FRACTURE TOUGHNESS VARIATIONS FOR ALLOY 718 BASE METAL AND WELDS

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ABSTRACT

This paper reviews the effects of microstructural variations on the fracture toughness properties of Alloy 718 base metal and welds at 24 to 538°C. Seven different base metal lots, including five base metal heats and three different product forms from one of the heats, were tested in both the conventional (ASTM B637) precipitation treatment condition and a modified treatment condition that was developed to improve the fracture resistance for this superalloy. A gas-tungsten-arc weld was tested in both heat treatment conditions and the as-welded condition. Significant heat treatment and heat-to-heat variations in fracture toughness were found and the results were analyzed statistically to establish minimum-expected toughness values for use in fracture control analyses.

In the conventional heat treatment (CHT) condition, the presence of coarse second phase precipitates, δ phase in the base metal and δ plus Lavesphase in the weld, controlled the fracture behavior by causing premature microvoid nucleation and growth. The higher annealing temperature used during the modified heat treatment (MHT) dissolved these coarse particles and suppressed premature microvoid coalescence. This accounted for the improved fracture resistance exhibited by MHT materials. Heat-to-heat variations in fracture toughness behavior were attributed to differences in precipitate morphology and alternate secondary fracture mechanisms.

INTRODUCTION

Alloy 718 is a high strength nickel-base superalloy that possesses excellent corrosion and oxidation resistance, coupled with good tensile and creep properties. As a result, this alloy is used extensively in structural applications in the aerospace, nuclear, cryogenic and petrochemical industries. In addition, Alloy 718 has been selected for several welded applications because it exhibits superior weldability relative to most superalloys. The enhanced weldability characteristics are associated with the sluggish precipitation kinetics of the primary strengthening $\gamma^{"}$ (bodycentered-tetragonal Ni₃Nb) phase[1,2]. This sluggish age hardening behavior results in a relatively high ductility heat-affected-zone and fusion zone during cooling and aging. This permits relaxation of residual stresses and thereby improves the strain-age cracking resistance.

Superalloy 718-Metallurgy and Applications

Edited by E.A. Loria The Minerals, Metals & Materials Society, 1989 For most structural components, this superalloy is given a conventional heat treatment (CHT) per ASTM B637: annealed 1 h at 954°C and air cooled to 24°C, aged 8 h at 718°C and furnace cooled to 621°C, aged at 621°C for a total aging time of 18 h and air cooled. This heat treatment, however, has contributed to a series of failures in welded structures by severely reducing the ductility and impact toughness of the weld fusion zone[3-5]. The inferior toughness was attributed to the presence of Laves phase in the CHT weld metal. To increase ductility and fracture properties, the Idaho National Engineering Laboratory[5,6] developed a modified heat treatment (MHT): annealed 1 h at 1093°C and cooled to 718°C at 55°C/h, aged 4 h at 718°C and cooled to 621°C at 55°C/h, aged 16 h at 621°C and air cooled. The slower cooling rate reduced thermal stresses and the high annealing temperature dissolved the Laves phase, which restored adequate impact toughness to the fusion zone.

Many Alloy 718 components are highly loaded during service, so fracture control is a primary design consideration. Structural integrity assessments of such high strength structures require a comprehensive understanding of the fracture toughness characteristics for this superalloy. Since this alloy is metallurgically complex, involving precipitation of several phases, its fracture resistance is expected to be strongly influenced by heat treatment, processing history, melt practice and alloy composition. A series of investigations[7-11] were conducted at Westinghouse Hanford Company to evaluate the influence of microstructural variations on the fracture properties for Alloy 718 base metal and welds and the results were reviewed in this paper. The parameters examined in these studies included the effects of heat treatment, product-form and heat-to-heat variations. Fracture toughness tests were performed using linear-elastic (K $_{\rm IC}$) and elastic-plastic (J $_{\rm IC}$) test techniques. Toughness values for the wrought material were statistically analyzed to establish minimum-expected toughness levels that account for material variability. Metallographic and fractographic examinations were also performed to relate key microstructural features and operative fracture mechanisms to macroscopic properties.

