Nickel alloys and stainless steels for elevated temperature service: weldability considerations

Reprinted with permission from
Proceedings from Materials Solutions '97 on
Joining and Repair of Gas Turbine Components
ASM International, September 1997, Indianapolis, Indiana

NiDI
NICKEL DEVELOPMENT INSTITUTE
NiDI Reprint Series Nº 14 053

D.J. Tillack

Proceedings from Materials Solutions '97 on Joining and Repair of Gas Turbine Components
15-18 September 1997, Indianapolis, Indiana

Nickel Alloys & Stainless Steels for Elevated Temperature Service: Weldability Considerations

D. J. Tillack
Tillack Metallurgical Consulting, Inc.
Huntington, WV

ABSTRACT:
The gas turbine is one of the most demanding applications for materials, particularly the hot sections of the turbine. The welds that join the various components of the turbine must withstand the same high pressures and temperatures. Complicating this requirement is the fact that gas turbine materials can be metallurgically complex and can be a challenge to weld. The actual welding of these materials is not particularly difficult as long as established guidelines are followed, but microfissuring and strain-age cracking can be a problem. By being aware of the role of residual elements and the effect of stress on the weld area, many of these materials can be welded successfully.

A MODERN GAS TURBINE ENGINE relies heavily on austenitic materials, primarily nickel-base alloys, to withstand the aggressive conditions encountered during operation of the turbine. Not only do these materials need to be strong at elevated temperatures, they also need to be able to withstand the corrosive environment of the combustion gas stream. There is also a need for metallurgical stability and reliability, since these components will be expected to provide trouble-free service for thousands of hours.

While these austenitic materials do possess very attractive properties, the expected design life of the components fabricated from these alloys can only be achieved if the welds used in their construction are of suitable quality. Not only must the design of the components be optimized to resist the stresses imposed during service, but the welds themselves need to be deposited and positioned properly to achieve desired results.

There are two main categories of weld-associated defects. The first group include those that occur during the welding operation, such as undercut, porosity, slag inclusions, lack of fusion and insufficient penetration. These are associated with the welding process but are also influenced by the material being welded and the filler metal material. An example would be the relatively sluggish nature of a nickel alloy weld puddle, which necessitates opening up of the weld joint to allow proper manipulation of the filler metal. The second group of weld-associated defects are metallurgical in nature, such as microfissuring in the HAZ or weld metal. There are times when a weld defect involves both categories, such as the centerline cracking of a weld which can be part welding procedure (improper weld bead contour, for example) and part metallurgical (high levels of deleterious trace elements, for example).

Austenitic Materials Used in Gas Turbines

The modern gas turbine engine has made tremendous strides since Sir Frank Whittle developed the first gas turbine in the late 1930s. Temperatures, pressures and stresses have increased steadily; these advancements have been made possible by improved design, new alloy developments and advances in metal processing/fabrication. An example of the influence of processing innovations is the improvements in alloy 718 properties created by triple melting.

Table 1 lists many of the alloys used in the compressor, combustor and turbine sections of a modern gas turbine engine. Included are designations as to whether the alloys are usually used in the wrought or cast condition and whether the compositions are solid solution or precipitation-hardenable.

Table 1a Compositions of precipitation-hardenable nickel and iron alloys for gas turbines

Alloy Name	UNS No.	Composition - Weight %											
		Ni	Co	Cr	Fe	Mo	Nb	Al	Ti	W	Zr	B	Other
706	N09706	42		16	40		2.9	.2	1.8				
718	N07718	52		19		3	5.2	.5	.9				
X-750	N07750	73		15.5	7		1	.7	2.5				
MA 754	N07754	78		20				.3	.5				.6 Y_2O_3
Re 41	N07041	55	11	19		10		1.5	3.1			.005	
Re 95		61	8	14		3.5	3.5		2.5	3.5	.05	.010	.15C
U-500	N07500	54	18	18		4		2.9	2.9		.05	.006	
U-700		53	18.5	15		5.2		4.3	3.5			.030	
Waspaloy	N07001	58	13.5	19.5		4.3		1.3	3		.06	.006	
214	N07214	75		16	3			4.5			.1*	.01*	
N-155	R30155	20	20	21	30	3	1	.8		2.5			1.5Mn, .15N
80A	N07080	76		19.5				1.4	2.4		.06	.003	
90	N07090	59	16.5	19.5				1.4	2.4		.06	.005	
105		53	20	15		5		4.7	1.2		.10	.005	
115		60	13	14		3.3		4.9	3.7		.04	.160	
Discaloy	K66220	26		14	54	2.7		.1	1.7			.005	.9Mn, .8Si
901	N09901	43		12	36	6		.2	2.8			.015	
903	N19903	38	15		41		3	.7	1.4				
909	N19909	38	13		42		4.7		1.5			.001	.4Si
783	R30783	29	34	3	26		3	5.4	.1				
A-286	S66286	26		15	54	1.3		.2	2			.015	1.3Mn, .5Si
17-4PH	S17400	4		17	75		.3						4Cu

Note: all carbon levels are 0.10% or below, unless otherwise noted * = maximum

Table 1b Compositions of wrought solid solution nickel, cobalt, and iron alloys for gas turbines

Alloy Name	UNS No.	Composition - Weight %										
		Ni	Co	Cr	Fe	Mo	Nb	Al	Ti	W	B	Other
Hast. S	N06635	67		16	1	15		.3				.02La
Hast. X	N06002	47	1.5	22	18	9		2		.6		
230	N06230	57	5*	22		2		.3		14	.005	.4Si, .02La
242		65		8		25						
556	R30556	21	20	22	29	3	.1	.3		2.5		.5Ta, .02La
600	No6600	76		15.5	8							
601	N06601	61		23	14			1.3				
617	N06617	55	12.5	22		9		1				
625	N06625	61		21.5	2	9	3.6	.2	.2			
L-605	R30605	10	51	20						15		1.5Mn
188	R30188	22	40	22						14		.03La
75	N06075	75		20	5				.4			
263	N07263	51	20	20			6		.4	2		.13C
HR-120	N08120	37		25	33		.7				.004	.2N
HR-160	N12160	37	29	28	2							2.75Si
800H	N08810	33		21	46			.4	.4			.8Mn
330	N08330	35		19	45							1.2Si
803	S65803	35		27	37			.3	.4			

