

Nickel Base Alloys and Associated Problems on Welding

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Abstract— To study the various types of nickel based alloys and effect of alloying element on properties of alloy, welding problems present in nickel based alloys. The Super alloys are primarily used in gas turbines, coal conversion plants and chemical processing industries and for other specialized applications requiring heat and corrosion resistance. The γ' phase in the Nickel base super alloys is reasonable for the elevated temperature strength of the material and its incredible resistance to creep deformation and grain boundary dislocation. The effect of alloying elements on the Ni base super alloys were studied. The possible welding problems of Ni based super alloys were studied. The Nickel base alloys are mostly prone to the strain age cracking, Solidification cracking, HAZ liquification cracking, Fusion Zone Solidification Cracking. The presence of the carbides MC, M₂C₆, M₆C, M₇C₃ and the laves phases are leads to hot cracking.

Key words: Nickel Based Super Alloys, SAC- Strain-Age cracking, Process Parameters, Hot Cracking Susceptibility, Effect of Alloy Elements

I. INTRODUCTION

Nickel base alloys are one of the most important classes of engineering materials, it can be used to a higher fraction of melting temperature and are therefore more favourable than cobalt-based and iron-nickel-based superalloys at service temperatures close to the melting temperature of the materials. they can also be used in a wide range of environments and applications. These alloys are selected for both aqueous and high temperature corrosion resistance, high strength at both ambient and high temperatures, ductility and toughness at low temperatures, specific electrical properties, and many other physical property - dependent applications. The Ni - base alloy welding consumables offer some properties in the as- welded condition that no other family of welding products can offer, such as the ability to be diluted by a number of diverse alloying elements while maintaining strength and ductility from cryogenic temperatures to temperatures approaching the solidus. For example, the family of Ni-Cr-Mo welding products is used to weld 9% Ni steel to produce excellent as-welded strength and impact toughness at liquid nitrogen temperatures. Nickel and nickel - iron alloys are used to weld cast irons because they can be diluted by iron and carbon while remaining ductile and providing good machining characteristics. They are also widely used in the power generation industry for dissimilar welds between carbon steels and austenitic stainless steels in order to provide a transition in coefficient of thermal expansion for elevated temperature service. Relative to steels, Ni-base alloys can be used both at cryogenic temperatures and temperatures approaching 1200°C because the matrix of the solid solution alloys remains austenitic from solidification to absolute zero. With appropriate alloying additions, these alloys provide useful corrosion resistance and have

applications in a wide range of industries, including Power Generation, Petrochemical, Chemical Processing, Aerospace, and Pollution Control.

II. NI-BASE ALLOY CLASSIFICATION

There is no systematic classification system for Ni-base alloys as there is for steels and aluminium alloys. For this reason, most Ni - base alloys are known by their trade names or by the alloy number that was originally assigned by the alloy producer. For example, INCONEL alloy 600 ii and HASTELLOY alloy C - 22 iii are also referred to as Alloy 600 and Alloy C- 22. Ni- base alloys are generally classified by composition, as shown in Figure 1.1

