Effect of Multiple Postweld Heat Treatment on the Weldability of Alloy 718

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Abstract

This paper describes the effect of multiple postweld heat treatment thermal cycles on the weldability of both cast and wrought Alloy 718. The liquation cracking susceptibility of specially-melted alloys containing a range of niobium and boron contents was evaluated using the hot ductility test. Cracking susceptibility increased as a function of both composition and thermal cycling for the cast materials, while the wrought materials showed only a small influence of composition and effectively no deleterious effects from the multiple thermal cycles.

Introduction

Nickel-base superalloys, and especially Alloy 718, are frequently used in high performance turbine engines. This is due to their ability to withstand severe mechanical stressing at elevated temperatures. Because of the nature of a turbine engine, many of the components are subjected to extensive wear and fatigue cracking, requiring the frequent use of repair welding to restore structural integrity. The high cost of most turbine engine components makes it more economical to repair rather than replace. Precipitation-hardenable “superalloys”, such as Alloy 718, must be heat treated after repair to restore the properties of the surrounding material. It has been observed that after an accumulation of repair/PWHT events the weldability of the material degrades and that multiple repairs may be required during a given engine tear down to restore the component to a serviceable (i.e. crack free) condition. As a result, the material will be subjected to a large number of heat treatment cycles during its lifetime. Previously, little attention was given to the influence of the number of heat treatment cycles on Alloy 718 weldability, despite the fact that there seems to be a definite relationship between the number of repair events and cracking susceptibility [1]. To study the influence of multiple heat treatments on weldability, and specifically HAZ liquation cracking, hot-ductility tests were performed on three specially-melted heats of Alloy 718 with low, medium, and high levels of niobium and boron. These materials were tested in both the cast and wrought conditions.

Procedure

Three heats of Alloy 718, with controlled levels of niobium and boron, were vacuum melted in nominally 300-lb ingots. The niobium and boron content was varied within the specification range with all other alloying elements nominally the same among heats. The compositions of these materials are listed in Table 1. The same heat was used to produce both wrought and cast forms in order to eliminate any compositional difference between different product forms.

Table 1, Alloy 718 Compositions (wt%)

Wrought	Nb	B	Cast	Nb	B
W997	4.4	0.003	C997	4.5	0.004
W995	4.9	0.004	C995	5.1	0.005
W996	5.5	0.005	C996	5.4	0.006
Bulk composition: 20.0Cr, 18.0Fe, 2.9Mo, 1.0Ti, 0.5Al, 0.014C, bal Ni					

The heats with medium Nb and B were subjected to 5 heat treatment cycles, and all heats were subjected to 20 and 40 heat treatment cycles. The heat treatment cycle consisted of a solutionizing treatment (950°C/15 min) followed by a standard aging treatment (775°C/5 hr + 665°C/ 1 hr). This particular heat treatment for Alloy 718 is used to minimize grain size and optimize mechanical properties, particularly fatigue.

The weldability of all these heats was evaluated using the Gleeble® hot ductility test. This test simulates the thermal cycle which the material in the heat-affected zone undergoes during welding [3]. Small tensile samples are fractured rapidly at specific temperatures during either the on-heating or the on-cooling portion of a duplicated weld thermal cycle. The ductility of the material during the thermal cycle is measured in terms of transverse

reduction in area and plotted versus temperature to represent the hot ductility “signature” of the material (Figure 1). When testing on-heating, the material will first exhibit a sharp reduction in ductility until reaching the nil ductility temperature (NDT). With continued increase in temperature the sample will eventually exhibit effectively no strength at the nil strength temperature (NST). Samples are then tested on cooling from the NST at progressively lower temperatures until ductility is recovered at the ductility recovery temperature (DRT).

Testing was accomplished using cylindrical test specimens nominally 6.4 mm in diameter, 100 mm in length and threaded on both ends. The sample free span (between Gleeble grips) was on the order of 20 mm. Chromel-alumel thermocouples were attached to the middle of the specimen for temperature measurement and control. The heating rate was 111°C/s. The cooling rate was 20°C/s between 1260-1200°C and 43°C/s below 1200°C. This cooling rate schedule closely simulates actual weld cooling conditions. Samples were pulled to failure at a rate of 25.4 mm/s.

A number of specimens were examined using optical microscopy and scanning electron microscopy (SEM). Before examination, the samples were polished and electrolytically etched with 10% oxalic acid. Also, a number of fracture surfaces was studied using the SEM.

Figure 1 Hot ductility “signature” showing on-heating and on-cooling ductility relative to temperature.

Results and Discussion

Hot Ductility Testing

Hot ductility test results are presented in Table II in terms of NDT, NST, DRT, and the temperature range between DRT and the peak temperature to which the sample was heated (1260°C for all materials except C996). This differential ($T_p - \text{DRT}$) is often used to assess susceptibility to liquation cracking, with larger values indicating increased susceptibility. For the wrought materials, the results indicate that there are only minor variations as a function of both composition and number of thermal cycles. There appears to be a slight decrease in weldability with increasing niobium and boron content. In general, the heat with the lowest niobium and boron content (W997) exhibits the lowest susceptibility, particularly in the as-received (0 cycle condition).