Note: all carbon levels are 0.10% or below, unless otherwise noted * = maximum

Table 1c Compositions of cast polycrystalline nickel and cobalt alloys for gas turbines

Alloy	Ni	Co	Cr	Mo	Al	Ti	W	Zr	B	C	Other
IN-100	64	15	10	3	5.5	4.7		.06	.014	.18	1V
IN-713	75		12	4.5	5.9	.6		.10	.010	.12	2Nb
IN-738	62	8.5	16	1.8	3.4	3.4	2.6	.10	.010	.17	1.8Ta, .9Nb
IN-792	61	9	13	2	3.2	4.2	3.9	.10	.020	.21	3.9Ta
IN-939	49	19	22		1.9	3.7	2	.10	.009	.15	1.4Ta, 1Nb
B-1900	65	10	8	6	6	1		.08	.015	.10	4.3Ta
X-40	10	56	25				8		.010	.50	
FSX-414	10	53	29				7		.010	.25	
MAR-M 247	60	10	8	.6	5.5	1	10	.09	.020	.16	3Ta, 1.5Hf
MAR-M 509	10	49	24			.2	7			.60	7.5Ta

Making the Weld

The shape and contour of the weld beads themselves can be instrumental in whether or not the weld will be crack free. Even if the welds are properly made, residual stresses from the welding operation may cause problems, such as distortion or cracking. The effect of welding on the base material is another potential source of problems, particularly with the metallurgically complex alloys that are often specified in gas turbines. Post-weld annealing or stress-relieving practices are another potential source for problems.

The welding process that is used the most in joining gas turbine components is the Gas Tungsten Arc process, or GTAW. This welding process allows greater control of the heat input and provides the ability to use very low heat input, which is often necessary when joining many of the more difficult-to-weld turbine alloys. Also, this process uses bare wire as the filler metal, which avoids slag-related problems associated with flux coated electrodes used in the Shielded Metal Arc process (SMAW). Another reason for the limited use of the SMAW process is the difficulty of transferring reactive elements, such as Al and Ti, across the arc of a flux-shielded electrode. Many of the nickel alloys used in a gas turbine rely on either Al or Ti, or both, to achieve their high strength levels via precipitation-hardening. The SMAW process is used to weld the solid solution nickel and stainless alloys, particularly on industrial gas turbine structural components.

The spray transfer mode of the Gas Metal Arc process, or GMAW, is seldom used for welding components used in a gas turbine because of the high heat input inherent in the process. The cracking problems that often occur when this process is used on the metallurgically complex alloys used in a gas turbine do not justify the attractive deposition rates possible. Limited use of the pulsed mode of GMAW transfer is found. The electron beam, or EBW, process is also used to a limited extent and is an attractive method because of high production rates, ability to make narrow, deep-penetration welds, and a lower heat input per unit length for a given depth of penetration. However, there have been HAZ microfissuring problems with high strength alloys, such as alloy 718, when using the EBW process, particularly with coarse grain material. Table 2 compares the effect of grain size on several welding processes (1). This table does not take into consideration the thickness of the parts being joined nor the effect of stress on cracking tendency. A highly restrained joint, such as with plate material, will be more sensitive to cracking than thin sheet, if both materials have the same large grain size.

Table 2. Effect of grain size on recommended welding processes[a]

Alloy	Grain Size[b]	Gas Metal Arc[c]	Electron Beam	Gas Tungsten Arc	Shielded Metal Arc
600	Fine	X	X	X	X
	Coarse	—	—	X	X
617	Fine	X	X	X	X
	Coarse	—	—	—	X
625	Fine	X	X	X	X
	Coarse	—	—	X	X
706	Fine	—	X	X	X
	Coarse	—	—	—	X
718	Fine	—	X	X	X
	Coarse	—	—	—	X
800	Fine	X	X	X	X
	Coarse	—	X	X	X
AISI Type 316 Steel	Fine	X	X	X	X
	Coarse	—	—	X	X
AISI Type 347 Steel	Fine	X	X	X	X
	Coarse	—	—	X	X

a. Processes marked X are recommended.

b. Fine grain is smaller than ASTM Number 5; coarse grain is ASTM Number 5 or larger.

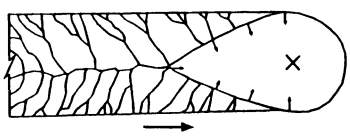
c. Spray transfer.

Susceptibility to weld cracking in austenitic stainless steels and nickel alloys is very dependant on the shape of the weld pool, which in turn is dependant on the welding process and technique. In general, high heat inputs are more conducive to solidification cracking but the shape of the weld pool is also very important. A tear-drop shaped weld pool, which is usually associated with automated welding, is more prone to solidification cracking because of the orientation of the solidifying grains. Figure 1 illustrates the difference between an undesirable tear-drop shaped weld pool and a desirable elliptical shaped weld pool (2). The solidification pattern of the tear-drop shaped weld pool results in centerline segregation that may be high in low melting point elements that will be deleterious to weld hot strength. On the other hand, the elliptical pool shape will have a much greater distribution of these elements, thus minimizing their effect.

