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A Gleeble®-based Method for Ranking the Strain-Age Cracking Susceptibility of Ni-Based Superalloys

Results of thermomechanical testing for cracking susceptibility were compared to those obtained from other investigations

BY D. A. METZLER

ABSTRACT. Gamma-prime strengthened Ni-based superalloys comprise a family of critical construction materials for modern gas turbines used in land-based power-generation and aviation applications. Strain-age cracking during postweld heat treatment remains a critical issue in the widespread use of higher-strength members of this alloy family. Previous work (Ref. 1) demonstrated that the load frame-based controlled heating rate test (CHRT) was capable of ranking the strain-age cracking susceptibility of these alloys. Various attempts have been made over the years to adapt this test methodology to the Gleeble® thermomechanical simulator using a variety of specimen geometries and thermomechanical cycles. In this study, the strain-age cracking susceptibility of six, gamma-prime strengthened Ni-based superalloys was evaluated using a simple (both from a specimen geometry and thermomechanical cycle perspective), Gleeble-based test method and compared to results obtained by previous investigators.

Introduction

Gamma-prime strengthened Ni-based superalloys embrace a unique combination of high-temperature mechanical strength, resistance to creep deformation, and oxidation resistance necessary for efficient operation of modern gas turbine equipment. Typically, aluminum, titanium, and possibly niobium are added to an oxidation-resistant Ni-Cr-Co-Mo matrix to develop higher strengths in the 538°–871°C (1000°–1600°F) operating range and, in the case of aluminum additions, further augment the alloy's resistance to oxidation. Because these elements precipitate from solid solution over that same temperature range, this alloy family can become susceptible to a postfabrication heat treatment cracking phenomenon known as strain-age cracking. This type of cracking can occur when alloys containing gamma-prime-forming elements in solid solution (such as mill-annealed products) are heated through the 593°–982°C (1100°–1800°F) temperature range during postfabrication solution-annealing heat treatment. During gamma-prime precipitation, alloy ductility may drop significantly. Cracking will occur if the alloy is subjected to strains that exceed its available ductility. In mechanically restrained parts, tensile stresses will develop because precipitation of gamma-prime from solid solution also produces a bulk volume contraction of the matrix (Refs. 2, 3). Cracking may also be aggravated by a coarse grain size, an oxidizing environment (Ref. 2) (producing oxygen embrittlement at grain boundaries) and by grain boundary carbide films produced by constitutional liquation. These factors are often associated with welding processes because of grain growth in the associated HAZ, constitutional liquation in the HAZ, and geometric stress concentration at the weld toe. Strain-age cracking, however, is not exclusively limited to weld HAZs and can occur in unaffected base material (Ref. 2).

As alloy designs become more sophisticated, secondary processing effects such as strain-age cracking are more likely to be examined in the development of new compositions. This highlights the need for a robust method for assessing a given alloy's susceptibility to strain-age cracking. Such a test method should 1) deliver a quantitative index of alloy susceptibility to strain-age cracking, 2) deliver a reproducible index of alloy susceptibility to strain-age cracking, and 3) deliver results that accurately reflect real-world experience.

The test method should be 1) mechanically simple, 2) usable with sheet and plate materials as well as bar stock, 3) require minimal setup/stabilization time, and 4) adaptable to existing mechanical test equipment.

A wide variety of strain-age cracking tests have been described in the literature and reviewed by Rowe (Ref. 1). He found that circular patch-type tests required considerable investment in automated weld equipment and, at best, produced semiquantitative results. Others (Refs. 4, 5) reported that this test did not consistently crack R-41 alloy (considered very strain-age-cracking susceptible) unless conducted as a repair weld simulation. This finding severely limits the usefulness of the circular patch test in ascertaining the strain-age susceptibility of less crack-prone alloys.

The Gleeble thermomechanical simulator offers considerable flexibility in ther-

KEYWORDS

Gleeble
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Heat Treatment
Ni-Based Superalloy
Strain-Age Cracking

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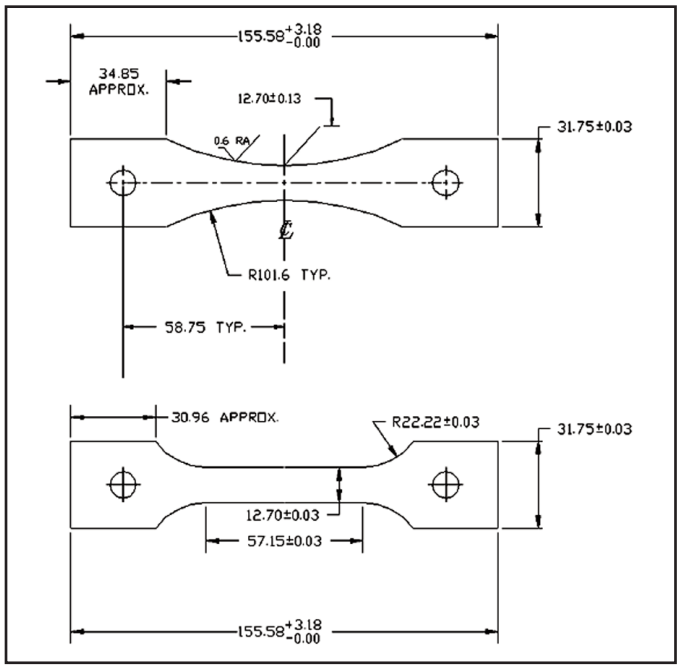

Fig. 1 — Modified and standard CHRT sample geometries (dimensions in mm).

