INTERGRANULAR LIQUID FORMATION, DISTRIBUTION, AND CRACKING

IN THE HAZ OF ALLOY 718 WELDS

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Abstract

The welding of alloy 718 is known to cause intergranular hot cracks in the heat affected zone adjacent to the fusion zone. These intergranular cracks form when liquid is present simultaneously with thermal contraction stress. The source of the liquid is the constitutional liquation of second phases such as NbC and Laves.

Three attributes of the intergranular liquid are suspected of controlling the alloy's susceptibility to intergranular hot cracking. These are the ability to wet the grain boundaries, the resistance of the liquid to isothermal-type resolidification and the eutectic temperature of the liquid. These attributes are examined through the study of intergranular liquid phase distributions as a function of temperature and time.

Correlations are reviewed between the liquid phase distribution and carbon content and sulfur content. Supporting data is examined that relates the effect of carbon to the volume fraction of NbC and the effect of sulfur to intergranular segregation. The effects of heat treatment, temperature, and primary processing on the liquid phase distribution are also reviewed.

Finally, a model is reviewed that relates the time for isothermal-type resolidification of a grain boundary liquid film to various material properties. These properties include the volume fraction of liquating phase and grain size.

Introduction

Jet engine fabricators have a problem welding nickel alloys due to strain age cracking which is related to $Ni_3(Al,Ti)$ precipitation hardening. This problem was overcome through the use of Nb-bearing alloys which harden by a sluggish Ni_3Nb precipitation. These alloys, however, are susceptible to intergranular hot cracking in the heat-affected zone (HAZ) adjacent to welds. The magnitude of this problem can be appreciated when one considers that a given weld may have multiple hot cracks and there are literally hundreds of welds in some jet engines. Although this problem has been studied extensively, a satisfactory solution has not been found.

The intergranular hot cracking observed in Nb-bearing nickel alloys can be traced to the constitutional liquation of NbC in wrought alloys and both NbC and Laves phase in cast alloys. Because these phases are found intergranularly, the liquid phase is also found along the grain boundaries after liquation.

A first approximation is to treat the intergranular liquid according to C. S. Smith [1]:

$$\gamma_{gb} = 2 \, \gamma_{SL} \cos \theta \qquad (1)$$

where:

θ is the wetting angle,
γ_{gb} is the grain boundary energy, and
γ_{SL} is the solid-liquid interface energy.

This treatment provides an easy explanation for the cause of hot cracking and the detrimental effects of impurities in terms of the liquid wetting angle. However, rapid thermal cycles such as found in welding create non-equilibrium chemical gradients across the grain boundaries and across the solid-liquid boundary which affect γ_{gb} and γ_{SL}. Since the chemical gradient is diffusion controlled, it changes continuously during the thermal cycle. Even if the intergranular liquid phase distribution (LPD) behaved in a generally equilibrium manner with respect to the γ_{SL}/γ_{sb} ratio, it would still be temperature and time dependent. Hence, hot cracking susceptibility would depend on the LPD at the precise time and temperature that strain, usually due to thermal stress, is applied to the HAZ.

The above situation is complicated further by the effect that various processing techniques have on the chemical composition of the intergranular liquid. Processing techniques such as casting, forging, rolling, and hot isostatic pressing strongly affect microstructure as does heat treatment and heat-to-heat chemistry variations. This determines the impurity concentration and second phase precipitation at grain boundaries which contribute to the liquid composition. The picture that develops is not one of a simple wetting angle correlation to hot cracking. It is a picture of a temperature-time dependent LPD that is influenced by alloy chemistry, primary and secondary processing, and the welding parameters.

Review of Liquid Phase Distribution and Intergranular Chemistry

The wetting angle of an intergranular liquid phase, as a function of time and temperature during cooling, controls the metallurgical contribution to hot cracking. In 1948, Smith [1] showed the role of dihedral wetting angle on the intergranular LPD. Since then, others have correlated the

dihedral angle with cracking [2], and shown it to be a function of temperature [3-6], stress [6,7], and trace elements [4]. Techniques for quantitative measurement of the dihedral angle have been presented by Stickels and Hucke [8], Riegger and Van Vlack [9], and, most recently (1986), Dehoff [10]. Measurements on eutectic systems were made in the 1960's [5-7] and more recently (1985) were made on eutectic and monotectic systems [11].