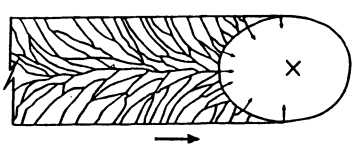
The welding process and welding conditions determine the shape of the weld pool, with higher travel speeds and higher heat inputs favoring the undesirable tear-drop shape weld pool. Accordingly, the GMAW and GTAW welding processes will more likely produce the tear-drop shape weld pool while the SMAW process will usually produce the favored

Table 3 Probable weld pool shape for different welding conditions

Condition	Process	Consumable Diameter, mm	Voltage V	Current A	Travel Speed, mm/min	Heat Input kJ/mm	Probable Pool Shape
1	SMA	3.2	20	100	180	0.7	Elliptical
2	SMA	4.0	21	145	130	1.4	Elliptical
3	GMA (spray)	1.2	31	320	450	1.3	Teardrop
4	GMA (spray)	1.2	27	240	230	1.7	Elliptical
5	GMA (globular dip)	1.2	23	180	125	2.0	Elliptical
6	Submerged-Arc	2.4	28	270	250	1.8	Elliptical
7	Submerged-Arc	2.4	32	370	620	1.1	Teardrop



Tear-shape weld pool: growth of columnar grains result in high solute centerline segregation concentration



Elliptical weld pool: more favorable orientation of grains, with less concentration of centerline solute segregation

Fig. 1 Effect of pool shape on solidification structure (arrows indicate welding direction)

elliptically-shaped weld pool. Table 3 lists several welding processes with varying parameters and their effect on pool shape. Travel speed stands out as a very influential parameter in controlling pool shape - in some cases, a higher heat input produces the more favorable elliptical pool shape when combined with lower travel speed. One factor not taken into consideration in this comparison is the effect of stress: some austenitic alloys are routinely autogenously welded at very high travel speeds that produce tear-drop shape weld pools, such as in automatic welding of alloy 800 (N08800) strip in the production of heater element strip for electric ranges. Assisting in this operation is the addition of compressive stresses provided by the forming rolls used to form the strip into the tubular shape immediately prior to welding. However, most welding operations do not benefit from such compressive stresses. Usually, the strong residual tensile stresses created during welding operations will cause cracking along the centerline condition formed in the tear-drop shaped weld pool.

One study of welding parameters and cracking resistance compared welding current with welding speed for the autogenous GTAW welding of austenitic stainless steels in sheet form. Cracks were found at low speeds and cracking became more pronounced as the speed was increased. At medium speeds they were replaced by center cavities. The tendency for crack and center cavity formation increased with sulfur content, with these defects being infrequent when the sulfur content was below 0.003%. A reduction in the phosphorous level resulted in fewer cracks but did not greatly affect center cavity formation. At lower welding speeds the weld puddle is elliptical and cracking is diminished, but when the welding speed is increased the puddle becomes tear-drop shaped; in these tests on 2mm (0.079 in.) austenitic stainless steel sheets this transformation occurred at a speed of about 200 mm/min (7.9 in./min) (4). Heavier sections would be able to withstand faster weld travel speeds before the weld pool transforms to tear-drop shape because of more favorable heat transfer characteristics.

In addition to the two-dimensional shape of the weld pool during welding, the cross-sectional contour of the weld needs to be convex to provide reinforcement during solidification. Concave weld beads may crack even though they have been welded with the desired elliptically shaped weld pool. For optimum results, the weld pool shape should be elliptical, or rounded, at the trailing edge, and the weld contour should be convex.

Problems Associated with Welding

Most gas turbine components need to be welded sometime during their service life. If welding is not required during initial construction of the turbine (such as single crystal blading) they often need to be repair welded to extend their service life. Following is a discussion of several commonly-encountered weld related problems.

Weld-associated cracking can be categorized by location or time of occurrence. Cracking usually occurs in the weld or HAZ:

-during welding
-during postweld heat treating
-during service.

Cracking during welding usually can be attributed to incorrect welding procedures, such as improper welding parameters or travel speed. When welding dissimilar metals, dilution problems that cause metallurgically crack-prone compositions can be a source of cracking.

Hot Cracking Hot cracking, which is also called solidification cracking, liquation cracking, fissuring, or microfissuring can be controlled in fully austenitic materials by paying attention to welding parameters (such as preheat, heat input, travel speed, etc.), joint design, joint restraint control and welding technique. There are numerous theories regarding the cause of hot cracking. One theory is that cracking usually occurs when liquid films are present between solidifying grains such that the weld cannot accommodate the strains caused by cooling (5). These liquid films are heavily influenced by trace elements in the base material or filler metal. Solidification cracking may be caused by segregation or by the formation of second phase films, which are bulk phases formed when the local impurity content is above the solubility limit. Films are particularly detrimental because they usually have lower inherent ductility than the surrounding region and they replace a single interface (the grain boundary) with two interphase boundaries that have lower interfacial tension or bonding (6). Other theories include: liquid grain boundary films due to the melting of intergranular precipitates, such as carbides; solid state deformation concentrated in the grain boundaries such as grain boundary sliding; constitutional liquation of sulfide inclusions; eutectic melting of the grain boundaries due to elements such as sulfur; and the effect of liquid-solid surface tension on grain boundary wetting (7).

A fusion weld is made by moving a very intense heat source along the material to be welded. In the immediate vicinity of the arc, the base metal is being heated rapidly. Expansion of this material is constrained by the cooler surrounding base material and the resulting compressive stress causes the HAZ to be upset. As the arc passes and the base material and fusion weld area cools and attempts to contract, it is again constrained. This results in residual tensile stresses. The magnitude of stress can be affected to some degree by the choice of welding process and heat input; however the thermal stresses cannot be eliminated.

One study (8) on the effect of microfissures on mechanical properties found that microfissures up to 0.070" in length had little or no effect on the fatigue strength of alloy 600 (N06600) weldments. No effects were noted until the fissures exceeded 5% of the cross-section. Work with alloy 617 (N06617) agrees with these data; little effect was seen with microfissures up to 0.030" in length. This is encouraging since most of the microfissures encountered are less than 0.015" in length. One reason for the minimal effect of microfissures is the excellent ductility and toughness of high nickel alloys. These results suggest that microfissures, although certainly undesirable, are not particularly detrimental unless they occupy a significant amount of the cross-section.