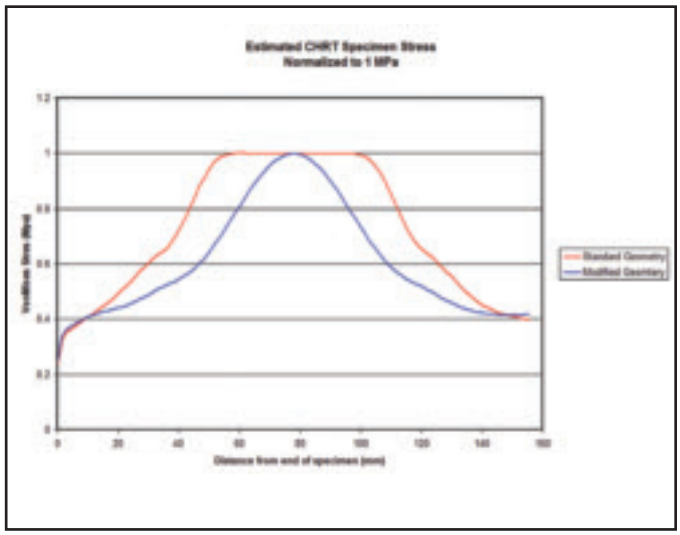

Fig. 2 — FEA stress analysis of modified and standard CHRT samples.

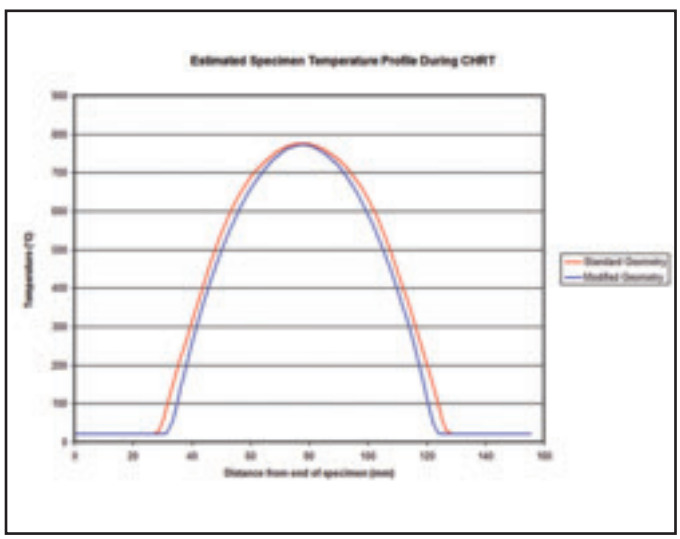

Fig. 3 — FEA thermal analysis of modified and standard CHRT samples.

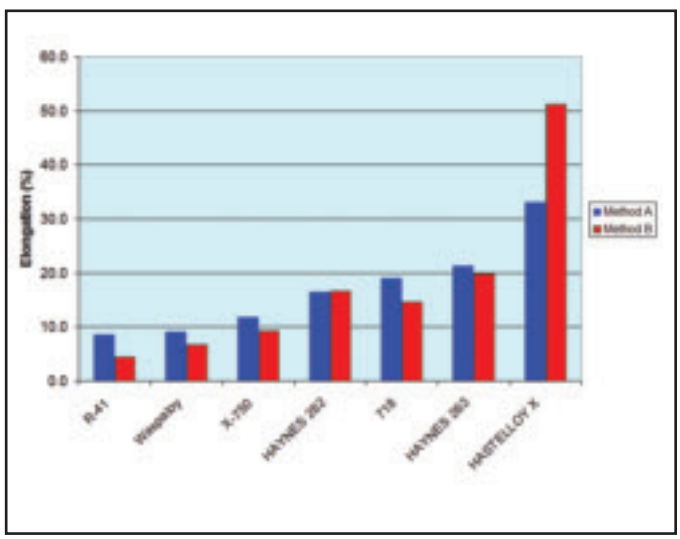

Fig. 4 — Methods A and B results (25.4-mm gauge length).

mal cycles and loads that can be applied to a number of different sample geometries. Various Gleeble-based test methods have also been reviewed (Refs. 1, 6, 8), but most methods were either confined to round bar samples or notched sheet samples. Additionally, various Gleeble-based methods reported in the literature appear as the subject of a single study/paper and thus have not been widely accepted in the welding arena. Little or no correlation of their results with real-world strain-age cracking incidents has been reported.

Among the many test methods described in the literature, the controlled heating rate test (CHRT) appears to have the greatest potential as an economical method for ranking the strain age cracking susceptibility of Ni-based superalloys. In this test, developed in the late 1960s by Prager et al. (Ref. 7), a solution-annealed (mill-annealed) tension test sample is heated at a controlled (usually constant) rate to a temperature in the gamma-prime precipitation range and pulled to failure at a controlled extension rate. Generally, this procedure is repeated over a range of temperatures in the gamma-prime precipitation range and the minimum elongation to failure taken as a given alloy's index of strain-age cracking susceptibility. This test method appears to meet a number of the desired test attributes listed below.

1) This test method is mechanically simple, essentially loading a simple sample geometry in tension, only.

2) The CHRT method appears adaptable to sheet and plate materials, as well as bar stock.

3) Existing mechanical test equipment (both conventional loading frames and Gleeble thermomechanical simulators)

Table 1 — Composition of Alloys Used in CHRT Trials

Alloy Element	Rene 41 wt-%	Waspaloy wt-%	HAYNES 282 wt-%	718 wt-%	X-750 wt-%	HAYNES 263 wt-%	HASTELLOY X wt-%
Al	1.49	1.35	1.46	0.63	0.58	0.57	0.00
B	0.01	0.00	0.00	0.00	0.00	0.00	0.00
C	0.09	0.08	0.06	0.05	0.04	0.06	0.07
Co	10.69	13.37	10.11	0.11	0.00	19.91	1.26
Cr	19.42	18.87	19.67	18.22	15.39	19.91	22.10
Cu	0.01	0.01	0.00	0.07	0.01	0.00	0.00
Fe	3.72	0.00	0.20	18.14	8.23	0.36	18.27
Mn	0.01	0.00	0.04	0.17	0.15	0.41	0.65
Mo	9.91	4.26	8.36	2.99	0.00	5.77	8.99
Si	0.05	0.05	0.04	0.13	0.14	0.23	0.48
Ti	3.11	2.99	2.08	1.01	2.55	2.19	0.00
V	0.02	0.00	0.01	0.00	0.00	0.00	0.00
W	0.05	0.00	0.00	0.00	0.00	0.00	0.75
Nb	0.00	0.00	0.00	5.31	0.96	0.00	0.00
Ni	51.43	59.02	57.97	53.17	71.95	50.59	47.43
Est. Gamma-Prime Vol. Fraction at 500°C	0.30	0.28	0.24	0.24	0.21	0.15	0.00
Atomic-% Al+Ti+Nb+Ta	6.93	6.49	5.63	5.88	4.91	3.86	0.00
Initial Grain Size (microns)	76	76	107	32	53	107	90

can be used to perform CHRT.