Table II, Results of Hot Ductility Tests

Alloy	Number of HT-cycles	NDT (°C)	NST (°C)	DRT (°C)	Peak (T _p) (°C)	T _p -DRT (°C)
W997	0	1200	1311	1150	1260	110
	20	1160	1306	1120	1260	140
	40	1160	1304	1140	1260	120
W995	0	1150	1299	1095	1260	165
	5	1150	1299	1090	1260	170
	20	1150	1294	1110	1260	150
	40	1140	1294	1120	1260	140
W996	0	1150	1283	1090	1260	170
	20	1145	1294	1090	1260	170
	40	1140	1289	1090	1260	170
C997	0	1160	1301	1080	1260	180
	20	1110	1309	1060	1260	200
	40	1120	1299	1070	1260	190
C995	0	1120	1293	1080	1260	180
	5	1110	1292	1080	1260	180
	20	1120	1299	1020	1260	240
	40	1105	1304	1050	1260	210
C996	0	1105	1293	1015	1260	245
	20	1095	1289	1070	1245	175
	40	1085	1289	1040	1230	190

The results for cast versions show a more pronounced effect of chemical composition. When comparing the different heats in the as-received condition there is only a slight decrease in weldability with increasing niobium and boron. The same trend is apparent when comparing the heats with the same number of heat treatments. The number of heat treatment cycles also has a distinct influence. The heat with low niobium and low boron content (C997) shows only slight deterioration in the weldability (especially in NDT) with increasing heat treatment cycles.

In the heat with medium niobium and medium boron content (C995), there is no influence on the weldability after 5 heat treatment cycles. With 20 and 40 heat treatment cycles there is some decrease in the DRT, but more importantly, there is an increase in the number of specimens cracking at peak temperature. In the case of the material with high niobium boron content (C996), the on-cooling test could not be performed at 1260°C for 20 and 40 heat treatment cycles because all specimens cracked at peak temperature. This is presumably because significant liquation occurred in these samples. In the case of 20 heat treatment cycles the peak temperature had to be dropped to 1245°C and for 40 heat treatment cycles even to 1230°C. The percentage of specimens cracking at peak temperature for the heat with high niobium and high boron content is shown in Table III. When comparing the DRT for as-received material and material with 20 heat treatment cycles, which both have had the same peak temperature (1245°C), then there is no

difference. Obviously the more heat treatment cycles the material experiences, the greater the amount of liquid formed at peak temperatures.

Table III, Percentage of Samples Breaking at Given Peak Temperature for the High Niobium and Boron Cast Material (C996)

Peak Temperature (°C)	0 HT-cycles	20 HT-cycles	40 HT-cycles
1260	0%	100%	-
1245	0%	57%	100%
1230	-	-	36%

Microstructure Analysis

Wrought Material. All of the wrought base materials exhibited a distribution of secondary constituents, with the amount increasing with niobium content. Because of this relationship to niobium content, it was presumed that this secondary constituent was delta phase, an orthorhombic Ni_3Nb phase known to form during elevated temperature exposure. An example of the widespread precipitation of δ -phase that is possible after multiple heat treatments is shown in Figure 2. This phase was examined in the SEM and its composition analyzed using the EDS detector. Typically, the δ -phase appears as high-aspect needles (on the order of 10 μm long). In the wrought material, it tends to be rather evenly distributed over the grains with no relationship to the grain boundaries. Along some grain boundaries there is a line of small δ -phase needles present, but the amount is always small. Also present in the material are carbides, mostly NbC but also some TiC, but they are limited in number. These carbides were present in the as-received material and did not appear to change in size, shape, or distribution as a result of the heat treatments imposed.

Figure 2 δ -phase formation in the high niobium wrought material (W996) that has undergone 20 heat treatment cycles, 400X.

Cast Material. In the cast material, there were no secondary phases present in the as-received material. The microstructure exhibited a normal coarse-grained appearance with evidence of segregation from the solidification process. With increasing heat treatment cycles, there was a clear increase in the amount of secondary phases (as shown in Figure 3). The nature of these secondary phases was also established using SEM/EDS, and was found to be mostly δ -phase. The δ -phase needles in the cast material are much longer than the needles in the wrought material (up to 40 μm), except along some grain boundaries where small needles are present.

All the δ -phase in the cast material was located at or near grain boundaries, particularly in the low and medium niobium heats. This is due primarily to the segregation which occurs during the casting process. These boundaries will be enriched in niobium, thus promoting the precipitation of δ -phase. In the medium and high niobium heats, some intragranular precipitation occurred after 20 or 40 heat treatment cycles. Small patches of Laves phase were found along the grain boundaries, with the amount increasing as the number of heat treatment cycles increased. As in the wrought material, there were some carbides present, primarily NbC and some TiC.

