Tracking heat-affected zone cracking susceptibility in standard and modified heat treated IN 738 superalloy welds

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Abstract

The conditions influencing the phenomenon of heat affected zone (HAZ) cracking during tungsten inert gas (TIG) welding of a gamma-prime (γ') precipitation strengthened nickel-base superalloy, IN 738, is discussed here. The discussion is reinforced with the dependence of the cracking on particle size variation of the γ' precipitates and base alloy hardness. Microstructural analysis of the magnitude of HAZ cracking in the alloy indicates that while resistance to cracking susceptibility is poor in the standard heat treatment (SHT) condition, the modified heat treated (MHT) alloy exhibits significantly improved resistance to HAZ cracking. The improvement in resistance to HAZ cracking in the MHT alloy is attributed to the lower hardness of the alloy which permits stress relaxation in the base alloy compared to the SHT alloy. In contrast to what is expected based on the recommended manufacturer SHT procedure, preweld microstructural modification presents a viable way of limiting cracking susceptibility in TIG welded IN 738 superalloy.

Keywords: joining, superalloy, HAZ cracking.

1 Introduction

Nickel-based superalloy IN 738 is a vacuum melted and investment cast alloy developed at the Paul D. Merica research laboratory of the International Nickel Company Canada [1]. The alloy contains a significant volume fraction of an ordered LI2 intermetallic $Ni_3(Al,Ti)$ γ' phase, that results from the additions of

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small amount of Al + Ti, ~ 6 wt.%. Gamma-prime is a highly effective strengthener principally responsible for excellent high-temperature strength and remarkable hot corrosion resistance in superalloy [2]. These unique properties make IN 738 a suitable material for the manufacture of hot-section components in aero and land-based power generation turbines where they are exposed to severe operating conditions [2, 3]. During fabrication and repair of turbine components, fusion welding techniques are preferentially employed compared to other joining techniques because such techniques guarantee seamless integration of components with minimal property disruption.

IN 738 alloy, like other precipitation-hardened nickel-based superalloys that contain a substantial amount of Al and Ti, is very difficult to weld due to its high susceptibility to HAZ during welding and post-weld heat treatment (PWHT) [4]. Nakkalil *et al.* [5] attributed the cause of this cracking during fusion welding processes, such as TIG and electron beam welding to the liquation of various phases in the alloy, subsequent wetting of the grain boundaries by the liquid and decohesion across one of the solid-liquid interfaces due to on-cooling tensile stresses [6–8]. Detailed mechanism of grain boundary liquation cracking in the alloy has been a subject of extensive investigation by several researchers [5–8]. These investigations suggest that grain boundary liquation cracking is influenced by a combination of several factors that include preweld initial microstructures, alloy composition, heat input and cooling rate amongst others. For instance, welding techniques with low heat input (high power density) such as laser and electron beam fusion processes are credited with good and deep weld joint. Similarly, the use of solid-state welding techniques (such as friction stir and linear friction welding) has been shown to ensure good weld-joint integrity without crack [9]. But these joining techniques require huge capital outlay and technical expertise. Thus, other welding techniques like TIG welding which requires less resource commitment are explored for the joining of nickel based superalloy. Though, the manufacturer of the alloy provided a pre-fabrication and welding SHT procedure to enhance good joint integrity free of inherent cracking defect in the alloy, field experience and available information in literature indicates that cracking yet manifests in the alloy during fabrication and/or repair welding.

Therefore, the current study sought to modify the microstructure of IN 738 superalloy via preweld thermal treatment outside the SHT procedure recommended by the manufacturer prior to TIG welding. The object was to track the effect of such microstructural modification on HAZ cracking susceptibility in IN 738.

2 Materials and method

Sample coupons of size 75 mm x 18 mm x 4 mm were machined from 238 mm x 58 mm x 14 mm Cast IN 738 superalloy plate supplied by PCC Airfoils, LLC Prototype foundry, OH USA. The nominal composition of the alloy provided by the cast alloy producer is listed in Table 1. The machined coupons were heat treated using two different heat treatment regimes – solution heat treatment (SHT 1120°C/2 hour/air cooled) and Modified heat treatment (MHT 1120°C/24 hour/furnace cooled).

Table 1: Nominal composition of cast IN 738 superalloy.