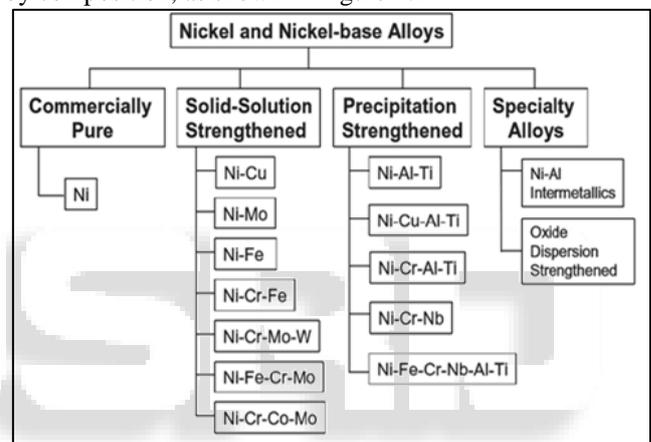


Fig. 1.1: Ni-Base Alloy Classification

A. Commercially Pure Nickel Alloys

Commercially pure nickel alloys are those that contain primarily nickel (> 99 wt.%). There is an entire family of commercially pure nickel alloys anchored by Alloys 200 and 201. These materials have low strength, hardness and are used principally for their corrosion resistance in caustic environments. Alloy 201 has a limit of 0.02 wt.% carbon so that it can be used above 315°C without the danger of being graphitized. Because carbon is relatively mobile in the nickel matrix above 315°C, carbon additions beyond the solubility limit (~ 0.02 wt.%) will result in precipitation of graphite particles that render the material brittle and weak. There are a number of additional commercially pure nickel alloys that exist for electrical or magneto strictive limited applications. These alloys have good weldability but are susceptible to porosity when welded. If these alloys are kept clean before and during welding, they will exhibit good resistance to cracking, but gas shielding or fluxing must be sufficient to prevent the formation of porosity. Both Ti and Al combine with oxygen to form oxides and nitrogen to form nitrides, thereby controlling porosity in the weld deposits.

B. Solid - Solution Strengthened Alloys

Nickel and copper are isomorphous (complete solid solubility), which allows production of single phase alloys

over the entire composition range. This family of materials generally exhibits good corrosion resistance to seawater and other general corrosion environments. The Ni- Cu alloys are usually quite weldable but may be susceptible to porosity if proper shielding or well- deoxidized consumables are not used. Other solid-solution strengthened Ni base alloys may contain only iron and most of these alloys are used for their particular coefficient of expansion or electrical properties. The Ni -36 wt.% Fe alloy commonly known as INVAR®iv exhibits the lowest coefficient of expansion of any of the Ni-base alloys and expands and contracts at a rate of less than 1.0×10^{-6} in/in/F. when heated and cooled over a range of several hundred degrees up to about 300°F. The Ni-Fe alloys have reasonable weldability, but the development of consumables with good solidification cracking resistance with near-matching expansion properties has presented a challenge to consumable manufacturers. The Ni-Fe alloys and their consumables may also be susceptible to ductility-dip cracking. Other solid-solution alloys contain a variety of substitutional elements including chromium, molybdenum, and tungsten. Each element imparts particular characteristics and has the ability to alter the welding response of each alloy. Maximum ultimate tensile strength values of the solid-solution strengthened Ni-base alloys approach 830 MPa with yield strengths in the range of 345 to 480 MPa. These alloys are used in a broad range of applications requiring good corrosion resistance. If higher strength levels are required, it is necessary to select precipitation- strengthened alloys.

C. Precipitation-Strengthened Alloys

The precipitation - strengthened Ni - base alloys contain additions of titanium, aluminium and/or niobium that form a strengthening precipitate with nickel after an appropriate heat treatment. Under most conditions, these precipitates are coherent with the austenite matrix, and thus strain the matrix such that the strength of the alloy increases substantially. The most common of these precipitates are called gamma prime [γ' - Ni₃Al, Ni₃Ti, and Ni₃(Ti, Al)] and gamma double prime (γ'' - Ni₃Nb). By optimizing alloying additions and heat treatment, these alloys can be strengthened to reach ultimate tensile strength values exceeding 1380 MPa with 0.2% offset yield strengths over 1035 MPa. The first precipitation-hardened Ni- Cr alloy (X- 750) is strengthened by gamma prime and exhibits the combination of good oxidation resistance and high temperature strength to near the gamma prime solvus temperature. Unfortunately, it is subject to post weld strain age cracking (SAC) when welded and direct- aged without an intervening annealing treatment. In an effort to improve weldability and avoid SAC, a second generation of precipitation- hardening Ni- Cr alloys were developed that are strengthened by gamma double prime. The most popular of these alloys is Alloy 718. Because the gamma double prime precipitate forms more slowly than gamma prime, Alloy 718 is generally immune to SAC during post weld heat treatment. One of the major applications for Alloy 718 is for aerospace gas turbine shafting and pressure containment. When melted properly to produce low levels of impurities, this alloy provides tremendous design opportunities with excellent fatigue life at service temperatures up to 760°C when properly designed. The precipitation- strengthened alloys are often referred to as “superalloys” based on their

