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Research Paper

STUDY OF MECHANICAL BEHAVIOR IN AUSTENITIC STAINLESS STEEL 316 LN WELDED JOINTS

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Low carbon grade type 316 austenitic stainless steel, alloyed with nitrogen designated as 316 LN SS exhibit superior strength at ambient and high temperatures, excellent corrosion resistance to replace other expensive materials. Austenitic stainless steel get sensitized during welding. The problem of sensitization can be over come by decreasing the carbon content. Reducing the carbon content would cause drastic reduction in mechanical properties. Replacing much of the carbon with nitrogen can offset this deterioration in mechanical properties. Nitrogen in solid solution is the most beneficial alloying element in promoting high strength in austenitic SS without sacrificing their good ductility and toughness. The upper limit of nitrogen is set on considerations of minimizing scatter in mechanical properties and improving weldability. Phosphorous, sulphur and silicon are treated as impurities having adverse effects on weldability. The selection of the material is based on its good combination of tensile, creep strength, ductility and its high resistance to stress corrosion cracking and sensitization. Nickel content has been known as an element that stabilizes austenite and increases toughness. Nitrogen has diverse effects on the microstructure and solidification cracking behavior of type 316 stainless steel. In the present paper an attempt has been made to present the influencing factors on the weldability of 316 LN SS. The welding processes such as single pass activated TIG (A-TIG), multi-pass activated TIG (MP-TIG) welding and Laser welding were discussed in relation to the 316 LN austenitic stainless steel. Residual stresses are a system of self-equilibrating stresses, which may exist when it is free from external loads or forces. The existence of such stresses have been analyzed and compared. Non-destructive techniques have been used for measuring residual stresses and defects in welding. The purpose of this review is to investigate weldability of 316 LN SS, study the beneficial effect on fully austenitic steels with particular emphasis on nitrogen-alloyed and stabilized stainless steels.

Keywords: Austenitic stainless steel, Sensitization, Weldability, Residual stresses, Hot cracking and nitrogen

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INTRODUCTION

Nitrogen added stainless steels exhibit several unique characteristics such as superior strength at ambient as well as high temperatures, excellent corrosion resistance in various media and are candidate materials to replace more expensive materials. In 316 LN SS for nuclear service, carbon has been reduced to decrease the propensity for inter granular corrosion. Nitrogen is added mainly to recover the high strength and elevated temperature properties associated with the loss of carbon in comparison with conventional 316 (Bandy and Van, 1985). Nitrogen is expected to have strong influence on the hot cracking behavior (Bhanu et al., 1993). Nitrogen in weld metal arises from various sources such as prior content in the base metal, intentional addition through the shielding gas or as inadvertent pick up due to inadequate shielding during welding. Nitrogen resides in weld metal in the following forms (Bhanu et al., 1994): (1) interstitial nitrogen dissolved in the lattice structure-may collect around lattice defects, (2) combined nitrogen present as nitrides, and (3) occlude nitrogen in pores. It is the first two forms of nitrogen that affect the metallurgical behavior of the weld metal. Nitrogen has strong stabilizing effect on austenite when present in iron Fe-N austenite is isomorphos with Fe-C austenite in the Fe-C-N system. The solubility of nitrogen in austenite is enhanced by several solutes such a Cr,Mn and to great extent by V and Nb.The Mo increases solubility slightly while C, Si and Ni tend to decrease it (Borland, 1960). The nitrogen activity in steel decreases when elements with N-affinity such as AI, Cr, or Nb are added, while it increases with additions of

