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Self-restrained testing for residual stress driven cracking in nickel-based alloys

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ABSTRACT

The use of nickel-based superalloys in the hottest sections of coal-fired power plants would enable efficiency gains and emissions reductions. In order to study the effect of residual stress on cracking susceptibility in some candidate superalloys, compact tension samples were pre-strained in compression beyond yield and tested without additional external loading. The testing parameters pre-strain and temperature were investigated and were found to have a minor effect on cracking relative to microstructure. The microstructural factors which had the strongest effect on crack length were grain size and degree of intragranular gamma prime precipitation.

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1. Introduction

In order to increase efficiency and reduce emissions, coal-fired power plant steam temperatures and pressures are continuously increased. The upper temperature limits of power plant steels are being approached and progress in power plant design will require nickel-based superalloys for some of the hottest components [1]. There are recent concerns regarding the effect of residual stress on the creep performance of these materials and the possibility of premature cracking driven by residual stress [2].

Residual stresses, which can arise due to welding or forming processes, can influence cracking in Nickel-based alloys. These stresses can be greater than the yield stress in situations of large enough constraint, and depending on their orientation, directly add to operating stresses. Residual stresses are most commonly relaxed with heat treatments, during which thermal energy allows local plastic flow to reduce residual stress levels to the yield stress of the material at the heat treatment temperature. Additional time at elevated temperatures will continue to relieve residual stress by creep.

Residual stress can cause or exacerbate cracking in steels and nickel-based alloys alike. The phenomenon may be called stress relief cracking, stress relaxation cracking, or strain-age cracking depending on the time and cause of cracking. Relief and strain-age cracking occur as stress is relieved on heating in steels and superalloys, respectively. Relaxation cracking occurs after the initial stress relief when stresses relax by creep. This occurs after some period of time in service. The common mechanism involves a strong precipitation reaction in the stress relief temperature range which strengthens grain interiors relative to boundaries, concentrating stresses at the boundaries which may crack. The mechanism responsible for stress relief cracking, strain age cracking, and stress relaxation cracking always produces intergranular cracking [3].

Many engineering materials are used in elevated temperature service where creep is the limiting factor. Creep occurs through a wide range of mechanisms, many of which are controlled by dislocation motion. One important obstacle to dislocation motion is the presence of other dislocations (other primary obstacles are precipitates and grain boundaries) and dislocations generated by plastic deformation can have a profound effect on creep performance, as reviewed by Li et al. [4]. This review provides the following main points: pre-strain can affect primary and secondary creep rate depending on the nature of dislocations produced (stable substructure vs mobile), there can be an increase in creep resistance or enhancement in creep rate depending on the direction of pre-straining (and direction of resultant residual stress), pre-strain can affect tertiary creep rate and ultimate creep strain depending on the damage created (void nucleation and distribution at grain or particle interfaces), and Monkman-Grant constant

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and creep ductility are typically negatively affected by any pre-straining. An aspect of pre-straining not covered in Li's review but pertinent to the present discussion is the possibility of increasing precipitation rate.

Testing for the effects of residual stresses on creep cracking has been investigated through many approaches such as a bolt-loaded C-Ring [5], thermal-expansion loaded three-point bend [6], thick section three-point bend [7,8], and several Gleeble-based methods [9–13]. One method which has received recent attention was developed by Turski in 2004 [14]. Compact tension samples are pre-strained in compression and then placed in a furnace with no additional loading. Cracking due to stress relaxation ensues. Studies involving this method [15–17] and other studies which use different methods to investigate creep cracking in compact tension samples [18–20] typically focus on modeling the fracture mechanics aspect of the problem with very little attention to metallurgy, even when a weld heat-affected zone is introduced [15,21]. These modeling efforts have been successful, which encourages further study of this method. Success has been found using both a ductility exhaustion approach applied by Turski, Bouchard, Steuwer, and Withers [22] and a creep fracture mechanics parameter approach (C^*) applied by Webster, Davies, and Nikbin [17]. Another appeal of Turski's approach is the simplicity and therefore accessibility by many researchers. The aim of the present work is to extend this approach to Nickel-based alloys and examine the resultant cracking from a metallurgical perspective.

1.1. Procedure

Samples were pre-strained in compression, leaving a yield-level residual stress. Plastic compressive displacement was measured after spring-back and held constant among all of the alloys, resulting in residual stresses different in magnitude but similar with respect to the yield point. Measurements were made at the step in the sample notch, as shown in Fig. 1.