4) Nonmaterial-related test input variables such as sample heat-up rate, steady-state temperature distribution, and strain rate to failure can be accurately controlled, reducing variation in test results.

Rowe (Ref. 1) was able to adapt this technique to a standard elevated-temperature tension testing load frame and resistance-heated clamshell furnace arrangement with relatively good success, reporting excellent correlation of minimum CHRT elongation with estimated volume fraction of gamma-prime over a range of alloy compositions. Rowe's technique, however, required testing each alloy composition at a minimum of three temperatures in the gamma-prime precipitation temperature range. Each test required considerable setup time, involving the installation and wiring of three control thermocouples to the samples and furnace control system. The vertically oriented tension test/heating furnace equipment also required careful insulation to avoid inadvertent thermal gradients produced by "chimney effects." Even with proper equipment setup, sample heating rates barely reached the desired 17°C (30°F) per minute into the gamma-prime range. These equipment constraints, coupled with the three-temperature per alloy testing requirement, limited the practical sample size per alloy composition to a relatively small number. This, in turn, made rigorous assessment of this test method's discrimination and inherent variability difficult.

The temperature/heating rate/loading rate versatility of the Gleeble, coupled with the simplicity of the CHRT test method, potentially offers a way to meet the broad strain-age cracking test requirements outlined above. Since samples are directly heated, electrically, under closed-loop, real-time temperature control using a sensing thermocouple, percussion welded directly to the test specimen, very high, yet controlled heating rates are available with this testing machine. Closed loop control obviates the need for careful insulation around the test specimen, reducing test setup time considerably, compared to conventional load frame arrangements. Similarly, loading rates are not necessarily confined to the relatively slow rates (≤ 500 mm/min) achievable with screw-driven, conventional load frames. Additionally, since test specimens are enclosed in a relatively small chamber in the Gleeble, this apparatus offers the possibility of performing CHRT tests in environments other than air (and perhaps quantifying the effects of oxygen-induced grain boundary embrittlement).

In spite of its apparent advantages, adapting the Gleeble for CHRT tests poses one serious challenge. Any sample geometry must be clamped in the machine's jaws at each end, to serve both as a contact for the electrical current used for heating and for introduction of the programmed mechanical load cycle. Contact at the jaws introduces a heat sink at each end of the sample and thermal conduction through the sample produces thermal gradients along the sample's axis. If a standard, constant gauge section sheet tensile sample were used as a CHRT specimen, this specimen would reach its peak temperature at its midspan. Gamma-prime precipitation would first occur at this location, relative to the rest of the sample. The sample becomes "strongest" at its midspan. If the CHRT load were applied over a short time period, the sample would likely fail off-center in the areas that have not yet undergone (or have undergone significantly less) gamma-prime precipitation. This could produce abnormally high or erratic elongation to break values in the CHRT test.

In this study, two approaches were explored in an effort to circumvent the effects of these undesired thermal gradients in CHRT test samples. They are as follows:

1. Modify the sample geometry so its cross section increases with distance from its axial centerline, forcing it to break at midspan. This approach, however, could introduce shear components to the applied stress that could possibly reduce sample elongation during CHRT loading.

2. "Pre-age-harden" CHRT samples before testing and add a solution-annealing treatment (before the classic CHRT heating cycle) to the beginning of the CHRT testing thermal cycle. Presumably this would "strengthen" those portions of the gauge section that did not reach peak temperature due to the axial thermal gradient along the sample's axis. Unfortunately, this approach more closely simulates a repair weld than an initial fabrication.

Table 2 — Modified CHRT Test Results and Tests for Significant Differences

Elongation in 25.4 mm - Method A

Alloy	Count	Mean	Homogeneous Groups
R-41	5	8.59	X
Waspaloy	5	9.07	X
X-750	5	11.85	X
HAYNES 282	5	16.42	X
718	5	19.00	X
HAYNES 263	5	21.35	X
HASTELLOY X	5	33.11	X

95% CL = +/- 0.885 (Fisher's least significant difference method)

Elongation in 38.1 mm - Method A

Alloy	Count	Mean	Homogeneous Groups
Waspaloy	5	6.83	X
R-41	5	6.91	X
X-750	5	8.04	X
718	5	12.84	X
HAYNES 282	5	12.97	X
HAYNES 263	5	18.85	X
HASTELLOY X	5	23.98	X

95% CL = +/- 0.763 (Fisher's least significant difference method)

Elongation in 50.8 mm - Method A

Alloy	Count	Mean	Homogeneous Groups
Waspaloy	5	5.94	X
R-41	5	6.13	X
X-750	5	6.37	X
718	5	9.99	X
HAYNES 282	5	10.47	X
HAYNES 263	5	16.21	X
HASTELLOY X	5	18.33	X

95% CL = +/- 0.722 (Fisher's least significant difference method)

Elongation in 25.4 mm - Method B

Alloy	Count	Mean	Homogeneous Groups
Waspaloy	5	4.42	X
R-41	5	6.64	X
X-750	5	9.28	X
718	5	14.62	X
HAYNES 282	5	16.57	X
HAYNES 263	5	19.80	X
HASTELLOY X	5	51.10	X

95% CL = +/- 2.103 (Fisher's least significant difference method)

Elongation in 38.1 mm - Method B

Alloy	Count	Mean	Homogeneous Groups
Waspaloy	5	3.04	X
R-41	5	5.08	X
X-750	5	8.83	X
718	5	11.70	X
HAYNES 282	5	18.17	X
HAYNES 263	5	22.04	X
HASTELLOY X	5	38.77	X