The sources of intergranular liquid during a rapid thermal cycle might include the following:

- constitutional liquation [12-16]

- eutectic melting of the terminal solidification products of a segregated, cast structure [17-19]

- mushy-zone type intergranular melting [20]

The above mechanisms have been used to discuss intergranular hot cracking since the 1960's. More recently (1981-1983), constitutional liquation of NbC has been discussed relative to intergranular HAZ hot cracking in nickel alloys [12] and stainless steel [16] while the liquation of TiC has been discussed in Inconel 800 [15]. Eutectic melting of Laves phase has been discussed for nickel alloys [17,21] as has the combined liquation of NbC and melting of Laves phase [17]. Grain boundaries which are very clean will not melt below the bulk solidus, nor will a molten grain boundary film penetrate along a clean grain boundary below its solidus [22]. It can be postulated that an intergranular liquid which captures impurities will penetrate along a boundary to a temperature determined by the instantanious microchemistry of the liquid. However, to the author's knowledge, such behavior has not been experimentally verified.

The intergranular liquid produced under non-equilibrium conditions may capture intergranular precipitates and segregated elements in the grain boundary region before they escape by diffusion. Thus, to identify the source of intergranular chemistry one must characterize intergranular precipitation and segregation prior to liquid formation. Comprehensive studies of this type are almost non-existent in engineering alloys. Hall and Briant [23] did this for several heat treatments of alloy 600. The importance of their work to the present study is their demonstration of the interrelationships between precipitation, segregation, and intergranular (region) chemistry. They also identified the effect of differing heat treatments on the segregation of various elemental species and precipitate morphologies. Such a comprehensive study is the only way to isolate the various contributions to intergranular chemistry. A convergent beam TEM study of alloy 718 microstructure was performed by Vincent [24] and surface segregation in alloy 718 has been reported by Thompson, et al. [25]. A limited study of both intergranular precipitation and segregation in alloy 718 was performed by Miglin and Domian [26] relative to stress corrosion cracking. Their study contains some interesting information including the percent of grain boundary coverage by Laves phase.

Studies of intergranular segregation are limited by the ability to obtain intergranular fracture using hydrogen charging techniques or in situ creep. A number of studies have looked at intergranular segregation in various nickel alloys including; Ni [27,28], Ni-Cr [29], B1900 and MAR-M 200 [30], and Hastellloy C-276 [31]. When intergranular fracture is not possible the alternatives are computer modeling of grain boundary segregation [32,33] or surface segregation studies. For an engineering alloy, the surface segregation approach, while not perfect, will probably

provide more insight than a computer modeling approach for estimating intergranular segregation. Seah and Lea have looked at the kinetics of surface segregation [34] and the relationship between surface segregation and intergranular segregation [35]. They concluded that surface segregation does not accurately reflect what may be happening at the grain boundaries. Larere, et al. [28] studied sulfur segregation in nickel and determined that the segregation free energy for surface segregation of sulfur was approximately twice that of the grain boundary. However, they also showed that ternary chemistry interactions [36] caused the surface and grain boundary to behave differently thus substantiating the conclusions of Seah and Lea [35]. Thus, if surface segregation data is used to argue grain boundary behavior, it must be done with caution.

The liquid solidification temperature is of primary concern to hot cracking. Not much is published concerning the solidification behavior of an intergranular liquid under rapid cooling conditions. Several studies have examined the terminal solidification products along a liquated grain boundary in alloy 718 [12,18,21,24]. Only Vincent [24] has identified the constituents using analytical TEM and then he did not do a chemical analysis of the constituents using EDS techniques. Cieslak, et al. [37,38] studied the general melting and solidification behavior of several nickel alloys including three Nb-bearing types; 625, 718 and 909. They used DTA techniques to identify eutectic temperatures for Laves formation and then characterized the Laves composition using combined TEM-EDS techniques. They also identified solidus and liquidus slopes and partition coefficients. One of the interesting results of their study was the dependence of the eutectic temperature during cooling on the chemical composition of the Laves phase.

Finally, literature on liquid phase sintering has looked at the problem of LPD when a non-equilibrium liquid composition exists [39]. It was reported that the transition from complete liquid wetting to partial liquid wetting occurred as the non-equilibrium liquid composition attained equilibrium. It was postulated that the change in the LPD was due to an increase in γ_{SL} as the liquid phase composition reached equilibrium.