EXPERIMENTAL MATERIALS AND PROCEDURES

Five heats of Alloy 718 were examined in the CHT and MHT conditions. In addition, three product forms (plate, round bar and upset forging) from a single heat were studied. Details of the material supplier, product form and melt practice for the seven material lots are summarized in Table I. A material lot is defined as a two letter code that designates a particular heat/product form combination: the first letter denotes the heat; the second letter indicates the product form (i.e., P for plate, B for bar and F for forging). A gas-tungsten-arc (GTA) weld was also tested in the as-welded, CHT and MHT conditions. The welding procedures used to manufacture this weld were detailed in Reference 7.

Chemical analyses and tensile properties for the test materials were reported in References 9-11. The CHT wrought metal exhibited 10-20% higher yield strength and 0-10% higher ultimate strength levels relative to its MHT counterpart. Ductility values were generally found to be insensitive to heat treatment. In contrast to the base metal response, yield strength levels for the age-hardened welds were relatively insensitive to heat treatment, while the ultimate strength for the CHT weld was slightly lower than that for the MHT weld. The two heat treated welds exhibited comparable uniform and total elongation values, but the reduction in area for the MHT weld was much greater than that for its CHT counterpart. The strength in the as-welded condition was approximately half of that for the aged materials.

Melt Material Product Form Lot Producer/Heat No. Practice Haynes 2180-1-9327 12.7 mm Plate VIM-ESR A-P Teledyne Allvac H-672 15.9 x 41.3 mm Bar VIM-VAR B-B C-P Huntington 52C9EK 12.7 mm Plate VIM-ESR D-F Cartech 84221 1.6 m Dia. x 66.7 mm VIM-VAR Forging Haynes 2180-6-9457 12.7 mm Plate VIM-ESR E-P 50 mm Dia. Bar VIM-ESR E-B Haynes 2180-6-9457 E-F Haynes 2180-6-9457 5:1 Upset Forging VIM-ESR Weld Huntington 60C7EY U-Groove Geometry GTA

Fracture tests were performed on deeply precracked (a/W = 0.65) compact specimens with a width (W) of 29.3 mm and thicknesses (B) ranging from 12.5 to 14.7 mm. Test details were provided in References 8-11. Values of J_{IC} were computed using the multiple-specimen J_R curve method given in ASTM Specification E813-87. J_R curves were constructed by plotting values of J as a function crack extension, Δa . J_{IC} was taken to be the value of J where a least-squares linear regression line of the following form

$$\ln J = \ln C_1 + C_2 \ln(\Delta a) \tag{1}$$

intersected the 0.2 mm offset blunting line

$$J = 2 \sigma_{e} (\Delta a - 0.2 \text{ mm})$$
(2)

where: C_1 , C_2 = regression constants σ_f = flow strength [σ_f = 0.5 (yield strength + ultimate strength)]

The ASTM specimen size criteria for valid J_{IC} determination were met for all specimens. Note that the ASTM E813-87 procedures for computing J_{IC} and individual J values differed somewhat from the earlier version given in E813-81; hence, toughness values reported herein were slightly different from those reported previously in References 8-11.

 $K_{\rm IC}$ fracture toughness tests, per ASTM Specification E399-83, were performed on the CHT welds tested at 24°C, because they exhibited a brittle fracture response. Specimen size requirements to assure valid plane strain conditions were satisfied for the four CHT weld specimens tested at 24°C. Equivalent $J_{\rm IC}$ values were then computed from experimental $K_{\rm IC}$ values by the following[12]

$$J_{IC} = K_{IC}^2 / E$$
 (3)

where E is Young's modulus.

