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METHOD TO IMPROVE THE WELDABILITY AND FORMABILITY OF NICKEL-BASE SUPERALLOYS

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ABSTRACT OF THE DISCLOSURE

A procedure for improving the capacity of the advanced nickel-base superalloys to be welded without cracking is described. Concurrently, the process also provides enhanced room temperature formability. Basically, the procedure involves an overaging heat treatment of the material to produce a coarse γ' precipitate distributed throughout the alloy microstructure, a condition of low strength and enhanced room temperature ductility.

This is a division of application Ser. No. 746,011, filed July 19, 1968, now abandoned.

BACKGROUND OF THE INVENTION

The present invention is most conveniently characterized as a metal processing technique and has particular application to these processes involving the fabrication or welding of the advanced nickel-base superalloys.

The typical nickel-base superalloy is essentially a nickel-chromium solid solution (γ phase) hardened by the addition of materials such as aluminum and/or titanium to precipitate an intermetallic compound (γ' phase). The predominant intermetallic compound precipitated, represented by the formula $Ni_3(Al, Ti)$, is an ordered face-centered-cubic structure with aluminum and titanium at the corners of the unit cell and nickel at the face centers. These alloys also normally contain cobalt to raise the solvus temperature of γ' phase, refractory metal additions for solution strengthening, and carbon, boron and zirconium to promote ductility and fabricability.

In the gas turbine engine industry in which the nickel-base superalloys are widely utilized, progressive increases in the powerplant performance requirements have led to increases in engine operating temperatures which have in turn imposed increasingly stringent demands on the turbine materials. Historically, the extent of the engine temperature increases has been limited by the physical characteristics of the hot section alloys, particularly those used in the highly stressed components such as turbine blades and vanes. Recent alloy developments, promising advances in coating technology, and the use of internal cooling techniques, however, will now allow turbine operating temperatures to be significantly increased. Such temperature increases mean that casements and other sheet metal structures, and numerous other components will also be subjected to higher service temperatures. As a consequence, currently available sheet materials formed from the lower strength materials, including certain of the low alloy content superalloys, such as Waspaloy, may prove inadequate under some engine operating conditions. The need for improved sheet alloys is thus suggested.

High temperature superalloy sheet materials have recently been derived from some of the forging superalloys. One such advanced superalloy that shows promise in terms of its physical properties is that identified in the industry as Udimet 700. Another is Astroloy. The properties of sheet material purchased to the Udimet 700 or Astroloy chemistry are superior to those of any of the

other nickel-base alloy sheet commercially available at the present time. Unfortunately, the use of these materials in sheet form has been virtually prohibited by two factors, i.e., poor weldability and limited formability.

Udimet 700, for example, is susceptible to hot cracking during welding. However, the greater weldability problem with this alloy is its tendency to crack during post-weld heat treatment under the influence of residual stress, i.e., strain-age cracking. Age-hardenable, nickel-base alloys must be heat treated after welding to restore mechanical properties and relieve residual stresses, and it has been found that restrained weldments of most of these alloys are susceptible to cracking during this heat treatment, particularly during the heat-up stage. Alloys of lower alloy content such as Waspaloy or René 41 generally exhibit strain-age cracking only when welded in the fully heat-treated condition and then taken directly to the aging temperature during the subsequent heat treament. Weldments made of the latter alloys, when initially solution heat treated, can usually be successfully heat treated after welding. Restrained Udiment 700 weldments, however, cannot be post-weld heat treated without cracking even when welded in the solutioned condition and then heated as rapidly as possible through the aging range up to the solution heat treatment temperature. This is due to their high alloy content which results in additional precipitation of the γ' phase and consequent hardening which cannot be prohibited despite rapid heating.

Another undesirable characteristic of the advanced superalloys is their relatively poor cold formability. The capability of being formed into hardware shapes at low temperatures is an essential characteristic for a useful sheet material.