Intergranular Liquid Phase Distributions in Alloy 718

A liquid phase can be produced along a grain boundary due to constitutional liquation of second phases in the matrix. Figure 1 shows a
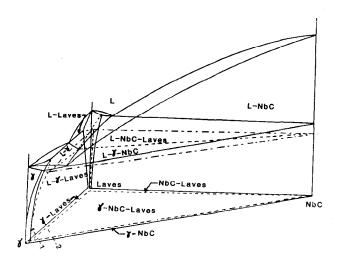

Fig. 1 - Space diagram for the pseudo-ternary γ-NbC-Laves of alloy 718 [40].

440

hypothetical alloy 718 ternary diagram involving γ-Laves-carbides [40]. Coring in a cast alloy moves the nominal composition from a section such as 1-1 in Fig. 1 to a section like 2-2 for the interdendritic region (Fig. 2). Constitutional liquation of the interdendritic γ-Laves-NbC eutectic occurs at the temperature T_{LCL} (Fig. 2) for the cast alloy. Constitutional liquation in the wrought alloy can also be conceptualized using Fig. 2. By rapidly heating γ + NbC together in the matrix without allowing NbC to dissolve, they behave like a γ + NbC field and liquation begins at T_{CL}. An example of this is shown in Fig. 3. A complete description of these liquation processes can be found in reference 40.
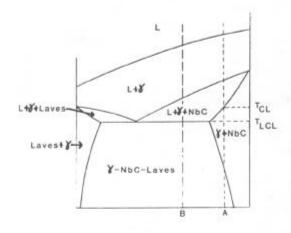

Fig. 2 – Vertical section through plane 2-2 of Fig. 1. Vertical A is liquation of a carbide which is typical of wrought alloys and results in a liquation temperature at T_{CL}. Vertical B is liquation of the eutectic involving Laves phase which is typical of cast alloys and results in a liquation temperature of T_{LCL} [40].
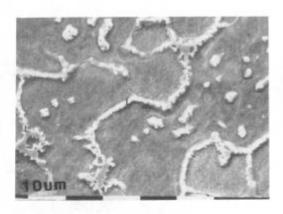

Fig. 3 – Fully liuqated wrought alloy 718. Note continuous grain boundary liquid film.

Once molten, the liquid chemistry of the liquating phases can be altered through interaction with elements segregated at the boundaries, phases precipitated along the boundary, and the matrix near the boundary. The distribution of this liquid, as a function of time and temperature (Fig. 4), dictates the material's resistance to intergranular cracking.

The situation described above becomes a significant technical problem in the HAZ of welds. The weld HAZ has the time and temperature characteristics which can lead to the co-existence of intergranular liquid films and stress, thus resulting in grain boundary cracking (Fig. 5). The slow diffusivity of Nb allows the intergranular liquid to persist for long periods of time in non-equilibrium situations. This makes these alloys very susceptible to hot cracking.
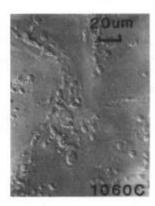
441

Fig. 4 - Liquid phase distribution during cooling of liquated cast alloy 718. Cooling rate was 30°C/sec. from a peak temperature of 1180°C. Note loss of the continuous liquid film along grain boundary during cooling [21].
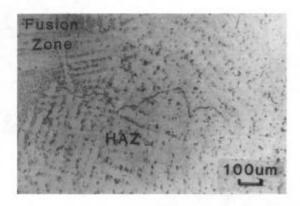

Fig. 5 - Intergranular hot crack in the HAZ of alloy 718 produced by the spot varestraint test.

Contributing phenomena that determine the LPD in a Nb-bearing nickel alloy like 718 can be identified as:

1) volume fraction and intergranular distribution of second phases which produce liquid during heating, usually NbC and Nb-rich Laves phase;

2) chemical composition of the intergranular region including elemental grain boundary segregation, precipitates like sulfides and borides, and matrix;

3) kinetics of the chemical composition change of the intergranular liquid during cooling; and

4) solidification of the intergranular liquid including its solidus or eutectic temperature $[T_{S(gb)}]$ and terminal solidification phases.

Studies of intergranular LPD are rare although the phenomena is quite important to a number of processes. It is believed that the most important aspects of the problem are the development and evolution of the intergranular liquid chemistry and its effect on the intergranular solidification temperature and LPD.