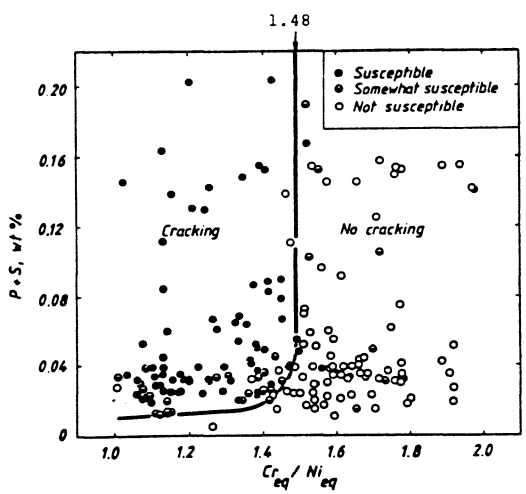
In addition to S and P, other elements known to adversely affect hot cracking resistance in nickel alloys are Si and B. Silicon is sometimes added to high temperature alloys to enhance oxidation and carburization resistance and for improved castability in cast alloys, but it has decidedly deleterious affects on weldability. Nb was first used in Ni-Cr-Fe weld filler metals many years ago to counter the influence of silicon on hot-cracking (9). Crack-free welds were obtained when the Nb/Si ratio for an alloy 600 (N06600) weld composition was 4.5. The critical Nb/Si ratio required to prevent cracking is not the same for all alloys, but generally tends to increase at constant Cr level, as the Ni content of the alloy is decreased. The improved melting techniques available today have enabled metallurgists to lower the levels of numerous deleterious elements that contribute to weld metal and base metal cracking.

techniques and therefore in some respects are not as dangerous as service-related cracks, which are often not discovered until failure occurs.

The Sulfur Paradox For many years, metallurgists were convinced that the lower the sulfur level of austenitic materials the better. This is true when considering the impact of sulfur on hot-cracking, but weldability studies conducted over the relatively recent past have shown that controlled amounts of sulfur can have a positive effect on weld penetration.

One study (16) showed that several minor elements have an effect on the penetration of GTAW welds on S30400 stainless steel. Al additions decreased penetration 14% when the Al was raised from 0.001% to 0.015%; sulfur effects depended on the Al content: with <0.004% Al, there was no sulfur effect, but with 0.02-0.04% Al, there was increased penetration with increasing S; both Si (with an optimum level of about 0.50%) and Mn (with an optimum level of about 1.0%) increased penetration; while P, Cu and Mo had no effect on penetration.

In a study of residual element effects on S30400 and S31600 (17), reduced weldability was attributed to a combination of low (0.001%) S levels and relatively high (0.002%) Ca levels. This study concluded that variations in heat-to-heat penetration could be reduced by optimizing the welding procedures by using short arc lengths, Ar-5%H_2 shielding gas, low arc currents and low welding speeds to maintain a given heat input.



Cr_{eq} = (%Cr) + (%Mo) + 1.5(%Si) + 0.4(%Nb)
Ni_{eq} = (%Ni) + 30(%C) + 0.5(%Mn)

Figure 2. Relationship between P + S content and Cr_{eq}/Ni_{eq} in austenitic stainless steel during welding

Segregation, particularly of harmful impurities such as S and P, has a large effect on how the solidifying weld metal is able to cope with the stresses of solidification. Figure 2 (18) shows one depiction of weld cracking sensitivity of austenitic steels with combined sulfur + phosphorous levels as related to the Cr/Ni equivalents ratio. (The Cr_{eq} and Ni_{eq} are values derived from the composition of the alloy. The Cr_{eq} adds factored values of the ferrite formers Cr, Mo, Si and Nb, while the Ni_{eq} adds factored values of the austenite formers Ni, C, and Mn).

In a study of the effects of sulfur content on the penetration of GTAW welds on thin 304SS sheet, a level of 50 ppm of sulfur was found to be necessary to change the surface tension from negative to positive, thus increasing the penetration of the weld (19). The changes in penetration are controlled by thermocapillary (Marangoni) flow arising from temperature and surface tension gradients over the surface of the weld pool (20). The surface tension is dependent on the concentration of soluble surface active elements, such as oxygen and sulfur, in the fusion zone. When the soluble surface active elements are above a critical level, such as 50ppm for S, the flow of liquid weld metal is swept to the bottom of the molten pool where melt-back produces a deeper weld. This is in contrast to lower sulfur levels, where the molten weld metal produces melt-back at the edges of the weld pool which results in a shallow weld. From a practical standpoint, higher sulfur contents in the area of 50ppm can allow higher welding speeds. However, the negative effect of sulfur on hot and cold formability must also be taken into consideration. There is also a problem with obtaining consistent sulfur levels in commercial heats of material. For example, if a GTAW welding procedure is established on material having an 0.008% sulfur content the welding parameters would not produce the same penetration characteristics on a heat of material containing 0.003% S. This can be a particular problem when welding the root pass of a pipe joint, where complete penetration is critical. Actually, with modern melting technology, it is unlikely that sulfur levels will reach the 0.008% or higher level in the austenitic stainless and nickel alloys unless that level is specifically requested, but a fabricator should be aware of the importance of minor element levels and how these elements interact with one another to influence weld characteristics and structures. As mentioned previously, penetration differences caused by variations in minor elements such as sulfur can be minimized by careful control of welding parameters, including heat input, gas shielding and arc length.

Selenium and tellurium are also able to increase the weld penetration pattern in austenitic alloys.

Additions of 40, 55, and 140ppm of Se were added to a 21Cr-6Ni-9Mn austenitic stainless steel and the resulting electron beam weld penetration increased 81%, 158% and 165% respectively (normal Se levels in the alloy were <20ppm). The sulfur analysis of the three tests were 34, 29, and 30ppm. When Te was added to the same 21-6-9 stainless steel base composition, there also was a large increase in weld penetration: when 43 and 62ppm of Te were added, the weld penetration increased 81% and 156% respectively (normal Te levels in the alloy were <5ppm). Oxygen has also been shown to increase weld penetration(21).