retention of unusually high strength and corrosion resistance at elevated temperatures. The term has been loosely applied to many other high strength complex alloys, but generally the term “superalloys” is used to describe Ni- base alloys with their superior strength properties provided by the gamma prime and gamma double prime phases. The use of “superalloys” for rotating gas turbine blades or “buckets” began with alloys such as IN713C. This alloy was similar to Alloy X-750 but was only produced as a casting and contained such high additions of aluminium and titanium that it age-hardened upon cooling from casting temperatures. Structural repairs were limited due to its extreme susceptibility to SAC, but blade tip build - up was performed without cracking. After decades of technological development, other members of the “superalloy” family were added to include the very high strength, corrosion- resistant single crystal turbine blade alloys. Welded blade tip builds up due to erosion during service is possible if the welding process is well controlled and the residual stresses are kept low. The primary challenges to weld repair of these materials include the avoidance of stray grains in the melt pool and elimination of cracking.

III. GENERAL INFLUENCE OF ALLOYING ADDITIONS

The general effect of various alloying elements on the phase stability in Ni base alloys is summarized in Table 1.1. The elements Co, Cr, Fe, Mo, and Ta are generally used as solid solution strengtheners. Elements that exhibit similar atomic radii, electronic structure, and crystal structure relative to Ni are most likely to remain in solid solution. On the other hand, the ability of a dissolved element to increase strength by solid-solution hardening can be assessed by its atomic size difference compared to that of Ni. Table 1.2 summarizes approximate atomic diameters and solubility data for various elements in Ni at 1000° C. These data show that Al, Ti, Mn, Nb, Mo, and W provide the best combination of atomic radii mismatch and appreciable solubility needed for solid-solution strengthening. The elements Cr, Mo, Mn and W are used in many single phase commercial alloys for solid-solution strengthening. Additions of Mo and W are also useful for increasing creep strength due to their low diffusivity in Ni. Although Ti and Al can be effective solid - solution strengtheners, they typically improve strength by precipitation of the γ' - Ni₃(Ti, Al) phase in Ni base superalloys.

Summary of the general effect of various alloying elements on phase stability in Ni-base alloys. (Courtesy ASM International)

Effect	Element
Solid Solution Strengtheners	Co, Cr, Fe, Mo, W, Ta
γ' - Ni ₃ (Al, Ti) Former	Al, Ti
Solid Solution Strengthening of γ'	Cr, Mo, Ti, Si, Nb
γ'' - Ni ₃ Nb Former	Nb
Carbide Formers:	
M ₂ C and M ₃ (C, N)	W, Ta, Ti, Mo, Nb
M ₇ C ₃	Cr
M ₂₃ C ₆	Cr, Mo, W
M ₆ C	Mo, W

TCP Phase (σ , P, μ , Laves)	Ti, V, Zr, Nb, Ta, Al, Si
Surface Oxide (Cr_2O_3 / Al_2O_3) Former	Cr, Al

Table 1.1: General effect of various alloying elements on phase stability

Niobium is also an effective strengthener and is used for both solid - solution strengthening and for its ability to strengthen by promoting precipitation of the γ'' -Ni₃Nb phase. γ' - Ni₃(Ti, Al) is an ordered L12 structure phase that has very good crystallographic matching with the Ni- f cc matrix. The Al atoms reside on the cube corners while Ni resides on the face center. In commercial alloys, elements such as Cr, Co and Fe can substitute for Ni, and Nb can substitute for Ti and Al in the Ni₃(Al, Ti) phase. The mismatch between the γ matrix and γ' precipitate is generally less than one percent. The coarsening rate of precipitates is directly proportional to the surface energy of the matrix/precipitate interface. The good crystallographic matching between the γ matrix and γ' precipitate in superalloys leads to a very low surface energy that, in turn, leads to very low coarsening rates of the precipitates. The γ' phase shows the remarkable effect of increasing yield strength with increasing temperature up to ~ 800°C.

summary of approximate atomic diameters and solubility data for various elements in Ni at 1000°C. (Courtesy ASM International)

Solute	Approximate atomic size difference compared to nickel, %	Approximate solubility in nickel at 1000 ° C, wt.%
C	+43	0.2
Al	- 15	7
Si	+6	8
Ti	- 17	10
V	- 6	20
Cr	- 0.3	40
Mn	+10	20
Fe	+0.3	100
Co	- 0.2	100
Cu	- 3	100
Nb	- 15	6
Mo	- 9	34
Ta	- 15	14
W	- 10	38