In order to investigate the effects of microstructure, degree of pre-strain, and temperature on stress relaxation cracking, combinations of the following samples and conditions were tested: Alloys 740H,[2] 617, 282, and 230, with plastic pre-strain displacements of 1, 2, and 4 mm were heated to 760 °C (1400 °F) and 620 °C (1150 °F) for 720–1600 h. Table 1 gives the nominal compositions of the alloys used in this study and Table 2 shows the matrix of samples to be discussed here.

Alloys 740H and 282 in the aged condition were given an aging treatment at 800 °C for 4 h. Alloys 740H and 230 were available after a very long age at 800 °C for 10,000 h. The annealed condition for alloys 282 and 617 refers to a mill anneal. The purpose of testing such a wide variety of microstructures was to evaluate this test's sensitivity to microstructure.

Heating samples bearing such large residual stresses produces cracking whenever the microstructure is unable to accommodate the necessary stress relaxation. All heat treatments were done in a large box furnace. Samples were placed in the furnace while it was at room temperature so heating rates were rather slow.

2. Results and discussion

2.1. Effect of initial microstructure

The initial microstructures of alloys 740H, 617, and 230 are

[2] In the alloy name 740*H*, *H* is an arbitrary designator used to distinguish the second generation of this alloy and does not indicate a specific alloy element level such as Carbon, as is the case in 316*H*.

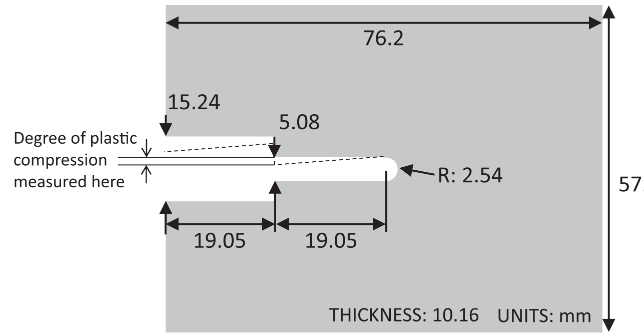

Fig. 1. Sample schematic showing pre-strain measurement location.

Table 1
Nominal compositions of alloys in this study.

Element	617 [23]	740H [24]	282 [25]	230 [26]
Ni	Balance	Balance	Balance	Balance
Cr	22	25	20	22
Co	12	20	10	5
Mo	9	0.5	8.5	2
Fe	3	0.7	1.5	3
Mn	1	0.3	0.3	0.5
Al	1.2	1.3	1.5	0.3
Ti	–	1.5	2.1	–
Nb	–	1.5	–	W = 14
Si	0.5	0.15	0.15	0.4
C	0.15	0.03	0.06	0.1
B	–	0.001	0.005	0.015

Table 2
Test matrix.

Temperature	Pre-strain	Samples			
760 °C (1400 °F)	4 mm	740H aged	740H 10k h age	230 10k h age	617 annealed
760 °C (1400 °F)	1, 2, 4, 0 - > 1	740H aged			
620 °C (1150 °F)	1 mm	740H aged	282 aged	282 annealed	617 annealed

shown in Figs. 2 and 3. These samples were pre-strained 4 mm before testing at 760 °C (1400 °F) for 1600 h. Fig. 2 shows grain size variations in optical micrographs. Large grains are normally desired for creep resistance but since relaxation cracking is a boundary phenomenon and large-grained microstructures have less boundary area than smaller grains in a given volume, the larger grains are expected to increase relaxation cracking susceptibility by concentrating stress relaxation at fewer boundaries. This concentration at boundaries is also expected to occur during pre-straining, with the most boundary damage occurring at the boundaries in large-grained material. Small grains in Alloy 230 are likely due to boundary pinning by extremely stable W, Mo carbides. In Alloy 740H aged for 10,000 h, the small grains are more likely due to recrystallization, which preferentially takes place at prior high angle grain boundaries, resulting in the non-uniform distribution of smaller grains seen in the micrograph. Fig. 3 shows the γ' precipitate size and distribution found in each sample before testing. Intragranular precipitation will strengthen grain interiors and force any stress relaxation (or stress accumulation during pre-straining) to be accommodated by grain boundaries. The sample with the largest grain size and most intragranular precipitation is expected to be the most susceptible to relaxation cracking, Alloy 740H in this case.

After pre-straining to a permanent deflection of 4 mm,

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