442

Cast and wrought material present two very different grain boundary environments due to their respective interdendritic and homogenized natures. Figure 6 shows that the intergranular liquid of the alloy 718 persists to a temperature below that of wrought alloy 718. This dramatic difference in $T_{S(gb)}$ for the intergranular liquid must be related to the initial chemistry of the liquid phases and the concentration gradient established between the liquid and solid. The compositions of the liquid phases in this case are obviously related to the local grain boundary environment prior to heating and melting. A more subtle difference in the LPD is caused by heat treating prior to the thermal cycle which produces the intergranular liquid. Figure 7 shows that heat treatment does not noticeably change the $T_{S(gb)}$, however, it does change the liquid coverage on the grain boundary as a function of temperature.
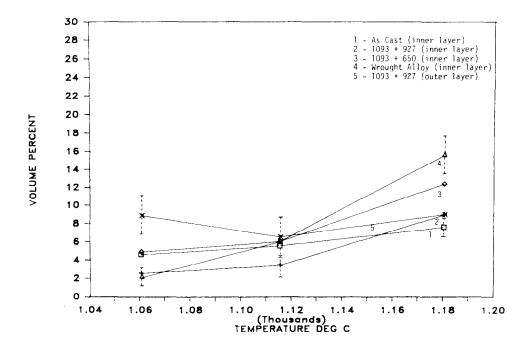

Fig. 6 — Volume fraction of liquid as a function of temperature during cooling (30°C/sec) from 1180°C for cast (#1) and wrought (#4) alloy 718. Note the relative difference in rate of liquid formation and resolidification [21].

Figure 8 shows an expanded description of the LPD problem. Current research has considered equilibrium, monolayer-type grain boundary segregation and its effect on the LPD [41]. However, 10µm thick liquid layers were formed on grain boundaries at 1230°C [21] and a mono-layer of segregation would contribute only 0.002 atomic percent to this liquid chemistry. Thus impurity segregation could only make a significant contribution to the bulk intergranular liquid chemistry of very thin liquid films (less than about 0.05µm). Intergranular segregation is expected to have its greatest effect on the LPD through the grain boundary and solid-liquid surface energy terms, γ_{gb} and γ_{SL}.

For all practical purposes, the chemistry of the intergranular liquid film will be equal to the chemistry of the region on either side of the grain boundary prior to welding. Diffusion kinetics, which are not well known for the complex nickel alloys, will determine if precipitates and

elements segregated near the grain boundary are trapped in a spreading intergranular liquid phase. For example, intergranular segregation kinetics for sulfur in nickel were shown by Larere, et al. [28] to obey the following relationship:

$$c_S^{gb} = \frac{4\ c_S^B\ \sqrt{Dt}}{\sqrt{\pi d}} \tag{2}$$

where:

c_S^{gb} and c_S^B are the intergranular and bulk S concentrations;

  d is the intergranular width;
  D is the diffusivity; and
  t is time.

Using their reported value of:

$$D\ (cm^2/sec) = 1.40\ exp\ (-52,000/RT) \tag{3}$$

and calculating D at a maximum heating temperature of 1450K (2150°F) gives:

$$D = 1.40\ exp\ (-52,000/1.98[1450]) = 1.90 \times 10^{-8}\ cm^2/sec \tag{4}$$

Since we have measured that liquid sweeps out an intergranular width of 10µm at 1230°C, then it would take sulfur about 13 seconds to escape 5µm on either side of the grain boundary (using an $x = \sqrt{Dt}$ relationship). Although this time is for a dilute Ni-S solution, it is clear that sulfur would not escape the liquid phase formation given our thermal cycle of about two seconds to heat to full liquation. This type of exercise suggests that much of the chemistry of the grain boundary region will be trapped in the liquating liquid.

A primary effect on $T_{S(gb)}$ is expected from the chemical composition of the final solidification phases, notably involving a Laves (A_2B) phase (where B is primarily Nb). It was shown [37] that the eutectic temperature
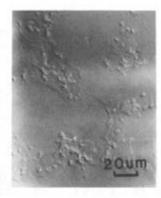

Fig. 7 - The effect of heat treatment on intergranular liquid phase distribution at 1060°C during cooling of cast alloy 718. Note that the heat treatment affects the wetting of the liquid along the grain boundaries producing a more (650°C) or less (927°C) crack sensitive microstructure. Treatments from left to right: as-cast, 1093C (1 hour) plus 650°C (1 hour), 1093°C (1 hour) plus 927°C (1 hour) [21].
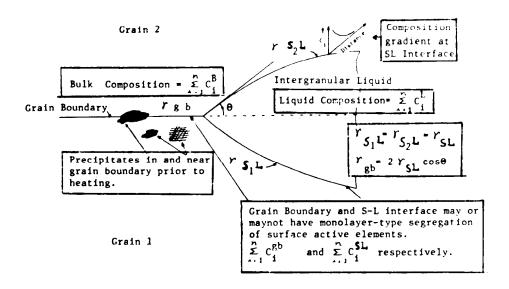
444