As is evident from the foregoing, the weldability and formability problems are not as significant in the case of certain of the low alloy content superalloys as in the case of the advanced nickel-base superalloys. Accordingly, the processes described herein are directed toward and have particular application to those alloys having more than 30 volume percent γ' precipitate at room temperature in the fully strengthened condition. Representative of the advanced alloys of this type are those identified in the industry as follows:

Description:	Composition (by weight)
Udimet 500	19% Cr, 17% Co, 3% Ti, 3% Al, 4% Mo, .07% C, .005% B, balance Ni.
Udimet 700	15% Cr, 18.5% Co, 3.3% Ti, 4.3% Al, 5% Mo, .07% C, .03% B, balance Ni.
Mar–M200	9% Cr, 10% Co, 2% Ti, 5% Al, 12.5% W, 1% Cb, .15% C, .015% B, .05% Zr, balance Ni.
Astroloy	15.5% Cr, 17% Co, 3.3% Ti, 4.5% Al, 5.3% Mo, .07% C, .03% B, balance Ni.
B–1900	8% Cr, 10% Co, 1% Ti, 6% Al, 6% Mo, 4.3% Ta, .11% C, .15% B, .07% Zr, balance Ni.

The above alloys, primarily as a function of their high temperature strength as related to their high aluminum and/or titanium contents, are classified as advanced nickel-base superalloys of the γ–γ' type. A particular characteristic of these alloys which contain a large volume fraction of the γ' phase, is that they undergo a hardening stage at a temperature intermediate room temperature and the solutioning temperature during the usual post-weld heat treatment. It is this phenomenon which establishes a proneness to strain-age cracking.

SUMMARY OF THE INVENTION

The present invention is directed toward improving the weldability and formability of the advanced nickel-base super-alloys of the γ–γ' type having more than 30 volume percent of the precipitate in the room temperature condition. The foregoing objectives are achieved through establishment by an appropriate heat treatment cycle of an alloy microstructure characterized by a coarse, stable γ' precipitate. Through a combination heat treatment, involving solutioning plus overaging a microstructure is obtained which provides not only a relatively low yield strength and high ductility at room and elevated temperatures, but also a microstructure which resists the precipitation of additional deleterious phases, particularly strengthening or hardening phases, during heat-up in a post-weld heat treatment.

In a preferred embodiment of the present invention, maximum resistance to weld cracking is obtained by a heat treatment sequence comprising solutioning followed by a two-stage overaging consisting of aging near the upper end of the aging temperature range for the alloy involved, followed by aging at a temperature near the lower end of the aging temperature range(but above the $M_{23}C_6$ carbide precipitation temperature range), with a slow cool to room temperature.

In another preferred embodiment, the same two-step overaging treatment is followed except that, instead of a slow cool, a subsequent oil quench is utilized to produce an additional improvement in room temperature formability.

DESCRIPTION OF THE PREFERRED EMBODIMENTS

The basic concept involved here relates to the use of specific heat treatments which selectively alter the microstructural characteristics of the alloy. These heat treatments overage or coalesce the strengthening γ' precipitates to produce a microstructure that is more amenable to fabrication than the solution heat-treated condition employed in the past, specifically one characterized by a temporary condition of lower room temperature strength and increased ductility. Because of high alloy content of the advanced nickel-base superalloys, the improved fabrication characteristics cannot be obtained by retaining the precipitate in solution, as is done in the case of the less highly alloyed materials.

In addition, to provide the desired resistance to strain-age cracking in these advanced alloys it has been necessary to provide a microstructure which does not go through a strengthening or hardening phase as the temperature of the alloy is raised after welding for stress relief or for resolutioning and conventional aging. The concept here is to slow cool from the overaging temperature to reduce the potential for additional precipitation of the strengthening phase during the initial portion of a post-weld heat treatment. Since this reduces the cold formability of the alloy somewhat, when maximum cold workability is desired a rapid cool from the overaging temperature is suggested to prevent the additional precipitation of fine γ' particles and possibly some grain-boundary carbides.

FORMABILITY EVALUATION

Specimens used for tensile, bend and hardness tests as well as for metallographic studies were produced from ⅝ inch diameter, hot-rolled, centerless-ground, Udimet 700 barstock. The weldability studies were made on 0.060 inch thick commercially rolled sheet of Astroloy chemistry.

Bend tests were conducted at room temperature using a guided bend fixture with a 0.062 inch radius plunger. Tests were terminated at the onset of cracking when possible, although several of the less ductile specimens failed completely before the test could be terminated, and the bend angle achieved before cracking was taken as the measure of relative formability.