Elemental Composition (wt. %)												
C	Cr	Co	W	Mo	Nb	Fe	Al	Ti	Ta	Zr	B	Bal
0.11	15.84	8.5	2.48	1.88	0.92	0.07	3.46	3.47	1.69	0.04	0.01	Ni

The heat treated samples (4 numbers) were autogenously welded with TIG welding machine using the parameters – current 60 Amp, voltage 10, shielding gas flow rate 8 litre/mm and welding speed 280–430 mm/min in argon shielding environment. Thereafter the quality of the weld was visually examined for defect. Representative metallographic specimens were sectioned from both the heat treated and welded coupons (10 section/sample) via electrical discharge machining (EDM) wire cut for microscopic examinations. Prior to microscopic examination, polished specimens were etched electrolytically in 12 mL H_3PO_4 + 40 mL HNO_3 + 48 mL H_2SO_4 solution at 6 volts for 5 seconds. Furthermore, sections of the heat treated and welded coupons were equally etched in Kallings reagent (2 grams $CuCl_2$ + 33 mL HCl + 33 mL Methanol) to reveal the grain boundaries and distributions of the γ’ particles present in the coupons. Microstructures of the as-received and variously treated coupons were examined and analyzed with an inverted optical microscope equipped with a CLEMEX Vision 3.0 image analyzer, JSM 5900 scanning electron microscope (SEM) equipped with an Oxford energy-dispersive spectrometer (EDS) and Inca analyzing software. The hardness of the coupons was measured with a Buehler microhardness tester using an indentation load of 300 g after the materials were polished to 1 µm surface finish.

3 Results and discussions

A typical optical microstructure of an electro-etched sample of the as-cast alloy shown in Figure 1 reveals a cellular dendritic cored structure with the interdendritic region enriched in solute elements due to solidification induced segregation; and a coarse grain structure with grains size averaging ~750 um. Higher magnification SEM examination reveal the presence of the main strengthening phase in the alloy, $Ni_3(Ti,Al)$ (γ') intermetallic phase (Figure 2). Other secondary phase particles observed in the microstructure include: MC-type carbides, M_2SC sulphocarbides, γ-γ' eutectic, and M_3B_2 borides. These secondary phases have equally been reported by other investigators [10–12]. The MC carbides, sulpho-carbides and M_3B_2 are considered as solidification products while the γ' is a product of solid state transformation from a supersaturated γ matrix [13].

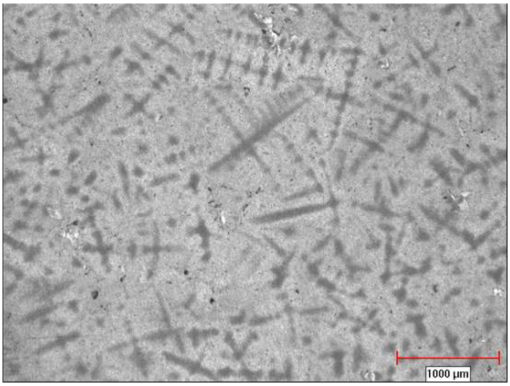

Figure 1: An optical image of as-cast IN 738 alloy showing the dendritic solidification microstructure.

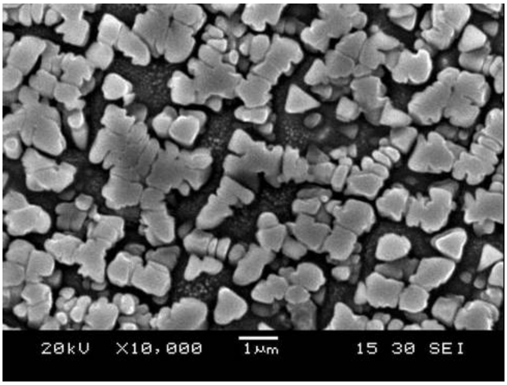

Figure 2: SEM image of as-cast IN 738 alloy showing the γ' particles.

Figure 3 shows the microstructure of the recommended manufacturer standard SHT preweld IN 738 which consists of coarse primary γ' precipitates, fine spherical secondary γ' precipitates of about 0.1 µm in diameter and other solidification products that formed during casting. This microstructure is consistent with previously reported investigation [10–12, 14]. In general, the excellent high temperature strength of nickel based superalloy is known to be strongly dependent on the morphology, distribution and size of the γ' precipitates in the matrix. More so, strength has been related to the interactions between γ' particles and moving dislocations [15]. Figure 4 shows the microstructure of the MHT coupons, with the primary γ' precipitates sizes comparatively larger and more in distribution than in the SHT coupons while the secondary γ' precipitates appears to have been completely dissolved. These apparent difference in the microstructural morphology of the two differently heat treated coupons may have significant influence on the cracking dynamics in the coupon in the welded condition.