C, Ni, P, S or Si (Bhanu, 1994). Nitrogen in solid solution is the most beneficial alloying element in promoting high strength in austenitic SS without sacrificing their good ductility and toughness as long as the solubility limit of nitrogen in austenite is not exceeded (Brooks, 1974). Together with chromium and molybdenum, nitrogen additions to austenitic SS also improve resistance to pitting and crevice corrosion (Borland and Younger, 1960; and Briant et al., 1982) sensitization (Dundas and Bond, 1975; Eckenrod and Kovach, 1979; and Dutta et al., 1993), and Stress Corrosion Cracking (SCC) (Eckenrod and Kovach, 1979; and Folkhard, 1988). Due to the above reasons, nitrogen-added austenitic SS are finding increasing applications in critical industries. During welding of austenitic SS, besides sensitization, the most commonly encountered problem is that of hot cracking of the weld metal. The most common method adopted is by intentionally rendering the austenitic SS weld deposit inhomogeneous by retaining some amount of high temperature d-ferrite to room temperature (Garner, 1977). Gill et al. (1987) reported that it was difficult to separate the beneficial effect of nitrogen on pitting corrosion resistance from its tendency to stabilize austenite and thereby prevent formation of d-ferrite. There is a general consensus in literature that nitrogen additions improve pitting and crevice corrosion resistance of austenitic SS and their weldments, provided the solubility limit of nitrogen is not exceeded (Hull, 1967a and 1967b; Jargelius and Vallin, 1986; Goodwin, 1988; Kamachi et al., 1996; and Kwok et al., 2006). The beneficial effect of nitrogen on pitting corrosion resistance of SS is attributed to improved passivity.

Chemical composition, mechanical properties of 316 LN SS were given in Tables 1 and 2.

Table 1: Chemical Composition (Wt%)					
Element	Minimum	Maximum			
Chromium	16.0	18.0			
Molybdenum	2.00	3.00			
Nickel	10.0	14.0			
Manganese		2.0			
Phosphorus		0.045			
Sulfur		0.030			
Silicon		0.75			
Carbon		0.030			
Nitrogen	0.10	0.16			
Iron	Balance				

Table 2: Room Temperature Properties				
Property	ASTM A 240			
Yield Strength, 0.2% offset	30 ksi 205 MPa			
Ultimate Tensile	75 ksi			
Strength	515 MPa			
Elongation in 2" (51 mm)	40%			
Hardness	217 Brinell 95 HRB			

WELDABILITY OF 316 LN SS

The austenitic stainless steels are considered the most weldable of the stainless steels. They are routinely joined by all fusion and resistance welding processes. Two important considerations for weld joints in these alloys are: (1) avoidance of solidification cracking, and (2) preservation of corrosion resistance of the weld and heat-affected zones. Type 316 LN stainless steel often is welded autogenously. If filler metal must be used for welding Type 316 LN, it is advisable to utilize the low carbon Types 316 L or E318 filler

metals. Contamination of the weld region with copper or zinc should be avoided, since these elements can form low melting point compounds, which in turn can create weld cracking. Fusion welding is widely used to fabricate the reactor vessel and piping due to its complexity and size involved. Service experience has shown that the cracking in weld joints is the life limiting factor (Manning et al., 1980; and Laha et al., 2007). Major welded component often have to be taken out of service at a point in time where, little damage would have occurred in the parent material. In general, Tungsten Inert Gas welding (TIG) process with and without filler metal are used to join this class of materials. However, the most constraints of TIG welding of stainless steel lie in the limited thickness of the material which can be welded in a single pass, poor tolerance to chemical composition of the deposited weld metal leading to the formation of undesirable phases and the low productivity of joining. In this process, weld penetration achievable in single pass welding of Stainless Steel (SS) is limited to 3 mm when using argon as shielding gas. The penetration capability of the arc in TIG welding can be significantly increased by application of a flux coating containing certain inorganic compounds on the joint surface prior to welding (Matsuda et al., 1982; and Mathew et al., 2007). A-TIG welding is nothing but an activated TIG process using multi-component flux. The flux in the form of paste is generally applied on the surfaces of the joint before welding. Using A-TIG welding, penetration upto 300% has been achieved in austenitic stainless steel. Use of activated fluxes during TIG welding has been found to reduce the magnitude of tensile residual stresses and distortion in welds. Multi-pass (TIG) welding process generally uses a V-groove for plates of 6mm thickness and a filler wire of 316 L SS of 1.6mm diameter. To reveal the ä-ferrite in the weld metal immersion etching in boiling murakkami etchant is generally used. This etchant consists of (10 gm potassium ferric cyanide, 10 gm potassium hydroxide and 100 ml water). The presence of σ -phase can be detected by etching with modified murakkami etchant (30 gm potassium ferric cyanide, 30 gm potassium hydroxide and 150 ml water). Wealth of information exist in the literature on the welding process effects of 316 LN material