Table 1.2: Atomic diameters and solubility data for various elements in Ni

This effect is thought to occur due to ordering effects and the relatively low mobility of super lattice dislocations that occurs with increasing temperature. Additional increases in the strength of γ' can be accomplished by solid-solution strengthening. In this case, the additional strengthening is obtained by dissolution of elements such as Cr and Nb directly within the γ' phase, with Nb being the most effective. The γ'' - Ni₃Nb phase can form in alloys with sufficient Nb additions. the γ'' phase is metastable and will generally be replaced with the orthorhombic δ phase with the same Ni₃Nb stoichiometric composition with long exposure times at elevated temperature. The δ phase is generally undesirable because it is incoherent with the Ni matrix and therefore not an effective strengthener. In addition, it can lead to

embrittlement associated with a loss in ductility. The effect of δ phase formation on the repair weldability of γ'' strengthened alloys Many Ni-base alloys with appreciable carbon levels can form various types of carbides, depending on the alloy composition, processing route, and service history. The MC- type carbide exhibits an FCC crystal structure and typically forms at the end of solidification by eutectic- type reactions with the γ matrix. The M(CN) carbonitrides are similar, except that appreciable levels of C are replaced by N. These eutectic- type reactions and the concomitant carbides and carbonitrides are promoted by the strong tendency of C, N, and some metallic elements (most notably Ti and Nb) to segregate to the liquid during solidification. As a result, MC carbides are typically distributed along the inter dendritic and solidification grain boundary regions. The MC carbides can often be replaced by M₂₃C₆ and M₆C carbides during thermal processing and high temperature service. The M₂₃C₆ carbides are generally Cr rich and form in the 760 – 980°C range with a complex cubic crystal structure. These carbides tend to form on grain boundaries and, when present as discrete particles, can improve creep strength by restricting grain boundary sliding. The M₆C carbides form in the range from 815 to 980 °C and also exhibit a complex cubic crystal structure. These carbides tend to form when the Mo and W content is greater than 6 - 8 atomic percent. The Topologically Closed Packed (TCP) phases such as σ , P, and μ can form in highly alloyed materials during thermal processing and during long term service. Some of these phases can also form at the end of solidification. For example, σ can form due to segregation of Mo, while Laves can form due to segregation of Nb. These phases exhibit complex crystal structures and, when formed in the solid state, their closed packed planes are parallel to the {111} planes of the austenite matrix. These phases are generally to be avoided because they typically produce a loss in strength and can promote premature failure. Loss in strength is promoted by depletion of solid-solution strengthening elements from the matrix, such as Cr, Mo, and W. Depletion of Cr from the matrix can also reduce corrosion resistance. Early failure can occur due to the high hardness and plate like morphology of these phases that cause premature cracking and brittle failure. Lastly, Cr and Al are important elements for providing high temperature corrosion resistance. Most commercial alloys contain 10 - 30 wt.% Cr, which leads to corrosion protection by forming a passive Cr₂O₃ surface film. This surface oxide provides protection by limiting inward diffusion of oxygen and Sulphur - bearing species as well as outward diffusion of alloying elements.

IV. PHASES AND MICROSTRUCTURE

Different phases are formed during the fabrication of nickel-based superalloys. The dominating phases are the gamma matrix and the gamma prime.

A. Gamma Matrix (γ) Phase

The foundation of all nickel-based superalloys is the gamma matrix, γ . The continuous phase is nonmagnetic and with a face centred cubic, FCC. Illustration is found in Figure 1.1. Nickel by itself has neither exceptional high elastic modulus

nor low diffusivity. On the other hand, nickel, has a nearly filled third electronic