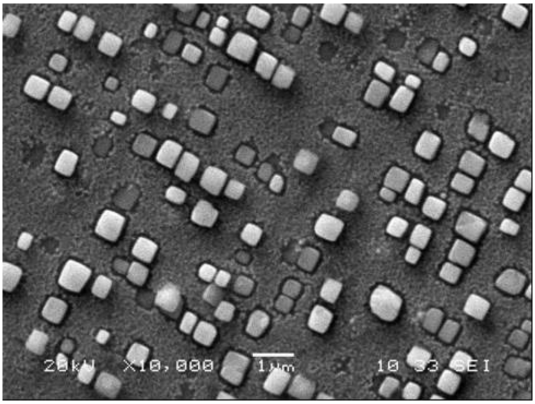

Figure 3: SEM image of γ' etched SHT IN 738 alloy showing the particles.

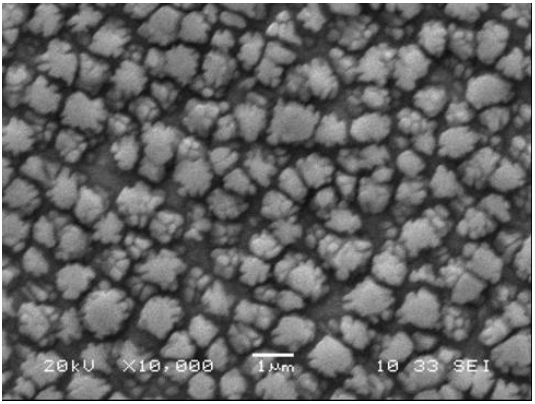

Figure 4: SEM image of γ' etched MHT IN 738 alloy showing the particles.

A micrograph showing a typical fusion zone (FZ) and (HAZ) in TIG welded IN 738 is shown in Figure 5. Microstructural examination of welded coupons reveals the occurrence of FZ and HAZ cracking irrespective of the preweld microstructural conditions. The cracks are intergranular in nature and consistent with the observed liquation cracks formed from grain boundary liquid film previously reported in Fe and Ni based alloy [16, 17]. FZ cracking is generally not considered a major weldability problem compared to HAZ cracking since it can be reasonably managed by the use of appropriate filler alloys unlike HAZ cracking [12]; hence, discussion is focused on HAZ cracking which in nickel based superalloys is generally presumed to be caused by the presence of liquid phase on grain boundaries at a time sufficient thermal and shrinkage tensile stresses have developed during weld cooling cycle, splitting apart weak intergranular liquid-solid bond [17].

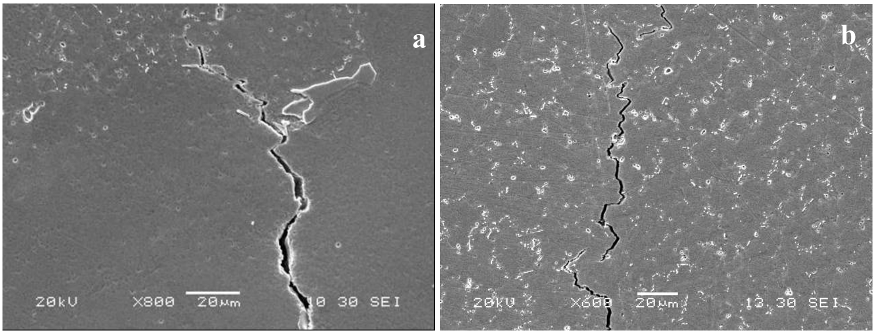

Figure 5: SEM image of (a) heat affected zone crack and (b) fusion zone crack in welded IN 738 alloy.

Liquating microconstituent that promotes HAZ grain boundary liquation cracking includes MC carbides and γ’ precipitates [18]. Whilst the carbides are non-equilibrium solidification product produced during ingot casting of the alloy, the γ’ particles are equilibrium strengthening phase of the alloy formed by solid-state precipitation reaction. It has been reported [19] that γ’ precipitates promotes susceptibility to HAZ microfissuring in superalloys through their rapid re-precipitation behaviour during cooling from the welding temperatures, which induces large shrinkage stresses along with a significant intragranular strengthening. Ojo *et al.* [19] further stated that besides the contribution to welding stresses, γ’ precipitates equally embrittle weld HAZ grain boundaries by persisting to temperatures at which they could react with the austenitic γ matrix to produce a liquid phase by a eutectic-type reaction. In the present work, similar to other reported investigations [20, 21] liquation of γ’ precipitates expectedly constituted a major contribution to grain boundary liquation in IN 738, due to their relatively higher volume fraction compared to other liquating constituents, such as, MC – type carbide and M_2SC sulpho-carbide.