Hardness measurements were made on metallographic samples using a Vickers hardness tester with a 30 kilogram load. Tensile tests were conducted utilizing a Riehle tensile machine at a strain rate of 0.033 min.$^{-1}$.

WELDABILITY EVALUATION

The weldability of Udimet 700 sheet in various conditions of heat treatment was investigated by means of restrained weld patch test. The restraint was provided by initially welding the test sheet specimen to a Hastelloy X strong-back before making the test weld on the specimen. Both longitudinal and transverse test welds were investigated, all welds being made by automatic gas tungsten-arc welding with filler wire additions.

HEAT TREATMENT EVALUATION

The first series of tests were designed to determine the conditions providing maximum room temperature ductility to the alloys. The desired heat treatments developed a coarse, overaged dispersion of the γ' precipitate which, when retained at room temperature, provides optimum ductility and low yield strength. The results of the various heat treatments on Udimet 700 are summarized in Table I.

TABLE I.—ROOM TEMPERATURE BEND DUCTILITY AND HARDNESS OF HEAT-TREATED UDIMET 700

History	Bend angle (degree)	Hardness (DPN)
Solution heat treated (2,140° F./4 hrs. plus forced air cool)	64	380
Fully heat treated (2,140° F./4 hrs. plus 1,975° F./4 hrs. plus 1,550° F./4 hrs. plus 1,440° F./16 hrs)	24	407
One-step overaging treatments:		
Solution heat treat plus following:		
1,975° F./1 hr. plus oil quench	115	
1,975° F./4 hrs. plus oil quench	137	
1,975° F./16 hrs. plus oil quench	147	320
1,975° F./64 hrs. plus oil quench	145	
1,975° F./16 hrs. plus furnace cool	73	
1,900° F./16 hrs. plus oil quench	80	
1,850° F./16 hrs. plus oil quench	47	
1,975° F./16 hrs. plus oil quench [1]	130	
Two-step overaging treatments:		
Solution heat treat plus 1,975° F./16 hrs. plus furnace cool to following:		
1,900° F./4 hrs. plus oil quench	142	
1,850° F./4 hrs. plus oil quench	168	290
1,850° F./1 hr. plus oil quench	121	
1,800° F./4 hrs. plus oil quench	154	
1,700° F./4 hrs. plus oil quench	106	
Two-step overaging treatment plus slow cool: Solution heat treat plus 1,975° F./16 hrs. plus furnace cool to 1,850° F./4 hrs. plus slow cool to room temperature (maximum 100° F./hr. to 1,050° F., furnace cool to room temperature)	74	344

[1] Fully heat treated prior to 1,975° F./16 hrs. plus oil quench.

As is evident from the foregoing table, the two-step over-aging treatment plus rapid cool develops the greatest room temperature ductility. The microstructure produced by this thermal treatment consists of coarse γ' particles dispersed throughout the grains and along grain boundaries. No fine γ' precipitate or $M_{23}C_6$-type grain-boundary carbides were revealed by electron microscopy studies of this structure. This is distinct from both the duplex (coarse+fine) γ' structure observed in the fully heat treated alloy and the very fine dispersion of γ' phase found in normal solution heat-treated Udimet 700.

The purpose of the overaging heat treatment is, of course, to form the coarse, stable γ' precipitate dispersed in the alloy microstructure. This particular heat treatment basically involves a temperature-time relationship. The object is to precipitate the maximum amount of γ' phase in agglomerate particles. In the aging process, the desired microstructure will form if sufficient time at temperature is allowed. The two-step overaging process is preferred, however, since it accelerates the formation of the coarse γ' phase and maximizes the quantity of the γ' phase precipitate within a reasonable time. The preferred heat treatment involves, in essence, a nucleation stage with some growth at a relatively high aging temperature followed by an agglomeration stage at a slightly lower temperature with sufficient time at each stage to allow growth of the γ' particles and precipitation of the major

portion of the γ' phase present in the alloy. Both stages are above the temperature for the precipitation of significant amounts of $M_{23}C_6$-type carbides.

In the following table, the room temperature tensile properties of Udimet 700 barstock are set forth for various conditions of heat treatment.