Grain 2

r S_2L

Composition gradient at SL Interface

Bulk Composition = $\sum_{i=1}^{n} C_i^B$

Grain Boundary r g b

θ

Intergranular Liquid

Liquid Composition = $\sum_{i=1}^{n} C_i^L$

Precipitates in and near grain boundary prior to heating.

r S_1L

$\dfrac{r_{S_1L} \cdot r_{S_2L} \cdot r_{SL}}{r_{gb} \cdot 2 r_{SL} \cos\theta}$

Grain Boundary and S-L interface may or maynot have monolayer-type segregation of surface active elements.

$\sum_{i=1}^{n} C_i^{gb}$ and $\sum_{i=1}^{n} C_i^{SL}$ respectively.

Grain 1

Fig. 8 – Aspects of liquid phase distribution problem.

involving a Laves phase varied by 40°C when the iron concentration in nickel alloys varied from 0.40 to 0.02. Significant differences are expected between the initial intergranular liquid composition of cast and wrought alloys and even in an alloy with different heat treatments. These chemistry differences may alter the composition of the Laves phase thus changing $T_{S(gb)}$.

The Effect of Carbon and Volume Fraction of NbC + Laves on the LPD

The bulk carbon content of as-cast alloy 718 has been shown [42] to have an effect on intergranular hot cracking (Fig. 9). The relative increase in hot cracking between the 0.02 w/o and 0.06 w/o carbon levels is about the same over a wide range of sulfur contents. This suggests that the effect of carbon is independent of the effects associated with impurity elements such as sulfur. The mechanism by which carbon affects hot cracking susceptibility appears to be its effect on the volume fraction (f) of NbC. The as-cast microstructure shown in Figs. 10(a) and 10(b) consists of (γ + NbC) dentrites with (γ + NbC + Laves) precipitated interdendritically. Volume fraction measurements (Table I) of the NbC (f_c) and Laves (f_L) phases showed that when the carbon content increased from 0.02 w/o to 0.06 w/o, the f_c increased by 70%. The f_L did not increase in a systemtatic manner with carbon content.

The best evidence that the volume fraction of second phase, in this case predominantly NbC, affects HAZ hot cracking is seen in Fig. 9, along with Table I. It was found that the higher f_c alloys maintained a consistently higher level of hot cracking than did the lower f_c alloys. The fact that sulfur additions also increased hot cracking is also evident from Fig. 9. Note that the relative increase in hot cracking associated with the higher f_c alloy was not affected by the sulfur level. This suggests that the effect of f_c can be evaluated as an independent mechanism from sulfur and possibly other surface active impurities.
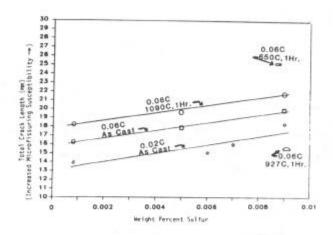
445

Fig. 9 – The effects of bulk carbon content, bulk sulfur content, and heat treatment on as-cast Alloy 718 [41,42].
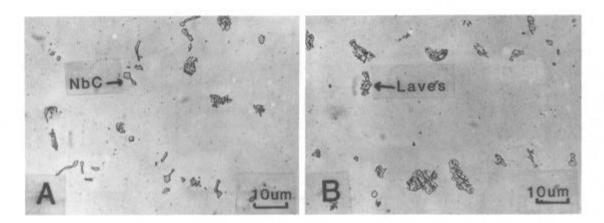

Fig. 10 – Microstructures of cast alloy 718 with high C/Nb ratio (A) and lower C/Nb ratio (B). Note the large fraction of carbides in A versus B [40].