95% CL = +/- 2.290 (Fisher's least significant difference method)

Elongation in 50.8 mm - Method B

Alloy	Count	Mean	Homogeneous Groups
Waspaloy	5	2.42	X
R-41	5	4.30	X
X-750	5	8.03	X
718	5	11.02	X
HAYNES 263	5	21.08	X
HAYNES 282	5	22.33	X
HASTELLOY X	5	31.40	X

95% CL = +/- 2.520 (Fisher's least significant difference method)

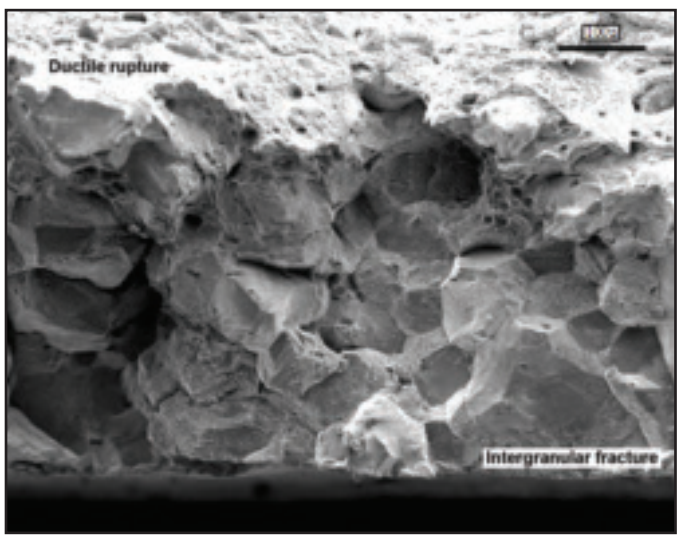

Fig. 5 — Post-CHRT (Method A) fracture surface (282 alloy).

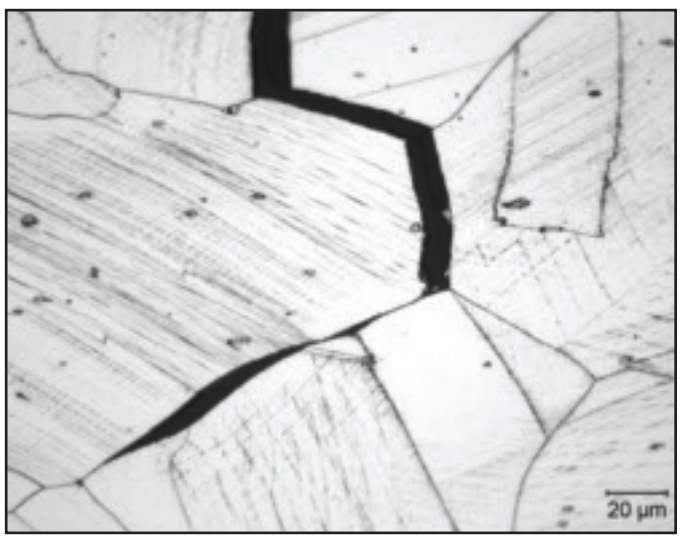

Fig. 6 — Intergranular fracture propagation in post-CHRT (Rene 41).

Preliminary Analysis of Sample Geometry

A simple modification to a standard sheet tension test sample was constructed and subjected to simple mechanical and thermal analyses by finite element analysis (FEA). The modified geometry was constructed by replacing the standard sample's constant gauge width with 101.6-mm (4.0-in.) radii tangent to the original gauge section. A 101.6-mm radius was chosen because, when tangent to the original 12.7-mm (0.5-in.) gauge width at the midspan of the specimen, its intersection with the tensile blank profile produced tabs very similar in size to a standard sheet tension test sample. This modified geometry (and standard geometry for comparison's sake) is depicted in Fig. 1.

The mechanical analysis was solved for the von Mises stress in both sample geometries under constant-stress, linear elastic conditions using FEA. Similarly, the thermal analysis was constructed by applying a constant heat flux to all sample surfaces except at the sample contact surfaces at the Gleeble jaws. Here, a constant temperature boundary condition was applied, approximating the heat sink effect of the water-cooled jaws. The steady-state temperature profile was calculated by FEA. Results for both analyses are shown in Figs. 2 and 3.

These results demonstrated that the modified sample geometry produced a narrower band of peak stress and temperature at the sample midspan relative to the standard geometry, thereby reducing its likelihood of off-center fracture. Based on these results, this study focused on comparing Gleeble-based CHRT test results for seven different Ni-based superalloys using both this modified sample geometry and a modified CHRT thermal cycle as described above.

Experimental Method

Seven different Ni-based superalloys were chosen for evaluation by two differently modified CHRT testing methods. Six of these alloys represented gamma-prime strengthened materials. One solid-solution strengthened alloy was included in this study as a baseline. These alloys and their compositions are listed in Table 1.

Gamma-prime-forming elemental con-

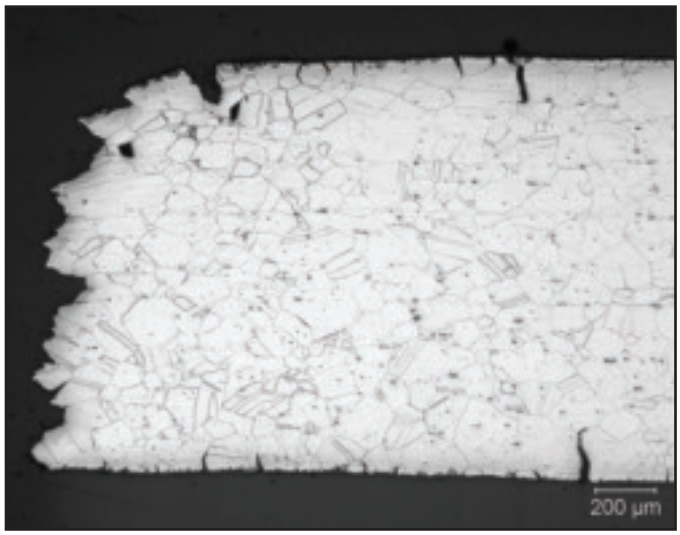

Fig. 7 — Surface-based fracture initiation (282 alloy).