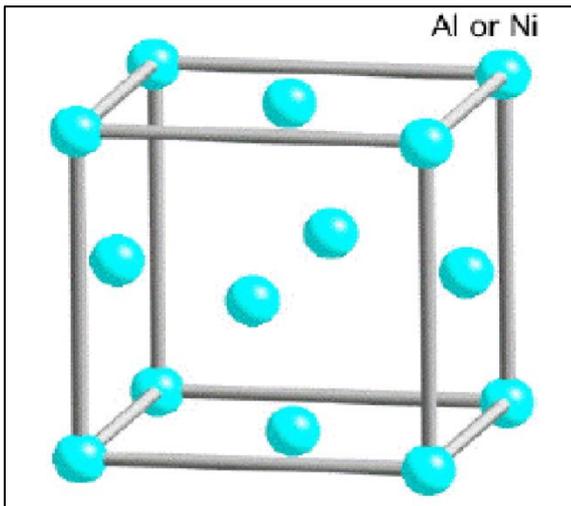


Fig. 1.2: FCC gamma structure.

shell which allows alloying with solid solution strengthening elements without losses in phase stability. The alloy elements composing the gamma matrix mainly belong to Group V, VI and VII and are cobalt, iron, chromium, molybdenum and tungsten.

B. Gamma Prime (γ') Phase

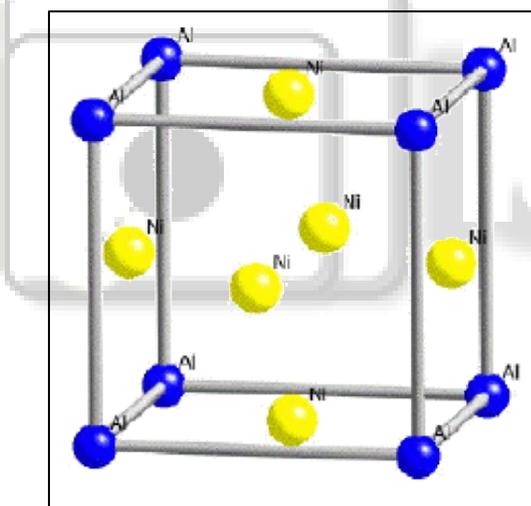


Fig. 1.3: FCC gamma prime structure.

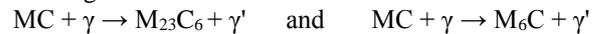
The precipitated phase gamma prime, γ' , wasn't identified until the 1940's. It is formed from elements from Group III, IV and V. The addition of for example aluminium and titanium, which are the essential solutes, results in a reaction with nickel, precipitating the γ' phase of the form Ni_3X , where X is an alloy element. The structure of the gamma prime is FCC. The atom ordering is illustrated in Figure 1.2. Other elements included in the γ' phase can be chromium, hafnium, niobium and tantalum. The γ' lattice parameter differs slightly from the one of gamma matrix. The mismatch is small, <0.2%, for spherical γ' . The close match makes it possible for the γ' to precipitate evenly in the matrix.

There are several benefits of the presence of γ' in the matrix. The coherence between γ' and γ results in a low surface energy and in an exceptional long-time stability. The phase is also the reason for the high temperature strength and creep resistance in most superalloys. The strength of the alloy

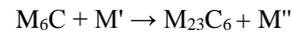
is strongly dependent on the volume fraction of γ' . Wrought alloys contain 20-45 % of γ' . Higher fractions will make the deformation too difficult. Cast superalloys can have a volume fraction of up to 60 %, which will increase the alloy strength compared with the wrought alloys.

C. Carbides

The addition of carbon, generally in amounts of about 0.02 to 0.2 %, will result in a union with refractory elements such as hafnium, niobium and titanium. Primary carbides of the type MC, where M can be one of the just previously mentioned metals, are formed during freezing of the alloy. The structure of these carbides is FCC. MC carbides are an essential source of carbon during heat treatment and service. On those occasions, the carbides tend to decompose into other, secondary, carbide variants, e.g. $M_{23}C_6$ and M_6C . The dominating formulas for formation of the two carbides are,



In some alloys the formation of M_6C from $M_{23}C_6$, according to the formula



has been observed, and in yet other alloys the reverse reaction occurs. M' and M'' can be replaced with chromium and cobalt, nickel or molybdenum, respectively. The most common M element in $M_{23}C_6$ is chromium but also iron, tungsten and molybdenum can be found in that position. The M elements generally found in M_6C are molybdenum and tungsten, but the carbide can contain chromium, cobalt and tantalum as well. [19] The existence of carbides plays an important role in polycrystalline superalloys. Both M_6C and $M_{23}C_6$ carbides are most likely to precipitate at grain boundaries. When properly formed, they strengthen the boundary and restrain grain boundary sliding. When $M_{23}C_6$ is formed in the grain boundaries, the chromium content in the matrix is reduced and the solubility for γ' is augmented in these zones. The shape of the carbides is crucial for the properties. Cellular shape of the $M_{23}C_6$ carbides can cause premature failure while irregular, blocky particles will strengthen the alloy. Fine, intergranular carbides will have a strengthening influence on the material.

