM 11. As the results, the mechanical properties of DP 718 were shown to be better than conventionally processed In718 and Super Waspaloy at low temperature [10]. From the environmental effect point of view, the delta phase in alloy 718 was found to decrease environmental effects due to its intrinsic oxidation resistance, by trapping oxygen at delta-matrix interfaces, or by depleting Nb in the matrix so that fewer NbC particles are present [11].

(a) A1 – 650°C, 3+100S

(b) B1 – 650°C, 3+100S

Figure 9: SEM fracture surface micrograph of A1 and B1 after tested at 650°C, 3+100S.

The beneficial effect of delta-phase on FCP resistance of 718Plus alloy was revealed by this investigation. Figure 9 shows the fractographical pictures of A1 (ASTM 8-10) and B1 (ASTM 4-6) after fatigue tested under 650°C, 3+100S condition. It is indicated that both samples show intergranular failure under this testing condition. However, further investigation on these micrographs revealed that there were numerous delta-phase precipitated along with grain boundaries of A1 sample, while the grain boundaries of B1 are “clean” with very few delta-phase. As mentioned in the above section, grain size effect cannot explain the difference of hold-time FCPRs between A1 and B1. Therefore, the better resistance of B1 to hold-time FCP can be safely attributed to the grain boundary precipitation of delta-phase.

To confirm the above discussion on delta-phase effect and to improve the alloy's hold-time FCP resistance by the optimization of delta-phase distribution, a pre-solution treatment (PST) at 857°C for up to 24 hours was added before the conventional heat treatment of the alloy. Figure 10 shows the PST alloy's FCP, as compared to that of conventional alloy. It can be seen in the figure that the steady state (stage II) FCP of PST alloy are the same as conventional alloy under 650C, 3S condition. However, pre-treatment does improve the hold-time FCP resistance of the alloy under 650C, 3+100S condition. Although the PST alloy has the same crack growth rate when ΔK close to ΔK_{th} , the slope of the crack growth curve for PST alloy is much lower than that of conventional alloy. The other effect of pre-treatment is the offset of stage III FCG under 650C/3S condition. Stage III FCG of the alloy with pre-treatment appears later than that of conventional alloy, which results in longer fatigue cracking life of the alloy.

Figure 10: FCP of the alloy with pre-solution treatment at 857°C, as compared to the conventional alloy.

Summary and Conclusions

The effect of various thermal-mechanical treatments, including direct aging, pre-solution treatment, and long-term exposure, on the hold-time fatigue crack propagation behavior of 718Plus alloy were studied and several conclusion can be drawn from the investigation:

1. The alloy shows elongated grain structure after direct aging. The hold-time fatigue crack growth rates of DA sample are almost the same as that of the alloy after conventional solution plus age treatment.
2. The long-term exposure tests show that the alloy's hold-time fatigue cracking resistance was improved after exposed at 760°C with 350 hours, which means that this alloy has good long-term structural stability.
3. The fine-grain alloy shows better hold-time FCP resistance than that of coarse-grain alloy, which is most likely attributed to the delta-phase effect.
4. The result shows that after pre-solution treatment at 857°C, the alloy's fatigue cracking resistance is improved because of delta-phase at grain boundaries. The optimum distribution of delta-phase at grain boundary needs to be determined by further study.

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