TABLE II.—DUCTILITY OF UDIMET 700

History:	Elongation (percent)
Solution heat treated	28.2
Solution+2-step overage+oil quench	39.0
Solution+2-step overage+slow cool	33.0
Normal full heat treatment	19.4
Solution+2-step overage+oil quench+full heat treatment	18.6

TABLE III.—STRENGTH CHARACTERISTICS OF UDIMET 700

History	Yield strength (k.s.i.)	Ultimate tensile strength (k.s.i.)
Solution heat treated	125	189
Solution plus 2-step overage plus oil quench	104	169
Solution plus 2-step overage plus slow cool	106	181
Normal full heat treatment	138	205
Solution plus 2-step overage plus oil quench full heat treatment	140	202

From the foregoing table, the desired ductility increase and the lower tensile strength of the alloy following overaging are evident. Equally important, however, is the demonstration of full strength recovery attainable with a final full heat treatment.

Although the two-step overaging treatment plus rapid cool develops the greatest room temperature ductility, the micro-structure created still has considerable potential for additional precipitation during the initial part of the necessary post-weld heat treatment. This subsequent precipitation potential forces the welded alloy through an intermediate hardening phase en route to the preferred post-weld temperature and, accordingly, makes the alloy prone to strain-age cracking. In order to reduce this subsequent precipitation potential and, hence, to improve strain-age cracking resistance, a slow cooling rather than an oil quench after overaging is preferred. The slow cooling (typically 50° F./hr. to 1650° F.; 100° F./hr. to 1050° F.; then furnace cooling to room temperature) induces additional γ' precipitation and reduces the ability of the specimen to precipitation γ' during subsequent heating. The bulk of the γ' phase, however, will necessarily have been precipitated during the overaging cycle. While slow cooling reduces the room temperature bend ductility and, hence, cold formability due to precipitation of fine γ' particles and some grain-boundary carbides, this modification is nevertheless preferred from a weldability standpoint.

Various samples were evaluated for different conditions of heat treatment to determine the weldability of the samples. The results of these tests are summarized in Table IV.

The increased resistance of these alloys, not only to hot cracking in the welding process, but more particularly to strain-age cracking in subsequent heat treatment operations, is clearly evident. Moreover, it has been previously demonstrated that the conventional heat treatment (solution) normally utilized to provide maximum workability to the nickel-base alloys, while satisfactory with those superalloys of low alloy content (less than about 30 volume percent γ' precipitate at room temperature), is actually the worst condition of heat treatment in which the alloy can be placed in terms of its susceptibility to strain-age cracking.

The methods disclosed herein provide both weldability and cold formability to the advanced nickel-base superalloys by means of heat treatment. Overaging heat treatments provide the advantageous alloy microstructure to enhance ductility and reduce strength. For maximum room temperature ductility a two-step overaging heat treatment followed by a rapid cool is preferred, while strain-age cracking resistance is enhanced using the two-step overaging with a slow cool.

The described overaging heat treatments are also applicable not only to the wrought alloys but also to the cast high-strength superalloys which are notoriously difficult to form. Wrought Udimet 700 barstock has been successfully warm rolled to 0.020 inch thick sheet at temperatures where the γ' precipitate is present in an overaged state. Through the use of the overaging heat treatment prior to working, B–1900 (a cast superalloy) has been successfully swaged to 0.060 inch diameter wire at room temperature. The particular inventive heat treatments, however, do appear confined in utility to the advanced nickel base superalloys.

Future improvements in alloy chemistry may further enhance the weldability and fabricability of the superalloys, although to date such improvements have been achieved only at the expense of high temperature strength. It is probable, however, that despite such compositional changes, the future nickel-base superalloys of the γ–γ' type will advantageously react to the thermal processing methods described herein. In any event, the present invention expands the scope of utility of today's advanced superalloys.

The principal heat treatments utilized in the practice of the present invention will, of course, vary somewhat with the alloy chemistry. The preferred heat treatments in terms of temperature vary not only with alloy composition but also with prior history and intended application. Such preferred heat treatment temperatures are readily available to those in the art either in the literature or through routine experimentation.