A common thread in this discussion of effects of minor elements on weld penetration is that the elements that increase weld penetration are Group 6A elements - O, S, Se, and Te. It would be interesting to see if the other Group 6A element, polonium, also has the same effect.

While slight increases in sulfur content may assist in weld penetration in an austenitic solid solution stainless steel, such as 316, it may not be advisable to use such an approach to improve weld penetration on a precipitation-hardenable alloy such as alloy 718. Sulfur segregation to grain boundaries in wrought alloy 718 has been cited as the cause of HAZ microcracking (22).

Increased penetration during GTAW welding can also be achieved by adding small amounts (less than 5%) of H to the shielding gas (23). Increases in penetration of over 50% are possible with H additions to argon shielding gas.

Strain-Age Cracking Cracking that occurs after a weld has been completed has been called various names, including relaxation cracking, postweld heat-treatment cracking and strain-age cracking. It is usually associated with precipitation-hardenable alloys and the cracking normally is located in the HAZ. Alloys differ greatly in their ability to resist this type of cracking, with alloy 718 being quite resistant. This single characteristic - of being able to be welded and heat treated without experiencing strain-age cracking - is a primary reason for its phenomenal success as a gas turbine material. (Approximately 35% of the total weight of a typical aero gas turbine consists of alloy 718.) The crack resistance of the alloy 718 HAZ arises from a combination of its relatively sluggish precipitation-hardening response, and lower strength and higher ductility at the start of aging compared to similar alloys, such as Waspaloy. This permits more rapid relaxation of stresses and crack-free accommodation of larger resulting strains. The stresses which produce the heat-treatment cracking result from residual welding stresses plus thermally induced stresses generated by differential thermal expansion in the weldment during heat treatment. Additional stresses also are generated by the dimensional changes caused by precipitation during postweld heat treatment. Restraint during welding imposes stresses which can be exceedingly large (24).

Strain-age cracking is very similar to stress-rupture fracture, and can actually be considered to be a very short-term stress-rupture failure. As carbides and hardeners precipitate, they generate large local microstresses due to coherency, lattice mismatch, or shrinkage. The problem can be viewed in terms of relaxation or stress relief. If the rate of relaxation is rapid in relation to the rate of strength change, stress relief occurs and there is no strain-age cracking. Relaxation may occur either through diffusion (creep), or by plastic flow (shear). Relaxation is retarded, however, by the strengthening produced by aging. Stress relief also occurs if the elastic modulus drops rapidly with temperature because the magnitude of the residual stress is proportional to the elastic modulus. Unfortunately, the elastic moduli of precipitation-hardenable, nickel-base alloys do not decrease rapidly with temperature. The rate of precipitation also plays an important part in strain-age susceptibility. Initially, during welding, the more-rapidly hardening alloys age in the HAZs and so there is a concurrent increase in the yield strength and hence in the maximum possible residual stress. During postweld stress-relieving operations, additional aging takes place and the resultant strengthening of the grains inhibits relaxation, or stress relief (25). To minimize the possibility of strain-age cracking in those alloys that precipitation-harden rapidly, such as Waspaloy, the material should be in the annealed condition prior to welding or repair welding and again annealed after welding has been completed. If these materials are not annealed after welding they will usually crack during the precipitation-hardening thermal cycle. Alloys that undergo a relatively sluggish precipitation response, such as alloy 718, are quite resistant to strain-age cracking, as mentioned previously, and can often be welded/repair welded and then directly aged without receiving an intermediate anneal.

This paper has concentrated on wrought material, because that's where most of the welding is done. Castings, however, comprise a large percentage of material usage in a gas turbine (approx 25-40% by weight in aero gas turbines). Most casting applications have little fabrication welding done on them but they are often repair welded. Castings, by their very nature, have poor weldability because of their structure and composition (relatively large levels of hardening elements). Figure 3 depicts the effects of Al and Ti on the weldability of several wrought and cast alloys (26).

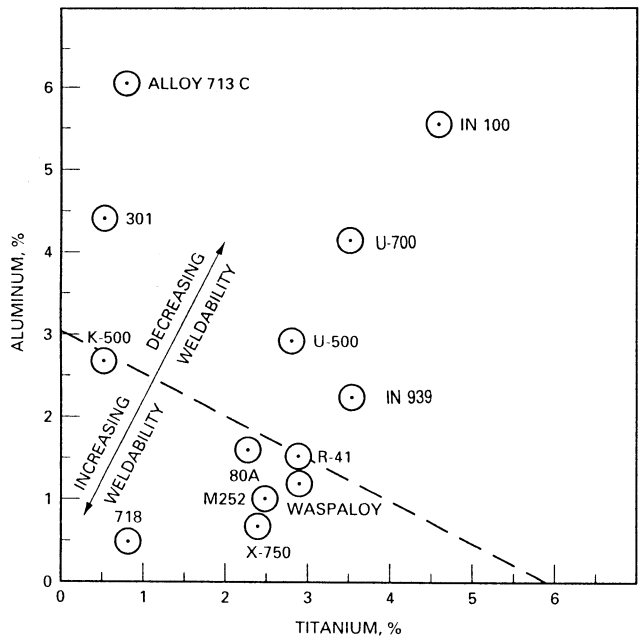


Fig. 3 Relationship between estimated weldability of precipitation-hardenable nickel alloys and their aluminum and titanium content, in wt %

Relaxation Cracking While strain-age cracking is usually associated with precipitation-hardenable alloys it also occurs in solid solution alloys. In such cases a more proper terminology would be 'strain-relaxation cracking', since an aging reaction plays a minimal role but excessive strain is the primary cause of the cracking. Depending on the level of strain, cracking can occur very quickly (in a matter of minutes or hours) or it may not occur for months or years. A good example of such cracking in a solid solution alloy is what has occurred with alloy 800H/800HT (N08810/N08811) on numerous occasions during exposure to the 540-650°C (1000-1200°F) range.