Light microscopy, SEM and surface replica TEM microscopy techniques were used to characterize the microstructure and fracture surface morphology for

IDENTIFICATION OF ALLOY 718 HEATS

the base and weld metals. To relate fracture surface appearance to key microstructural features, selected areas of fracture surfaces were electropolished so the fracture surface topography and underlying microstructure could be studied simultaneously. Detailed metallographic and fractographic procedures were given in References 7 and 8.

RESULTS AND DISCUSSION

Fracture Toughness Response for Alloy 718 Base Metal

The effect of heat treatment on the $J_{\rm R}$ curve response for the seven base metal lots tested at 24, 427 and 538°C is illustrated in Figures 1-3. Results for the CHT material are shown on the left, while those for the MHT alloy are shown on the right. Comparison of the fracture resistance for CHT and MHT conditions on a lot-to-lot basis revealed that the MHT improved the $J_{\rm IC}$ and stable tearing properties for each material over the entire test temperature range. This improvement was most apparent in heats A, B and D, whereas heats C and E showed only modest increases in fracture resistance.

Systematic heat-to-heat variations were displayed by both heat treatments with the same ordering of heats observed over the entire test temperature regime (i.e., decreasing fracture toughness values were observed in essentially the same sequence: MHT lots B-B, D-F, A-P, E-F, E-P, C-P and E-B; CHT lots A-P, B-B, E-P, C-P, D-F, E-F AND E-B). Note that heats with the highest J_{IC} initiation values also yielded the highest stable tearing resistance. The microstructural features responsible for the heat treatment and heat-to-heat variations in fracture toughness will be discussed later.

The fracture toughness response for the plate, bar and forging from Heat E, represented by the solid symbols in Figures 1-3, revealed significant product-form differences for both heat treatments. In the CHT condition, $J_{\rm IC}$ values for the plate were consistently higher than those for the bar and



Figure 1. J_R curves for CHT and MHT Alloy 718 tested at 24° C.



Figure 2. JR curves for CHT and MHT Alloy 718 tested at 427° C.



Figure 3. JR curves for CHT and MHT Alloy 718 tested at 538° C.

forging. Comparison of the present results with those reported elsewhere[13-17], however, revealed no consistent trends with respect to product form (Figure 4). While the CHT round bar and forgings studied here and the forgings studied by Logsdon[13] had relatively low fracture resistance, the rectangular bar (lot B-B) and forging in Reference 14 displayed relatively high J_{IC} values. Similarly, the high toughness CHT



Figure 4. Summary of J_{IC} values for CHT and MHT materials. Solid lines denote mean J_{IC} levels for each statistically distinguished population of heats. Dashed lines represent lower tolerance limits for the lowest toughness populations.

plates (lots A-P, C-P and E-P) were contrasted by the low toughness exhibited by Stout and Gerberich's plate[16]. This behavior was also observed in the MHT condition. Hence, while different thermo-mechanical processing histories apparently resulted in fracture toughness variations, a unique relationship between a particular product form and $J_{\rm IC}$ response did not exist.

Test results for the seven material lots were statistically analyzed to evaluate the significance of heat-to-heat and product-form variations over the test temperature range and to establish minimum-expected toughness values. Multiple comparison procedures, as described in Reference 18, yielded the results displayed in Figure 4. For the CHT material, there were three distinguishable J_{IC} levels for the seven lots. Mean toughness values for these three levels were represented by the solid lines plotted on Figure 4. At each level, J_{IC} was independent of temperature over the 24-538°C range. Tobler's fracture toughness data[14] with a variance that was equivalent to that measured here, demonstrated that the temperature-independent behavior continued to -269°C. Tobler actually measured K_{IC}, but the corresponding J_{IC} values (per Equation 3) agreed very well with toughness levels reported here. Multiple comparisons for the MHT alloy revealed three distinguishable toughness levels and showed a significant reduction in J_{IC} with increasing temperature.