on the low temperature and high temperature mechanical properties. A-TIG and multipass TIG welding processes have been used to join 316 LN material and evaluated and compared the mechanical properties, microstructures of these weldments.

The chemical compositions 316 L(N) SS, 316 L(N) weld metal of MP (TIG), and 316 L(N) weld metal of (A-TIG) have been given in Table 3.

The welding parameters of A-TIG and multipass TIG welding of 316 LN have been given in Table 4.

Table 3: The Chemical Compositions 316 L(N)SS, 316 L(N) Weld Metal of MP (TIG), and 316 L(N) Weld Metal of (A-TIG)									
Material	С	Cr	Ni	Мо	Mn	Si	S	Р	N
316 L(N) SS Base Metal	0.028	17.5	12.2	2.3	1.65	0.44	0.01	0.024	0.085
316 L(N) Weld Metal (MPTIG)	0.026	19	11	2.3	1.5	0.53	0.01	0.025	0.087
316 L(N) Weld Metal (ATIG)	0.030	17.5	12.0	204	1.45	0.5	0.005	0.021	0.09

Table 4: The weiding Parameters of A-TIG and Multi-Pass TIG weiding of 316 LN					
Welding Parameters	A-TIG	Multi-Pass TIG			
Current (A)	200	110			
Voltage (V)	12.5	12			
Travel Speed (mm/min)	120	80			
Number of Layers	Single Pass	7 Passes			
Heat Input (KJ/mm)	1250 J/mm	990 J/mm			
Plate Thickness	6 mm	6 mm			

HYBRID WELDING

In hybrid tests, torch of GMAW (gas metal arc welding) was placed next to the laser welding head with a handling system allowing movements necessary for changing interaction parameters. The GMAW machine had a modified contact nozzle. The shielding gas was guided to the melt via an extra nozzle. The parameters of GMAW were kept in a moderate level, i.e., the arc current and the voltage were selected in order to stay in a so-called "coldarc" field. The angle of the torch and the impingement point of filler wire were varied. The material in hybrid experiments was an austenitic stainless steel AISI 316 LN. Test pieces were as in filler wire experiments, but one or two root welds were made using only with filler wire and laser beam. The main interest in welding by a hybrid process for the narrow gap configuration was

to ascertain if it is possible to have the arc inside the groove (Figure 1).



SOLIDIFICATION CRACKING

Solidification cracking occurs predominantly by the solutes to form low-melting phases, which under the action of shrinkage stresses accompanying solidification cause cracking. The initial theories (Medovar, 1954) took into account the fact cracking is associated with segregation; the wider the liquid-solid range of the alloy, the grater the susceptibility. The 'Generalized Theory' of cracking was proposed by Menon and Kotecki (1989). According to Borland, solidification involves four stages that are classified according to the distribution of the liquid and solid phases, as shown in Figure 2 for a binary alloy. In stage 1, the solid phase is dispersed, the liquid phase is continuous and both the phases can accommodate relative movement. In stage 2, both the liquid and solid phases are continuous, but the solid dendrites are interlocked and only the liquid is capable of movement. At this stage, the liquid can heal



any cracks formed. In stage 3, the solid crystals are in an advanced stage of development and the free passage of liquid is prevented. The liquid is present in very small quantity. At this stage, If stress is applied which cannot be filled by the remaining liquid phase. This stage, during which much of the cracking occurs, is called the Critical Solidification Range (CSR). In stage 4, solidification is complete and no cracking involving the liquid phase is possible. Borland stated that for high cracking susceptibility, in addition to a wide freezing range, the liquid must also be distributed in a way that allows high stresses to be built up between grains. Hot cracks are believed to initiate during the later stages of solidification when most of the liquid has solidified.