Residual stresses are often the ultimate cause of cracking in components exposed to intermediate temperatures. Too often residual stresses are either ignored or dismissed, particularly if the components are being used in elevated temperature service, where it is assumed that they will be eliminated during start-up of the unit. This may or may not be true, depending on the service temperature, the strength of the material and the level of the residual stresses.

Residual stresses develop during welding due to the uneven distribution of nonelastic strains, and three sources of residual stresses can be identified. The first source is the difference in shrinkage of differently heated and cooled areas of the weld joint. The weld metal, which is subjected to the highest temperature, will contract more than all other areas in the joint during cooling. This contraction is hindered by the cooler portions of the joint, resulting in high stresses in the weld metal. The second source is in the thickness direction of the weld, with the surface layers cooling more rapidly than the interior. This leads to thermal stresses which can cause nonuniform plastic deformations and thus to residual stresses, compressive at the surface and tensile ones in the interior. This is a factor in thick sections but not in thinner sheet welds. The third source is from phase transformations that could occur during cooling, which are often accompanied by an increase in volume of the material being transformed. This expansion is restrained by the cooler material and thus causes residual stresses (27).

Even if the welding operation is entirely satisfactory and there are no hot cracks, porosity, undercuts, etc., the expected design life of the welded structure may not be realized if excessive stress is induced during the welding and fabrication operation. This is particularly true if the structure will be operating in an intermediate temperature region where little stress-relief will occur during service. In these situations, a stress-relief anneal will provide much needed relief of life-shortening stresses (28).

A specific example is in a styrene production component where the operating temperatures are in the 600-650°C (1100-1200°F) area and alloy 800H (N08810) is commonly used because of its superior strength compared to 304SS (S30400). One of the problems with alloy 800H, however, is that its high creep strength in this intermediate temperature range prevents it from relaxing during exposure to this temperature range. There have been weld-related failures that have been attributed to the inability of the un-stress relieved alloy to relieve itself of the stresses, induced during fabrication and welding, during exposure in this 600-650°C (1000-1200°F) range. The combination of fabrication stresses, often aggravated by forcing components into alignment, and welding stresses are often higher than the creep strength of the alloy at service temperature. With weaker alloys, such as 304SS, these fabrication/welding stresses would be relieved when first exposed to these operating temperatures. But since alloy 800H is not relieved of stresses at these temperatures, the alloy relieves itself by cracking. Documented cases of problems involving this alloy in these situations have existed for decades (29), but there also have been millions of pounds used successfully in these applications.

Avoidance of this problem lies in controlling and relieving stresses prior to putting the components into

service. Parts should not be forced into alignment or otherwise mis-handled. After welding, stress relieving or annealing should be performed. Stress relieving can be done at 900°C (1650°F) for an hour, with rapid heat-up and cool-down. There should be a minimum of time spent in the carbide precipitation range of 540-815°C (1000-1500°F) to avoid the precipitation of fine carbides (30). Carbides are inevitable in alloys containing chromium and carbon, but their morphology is critical. Carbide films are to be avoided because of their propensity for crack initiation and propagation. The optimum morphology for carbides in material exposed to intermediate temperatures is large and discrete.

Annealing to remove fabrication stresses in alloys used in the creep range can cause degradation in creep and rupture strength unless proper evaluation of the microstructure is performed to assure suitability for elevated temperature service. Solid solution alloys used in high temperature applications are often given a high temperature solution anneal at the mill, which provides a large grain size conducive to optimum creep/rupture strength. If the alloy is given a fairly large amount of cold straining during fabrication, an anneal can cause recrystallization and the resulting fine grain size can cause a large reduction in rupture strength. For example, 10% cold work, in the form of bending, will result in a recrystallization temperature of about 940°C (1750°F) for alloy 800H (N08810) (31). If material in this condition is annealed at 980°C (1800°F), recrystallization will occur in the cold worked area without grain growth, resulting in greatly reduced rupture strength. In this scenario, the material would need to be solution annealed in the 1150°C (2100°F) area to assure sufficient grain growth in order to achieve the expected design strength. However, on-site annealing is difficult and often impossible. Therefore, fabrication sequences should be properly planned to allow for adequate postweld heat treatments.

Conclusions

The austenitic stainless steels and nickel alloys offer advantages of strength and environmental resistance in high temperature applications. To achieve design life, weldability of the material must be understood and optimized. The design of weldments needs to take into consideration the sluggish nature of the molten austenitic weld metal and the lower weld penetration compared to ferritic materials. The welding operation must be controlled to minimize weld cracking: an elliptical, or rounded, weld puddle trailing edge will minimize centerline segregation; and a convex bead profile will minimize centerline cracking that can occur with a concave profile. Weld penetration can be influenced by trace element variations. Sulfur, in particular, can dramatically increase weld penetration in austenitic alloys if present in quantities exceeding 50ppm. Precipitation-hardenable alloys are a particular challenge because of the intermediate temperature transformations that occur during welding and heat treating. Microfissuring, which usually occurs in the HAZ, is promoted by large grain sizes and high heat input during welding. Strain-age cracking usually occurs during heat treatment and is influenced by alloy chemistry and precipitation reaction time. In all weld or HAZ cracking situations, stress plays a dominant role. Whether the material is solid solution or precipitation-hardenable, stress cannot be eliminated during welding. However, it must be minimized if crack-free, long-life weldments are to be realized.