Lower bounds for tolerance limits bracketing 90% of a total population at a 95% confidence level were evaluated for the CHT and MHT populations with the lowest fracture resistance. These lower bounds are denoted by the broken lines on Figure 4. For the CHT material, only two base metal heats exhibited toughness values that fell below the lower bound of 36 kJ/m^2 and the inferior fracture resistance for these materials was believed to be associated with anomalous microstructures: Logsdon's[13] low toughness forging had a contiguous grain boundary carbide network that induced intergranular fracture; and Stout's[16] plate received a high temperature duplex age (760 and 648°C) that caused its yield and ultimate strength levels to exceed those for the other CHT heats. These observations indicate that care should be taken to avoid anomalous microstructures, particularly contiguous intergranular precipitate networks, because they can degrade the toughness below the minimum-expected $J_{\rm IC}$ level.

The high solution annealing temperature employed during the MHT tends to homogenize the structure, so anomalous microstructural features are not expected to be present in this condition. Therefore, toughness values for MHT Alloy 718 are consistently expected to fall above the lower bound curve. Lower bound J_{IC} values (80-59 kJ/m² at 24-538°C) for the MHT materials were 60 to 100% higher than their CHT counterpart over the entire temperature range, demonstrating that the MHT should be used for applications where fracture is a primary concern.

The microstructure and fracture surface appearance for the seven material lots were studied to relate key microstructural features and operative fracture mechanisms to macroscopic fracture toughness behavior. A summary of these findings is given in Table II (material lots were separated into distinguishable toughness populations as determined above). Representative microstructures for the CHT and MHT materials are displayed in Figures 5 and 6. All lots exhibited large niobium-rich and titanium-rich MC-type carbides regardless of heat treatment. Coarse orthorhombic & precipitates and a few Laves particles decorated the CHT grain boundaries and annealing twins. Intragranular & particles were also observed in some CHT heats. Grain sizes for the CHT material ranged from ASTM 5 to 12. The high temperature MHT anneal resulted in considerable grain growth (ASTM 2-1/2 to 4) and it completely dissolved the coarse & and Laves particles, with the exception of lot E-B in which ghost boundaries containing small & remnants survived. Unique microstructural features and qualitative ranking of the carbide and & phase densities are listed in Table II. Typical low, medium and high densities of carbides and & precipitates were illustrated in Figures 5 and 6 of Reference 10.

Fractographic examinations revealed that primary dimples nucleated at failed MC carbides early in the plastic straining process, regardless of heat treatment. Separation of the ligaments between primary microvoids then



Figure 5. Typical microstructures for wrought Alloy 718. (a) CHT Lot E-P. (b) MHT Lot E-P.



 Figure 6. TEM surface replica micrographs showing precipitate morphology.
(a) CHT Lot D-F. MC-type carbides and acicular & precipitates were observed throughout the matrix.
(b) MHT Lot E-P. The MHT solution anneal dissolved the & particles, but it had no effect on the MCtype carbides.

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SUMMARY OF MICROSTRUCTURE AND FRACTURE SURFACE NORPHOLOGY FOR MRDUGHT ALLOY 718

Heat- Prod. Form	Heat Treat	ASTM Grain Size	Carbide Density	Coarse & Density	Channel Fracture Density	Dimple Size (µm)	Secondary Disple Size (µm)	Comments
B-B	CHT	11}	Low	Interm.	None	20±6	3.1±1.0	
A-P	CHT	81	Interm.	Interm.	None	20±5	2.6±0.8	Elongated grains with no $\boldsymbol{\delta}.$
E-P	CHT	9/12	Intern.	Interm.	None	21±6	3.4±1.2	
C-P	CHT	5	Interm.	Interm.	None	14±5ª/ 28±8 ^b	1.4±0.5ª	Severe & banding; tearing topography surface
E-P	CHT	5/11	Interm.	High/ Interm.	None	14±6	1.9±0.6	Extensive δ in necklace structure; localized carbide clusters.
D-F	CHT	101	Low	High	None	14±6	2.5±0.6	Tearing topography surface in regions with high δ density.
E-8	СНТ	6/11	High	High/ Interm.	None	21±8	2.2±0.8	Extensive carbide clusters; stringers of $\boldsymbol{\delta}$ and carbides.
8-8	HHT	4	Low	None	Very Low	27±10	v. s. e	Elongated tear ridges.
0-F	MHT	3	Low	None	Very Low	34±13	v. s.	
A-P	мнт	23	Intera.	None	Low	29±10	v. s.	
E-F	MHT	3	Interm.	None	Interm.	21±11	v. s.	Localized carbide clusters.
E-P	MHT	з	interm.	None *	Interm.	21±7 *	v. s.	
E-8	NHT	J	High	Low	None	20±7 */ 33±16 ^d	v. s.	Extensive carbide clusters; & remnants; secondary dimples.
C-P	NHT	21	Intern.	None	High	24±9	None	
⁸ Inside banded regions. ^b Outside banded regions.					^d Regions with intermediate carbide density. [®] V. S Void Sheets.			