DIFFERENT MANIFESTATIONS OF SOLIDIFICATION CRACKING

Solidification cracking is observed as: (a) gross cracking occurring at the junctions of dendrites with differing orientations, detectable by visual and liquid penetrate testing, and (b) micro fissuring in the interdendritic regions which are revealed only by applications of strain to the cracked region or at high magnifications.

The heat input during welding also affects the hot cracking behavior significantly, primarily by affecting the amount and scale of segregation (Mishra *et al.*, 1997). Using the Sigmajig test evaluate GTAW, laser and electron beam welds found that increased heat inputs decreased the threshold stress for cracking, thereby increasing cracking susceptibly. Higher energy densities decreased cracking, cracking resistance was progressively higher for GTAW, electron and laser beam welding, in the order of decreasing heat inputs.

EFFECTS OF NITROGEN ON SOLIDIFICATION CRACKING

In Fully Austenitic mode (FA) mode compositions, the most direct effect of nitrogen

on cracking is through in weld metal ferrite content. Nitrogen, by stabilizing the austenite phase, decreases the occurrence of primary delta-ferrite, which results in reduced resistance to cracking. In recent work (Moskowitz, 1967), N additions enhanced cracking in fully austenite 316L weld metal with S (0.012%) but did increasing cracking in low-S (0.001%) weld metal. In summery, it appears that nitrogen has diverse effects on the microstructure with corresponding consequences on the cracking behavior of type 316 stainless steel.

- In FA mode alloys, a reduction in ferrite content and increase in segregation, leading to an increase in cracking,
- In austenitic-mode alloys, retardation of polygonisation, which decreases wettability and hence cracking,
- In fully austenite alloys, a refining effect on the solidification leading to decrease in the cracking at moderate levels (up to 0.16 wt. %)
- N appears to act synergistically with S to increase cracking in fully austenitic 316 L weld metal; low-S weld metal shows no effect on increasing N levels.

EFFECT OF \triangle -FERRITE ON SOLIDIFICATION CRACKING

The effect of varying delta-ferrite contents on cracking in stainless steels was studied by (Mozhi *et al.*, 1987) using the cast pin tear test. Hull found that, while the cracking susceptibility was high for fully austenitic compositions, whereas specimens with 5-30% ferrite were quite resistant to cracking. A number of factors have been proposed to

explain the beneficial effects of delta-ferrite on cracking behaviour:

- The higher solubility for impurity elements in delta-ferrite leads to less interdendritic segregation and reduces cracking tendency (Mozhi *et al.*, 1985).
- The presence of ferrite results in a larger interface area due to the solid state transformation to austenite that begins soon after solidification. The increased area disperses the concentration of impurity elements at the grain boundaries.
- The presence of ferrite refines the grain size of the solidified metal, which results in better mechanical and cracking resistance.
- The ductility of ferrite at high temperatures is greater than that of austenite, allowing relaxation of thermal stresses.
- The lower thermal expansion coefficient of ferrite as compared to austenite results in less contraction stresses and fissuring tendency.