References

1 AWS Handbook, 8th Ed., Vol. 3, pg. 232, table 4.8
2 Davies, G. J. and Garland, J. G., Int. Met. Rev., 20 (197), 83-106, 1976
3 Gooch, T. G., Conf. Proceed., Weldability of Materials, ASM, pg. 34, 1990
4 Kujanpää, V. P., Conf. Proceed., Advances in Welding Science & Technology, ASM, pg. 133, 1986
5 Gooch, T. G., Conf. Proceed., Weldability of Materials, ASM, pg. 31, 1990
6 White, C. L. , Conf. Proceed., Weldability of Materials, ASM, pg. 65, 1990
7 "Microfissuring in Ni-Based Alloy Welds", Marshall Space Flight Center, MFS-25815, 1983
8 Yeniscavich, W., Welding Journal, 3/66
9 Kihlgren, T., & Lacy, C., Welding Journal, 11/46, pg 769s
10 Bologna, D., Metals Eng. Quarterly, ASM, 11/69
11 Owczarski, W. A., Duvall, D. S., & Sullivan, C. P., Welding Journal, pg. 152s, 4/66
12 Duval, D. S., & Owczarski, W. A., Welding Journal, pg.431s), 9/67
13 Lucas, & Jackson, Welding Journal, pg. 53s, 2/70
14 Valdez, P. J., Symposium Proceed., Effect of Minor Elements on the Weldabiilty of High-Ni Alloys, Welding Research Council, 1969
15 Thompson, E. G., Welding Journal, 2/69
16 Pollard, B., Welding Journal, pg. 202s, 1988
17 Lambert, J. A., Welding Journal, pg. 41, 5/91
18 Matsuda, F., Conf. Proceed., First US-Japan Symposium on Advances in Welding Metallurgy, pg.31, 1990
19 Scheller, Brooks and Mills, Welding Journal, pg. 69s, 1/95
20 Heiple, C. R., and Roper, J. R., Welding Journal, 4/82
21 Heiple, C. R., and Roper, J. R., Conf. Proceed.,Trends in Welding Research, pg. 489, 1981

22 Shira, C. S., Morrison, T. J., and Weisenberg, L. A., Symposium Proceed., Effects of Minor Elements on the Weldability of High-Nickel Alloys, Welding Research Council, pg. 47-68, 1969
23 Onsoien, Peters, Olson and Liu, Welding Journal, pg. 10s, 1/95
24 Duvall, D. S., & Owczarski, W. A., Welding Journal, pg. 10s, 1/69
25 Prager, M. and Shira, C. S., Welding Research Council Bulletin 128, 2/68
26 Prager, M. and Shira, C. S., Welding Research Council Bulletin 128, 2/68
27 Masubuchi, K. and Agapakis, J., Conf. Proceed., Trends in Welding Research in the US, pg. 209, 11/81
28 Tillack, D. J. , Conf. Proceed., 2nd International Conf. on Heat-Resistant Materials, ASM, pg. 177, 9/95
29 Kohut, G. B., Journal of Engineering Materials & Technology, pg. 316, 10/75
30 Private communication, C. S. Tassen, Inco Alloys International, Huntington, WV
31 Avery, R. E. and Valentine, H. L., Chemical Engineering Progress, pg. 89, 1/68

service. Parts should not be forced into alignment or otherwise mis-handled. After welding, stress relieving or annealing should be performed. Stress relieving can be done at 900°C (1650°F) for an hour, with rapid heat-up and cool-down. There should be a minimum of time spent in the carbide precipitation range of 540-815°C (1000-1500°F) to avoid the precipitation of fine carbides (30). Carbides are inevitable in alloys containing chromium and carbon, but their morphology is critical. Carbide films are to be avoided because of their propensity for crack initiation and propagation. The optimum morphology for carbides in material exposed to intermediate temperatures is large and discrete.

Annealing to remove fabrication stresses in alloys used in the creep range can cause degradation in creep and rupture strength unless proper evaluation of the microstructure is performed to assure suitability for elevated temperature service. Solid solution alloys used in high temperature applications are often given a high temperature solution anneal at the mill, which provides a large grain size conducive to optimum creep/rupture strength. If the alloy is given a fairly large amount of cold straining during fabrication, an anneal can cause recrystallization and the resulting fine grain size can cause a large reduction in rupture strength. For example, 10% cold work, in the form of bending, will result in a recrystallization temperature of about 940°C (1750°F) for alloy 800H (N08810) (31). If material in this condition is annealed at 980°C (1800°F), recrystallization will occur in the cold worked area without grain growth, resulting in greatly reduced rupture strength. In this scenario, the material would need to be solution annealed in the 1150°C (2100°F) area to assure sufficient grain growth in order to achieve the expected design strength. However, on-site annealing is difficult and often impossible. Therefore, fabrication sequences should be properly planned to allow for adequate postweld heat treatments.

Conclusions

The austenitic stainless steels and nickel alloys offer advantages of strength and environmental resistance in high temperature applications. To achieve design life, weldability of the material must be understood and optimized. The design of weldments needs to take into consideration the sluggish nature of the molten austenitic weld metal and the lower weld penetration compared to ferritic materials. The welding operation must be controlled to minimize weld cracking: an elliptical, or rounded, weld puddle trailing edge will minimize centerline segregation; and a convex bead profile will minimize centerline cracking that can occur with a concave profile. Weld penetration can be influenced by trace element variations. Sulfur, in particular, can dramatically increase weld penetration in austenitic alloys if present in quantities exceeding 50ppm. Precipitation-hardenable alloys are a particular challenge because of the intermediate temperature transformations that occur during welding and heat treating. Microfissuring, which usually occurs in the HAZ, is promoted by large grain sizes and high heat input during welding. Strain-age cracking usually occurs during heat treatment and is influenced by alloy chemistry and precipitation reaction time. In all weld or HAZ cracking situations, stress plays a dominant role. Whether the material is solid solution or precipitation-hardenable, stress cannot be eliminated during welding. However, it must be minimized if crack-free, long-life weldments are to be realized.