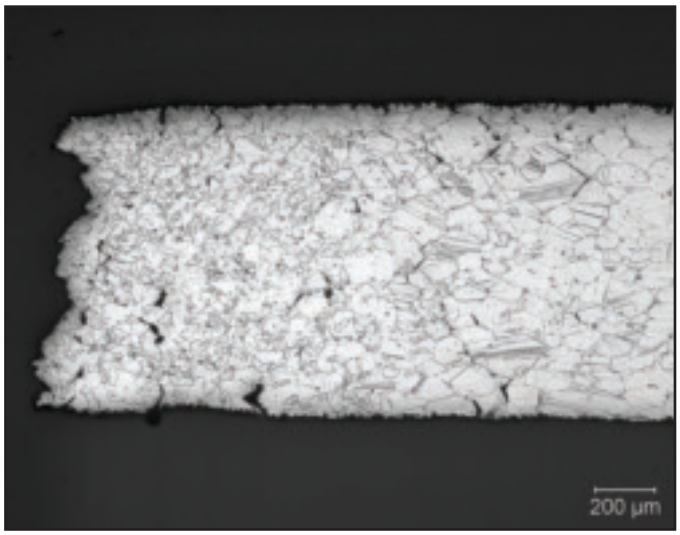

Fig. 8 — Surface/interior-based fracture initiation (263 alloy).

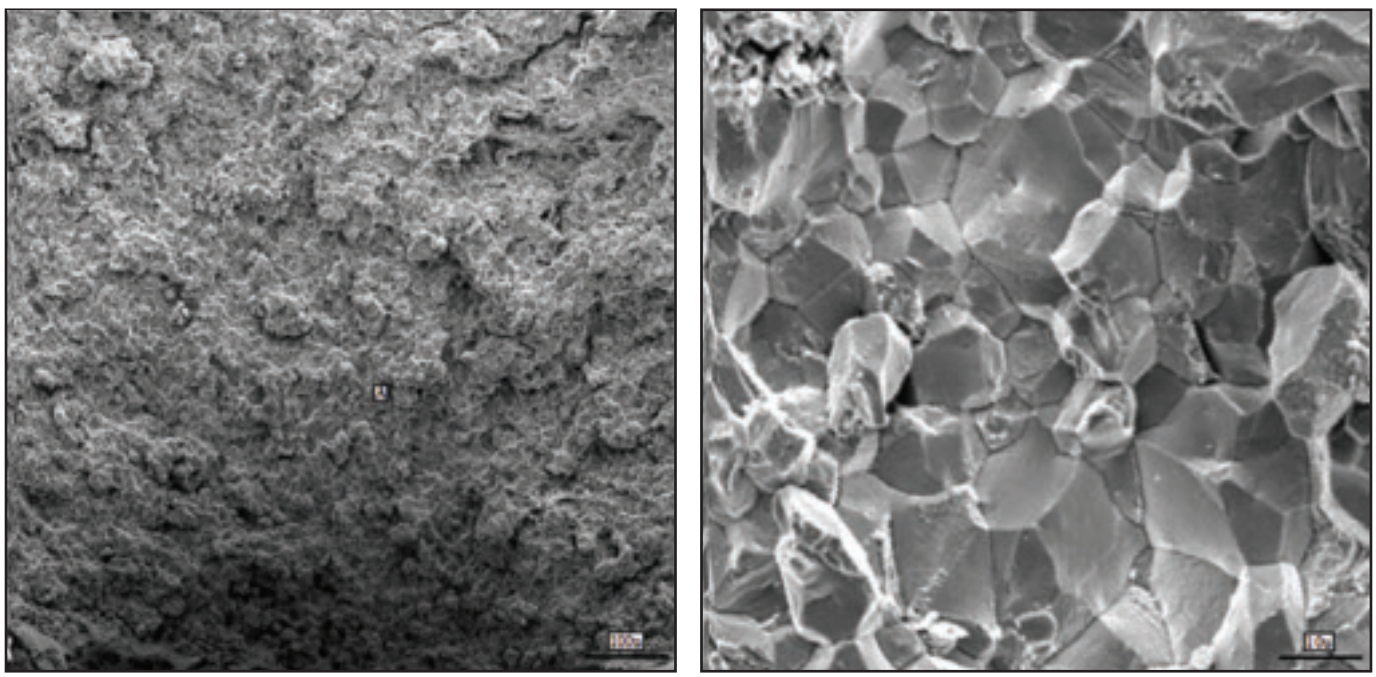

Fig. 9 — Post-CHRT intergranular fracture surface (fine-grain 718 alloy).

tent as a total atomic percent was calculated for each material. Gamma-prime volume fraction was estimated from thermodynamic phase stability calculations performed with Pandat™ software and v7.0 of the Ni-Data database of thermodynamic properties published by Thermotech, Ltd. Calculations were performed at 500°C (931°F) to represent the maximum fraction of gamma-prime that each material could potentially precipitate. Other phases, expected to form at time scales well beyond CHRT test duration were suspended from the calculations.

Modified CHRT Sample Geometry (Method A)

Five samples from each of the alloys under test were cut from mill-annealed sheet material ranging from 1.0 to 2.5 mm (0.040 to 0.100 in.) thick using a mechanical shear. All samples were oriented transverse to the final sheet rolling direction and contained either bright-annealed or annealed and pickled surfaces. Modified CHRT samples were machined from each group according to the dimensions detailed in Fig. 1. Prior to CHRT tests (in the Gleeble), each sample was inscribed with three sets of gauge marks at 25.4, 38.1, and 50.8 mm (1.0, 1.5, and 2.0 in.) separation. A single Type K thermocouple was percussion welded to the center of each sample on one of the broad faces. Samples were mounted between austenitic stainless steel (low thermal conductivity) flat jaws in a PC-controlled Gleeble 1500D. Each sample was heated to 593°C at 56°C/s (1100°F at 100°F/s), then heated to 788°C (1450°F) at 17°C/min (30°F/min), and finally pulled to failure (holding 788°C) at 1.60 mm/min (0.063 in./min). The 788°C (1450°F) test