Doutside banded regions.

^CRegions with very high carbide density.





(c)

Electron fractographs of CHT material. (a) Primary dimples surrounded by secondary dimple rupture network. Note that the primary dimples were nucleated by broken carbides (b)

Figure 7

Note that the primary dimples were nucleated by broken carbides. (b) Metallographic-fractographic profile showing that the acicular & particles served as microvoid nucleation sites for the secondary dimples. (c) Metallographic-fractographic interface for CHT lot E-B illustrating the dominant primary dimples in regions with very high carbide density.

resulted in an advance of the crack front; hence, the secondary cracking mechanism responsible for linking the primary microvoids controlled the overall fracture behavior. In the CHT material, primary dimple growth was pre-empted by nucleation of a secondary microvoid population at the δ particles, as shown in Figures 7a and 7b. Table II shows that the fracture toughness was related to the inverse of δ phase density.

The close proximity of primary dimples inside the carbide clusters in CHT lot E-B, and to a lesser extent in lot E-F, kept δ particles from participating in the fracture process (Figure 7c). As a result, premature failure of carbides dominated the local fracture process inside the clusters, whereas δ precipitates controlled the fracture response elsewhere. The particularly high density of carbide clusters in CHT lot E-B was responsible for its inferior toughness.

Separation of ligaments between primary dimples in the MHT alloy involved three mechanisms (Figure 8): coalescence of microvoids, void sheet formation[19] initiated by the γ " phase, and channel fracture[20] associated with localized tearing along planar slip bands. These mechanisms usually operated during the final stages of fracture as the primary microvoids were about to impinge; hence, they did not necessarily degrade an alloy's fracture resistance. Because primary microvoid growth was not interrupted prematurely, MHT alloys exhibited superior fracture resistance and larger primary dimples





(b)



(a)

(c)

Figure 8 SEM fractographs illustrating fracture mechanisms in MHT material. (a) Coalescence of the primary microvoids. Slip markings on dimple walls indicate that microvoid growth involved extensive plastic straining. (b) Primary dimples separated by small regions of void sheets nucleated by the r" precipitates. (c) Lot C-P. Channel fracture between the primary dimples.

(Table II) relative to their CHT counterparts. An exception was noted in MHT lot C-P where extensive channel fracture prematurely interrupted microvoid growth. This accounted for the relatively small improvement in toughness for this lot, relative to the CHT response.

Since primary microvoid coalescence was the dominant fracture mechanism, it is not surprising that lots containing low carbide densities exhibited comparatively high fracture tougnness and the greatest improvement in overall fracture resistance relative to the CHT response. This observation was also supported by the larger dimple size found in the higher toughness MHT heats. In MHT lot E-B, the very high density of carbide clusters coupled with δ remnants contributed to the inferior toughness behavior exhibited by this material.