V. WELDING PROBLEMS

A. Fusion Zone Solidification Cracking

As is characteristic of weld solidification cracking in other systems, cracks form during the terminal stages of solidification when liquid films are distributed along solidification grain boundaries and, in some cases, interdendritic sites. At this stage, shrinkage strains across the partially solidified boundaries can become appreciable. If the terminal liquid is distributed along the boundaries as a continuous film, the strains cannot be accommodated and the boundaries separate to form a crack. The solidification behaviour of Nb-bearing alloys is very sensitive to the relative amounts of Nb and C and the corresponding type of terminal phase that forms, i.e., NbC or Laves. These trends hold whether Nb is added for solid-solution strengthening or precipitation hardening by formation of the γ'' - Ni₃Nb phase. Since most commercial alloys have Nb additions for precipitate strengthening, The solidification cracking response of γ' strengthened alloys is strongly affected by the

level of minor elements such as P, S, B, C, and Zr. The impurity elements P and S should be held as low as possible for improved cracking resistance since they generally provide no benefit to alloy performance

B. HAZ Liquation Cracking

All engineering alloys melt and solidify over a range of temperatures. In general, the higher the alloy content, the larger the melting/solidification temperature range. Thus, solid-solution strengthened Ni- base alloys can exhibit particularly wide melting and solidification temperature ranges due to their high alloy content. During welding, the base metal just adjacent to the fusion zone will experience a range of peak temperatures that is between the liquidus and effective (non- equilibrium) solidus temperature of the alloy. The microstructure within this region will therefore undergo partial melting and is described as the partially melted zone (PMZ) region of the HAZ. Liquation cracking can occur in the PMZ in Ni- base alloys when the liquid within the locally melted region cannot sustain the applied strain and forms a crack, usually along a grain boundary. This tendency for HAZ grain boundary liquation is increased with welding processes that are made at high heat input or alloys that are single phase, the segregation of alloy and impurity elements (S, P, and B) to the grain boundaries can cause a local depression of the melting temperature and promote the formation of continuous liquid films along these boundaries. Thus, the grain boundaries typically undergo some degree of liquation within the PMZ of single phase materials. It has been suggested that grain boundary segregation can be increased during grain growth within the HAZ as solute and impurity elements are swept into and accumulate in the migrating boundary. It has been shown that fine grained materials are more resistant to HAZ liquation cracking than coarse grained materials. It has been suggested that larger grain sizes increase cracking susceptibility by promoting more extensive grain boundary wetting by the liquid films and inducing higher stress concentrations. Unfortunately, many nickel alloys are used at high temperature where large grain size is favored for better creep resistance. Thus, as with alloy composition, grain size must be balanced for both weldability and performance. In some cases, when coarse grained material is desired for creep resistance, welding can be performed on fine - grained material followed by a grain- coarsening solution annealing treatment.

C. Ductility- Dip Cracking

Ductility dip cracking (DDC) is a solid- state phenomenon that has been observed in a number of structural materials, including austenitic stainless steels, copper alloys, titanium alloys, and Ni - base alloys. In materials susceptible to DDC, a precipitous drop in ductility occurs in the temperature range between the solidus (T_S) and approximately $0.5 T_S$. Often this drop in ductility occurs over a very narrow temperature range. Since contraction stresses during welding can be large in this temperature range, this dip in ductility can result in local exhaustion of ductility and subsequent DDC. Haddrill and Baker defined DDC as the loss in ductility, over a temperature range below the solidus that is sufficient to produce cracking under the influence of thermal strain caused by welding.