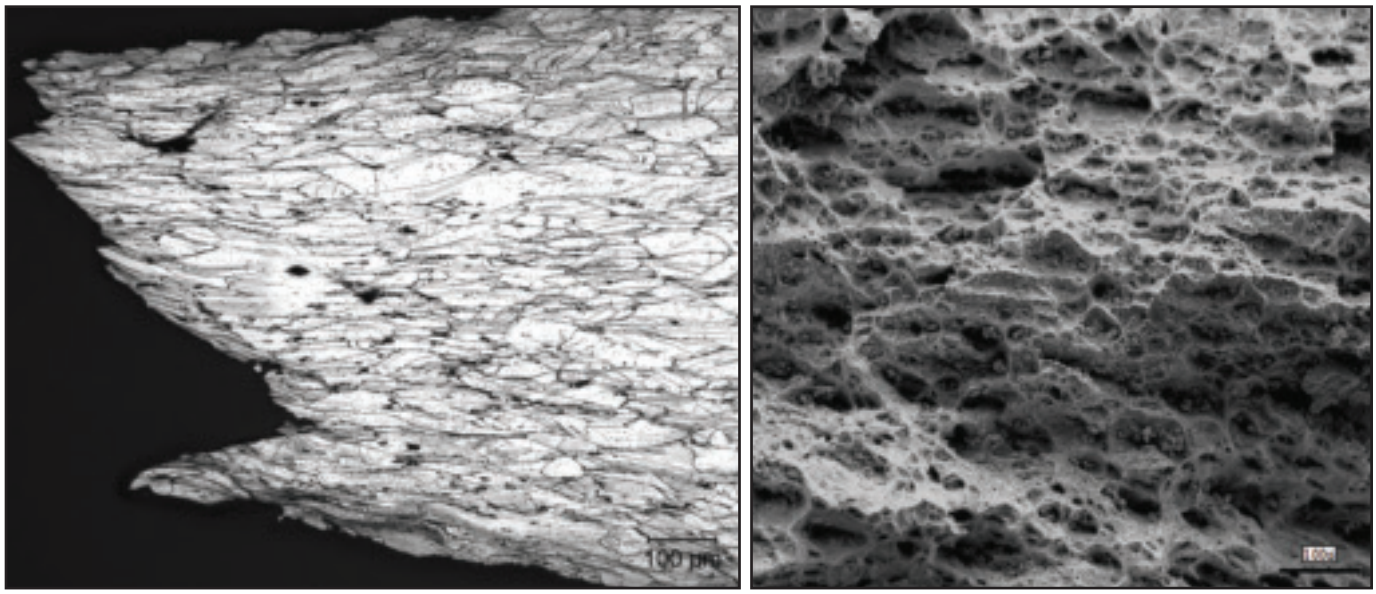

Fig. 10 — Post-CHRT fracture surface in solid solution strengthened (Alloy X).

Table 3 — Pre and Post CHRT Specimen Grain Size

	Method A				Method B			
	Pre-Test Grain Size		Post-Test Grain Size		Pre-Test Grain Size		Post-Test Grain Size	
Alloy	ASTM	microns	ASTM	microns	ASTM	microns	ASTM	microns
Rene 41	4.5	76	4.5	76	4.5	76	1.5	214
Waspaloy	4.5	76	4.5	76	4.5	76	4.5	76
X-750	5.5	53	5.5	53	5.5	53	3.5	107
HAYNES 282	3.5	107	3.5	107	3.5	107	3.5	107
718	7.0	32	7.0	32	7.0	32	3.5	107
HAYNES 263	3.5	107	3.5	107	3.5	107	3.5	107
HASTELLOY X	4.0	90	4.0	90	4.0	90	3.5	107

temperature was chosen as a compromise between CHRT minimum ductility temperatures exhibited by classic gamma-prime-forming (Al/Ti-containing) alloys and Nb-modified alloys (such as 718 and X-750) in Rowe's (Ref. 1) earlier load frame-based work. Postfailure, elongation to break was measured using each of the three sets of gauge marks. These groups of data were compared using standard statistical methods for normally distributed data. Typical fracture surfaces from each material were examined via optical and electron metallography (SEM).

Standard CHRT Sample Geometry (Method B)

Five samples from each of the alloys under test were cut from mill-annealed sheet material ranging from 1.0 to 2.5 mm (0.04 to 0.10 in.) thick using a mechanical shear. All samples were oriented transverse to the final sheet rolling direction. Each group of blanks was aged to peak hardness following the manufacturer's recommended cycle and then machined into standard tension test blanks (lower part of Fig. 1). Sample broad faces were not further machined or pickled. Each sample was prepared and mounted in the Gleeble as described above (Method A). Each sample was heated to 1094°C (2000°F) at 56°C/s (100°F/s), held at 1094°C for 60 s, then allowed to cool (below 300°C) in the Gleeble. Following this in-situ solution anneal, each sample was subjected to the same CHRT thermomechanical cycle described above (Method A). Postfailure, elongation to break was measured using each of the three sets of gauge marks. These groups of data were compared using standard statistical methods for normally distributed data. Typical fracture surfaces from each material were examined via optical and electron metallography (SEM).

Results

Modified CHRT test results (average of five replicate specimens) for test methods A and B and the results of tests for statistically significant differences (Fisher's least significant difference method at a 95% CL) are illustrated in Table 2.