Fracture Toughness Response for Alloy 718 Welds

The fracture response for the weld in the as-welded, CHT and MHT conditions is illustrated in Figures 9-11. At 24°C, the CHT weld failed via an unstable cracking mode at K_{IC} levels between 62 to 73 MPa/m, corresponding to J_{IC} levels between 19 to 27 kJ/m². The unstable fracture behavior was represented by the horizontal J_R curve in Figure 9. The K_{IC} values for the CHT weld were slightly higher than the 55 MPa/m obtained by Forman[21] and Lorenz[22] for GTA welds containing surface flaws. At elevated temperatures,



Figure 9. J_R curves for welds tested at 24° C. J_{IC} for the CHT weld was calculated from experimental K_{IC} values, per Equation 3. The horizontal J_R curve indicates that the CHT weld had essentially no tearing resistance once J_{IC} was exceeded.



Figure 10. $J_{\mbox{R}}$ curves for Alloy 718 welds at 427°C.



Figure 11. J_R curves for Alloy 718 welds at 538°C.

the CHT material displayed a stable tearing mode, but its overall fracture resistance remained very low (J $_{IC}$ = 30 - 31 kJ/m²).

The as-welded and MHT welds displayed stable tearing and superior fracture properties at all test temperatures. $J_{\rm TC}$ values ranged from 208 to 269 kJ/m² for the as-welded material and 105 to 132 kJ/m² for the MHT material. In addition to exhibiting high initiation toughness values, the as-welded and MHT welds were very resistant to tearing, as evidenced by their steep $J_{\rm R}$ curves. These results demonstrate that in the as-welded condition, Alloy 718 welds possess adequate fracture resistance to preclude fracture concerns for most low-strength structural applications. For high strength applications where welds must be age hardened, the MHT is preferred over the CHT because it dramatically improves toughness properties without sacrificing strength.

Metallographic and fractographic examinations were performed to characterize the effect of microstructure on fracture toughness response. Typical microstructures for the three weld conditions are shown in Figure 12. The as-welded structure exhibited extensive interdendritic Laves phase in the microsegregated regions enriched in niobium content. CHT welds exhibited a blocky Laves phase in the interdendritic regions which was surrounded by Widmanstatten δ precipitates. The 954°C CHT anneal did not homogenize the interdendritic structure in the fusion zone; moreover, significant amounts of δ precipitated in Nb-rich regions at this temperature. Unfortunately, Laves and δ precipitates contributed little to the strength of the welds and they acted as localized crack initiation sites that degraded ductility and toughness. The MHT, with a 1093°C solution anneal, homogenized the interdendritic structure and drastically reduced the amount of Laves phase in the fusion zone. Only traces of randomly distributed Laves particles remained within the MHT weld, which is consistent with results from References 5 and 6.



(b)



(c)

Figure 12 Microstructures for the welds. (a) Aswelded condition exhibiting Laves particles in the interdendritic regions. (b) CHT weld. Evidence of blocky Laves particles (white) surrounded by & needles (dark region). (c) MHT weld. Traces of Laves phase remained in the weld zone.

Representative fractographs of the weld toughness specimens are displayed in Figure 13. In the as-welded condition, fracture surfaces consisted of a dimple rupture network initiated by fractured Laves particles. The alignment of Laves particles in the interdendritic regions was responsible for the well-defined rows of dimples evident in Figure 13a. Metallographic profiles illustrated that fractured Laves particles nucleated microcracks away from the primary fracture surface, but these microcracks were arrested in the surrounding ductile nickel matrix. This observation demonstrated that the fracture-resistant matrix in the unaged weld accommodated cracking of the Laves phase, thereby accounting for the superior fracture properties exhibited by the as-welded material.