EFFECTS OF IMPURITY AND ALLOYING ELEMENTS ON CRACKING

The elements such as, sulphur, phosphorous, boron, niobium, titanium, and silicon were identified as most harmful. Among the elements whose effects on cracking were investigated, Hull found molybdenum, manganese and nitrogen beneficial. Sulphur is known to be an undesirable impurity in welding of stainless steels due to the formation of low-melting sulphide films along the interdendritic and grain-boundary regions. Sulphur is almost insoluble in all three major constituents of stainless steel, viz., iron,

chromium and nickel. Sulphur is strongly rejected at the liquid during solidification of austenite, rapidly lowering the melting point interdendritic liquid. On the other hand, deltaferrite shows higher solubility for elements like S, P, Si and Nb. Phosphorous ranks next to sulphur in the list of elements detrimental to good weldablity in stainless steels. Like, sulphur, P forms low-melting eutectics with iron, chromium and nickel (Ogawa et al., 1982) identifies phosphorous as particularly harmful in fully austenitic stainless steels since it has a strong tendency to spread as liquid films. Further, it has been stated that the low diffusivity of P in both austenitic and ferritic phases even at high temperatures virtually precludes homogenization (Osozawa and Okada, 1976). Further, the addition of lanthanum and other rare-earths has been found highly effective in binding the P and S as stable compounds (Parvathavarthini et al., 1994). Ti and Nb increase cracking by forming various eutectics in conjunction with S, N and C. The most deleterious phases in such steels are probably the carbosulphides such as titanium, followed by the carbonitrides and the carbides. In Ti-bearing stainless steels, the amount of these phases increases with increasing stabilization ratio, i.e. (Ti, Nb)/ (C, N). A lower ratio is desirable for minimizing the cracking propensity. The Ti-bearing fully austenitic stainless steel Ti/(C + N) between 2 and 3 was found optimal.

SENSITIZATION

Material degradation of stainless steels begins whenever they experience temperature in the range 450-800 °C. After sufficient time at temperature, intermetallic compounds (principally Cr carbides) precipitate in grain boundaries and (later) inside the grains as well. This precipitation creates a condition known as sensitization that ruins corrosion resistance while reducing cryogenic ductility and toughness. With continued aging, property degradation at 4 K is progressive. The toughness at 4 K declines gradually at first and falls more abruptly as intergranual fracture begins to dominate. Service embrittlement results if the precipitates become extensive enough to establish a continuous, low-energy path for cracks. A transgranular-to-intergranular fracture transition then occurs, constituting unacceptable embrittlement. Sensitization effects are determined by the diffusion rates of the relevant atomic species in particular steels. The primary cause of sensitization in stainless steel is C coming out of solution and forming the intermetallic compound, M23C6. Nitrogen comes out of solution at a slower rate, so that in micro alloyed 316 LN steels (described below) the formation of nitrides such as Cr₂N becomes an important secondary factor. The aging time required to sensitize a steel is fixed by the details of alloy composition, condition, and temperature. Time-temperature-sensitization curves were constructed recently to predict the aging time for some 316 LN-type steels. The curves purport to illustrate basic compositional effects and provide guidance to avoid sensitization. Because of uncertainties, published data for individual alloys must be verified before any application is made.

GENERAL CORROSION PROPERTIES OF 316 LN

Pitting Corrosion

Austenitic SS are susceptible to a number of manifestations of localised corrosion attacks,

of which the major problem is caused by sensitization which can cause failure of welded components by Intergranular Stress Corrosion Cracking (IGSCC) in the presence of residual stresses. Sensitization occurs due to Cr₂₃C₆ precipitation, when the austenitic SS is heated in the temperature range of 723 to 1123 K. which causes depletion of Cr from regions adjacent to the grain boundary to a level thus making the steel susceptible to Intergranular Corrosion (IGC) and IGSCC attacks. Austenitic SS get sensitized during service in the above temperature range or during welding. High heat input during welding increases the risk of sensitization because the slow cooling rates lead to longer residence time in the sensitization temperature range. The problem of sensitization can be overcome by decreasing the carbon content. However, reducing the carbon content would cause drastic reduction in mechanical properties. Replacing much of the carbon with nitrogen can offset this deterioration in mechanical properties. The presence of δ -ferrite in the weld metal causes partitioning of alloying elements, and extensive segregation of sulphur and phosphorous at the δ -ferrite/austenite interfaces. These micro-chemical heterogeneities in the weld metal result in preferential corrosion attacks. Presence of δ -ferrite reduces the critical pitting potential of the austenitic SS by providing susceptible pitting sites (Pehlke and Elliott, 1960; and Pickering, 1988). In the case of type 304 SS weld metal, the most susceptible sites for pitting corrosion were the $\partial \gamma$ interfaces, where extensive segregation of S and P caused difficulty in passivation (Stevens, 1989). In type 316 SS weld metal, the preferential sites for pit initiation were found