All "Method A" specimens failed at midspan (as expected). Some "Method B" samples failed off-center, but all broke within the 25.4-mm gauge sections. The non-gamma-prime strengthened material (HASTELLOY® X) produced significantly higher elongation to break using both test methods A and B and all three sets of gauge marks. The gamma-prime and gamma-double-prime strengthened alloys yielded elongation results that appeared consistent with estimated volume fraction of the strengthening phase — elongations generally decreased as the volume fraction of strengthening phase increased. The 25.4-mm and 38.1-mm gauge marks yielded apparently good elongation discrimination among alloys, but those differences became less significant when the 50.8-mm gauge marks were used to calculate elongations to break. Method A produced a narrower range of elongations over the range of alloys tested compared to method B, but the amount of sample-to-sample variation was considerably

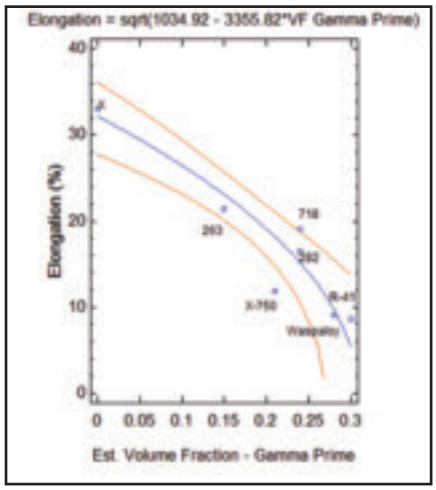

Fig. 11 — CHRT elongation vs. estimated volume fraction gamma-prime/gamma double prime.

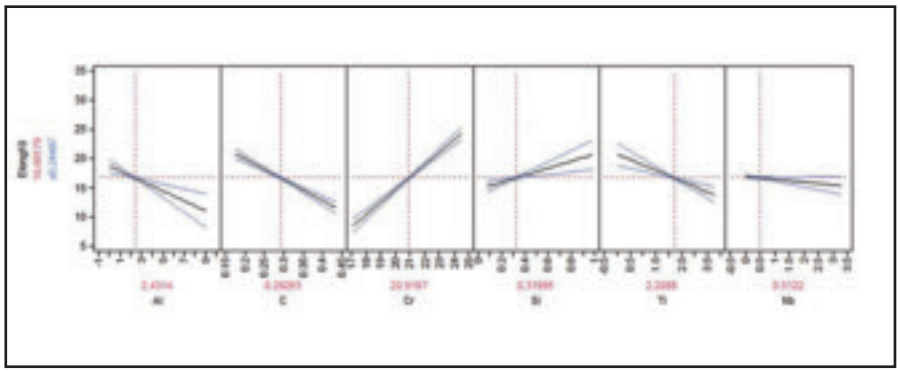

Fig. 12 — Individual element effects on CHRT elongation.

lower. Method A vs. Method B comparative results for 25.4-mm gauge length specimens are depicted in Fig. 4. Pretest and posttest grain sizes (both Method A and Method B) are listed in Table 3.

Specimen grain size did not change during CHRT using Method A. X-750 and 718 alloys exhibited significant grain size increases during CHRT, using Method B. The only apparent difference in ranking between the two methods was that 718 and HAYNES® 282® swapped positions, with 718 yielding higher elongation using Method A. Since mill-annealed material was used in these tests, the 718 specimens contained a finer grain size (ASTM 6-7) than the other groups (typically ASTM 3 - 4). Alloy 718 is intentionally produced this way. This grain size was retained in Method A, but coarsened in Method B by the pre-CHRT cycle, 2000°F simulated solution anneal. Thus, 718 should have performed better (and, in fact, did exhibit higher elongation) in Method A. This increase in Alloy 718 hot ductility with decreasing grain size was also observed by Norton and Lippold (Ref. 6) during development of a different Gleeble-based PWHT cracking sensitivity test. In both methods evaluated, here, the overall ranking of materials was virtually identical to that produced by Rowe's (Ref. 1) earlier load-frame-based CHRT tests. The only exception was Alloy X-750, which produced lower elongations than Rowe measured. This result was evident in both "Method A" and "Method B" results and is probably characteristic of the particular heat of X-750 tested, rather than a product of the test methods, themselves.

Fractography and Metallography: Typical Examples

The post-CHRT fracture surfaces of all gamma-prime/gamma-double-prime-strengthened materials, whether tested by Method A or Method B, exhibited areas of both ductile rupture and intergranular fracture. A typical example (HAYNES® 282®) is depicted in Fig. 5.

A typical example of crack propagation by intergranular fracture in Rene® 41 is illustrated in Fig. 6.

In the higher-strength materials, fracture appeared to nucleate at the specimen surface and propagate inward (Fig. 7), while the lower-strength materials exhibited signs of crack nucleation at both their surfaces and interiors — Fig. 8.

Intergranular fracture was readily evident even at very fine grain sizes (718 alloy) as shown in Fig. 9. This suggests the intergranular fracture mode observed in these tests was enhanced by gamma-prime/gamma-double-prime precipitation rather than being produced by environmentally driven grain boundary embrittlement. In contrast, HASTELLOY® X, a solid-solution strengthened alloy, exhibited only ductile rupture in post-CHRT fracture surfaces — Fig. 10.

Correlation of CHRT Results with Alloy Composition

Previous studies examined the relationship of alloy composition and estimated gamma-prime/gamma-double-prime phase fraction with CHRT minimum elongation over the 760°–871°C (1400°–1600°F) test temperature range. In the interest of standardizing and simplifying the CHRT, this study confined testing to one temperature 788°C (1450°F), only. Because of this change and modification of the specimen geometry, the CHRT elongation vs. estimated gamma-prime/gamma-double-prime phase fraction was reexamined for Method A, 25.4-mm gauge length results. Best fit results are shown in Fig. 11.

This analysis yielded a correlation coefficient (adjusted for degrees of freedom) of 90%, indicating a statistically strong correlation between CHRT elongation and an alloy's capability of precipitating gamma-prime/gamma-double-prime.

Based on these results, an attempt was made to fit CHRT response (Method A, 25.4-mm gauge length) to alloy composition expressed in at.-%. A simple (first order) linear model was constructed using at.-% Al, Ti, Nb, C, Cr, and Si as the inputs, and CHRT elongation as the response. 788°C CHRT elongations for 282 alloy (16.4%) and HAYNES® 214® (tested outside this study – 12.1% elongation) were reserved as validation data sets for the model. A least squares model fit produced the following predictor equation:

$$\text{CHRT elongation} = -15.374 - 0.305\,(\%\text{Al}) - 35.486\,(\%\text{C}) + 2.037\,(\%\text{Cr}) + 10.626\,(\%\text{Si}) - 1.277\,(\%\text{Ti})$$

with an $R^2 > 99\%$.