Figures 11(a) and 11(b) show the LPD and the relative liquid film stabilities for the low and high carbon alloys as a function of cooling temperatures above the system eutectic (≈ 1200°C). These structures were produced by rapid heating to liquate the NbC and Laves phases and then quenching after some cooling time to evaluate the liquid distribution. The two alloys were essentially the same except for the carbon content. Note that the high carbon alloy maintained an intergranular liquid at a time and temperature when the liquid of the low carbon alloy was essentially gone. This seems to verify the concept that in the as-cast condition the susceptibility to hot cracking is proportional to the volume fraction of liquating phase which is proportional to the volume fraction of NbC and/or Laves phase. However, it has been shown that this is not the case after heat treatment [42].

446

Table I - Volume Fractions of Carbide and Laves Phases in
Cast Alloy 718 [42]

Wt. % Carbon	Wt. % Sulfur	Heat Treatment	% Carbide*	% Laves Phase*
0.021	0.008	As Cast	1.17 ± 0.38	1.23
0.021	0.006	As Cast	1.14 ± 0.38	0.384
0.021	0.007	As Cast	1.17 ± 0.37	0.637
0.021	0.009	As Cast	1.17 ± 0.38	0.545
0.06	0.0008	As Cast	2.00 ± 0.46	1.33
0.06	0.005	As Cast	2.04 ± 0.53	0.461
0.06	0.009	As Cast	1.99 ± 0.39	1.44

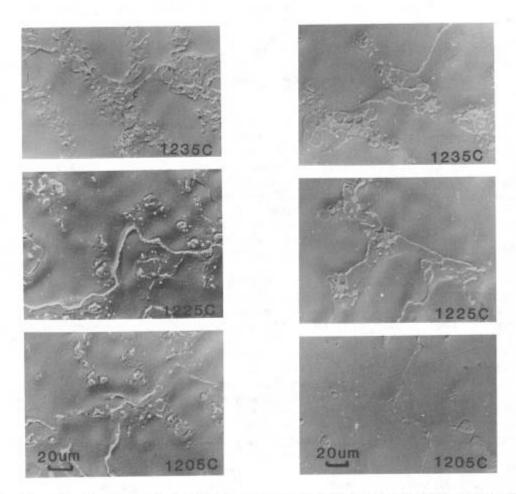*Ranges indicated are for a 95% confidence interval.

Fig. 11 - Liquid phase distribution as a function of cooling from 1235°C for cast alloy 718 with a bulk carbon content of 0.06 w/o (left) versus one with 0.02 w/o (right). Note that liquid produced in the lower carbon alloy completely resolidifies before the higher carbon alloy [42].

Effect of Sulfur on the Intergranular LPD

Sulfur is often mentioned as a contributing factor in HAZ hot cracking of nickel alloys [19,43-46]. Canonico et al. [46] used Borland's [47] relative potency factor to propose that sulfur should be more potent than other elements in promoting hot cracking. Indeed, studies [19,45] have indicated that high bulk sulfur levels cause increased hot cracking. The increase in hot cracking could result from either a lowering of the eutectic temperature of the intergranular liquid or an increase in wetting of the liquid along the grain boundaries. The effect of the latter on hot cracking is somewhat easier to interpret than the former because of the literature which correlate cracking with wetting angle [2] and shows the wetting angle to be a function of temperature [3-6], stress [6,7], and impurity elements [4].

The spot varestraint test [48] was used as a means to evaluate the susceptibility of a series of 718 alloys to HAZ hot cracking [41]. This series of alloys varied in sulfur content from 8 ppm to 90 ppm (w/o) and every attempt was made to keep the concentration of the other elements constant. Hot cracking as a function of sulfur content for the as-cast condition is shown in Fig. 9. A series of heat treatments were also used to vary the hot cracking susceptibility and possibly the intergranular impurity concentration. Hot cracking as a function of sulfur concentration for various heat treatments is also shown in Fig. 9.

It is evident from Fig. 9 that the bulk sulfur content had some effect on hot cracking susceptibility both in the as-cast and heat treated (1090°C) condition. Various heat treatments given to the 90 ppm sulfur alloy also affected the hot cracking susceptibility of the alloy. The change in hot cracking with heat treatment could not be correlated to the volume fraction of carbide and/or Laves phases. A correlation with intergranular segregation would be desirable but the grain boundaries could not be exposed by hydrogen changing techniques for a study of grain boundary impurities. An alternate, but less desirable technique of studying free surface segregation as a function of temperature was used to garner some information on the surface activity of impurities during heat treatment [41]. Tables II(a) and II(b) show the free surface segregation of several elements of interest. It is seen that some carbon had absorbed onto the freshly cleaned surface (room temperature, time-0) from the vacuum system but nothing else other than the standard alloying elements (Ni, Cr, Fe) was present. When the sample was heated to 1090°C, sulfur strongly segregated to the surface, i.e., the sulfur peak was 1.5 times as high as the nickel peak. Titanium also showed some surface segregation. The influx of these elements was so strong that the surface segregation increased 33% upon cooling from 1093°C to 650°C [Table II(a)]. Surface segregation as a function of time at 650°C is given in Table II(b). No sulfur segregation occurred when 650°C was the peak temperature in contrast to the segregation that occurred at 650°C during cooling from 1090°C. In the absence of sulfur segregation, carbon and niobium surface segregation was observed during the isothermal treatment at 650°C.