In the case of Astroloy, for example, the preferred heat treatments are as follows:

SOLUTION HEAT TREAT

2140° F./4 hours+forced air cool.

TABLE IV.—WELDABILITY OF ASTROLOY SHEET

History	Relative crack resistivity	Relative degree of cracking: During welding	Relative degree of cracking: Strain-age cracking
Solution heat treated	Severe	High	High.
Full heat treatment	Severe [a]	Intermediate	Intermediate.
	Moderate [b]		
	Slight [c]		
Solution plus 2-step overage plus oil quench,	Severe	Low	High.
Solution plus 2-step overage plus slow cool.	Moderate [d]	Low	Low.
	Slight [a b c]		
	None [e]		

[a] Simulated repair weld (Waspaloy Filler).
[b] Initial weld (Waspaloy Filler) (1,000° F./hr. heating rate).
[c] Initial weld (Waspaloy Filler).
[d] Initial weld (Waspaloy Filler) (full penetration weld).
[e] Initial weld (Inconel 718 Filler).

FULL HEAT TREATMENT

2140° F./4 hours+forced air cool, plus 1975° F./4 hours+forced air cool, plus 1550° F./4 hours+forced air cool, plus 1400° F./16 hours+forced air cool.

TWO STEP OVERAGE FOR MAXIMUM FORMABILITY

2140° F./4 hours+forced air cool, plus 1975° F./16 hours+furnace cool at 100° F./hr., to 1850° F./4 hours+oil quench.

TWO STEP OVERAGE FOR MAXIMUM WELDABILITY

2140° F./4 hours+forced air cool, plus 1975° F./16 hours+furnace cool at 100° F./hr., to 1850° F./4 hours+slow cool at 50° F./hr. from 1850° F. to 1650° F., at 100° F./hr. from 1650° F. to 1050° F., and furnace cool from 1050° F. to room temperature.

For other materials of particular interest representative heat treatment temperatures are as follows:

Material	Heat treatment (° F.)		
	Solidus, lowest melting point	Solution	Aging
B-1900	2,300	2,200-2,300	1,850-2,050
MAR-M200	2,300	2,250-2,300	1,850-2,100
Udimet 500	2,230	1,975-2,230	1,825-1,900
Udimet 700	2,220	2,100-2,220	1,800-1,975

The B-1900 alloy and the conventional heat treatments therefor are described in detail in the United States patent to Baldwin 3,310,399.

While the present invention has been described in connection with certain examples, materials and process parameters, these will be understood to be illustrative only. This invention is generally applicable to all of the nickel base superalloys wherein the γ' phase is present at room temperature in an amount exceeding about 30 volume percent. Furthermore, considerable variation in the selection of temperatures, time and cooling rates is possible in the processes to achieve the same basic microstructures which result from employment of the more preferred parameters. Accordingly, in accordance with its true spirit, this invention will be measured not by the illustrative material but by the appended claim.

What is claimed is:

1. The method of welding articles formed of the age-hardenable, nickel-base superalloys of the γ–γ' type containing at least about 6 weight percent of hardening elements selected from the group consisting of aluminum and titanium having a γ' precipitate phase in a quantity of at least about 30 volume percent at room temperature in the fully strengthened condition which comprises:

solution treating each such article at a temperature generally within the range of about 1975°–2300° F., above the γ' solvus temperature but below the superalloy solidus temperature, for a period of time sufficient to solution the γ' phase;

heat treating the article at a temperature generally within the range of about 1800°–2100° F., below the γ' solvus temperature but above the carbide precipitation temperature, the temperature and time at temperature being selected to maximize the γ' particle size and precipitate most of the γ' phase as coarse particles, providing an overaged alloy condition of relatively low hardness having coarse-stable γ' dispersed in the alloy microstructure;

slowly cooling the article to a temperature sufficiently low to insure the precipitation of substantially all of the γ' phase remaining in solution after overaging;

welding at least two of the articles together into an assembly;

heating the assembly to a temperature generally within the range of about 1975°–2300° F., above the γ' solvus temperature, for a period of time sufficient to solution the γ' phase;

and subsequently aging the assembly generally within the range of about 1800°–2100° F. to precipitate the γ' phase as fine particles providing an alloy condition of high strength and hardness.

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