to be the core of the austenite cells where depletion of Mo occurred (Simmons, 1996). Sasikala *et al.*, (1999) reported a decrease in the pitting corrosion resistance of types 304 and 316 SS weld metals made by laser penetration welding with a CW Nd:YAG laser in an argon atmosphere, as compared with the values of unwedded specimens. This decrease in pitting corrosion resistance of the welds was attributed to microsegregation in the weld zone of type 316 SS, and to the presence of δ -ferrite in type 304 SS.

GENERAL CREEP PROPERTIES OF 316 LN

Microstructure and Hardness Variation Across Joints

The 316 LN SS base metal had an equiaxed grain structure with grain size of around 75 µm. Heat Affected Zone (HAZ) having coarser grain size was observed in the A-TIG joints; whereas little grain coarsening was observed in the HAZ of MP-TIG joint. Weld metal of both the joints had dendritic and equiaxed grain structures. Weld metal was found to grow epitaxialy from the base metal in austenitic ferritic solidification mode. In the MP-TIG joints, continuation of epitaxial growth of dendrites from the previous weld bead to subsequent weld bead was observed. Adjacent to the base metal, the dendrites oriented more or less perpendicular to the welding direction in the A-TIG joints, whereas in the case of MP-TIG joint dendrites oriented more or less parallel to the welding direction (Figure 4). Coarser dendritic and equiaxed structures were observed in A-TIG joints than in the MP-TIG joint, because of higher heat input during welding of A-TIG joint. In the A-TIG joint, the

equiaxed zone in weld metal was found to grow continuously along the joints, whereas it is discontinuous in the MP-TIG joint (Figure 4). The mechanisms and conditions for the formation of dendritic and equiaxed grains in weld metal have been discussed by Shaikh et al. (2000). The epitaxial and competitive growth of solidified weld metal from the base metal leads to the formation of dendritic structure in weld metal when heat input is relative low and cooling rate is relatively high. The formation of equiaxed grain in weld pool will be favored under the condition of high heat input and low cooling rate leading to the dendrite fragmentation, grain detachment and heterogeneous nucleation. In both the joints, delta-ferrite of vermicular morphology was observed in between the dendrites and equiaxed grains. Orientation of the ä-ferrite of vermicular morphology was observed in between the dendrites and equiaxed grains. Orientation of the δ -ferrite was nearly perpendicular to the welding direction in the A-TIG joints, whereas it was parallel to the welding direction in the MP-TIG joint (Figure 5). More amount of δ -ferrite was observed in the MP-TIG joint (9 FN) than in the A-TIG joint (1.3 FN). The variations of hardness across the joints are shown in Figure 3. Hardness variation across the joints is reflected in the microstructural variations. Coarse grain HAZ in the A-TIG joint displayed lower hardness than that in the MP-TIG joint. Hardness variation across the weld metal was found to depend on microstructure. The equiaxed weld possessed lower hardness than the dendritic weld metal. Relatively more uniform hardness variation was observed in the A-TIG joint than in the MP-TIG joint.