CHT welds exhibited rows of shallow, ill-defined dimples that initiated at interdendritic Laves and δ particles (Figures 13b and 13c). The poorly defined dimple structure, reminiscent of a tearing topography surface,[23] resulted from a combination of the high density of microvoid nucleation sites and the high strength matrix that suppressed plastic deformation around the growing microvoids. The limited amount of void growth precluded development of well-defined dimples. Metallographic profiles of CHT weld specimens revealed that fractured Laves and δ particles were confined solely to the region immediately adjacent to the primary fracture surface. The low fracture-resistant CHT matrix was unable to arrest the microcracks initiated by the coarse second phase particles. As a result, these microvoids coalesced





(c)

(d)

Figure 13. Electron fractographs of the Alloy 718 welds. (a) As-welded condition. Rows of dimples nucleated by the interdendritic Laves particles. (b) CHT weld. Rows of ill-defined dimples associated with the tearing topography surface mechanism. (c) CHT weld. Metallographic-fractographic profile illustrating that the high density of Laves and & particles provided closely spaced microvoid initiation sites. (d) MHT weld. Metallographic-fractographic interface showing equiaxed dimples nucleated by Laves remnants.

causing an overall advance of the macroscopic crack before other particles further away from the crack plane could fail. Moreover, the high density of Laves and δ particles, coupled with the low fracture resistance of the matrix, resulted in the very low toughness displayed by the CHT weld.

The MHT weld fracture surfaces exhibited well-defined, equiaxed dimples (Figure 13d) that were nucleated by the Laves phase remnants. The microvoids initiated as a result of decohesion of the particle/matrix interface, in contrast to the catastrophic fracture of large Laves particles in the as-welded and CHT welds. The Laves phase remnants in the MHT weld were more resistant to fracture than the larger particles present in the other two conditions. Hence, the reduction of the size and density of coarse particles in the fusion zone was responsible for the drastic improvement in fracture resistance for the MHT weld.

CONCLUSIONS

The effects of microstructural variations on the fracture toughness of Alloy 718 base and weld metals were reviewed. Specifically, the toughness responses for a GTA weld and seven base metal lots were evaluated and a summary of the findings is given below.

- 1. Fracture properties for the MHT base metal and weld were superior to those for their CHT counterparts, thereby supporting the use of the MHT when fracture resistance is a primary design concern. In the as-welded condition, the weld possessed adequate toughness to preclude fracture for most low-strength applications.
- Heat-to-heat and product-form variations in fracture toughness were associated with different microstructures resulting from variations in thermo-mechanical history. Comparison of the present results with those in the literature, however, revealed no unique relationship between a specific product form and toughness response.
- 3. Statistical analysis of toughness results identified three distinguishable J_{IC} levels for both the CHT and MHT materials. J_{IC} for the CHT material was independent of temperature, but toughness values for the MHT alloy decreased by 20% as temperature increased from 24 to 538°C. Minimum-expected J_{IC} levels for the MHT alloy, which ranged from 80 kJ/m² at 24°C to 59 kJ/m² at 538°C, were 60 to 100% higher than that for the CHT material (36 kJ/m²).
- 4. In all base metal lots, primary microvoids nucleated at broken MC-type carbides during the early stages of plastic straining, regardless of heat treatment. A second population of microvoids, nucleated by coarse δ particles, pre-empted primary microvoid growth in the CHT alloy and thereby caused premature crack extension.
- 5. By dissolving the coarse δ particles, the MHT suppressed secondary microvoid formation and enhanced the fracture resistance for this alloy. During the final stages of microvoid coalescence, primary dimple growth was sometimes interrupted by void sheet formation or channel fracture, but these mechanisms generally occurred too late in the fracture process to impact fracture properties. Exceptions were noted in the two lowest toughness MHT lots. These materials exhibited secondary fracture mechanisms that prematurely pre-empted primary microvoid coalescence: extensive channel fracture occurred in lot C-P and δ remnants nucleated a secondary dimple rupture network in lot E-B.
- 6. The inferior fracture properties for the CHT weld were associated with the low fracture-resistant matrix coupled with the high density of Laves and & precipitates that served as microvoid nucleation sites. The high temperature MHT anneal significantly reduced the size and density of second phase particles within the fusion zone and thereby enhanced fracture properties without sacrificing strength. In the as-welded condition, the highly fracture-resistant weld matrix arrested microcracks initiated by the coarse Laves phase. This behavior resulted in the superior fracture properties displayed by the as-welded material.

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