Calculated CHRT elongations for the 282 and 214 alloys were 15.6 and 13.8%, respectively. Refitting the linear model, with all data included, yielded a slightly modified elongation predictor

$$\text{CHRT elongation} = -12.626 - 0.839\,(\%\text{Al}) - 35.784\,(\%\text{C}) + 2.111\,(\%\text{Cr}) + 6.091\,(\%\text{Si}) - 1.843\,(\%\text{Ti}) - 0.473\,(\%\text{Nb})$$

Individual element effects are illustrated graphically in Fig. 12. Each panel depicts the linearly decoupled CHRT elongation response to changes in the levels of individual alloying elements — the blue lines represent the 95% confidence limits and the black lines represent the mean response for each alloying element, respectively.

This approach was not intended to address how each constituent affects alloy CHRT behavior from a detailed physical metallurgy perspective, but rather to suggest general trends and possibilities for future work. As expected, gamma-prime/gamma-double-prime forming elements reduced CHRT elongation, with the effects of Al and Ti approximately

twice as strong as Nb's effect. Cr and Si appeared to mitigate the effects of these gamma-prime formers, while C appeared to exacerbate their effects. This model appears valid within the envelope of data generated by this testing program, but should only be extrapolated with extreme caution.

Discussion

Strain-age (SA) cracking requires the simultaneous presence of (tensile) stresses and a ductility decrease (produced by precipitation of gamma-prime/gamma-double-prime) in the material under test. Significant precipitation hardening must occur before significant stress relaxation takes place by shear, creep, or any other thermally activated process. An ideal choice of heat-up rate and test temperature ensures that these conditions are met. The six age-hardenable alloys evaluated here typically are held for 4 to 6 hours between 720° and 900°C to produce peak strength (although some employ more complex, multistep heat treatments). Across the spectrum of Ni-based alloys examined, the mean (primary) age hardening temperature was ~790°C, hence its choice as a "universal" test temperature. The choice of test heating rate (17°C/min) was largely based on two considerations:

1) The ability to directly compare Gleeble-based CHRT results with prior load-frame based CHRT results reported by Rowe (Ref. 1) and Prager et. al. (Ref. 2) (who also reported that this heating rate was approximately representative of those likely to be experienced by fabricated parts undergoing postweld solution annealing).

2) The necessity that all alloys under test undergo partial age hardening during CHRT. This condition essentially combines precipitation reaction kinetics with estimated gamma-prime/gamma-double-prime volume fraction as a single, integrated material property that determines that material's propensity toward fracture during PWHT.

The choice of specimen extension rate (1.60 mm/min) was also based upon these two considerations.

Based on these requirements, a single heating rate (17°C/min) into the gamma-prime and/or gamma-double-prime precipitation temperature range (593°–788°C), coupled with a specimen extension rate of 1.6 mm/min was chosen for this series of experiments. The Gleeble's ability to reproduce a wide range of heat-up and strain (extension) rates, however, provides an opportunity to optimize CHRT test conditions with respect to heating rate, test temperature and strain rate. Such studies are currently in progress.

In this study, Gleeble-based CHRT results (using the test condition described above) provided clear discrimination among the SA cracking behaviors of six different age-hardenable Ni-based superalloys, using both modified (Method A) and standard (Method B) specimen geometries. While both methods are capable, in a statistical sense, Method A appeared to offer several advantages over Method B as noted below.

1) The modified stress distribution ensured that failure occurred at (or very near) midspan, within the prescribed (25.4-mm) gauge length.

2) Consequently, failure always occurred at or near the site of peak specimen temperature, ensuring that failure occurred in a region where gamma-prime/gamma-double-prime precipitation was ongoing, thus adequately simulating SA cracking field failures.

3) Since no prior aging and in-situ solution annealing thermal cycles were needed, grain size modification did not occur prior to or during CHRT and add additional variation to CHRT results. Changes in alloy grain size can (and usually will) produce corresponding changes in material creep response, with larger grain sizes (less grain boundary area available to accommodate sliding and related short-circuit diffusion processes) favoring decreased creep rates. Since creep-related processes are expected to provide some reduction of stresses available to promote SA cracking, it follows that uncontrolled grain size changes during CHRT could skew results in unexpected ways.

4) Similarly, no significant pre-CHRT surface oxidation was present in Method A. No decrease in CHRT elongation to break would have resulted from oxygen penetration into grain boundaries (Ref. 2). Since no specimen surface preparation or cleaning was needed, no surface residual stresses would have been present that could have added additional variation to test results.

5) Experimental results strongly suggested that Method A yielded lower variation in test results (elongation to break), compared to Method B (Table 2).

6) In those cases where SA cracking behavior during PWHT after repair welding is to be assessed, Method A's specimen geometry could be combined with Method B's thermomechanical cycle to offer potentially lower variation in material response.

Conclusions

1. Gleeble-based CHRT methods reproduce the results of earlier load-frame based CHRT method.

2. Simple specimen geometry modifications can be used to reduce test variation and ensure that samples fail at the center of their gauge section.

3. A single test temperature (788°C), heating rate (17°C/min), and specimen extension rate (1.6 mm/min) provide adequate discrimination of strain-age cracking susceptibility and considerably simplify implementation of Gleeble-based CHRT methods.

4. Gleeble-based CHRT can be performed with mill-annealed sheet stock.

5. Relatively small differences in SA cracking susceptibility can be distinguished by this test, using a radiused specimen gauge section and 25.4-mm gauge length. Less than 1.0% elongation difference is significant at the 95% CL.

6. Post-CHRT (Gleeble-based) fracture surfaces exhibited areas of intergranular fracture in all gamma-prime/gamma-double-prime containing alloys. HASTELLOY® X, which precipitates no gamma-prime or gamma-double-prime, exhibited only ductile rupture.

7. Alloy CHRT (Gleeble-based) behavior appears describable by a first-order linear function of composition.

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