The thickness of grain boundary liquid film can affect the resistance of an alloy to intergranular liquation cracking through its influence on the magnitude of stress required to cause microfissuring at a given temperature during weld cooling. Miller and Chadwick [22] related the tensile stress (σ) required to cause cracking, by overcoming surface tension at the solid-liquid interface on liquated grain boundaries (γ_{sl}), to the liquid film thickness (h) with eqn (1) which is equally referred to as the Chadwick’s relationship.

$$\sigma = {}^{2\gamma_{sl}}\!\big/_{h} \tag{1}$$

The implication of this relationship is that an increase in grain boundary liquid film thickness should reduce the stress required to cause microfissure by separation along intergranular solid-liquid interface. Accordingly, any factor that reduces the thickness of intergranular liquid film could presumably reduce the susceptibility of the alloy to HAZ cracking. Considering that γ’ particles are presumed as the major liquating phase in the alloy, minimizing or preventing their liquation can significantly reduce grain boundary liquid film thickness; and in relation to eqn 1, this enhances resistance to cracking.

A deviation from the equilibrium precipitate dissolution behaviour occurs during rapid heating, such as, that experienced during welding. Soucail and Bienvenu [23] have shown that the solvus temperature of γ’ particles may markedly depart from its equilibrium value during rapid heating, with the precipitates surviving to significantly higher temperatures than those predicted by equilibrium condition. They equally reported that this departure from equilibrium condition is dependent on the initial size of the precipitate particles; with the temperature at which complete γ’ dissolution occurred above the solvus temperature increasing as the particle size increases. This suggests that the size of γ’ particles influences the temperature at which complete solid state dissolution occurs; and thus, may reduce/increase their susceptibility to constitutional liquation. Moreover, Wang *et al.* [24] studied the dissolution kinetics of γ’ particles with different initial particle distributions and reported that the rate of γ’ dissolution depends strongly on the initial particle distribution; and that uniform size distribution produced the fastest dissolution kinetics. Consequently, this suggest that it may be possible to reduce grain boundary liquation in IN 738 alloy by ensuring a reduction in the size of γ’ particles which may be achieved through a preweld microstructural modification to obtain fine and uniformly distributed precipitates. Such treatment may yield a faster dissolution kinetics and lower complete solid-state dissolution temperature during the heating cycle of welding.

To investigate this possibility, IN 738 subjected to the SHT condition that produced mostly uniformly distributed fine γ’ precipitates was welded along with coupons given the MHT to compare their cracking susceptibility. The result of HAZ crack length measurements in the welded specimens presented in Figure 6 shows that the HAZ cracking was higher in the SHT condition compared to the MHT condition where the size of the γ’ particles is comparably larger and appear relatively more in number. The total crack length in the HAZ of the SHT materials is 1.1 mm which is comparatively significantly more than 0.2 mm in the MHT materials. Generally, HAZ cracking results from competition between the tensile driving stresses for cracking and intergranular liquation which embrittles grain boundary. Chadwick’s relationship suggests that decrease in intergranular liquid film thickness can be expected to enhance resistance to HAZ cracking by increasing the stress required to cause liquation microfissuring. However, in the present work, SHT produced a higher level of HAZ cracking compared to MHT condition. This is in spite of the supposedly minimal grain boundary liquation as a result of the reduced and uniformly fine sizes of the γ’ particles in the SHT condition. This cracking susceptibility observation is at variance with the expectation based on the Chadwick’s relationship. It should be noted, however, that such an expectation is essentially premised on the assumption that the HAZs in the preweld heat treatment conditions experience similar magnitude of welding stress during cooling because they were subjected to the same welding heat input parameter.

The magnitude of welding stress in the HAZ, however, can be significantly influenced by ability of the base alloy to relax part of the on-cooling generated welding stresses, which can be related to the hardness of the material prior to welding. The variation of base-alloy hardness in the two preweld heat treated material is shown in Figure 7. The MHT material produced a lower base-alloy hardness (380 VHN) compared to the SHT condition (442 VHN). A considerably hard base alloy that resists stress relaxation can cause a large portion of welding stress to be concentrated on a relatively weakened and liquated HAZ.