Figure 4: Microstructures of (a) A-TIG and (b) MP-TIG Weld Metal in the as—Welded Condition



Figure 5: Microstructures of (a) A-TIG and (b) MP-TIG Joint Containing Delta Ferrite, in the as—Welded Condition



CREEP DEFORMATION

Typical creep curves for type 316 L(N) SS base metal and the A-TIG and MP-TIG joints at 923 K and 240 and 160 MPa are shown in Figure 6. The creep curves are characterized by small strain on loading, followed by decelerating primary, a distinct steady state and an accelerating tertiary creep regime. Variations of steady state creep rate (ε m) with applied stress (σ) for the base metal and weld joints are shown in Figure 7. The variation obeyed power law relation as $\varepsilon m = A \sigma n$, where *n* is the stress exponent and *A* is constant. There was no appreciable difference in steady state creep rate between the base metal and both weld joints. The tertiary stage of creep deformation started earlier in both the joints than the base metal and was much earlier in the case of MP-TIG joint than in the A-TIG joint (Figure 6). The weld joints were found to deform nonuniformly due to the microstructural in homogeneity across it. Creep strain





accumulation across the different constituents of the weld joints with creep exposure was assessed by conducting interrupted creep test on joint specimens with grid marking inscribed on them prior to creep testing (Sindo, 1989; and Shankar *et al.*, 2000) and measuring the dimensions of the grids. Figure 8 illustrates the creep stain accumulation across the joint at 923 K and 190 MPa. In both the weld joints creep deformation was found to concentrate progressively in the weld and the equiaxed zone of the weld metal was found to deform at higher rate than the dendritic zone. Dendritic and equiaxed weld metal of the joints had finer microstructure than base metal, leading to the higher creep deformation rate in the weld metal. Finer microstructure was observed in the weld metal of MP-TIG joint than that in the A-TIG joint. Higher localized deformation rate in the MP-TIG joint weld metal than in the weld metal of the A-TIG joint was due to presence of finer dendritic and equiaxed structure in the MP-TIG joint (Figure 9). More amount of δ ferrite was observed in the MP-TIG joint than in the A-TIG joint. The δ -ferrite having more open BCC structure could increase the creep





rate of weld metal in the MP-TIG joint than in the A-TIG joint weld metal. The δ - γ interface can also act as effective diffusion path to increase creep deformation rate. Even though the localized preferential deformation in weld metal contributed negligibly to the over all creep deformation of the joint specimen (Figure 7), it has a significant role in the premature failure of the joints under creep condition, as discussed subsequently.

CREEP RUPTURE PROPERTIES AND CREEP DAMAGE

The variations of rupture life with applied stress for the base metal and the weld joints at 923 K are shown in Figure 10. The variation obeyed a power law relation as observed in the relation between creep rate and applied stress (Figure 7). Both the weld joint possessed lower creep rupture life than the base metal. The A-TIG joint had higher rupture life than the MP-TIG joint. The difference in rupture strength between base metal and weld was found to increase with creep exposure in the case of MP-TIG joint; whereas it decreased in the case of A-TIG joint. Delta ferrite in weld metal has significance roles influencing creep deformation and rupture behavior of the joints. As discussed earlier, the presence of δ -ferrite in the weld metal had increased creep deformation rate because of its relatively open BCC structure than FCC base metal and also the interface between matrix and δ -ferrite facilitates faster diffusion. The preferential deformation concentration in the weld metal led to the necking resulting in the premature failure in the weld joints. The delta ferrite changed into brittle intermetallic σ -phase on thermal and creep exposure. Complete transformation of δ -ferrite was observed within about 100 hours in creep exposure in both the weld metals (Figure 11) a in agreement with other investigators (Manning et al., 1980; Valsan et al., 1995; and Laha et al., 2007). However, the transformation of δ -ferrite into σ -phase was faster in the creep condition than the thermal exposure. Nucleation of creep





Figure 12: Creep Cavitations in the Weld Metal of (a) MP-TIG and (b) A-TIG Joint at 160 MPa





cavities occurred at the brittle σ -phase on the boundary of dendritic and equiaxed grain leading to the premature creep failure. Higher amount of δ -ferrite which had ultimately transformed into brittle σ -phase in the MP-TIG joint than in the A-TIG joint led to more cavitation in the MP-TIG joint to reduce the creep rupture life of the MP-TIG joint than in the A-TIG joint more than the A-TIG joint (Figure 12). Greater alignment of the dendrites and a-ferrite along the applied stress direction in the case of A-TIG joint than in the MP-TIG joint might be also the reason for lower cavitation in the A-TIG joint. Careful metallographic investigation revealed that pronounced cavitation occurred at the transition region between the dendritic and equiaxed constituents of the weld metal (Figure 13). Creep strength mismatch between the two constituents of weld metal might have produced stress concentration to induce preferential creep cavitation at the interface between two zones. More creep cavitations was observed in the MP-TIG joint than in the A-TIG joint.