Figure 3 δ -phase formation in the medium niobium cast material (C995) after A) 20, and B) 40 heat treatment cycles, 400X.

Hot Ductility Samples. The results of the hot ductility tests can be explained by microstructural analysis of samples which have been heated through the peak temperature (nominally 1260°C). For the wrought material, the δ -phase begins to dissolve in the temperature range from 925-1010°C [4]. As shown in Figure 4, for the medium niobium and boron wrought material (W995) subjected to 5 heat treatment cycles, the δ -phase has completely dissolved. Most of the δ -phase dissolves during the on-heating portion of the thermal cycle. At temperatures above the NDT, liquid “pockets” will begin to form. Some of these pockets form as a result of constitutional liquation of NbC, but their number is limited [5].

Figure 4 Medium niobium and boron wrought hot ductility specimen (W995) subjected to 5 heat treatment cycles, after heating through a peak temperature of 1260 °C, 400X.

In other areas, liquation is observed even though there is no obvious carbide present. This liquid pattern appears to mirror the morphology of the pre-existing δ -phase, suggesting that the dissolution of δ -phase contributes to the liquation. This could potentially result from the formation of Laves phase, which then reacts with the austenite matrix to form a liquid above approximately 1200 °C. This reaction is described in Figure 5 [6], which shows a pseudo-binary phase diagram for Alloy 718. In order for this liquation mechanism to be operative, Laves must form rapidly on-heating following the dissolution of δ -phase. It is unlikely that this reaction can occur this rapidly, considering the heating rates involved. Alternatively, it is possible that local melting occurs in the niobium-enriched regions of the austenite resulting from δ -phase dissolution. Both these mechanisms are the subject of a continuing investigation. The distribution of the liquid pockets has little or no relation to the grain boundaries, even with higher volume fractions of δ -phase in samples exposed to 20 heat treatment cycles (see Figure 6). The amount of liquid formed increases with increasing amount of secondary phases but liquid films along grain boundaries were not prevalent. This is also related to the small wrought grain size, which increases grain boundary area and reduces the liquid available per surface area [7, 8].

For cast material some investigations showed that liquation of Laves phase is the primary cause of liquation [9,10], but also mentioned sometimes is δ -phase [11]. In the cast material subjected to multiple thermal cycles, significant δ -phase dissolution also occurs on-heating during the hot ductility test. Again, liquation is associated with constitutional liquation of NbC and liquation of niobium-enriched regions resulting from δ -phase dissolution. Since the δ -phase in the cast materials was concentrated at grain boundaries (Figure 3), continuous liquid films were commonly observed in the medium- and high-niobium alloys (see Figure 7).

Figure 5. Pseudo-binary phase diagram for Alloy 718 [6].

For the cast material with low niobium and low boron (C997), isolated liquid pockets were observed at grain boundaries, but they never form a continuous film along the grain boundary. In the cast materials with high/medium niobium and boron (C995 and C996), the amount of secondary phases increases with number of heat treatments. For the as-received material (no secondary phases present), there was only a very limited amount of pockets filled with liquid near the fracture surface (probably due to NbC constitutional liquation). After 20 heat treatment cycles there is considerable δ -phase present (but mostly near the grain boundaries), and some grain boundaries are covered with Laves phase. In the

Figure 6 Medium niobium and boron wrought hot ductility specimen (W995) subjected to 20 heat treatment cycles, after heating through a peak temperature of 1260 °C, 400X.

Figure 7 High niobium and boron cast hot ductility specimen (C996) subjected to 20 heat treatment cycles, after heating through a peak temperature of 1260 °C, 400X.

material near the fracture surface there are liquid pockets everywhere. Along some grain boundaries the pockets connect to form a continuous film (see Figure 7). This progressive increase in grain boundary liquid film with increasing δ -phase along the grain boundaries, explains the degradation in weldability in the medium- and high-niobium cast materials.

Summary

Wrought Alloy 718 showed little influence of either chemical composition or number of postweld heat treatment cycles on hot ductility behavior. δ -phase which formed during multiple heat treatment was homogeneous in the wrought microstructure and dissolved readily upon heating through a simulated weld thermal cycle. This dissolution resulted in the formation of niobium-rich regions in the microstructure which were subject to local melting. Because of the homogeneous distribution of this liquid and the small wrought grain size, there was little effect on the wrought material weldability. The cast material was influenced by both the chemical composition and the number of heat treatment cycles. Higher levels of niobium and boron, and increasing heat treatment cycles resulted in a loss of the hot ductility of the material during the welding cycle. This is due primarily to the existence of a large-grained, highly segregated microstructure which promotes significant δ -phase formation along the grain boundaries during postweld heat treatment. The dissolution of this phase in the HAZ during welding increases the niobium concentration in the area near the grain boundaries, resulting in low melting liquid films at the grain boundaries.

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