The surface segregation results suggest that homogenization or solution annealing close to 1090°C increases hot cracking

Table II - Free Surface Segregation in Cast Alloy 718.

Table II(a) - Auger Peak Height Ratios (on cooling from 1089°C
in Auger System)

T(°C)	S	Nb	Ti	Cr	Fe	C	t (hrs).
RT	0.00	0.064	0.032	0.25	0.322	0.288	0.00
1089	1.47	---	0.352	0.35	0.29	0.000	0.00
1087	1.41	---	0.294	0.195	0.292	0.000	0.16
1080	1.55	---	0.263	0.184	0.289	0.000	0.38
1081	1.57	---	0.263	0.168	0.289	0.000	0.5
922	1.86	---	0.369	0.130	NM	0.000	0.58
923	1.78	---	0.339	0.130	0.26	0.000	1.5
670	1.95	---	0.375	0.130	0.25	0.000	1.58
653	2.19	---	NM	0.146	0.24	0.000	4.91

Table II(b) - Auger Peak Height Ratios (Isothermal Treatment at
650°C in Auger System)

T(°C)	S	Nb	Ti	Cr	Fe	C	t (hrs).
RT	0.00	0.104	0.059	0.074	0.268	0.298	0.00
651	0.00	0.032	0.35	0.105	0.25	0.6	0.00
645	0.00	0.4	0.32	0.105	0.25	0.65	0.06
645	0.00	0.414	0.365	0.146	0.24	0.68	0.16
657	0.00	0.52	0.4	0.2	NM	0.72	0.33
661	0.00	0.41	0.3	0.13	0.26	0.6	8.33
657	0.00	0.39	0.3	0.15	0.3	0.6	19.5
660	0.00	0.38	0.25	0.14	0.28	0.5	24.00

*Segregation during step cooling from 1089°C is shown in A while
segregation during an isothermal treatment at 650°C is given in B.
All data is reported as a relative concentration by dividing the dN/dE
peak of each element by the 848eV Ni peak [25,41]

susceptibility due to the surface activity of impurities according to
McLean's model [36]:

$$c_{GB}^{I}/c_{GB}^{M} = c_{B}^{I}/c_{B}^{M} \, e^{(-Q/RT)} \tag{5}$$

where:

c_{GB}^{I} and c_{B}^{I} are the impurity concentrations at the grain boundary and
in the bulk alloy respectively;

c_{GB}^{M} and c_{B}^{M} are the solvent concentrations at the grain boundary and
in the bulk respectively; and

Q, R, and T are the segregation activation energy, gas constant, and
absolute temperature respectively.

The two important attributes of the above equation relative to Fig. 9 and Table II are: C_{GB}^I increases as C_o^I increases from 8 ppm to 90 ppm and segregation increases with decreasing temperature, i.e., as the alloy cools from 1090°C. If it is assumed that the grain boundary is behaving in a manner analogous to the free surface, then the 1090°C heat treatment can be seen as one that releases sulfur for intergranular segregation to the grain boundaries and thereby increases hot cracking susceptibility. Increasing the bulk sulfur level proportionally increases the potential grain boundary sulfur concentration thus also increasing hot cracking susceptibility. Caution, however, should be used in accepting the assumption that surface segregation is on analog of grain boundary segregation potential.