DDC has become a particularly difficult problem in applications where high- C_r (~ 30 wt%), Ni - base consumables are required to provide corrosion compatibility with high - Cr base metals, such as Alloy 690. In applications, that do not require this level of corrosion resistance, consumables such as Filler Metals 82 and 625 will usually be effective in avoiding DDC. (Note that this may not be true in heavy section, multi pass welds.) Both these consumables contain sufficient Nb to ensure that NbC forms at the end of solidification, thus promoting weld metal microstructures with tortuous grain boundaries that are resistant to DDC. When use of resistant filler metals is not an option, precautions should be taken to minimize weld restraint through either joint design or process variables. While not particularly attractive from a productivity standpoint, the use of small stringer beads that minimize the overall heat input in large weldments can help reduce susceptibility to DDC. As noted in the previous section, the use of argon - hydrogen shielding gases tends to increase susceptibility to DDC. These shielding gas mixtures have been recommended because they improve the wetting characteristics of many Ni - based filler metals and help prevent lack- of- fusion defects. The use of pure argon or helium shielding gases will generally improve resistance to DDC, but may not be effective in totally eliminating it. Ultimately, filler metals must be designed that are resistant to DDC by the formation of weld metal microstructures that resist elevated temperature grain boundary sliding. The 52MSS (Ni - 30Cr - 4Mo - 2.5Nb) filler metal discussed in the previous section shows considerable promise in terms of generating such a microstructure, but research must continue to better understand the underlying mechanism. It should be noted that while the addition of Nb appears promising with regards to the improvement in DDC resistance, the formation of low melting Nb - rich eutectic constituents potentially increases susceptibility to weld solidification cracking. It is likely that a careful balance will have to be maintained so as not to solve one weldability problem while creating another.

D. Strain-Age Cracking

Strain - age cracking (SAC) is a form of reheat, or postweld heat treatment, cracking that is specific to the precipitation strengthened Ni - base alloys. It is a solid- state cracking phenomenon that is most often observed in the HAZ just adjacent to the fusion boundary, although it is possible for SAC to occur in the weld metal of these alloys. In most cases, it occurs during postweld heat treatment, but is also possible (although unlikely) during reheating in multipass welds. This form of cracking is most prevalent with the γ' , $Ni_3(Al,Ti)$, strengthened alloys and many of these alloys are considered "unweldable" because of this cracking mechanism. the term strain - age cracking (SAC) is derived from the fact that both local strain and aging must occur nearly simultaneously, the stresses which cause cracking may have three origins:

- 1) weld residual stress,
- 2) thermally - induced stresses arising from the difference in the coefficient of thermal expansion between the base material and weld metal, and
- 3) Stresses from dimensional changes caused by precipitation.

It is widely believed that PWHT cracking is the result of stresses developed during post weld heat treatment that are relaxed preferentially in the HAZ at the same time that the ductility of the HAZ is reduced by metallurgical reactions. The stresses contributing to cracking have been suggested to include residual welding stresses, stresses due to localized thermal expansion, and aging contraction stresses. The most effective method to avoid SAC is through alloy selection. Alloys that are strengthened by the γ' , $\text{Ni}_3(\text{Ti},\text{Al})$, precipitate are the most prone to SAC. As shown in Figure 4.60, as Ti and Al content of the alloy increases, susceptibility to SAC increases. By reducing (Ti + Al) content or strengthening the alloy by precipitation of γ'' (Ni_3Nb), susceptibility to SAC can be reduced or eliminated. Alloy 718, which is strengthened by γ'' , was developed as a SAC-resistant superalloy and is widely used in welded applications. Many of the high-strength superalloys (those with high Ti + Al content) are effectively “unweldable” due to SAC. The high levels of (Ti + Al) result in a high volume fraction of γ' , providing high strength at elevated temperatures, but also exacerbating SAC during PWHT. Welding in the solution annealed or overaged condition may provide some relief from SAC in these high-strength superalloys. The use of low-heat input welding processes and fine-grained base metal will also improve resistance to SAC, but avoiding cracking in highly-constrained structures can be challenging.

VI. CONCLUSION

This study of the various types of nickel-based alloys and the effect of alloying elements on the properties of the alloy, welding problems present in nickel-based alloys is carried out. Based on the above study, various types of welding problems were found to occur in nickel-based alloys, and remedies or steps to be followed to avoid these welding problems are discussed respectively.

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