GENERAL FATIGUE PROPERTIES OF 316 LN SS

Further a comparative evaluation of Low Cycle Fatigue behavior of 316 L(N)SS base metal, 316 SS weld metal and 316 L(N)/316 weld joint has also been conducted at 773 and 873 K (Vilpas and Hanninen, 1999; Vasudevan, 2007; and Vasudevan *et al.*, 2008). A detailed examination of the microstructural charges and crack initiation and propagation behavior has been studied with a view to understanding the features which influence the cyclic stress response and fatigue lives of base metal, weld metal and composite specimens. In particulars, the role of delta ferrite on the LCF life has been examined. 316 L(N)/316 weld joints displayed the least fatigue resistance. Detailed investigations have also been performed for assessing the importance of weld displayed the least fatigue resistance. Porosity on the specimen surface has been found to be particularly harmful and caused a life reduction by a factor of seven relative to sound weld metal. Defect combination of porosity and slag inclusions was found to be more deleterious that the case when either the slag inclusions or porosity was present alone. The higher volume fraction of deltaferrite in weld metal was found to be harmful for fatigue life.

RESIDUAL STRESSES MEASUREMENT OF 316 LN SS WELDMENTS

Figure 14 shows the variation in the longitudinal residual stresses obtained across the multipass weld joint of AISI type 316 LN stainless steel made by TIG welding process. Figure 15 shows the variation in the longitudinal residual stresses obtained across the weld centre of AISI type 316 LN stainless steel weld joint made by A-TIG welding process. The maximum tensile stress noticed here is 45 MPa at the weld centre and the maximum compressive stress observed is around 12 MPa present at about 10 mm away from the weld centre line. It is clear from the residual stress profiles that the weld regions exhibited maximum tensile stresses. Tensile residual stress present in the A-TIG weld joint is significantly lower than that of the weld joint made by TIG welding. Being austenitic stainless steel, the residual stresses generated in the weld joints are mainly due to shrinkage. Therefore, It is expected that shrinkage in the







weld metal of A-TIG welded joint should be less and hence the residual stress value compared to that of the multipass weld where the shrinkage should be more due to higher weld metal volume. Differences in the weld metal volume was due to the fact that the A-TIG weld joint was made in single pass without filler metal addition while the multipass TIG weld was made

using 6 passes with filler metal addition in Vgroove joint configuration. The benefit in terms of the reduction in the residual stresses in A-TIG weld joint is very significant. There was no distortion in the joint made A-TIG welding while angular distortion was evident in the joint made by TIG welding. The compressive residual stresses are smaller and their profiles are comparable in both the weld joints.

CONCLUSION

Very vaguely studied on the residual stress measurements of weldments fabricated by laser welding and hybrid welding. Activated flux TIG welding showed considerable reduction on the residual stresses as comparative MP-TIG welding. 316 LN SS was greatly accepted as a candidate material for structural applications of nuclear service due to its better in service properties both room temparature as well as in service properties. Both the joints possessed lower creep rupture strength than the base metal. The A-TIG joint had higher creep rupture strength than the MP-TIG joint. Failure in the weld joints occurred in the weld metal. Progressive localization of creep deformation in the weld metal led to the premature failure of the joints in the weld metal. Higher amount of ä-ferrite in the MP-TIG joint than the A-TIG joint induced higher creep deformation rate and higher accumulation of creep damage in MP-TIG joint to reduce creep rupture strength.

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