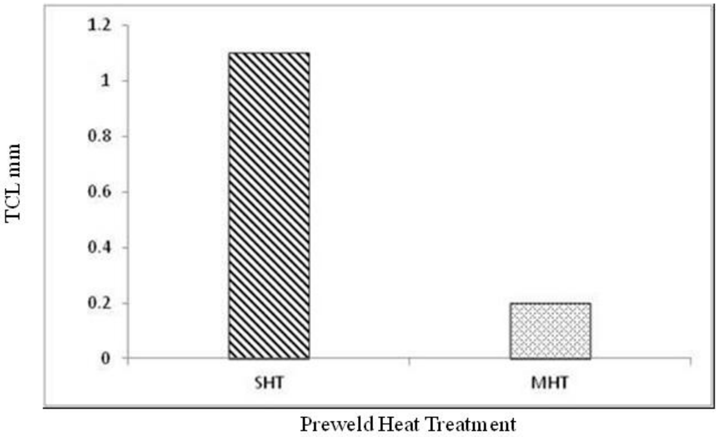

Figure 6: Variation in total crack length (TCL) with preweld heat treatment.

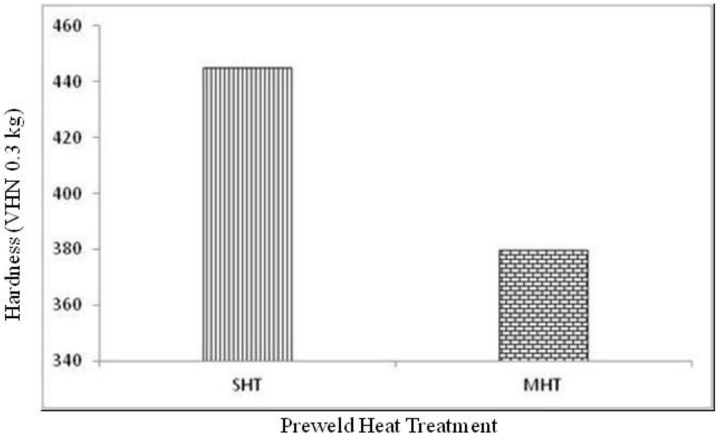

Figure 7: Variation in preweld heat treated base alloy micro-hardness.

Conversely, a base alloy with lower hardness capable of substantial stress relaxation can reduce the stress burden on HAZ to such an extent that could impart the driving force for intergranular liquation cracking. This is quite important in view of the fact that during TIG welding, considerable on-cooling stresses are generated due to the inherent rapid cooling rate of the process. In spite of the fewer number density/sizes of the γ' particles and the consequent lower grain boundary liquation expected in the SHT treated coupons, the comparatively higher hardness of the base alloy in the SHT condition induces corresponding resistant to welding stress relaxation; and this presumably caused high stress concentration on its susceptible HAZ regions which resulted in the level of HAZ cracking observed in the welded SHT treated coupons. On the other hand, the lower degree of cracking observed in MHT welded coupons can be related to a better capability of the base alloy in this condition to accommodate welding stress generated during weld cooling through stress relaxation. This is facilitated by the lower hardness of the base alloy in the MHT condition unlike in the SHT condition. Therefore, prevention or reduction in HAZ microfissuring in IN 738 requires not only a decrease in the magnitude of the thickness of the liquid film formed during welding but also a base-alloy condition with adequate capability for relaxing welding stresses.

4 Conclusions

The microstructure of TIG welded IN 738 superalloy with two different type of preweld microstructural morphology was carefully studied to better understand factors that mitigate HAZ cracking in the material and the followings emerged from the investigation:

1. In contrast to the recommendation of the manufacturer, preweld SHT, presumably designed to limit magnitude of the thickness of grain boundary liquid film, and hence reduced the extent of HAZ cracking during welding, resulted in increased HAZ cracking compared to a MHT condition.
2. Application of microstructural modification that induced moderate base-alloy hardness can considerably limit the occurrence of HAZ cracking in the IN 738 superalloy.
3. The improvement in resistance to HAZ cracking is related to a considerable decrease in the base-alloy hardness that accompanied the MHT. High base-alloy hardness reduces the ability of the material to relieve generated stresses, which can result in stress accumulation on embrittled liquated grain boundaries, resulting in microfissuring.

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