References

1 AWS Handbook, 8th Ed., Vol. 3, pg. 232, table 4.8
2 Davies, G. J. and Garland, J. G., Int. Met. Rev., 20 (197), 83-106, 1976
3 Gooch, T. G., Conf. Proceed., Weldability of Materials, ASM, pg. 34, 1990
4 Kujanpää, V. P., Conf. Proceed., Advances in Welding Science & Technology, ASM, pg. 133, 1986
5 Gooch, T. G., Conf. Proceed., Weldability of Materials, ASM, pg. 31, 1990
6 White, C. L. , Conf. Proceed., Weldability of Materials, ASM, pg. 65, 1990
7 "Microfissuring in Ni-Based Alloy Welds", Marshall Space Flight Center, MFS-25815, 1983
8 Yeniscavich, W., Welding Journal, 3/66
9 Kihlgren, T., & Lacy, C., Welding Journal, 11/46, pg 769s
10 Bologna, D., Metals Eng. Quarterly, ASM, 11/69
11 Owczarski, W. A., Duvall, D. S., & Sullivan, C. P., Welding Journal, pg. 152s, 4/66
12 Duval, D. S., & Owczarski, W. A., Welding Journal, pg.431s), 9/67
13 Lucas, & Jackson, Welding Journal, pg. 53s, 2/70
14 Valdez, P. J., Symposium Proceed., Effect of Minor Elements on the Weldabiilty of High-Ni Alloys, Welding Research Council, 1969
15 Thompson, E. G., Welding Journal, 2/69
16 Pollard, B., Welding Journal, pg. 202s, 1988
17 Lambert, J. A., Welding Journal, pg. 41, 5/91
18 Matsuda, F., Conf. Proceed., First US-Japan Symposium on Advances in Welding Metallurgy, pg.31, 1990
19 Scheller, Brooks and Mills, Welding Journal, pg. 69s, 1/95
20 Heiple, C. R., and Roper, J. R., Welding Journal, 4/82
21 Heiple, C. R., and Roper, J. R., Conf. Proceed.,Trends in Welding Research, pg. 489, 1981

22 Shira, C. S., Morrison, T. J., and Weisenberg, L. A., Symposium Proceed., Effects of Minor Elements on the Weldability of High-Nickel Alloys, Welding Research Council, pg. 47-68, 1969
23 Onsoien, Peters, Olson and Liu, Welding Journal, pg. 10s, 1/95
24 Duvall, D. S., & Owczarski, W. A., Welding Journal, pg. 10s, 1/69
25 Prager, M. and Shira, C. S., Welding Research Council Bulletin 128, 2/68
26 Prager, M. and Shira, C. S., Welding Research Council Bulletin 128, 2/68
27 Masubuchi, K. and Agapakis, J., Conf. Proceed., Trends in Welding Research in the US, pg. 209, 11/81
28 Tillack, D. J. , Conf. Proceed., 2nd International Conf. on Heat-Resistant Materials, ASM, pg. 177, 9/95
29 Kohut, G. B., Journal of Engineering Materials & Technology, pg. 316, 10/75
30 Private communication, C. S. Tassen, Inco Alloys International, Huntington, WV
31 Avery, R. E. and Valentine, H. L., Chemical Engineering Progress, pg. 89, 1/68

www.nidi.org

The Nickel Development Institute is an international nonprofit organization serving the needs of people interested in the application of nickel and nickel-containing materials.

Members of NiDI

Centaur Mining and Exploration Limited
Codemin S.A.
Falconbridge Limited
Inco Limited
Inco TNC Ltd.
Nippon Yakin Kogyo Co., Ltd.
Outokumpu Oy
P.T. International Nickel Indonesia
QNI Pty. Ltd.
Sherritt International Corporation
Sumitomo Metal Mining Co., Ltd.
Titan Resources N.L.
WMC Limited

North America

Nickel Development Institute
214 King Street West - Suite 510
Toronto, Ontario
Canada M5H 3S6
Telephone 1 416 591 7999
Fax 1 416 591 7987
E-mail nidi_toronto@nidi.org

Europe

Nickel Development Institute
42 Weymouth Street
London, England W1N 3LQ
Telephone 44 20 7493 7999
Fax 44 20 7487 4964
E-mail nidi_london_uk@nidi.org

Nickel Development Institute
European Technical Information Centre
The Holloway, Alvechurch
Birmingham, England B48 7QB
Telephone 44 1527 584777
Fax 44 1527 585562
E-mail nidi_birmingham_uk@nidi.org

Japan

Nickel Development Institute
11-3, 5-chome, Shimbashi
Minato-ku, Tokyo, Japan
Telephone 81 3 3436 7953
Fax 81 3 3436 7734
E-mail nidi_japan@nidi.org

Central & South America

Nickel Development Institute
c/o Instituto de Metais Não Ferrosos
Rua Coronel Paulino Carlos, 194
04006-040 São Paulo-SP, Brazil
Telephone 55 11 3887 2033
Fax 55 11 3885 8124

India

Nickel Development Institute
55A Uday Park (First Floor)
Khel Gaon Marg
New Delhi 110 049
India
Telephone 91 11 686 5631
Fax 91 11 686 3376
E-mail nidi_india@nidi.org

Australasia

Nickel Development Institute
150 Drummond Street, Suite 3
Carlton, Victoria 3053
Australia
Telephone 61 3 9650 9547
Fax 61 3 9650 9548
E-mail nidi_australia@nidi.org

South Korea

Nickel Development Institute
Olympia Building, Room 811
196-7 Jamsilbon-Dong, Songpa-Ku
Seoul 138 229, South Korea
Telephone 82 2 419 6465
Fax 82 2 419 2188
E-mail nidi_korea@nidi.org

China

Nickel Development Institute
Room 677, Poly Plaza Office Building
14 Dongzhimen Nandajie
Beijing, China 100027
Telephone 86 10 6501 1188
6501 3677
Fax 86 10 6501 0081
E-mail nidi_china@nidi.org