The liquid phase distributions which resulted from high and low sulfur content is shown in Fig. 12 [41]. The series of microstructures show that a high sulfur level tends to increase the liquid coverage on the grain boundaries at low temperature. At high temperature, no difference is seen between alloys of different sulfur content. This suggests that the sulfur level affects the intergranular wetting angle as a function of temperature.
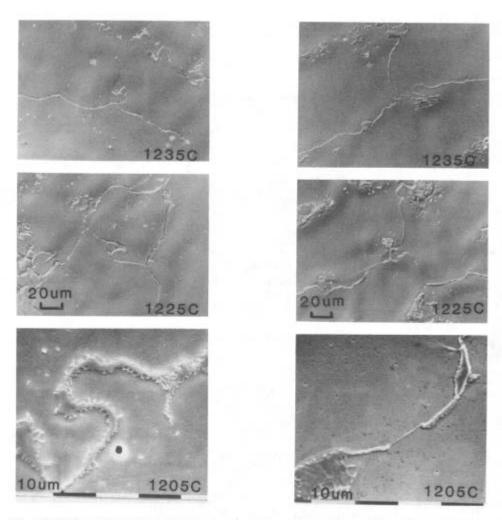

Fig. 12 -Liquid phase distribution as a function of cooling from 1235°C for cast alloy 718 with a bulk sulfur content of 90 ppm w/o (A) versus one with 8 ppm w/o (B). Note that at the lowest temperature the 8 ppm alloy begins to lose its ability to wet the gran boundaries [41].

450

A number of mechanisms have been presented in this paper to explain the formation, wetting and minimum solidification temperature of intergranular liquid films in alloy 718. However, the isothermal stability of the liquid film is of great importance to the formation of intergranular cracks. These liquid films are inherently unstable and equilibrium solubility will drive the liquid to resolidification as quickly as the excess solute can be removed from the liquated region. Intergranular hot cracks can not form unless the liquid is able to survive long enough to experience the stresses associated with thermal contraction.
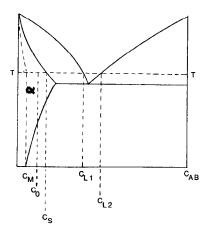

Fig. 13 - Hypothetical phase diagram between alpha and compound AB.

A model was recently proposed by Radhakrishnan and Thompson [40] to predict the isothermal stability of an intergranular liquid film. Consider the hypothetical phase diagram of Fig. 13. Let T be the isothermal temperature at which liquation occurs. Let the compositions C_S, C_{L1}, C_{L2}, and C_{AB} be as shown and C_m be the concentration of the matrix in equilibrium with AB at room temperature and C_0 be the average composition of the alloy. Assume when the alloy is rapidly heated to temperature T that AB liquates without any appreciable diffusion into the matrix and that the grain boundaries are completely wet by the liquating phase. Through a series of relationships involving tie-line constructions, quantitative microscopy, and diffusion approximations, it was shown that:

$$t = K f_{AB}^2 p^2 d^2 D^{-1} \qquad (6)$$

where:

　　t is the solidification time,
　　D is the diffusivity,
　f_{AB} is the volume fraction of phase which liquates,
　　p is the fraction of AB on the grain boundary,
　　d is the grain size, and
　　K is a constant which includes the tie line compositions at the temperature T.

Equation (6) predicts that the liquid film stability increases with more second phase (AB) on the grain boundary and with increasing grain size. This prediction correlates well with studies that show increased hot

451

cracking with increased volume fraction of NbC in alloy 718 [42] and increased grain size [49]. These correlations suggest a further correlation between the stability of the intergranular film and increased intergranular hot cracking.

Conclusions

Alloy 718 is susceptible to intergranular hot cracking in the HAZ of welds due to the simultaneous existance of intergranular liquid films and thermally induced stress. Both of these are functions of the welding thermal cycle and thus welding procedures are an important part of hot cracking prevention.

Intergranular liquid films in alloy 718 are also dependent on the metallurgy of the system including alloy composition and primary and secondary processing. The most characterizable attribute of the intergranular liquid film is its grain boundary wetting as a function of temperature and time. Available theories, segregation studies, and liquid phase distributions suggest that differences in intergranular wetting of the liquid is closely related to the effect of impurities on interfacial energy. The wetting angle also appears to bear some relationship to the volume fraction of liquid that is present as a function of temperature.

The volume fraction of liquid as a function of temperature and time and a similar characteristic, the length of time required for resolidification, play an important role in intergranular hot cracking. The length of time that the liquid exists in the HAZ is especially significant because it can be used with data on cooling stress as a function of time to predict intergranular hot cracking as a function of material properties and welding parameters. The relationship presented in Equation (6) for the solidification time of an intergranular liquid film is an important step in being able to predict the susceptibility of an alloy to HAZ hot cracking.

Acknowledgments

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