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ABSTRACT

A number of materials issues related to the design of piping and support components in high-temperature fluidized bed combustor systems were examined. These issues included the availability of long-time design data on structural materials, the general character of the creep and stress rupture behavior, the performance of weldments, and the assessment of damage accumulation. Emphasis was placed on alloy 800H, but several other alloys were briefly examined for use at temperatures above 816°C (1500°F). It was concluded that the character of the creep curve ranged significantly with chemistry, processing variables, and environment, and that the specification of design allowable stresses and life estimation techniques must be approached with caution for service above 816°C (1500°F).

INTRODUCTION

The development of combined-cycle power technology is being aggressively pursued throughout the world, and in the U. S. several systems of interest have been described by Bajura and Webb (1991). A schematic diagram of a second-generation pressurized fluidized bed combustor (PFBC) is shown in Fig. 1. Here, coal is fed to a carbonizer which produces fuel gas, and char from the carbonizer is fed to a PFBC where complete combustion occurs. A calcium-based sorbent is used in both the carbonizer and PFBC to reduce sulfur dioxide emissions by 90 to 95%. Both gas streams are filtered to remove ash particles and fed to a topping combustor that delivers gas to the turbine at temperatures exceeding 1150°C (2100°F) (Heat Engineering, 1986). A heat recovery system on the turbine exhaust raises steam for additional power.

Fig. 1. Schematic drawing of a second-generation combined-cycle power plant.

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Gasifiers, carbonizers, and fluidized bed combustors are being designed for combined-cycle applications, and decisions are being made regarding the materials of construction for vessels, piping, and hardware for hot-gas cleanup systems. Issues that are being addressed also include design configurations of pressure vessels and piping to accommodate thermal stresses that result from cyclic operation to very high temperatures. This paper considers some of the materials issues that must be resolved for safe and reliable long-time operation of combined-cycle plants. The focus is on the PFBCs, but many of the issues are important to combined-cycle systems that incorporate gasifiers and carbonizers. The design and construction methods used in second-generation combined-cycle systems are similar to those incorporated into current PFBC, and many of these are outlined in a recent publication (ASME, 1991). Current designs for combined-cycle plants call for refractory-insulated vessels and piping to permit the use of structural steel operating at near ambient temperature, and provisions are made to accommodate the differential thermal expansion between the structural steel, the insulation, and the liner. When it is practical, components are water cooled to reduce temperatures and thereby eliminate hot corrosion and severe thermal stresses. Often, it is difficult to avoid the use of high-strength, corrosion resistant alloys in critical structural components. This paper reviews the available pressure vessel materials, reviews alternate structural materials that may be candidates for pressure boundary and structural applications, and summarizes some material performance considerations that bear on the issue of reliability.

**ALLOYS FOR HIGH-TEMPERATURE PRESSURE BOUNDARY APPLICATION**

Pressure-bearing components are required to meet construction codes such as the American Society of Mechanical Engineers (ASME) B31.1 piping code or the ASME Boiler and Pressure Vessel (BPV) Code. If the piping is not insulated from the process stream temperature, then the materials selection is limited for service above 816°C (1500°F). In Fig. 2, a comparison is made of the allowable stresses for alloys currently approved for construction under the rules of Sect. VIII, Div. 1 of the ASME BPV Code. The stress tables are provided in various Code Cases or in Sect. II, Part D. Alloys include two cobalt-bearing alloys (alloy 617 and alloy 556), two high-strength nickel base alloys (alloy 230 and alloy 625), and two grades of alloy 800H (alloy 810 and alloy 811). Also, in the ASME BPV Code Sect. I, one stainless steel (253MA®) is approved for use to 900°C (1650°F). Only two materials (alloy 617 and alloy 810) have allowable stresses to 982°C (1800°F). The cobalt-bearing alloys and the nickel-base alloys have the highest strengths, alloy 810 and alloy 811 are intermediate in strength, and 253MA stainless steel is the weakest.

The allowable stresses plotted in Fig. 2 are derived from one of the strength criteria identified in appropriate construction code, and in most cases the stresses at 816°C (1500°F) and above are based on 67% of the average stress to produce rupture in 100,000 h at the design temperature. Concerns exist regarding the adequacy of this criterion for very long time service, and these concerns are described later in the paper. The selection of an alloy from available group, of course, depends on such issues as cost, fabricability, corrosion resistance, and strength requirements. A number of reviews to guide in the materials selection are available, and a recent paper by Natesan and Podolski (1991) provides information on corrosion behavior of 29 alloys in FBC cogeneration environments. The Code-approved alloys included in their evaluations are alloy 800H, alloy 556, and 253MA stainless steel. These and other alloys were exposed in laboratory tests and in a variety of locations in FBC systems. Natesan and Podolski (1991) showed that the relative ranking of the materials was highly variable and changed greatly with temperature and gas composition.
ALTERNATE HIGH-TEMPERATURE STRUCTURAL ALLOYS

The alloys that are available for structural applications, apart from pressure boundary applications, that are included in the corrosion evaluations of Natesan and Podolski (1991) range from cobalt-based alloys to nickel-based alloys and to cast and wrought stainless steels. The selection is too great to be covered here in any detail, so only representative alloys are discussed. For oxidizing environments containing sulfur, cobalt-based alloys are found to perform relatively well. For the oxidizing environments expected for the PFBC, the high nickel-chromium-iron alloys serve well. If strength is not of great concern, less expensive materials such as types 304 and 310 stainless steels are adequate. Although the extension of type 304 stainless steel for pressure containment above 816°C (1500°F) was judged to be impractical by Bynum, et al. (1992) because of poor strength, interest remains in extending the use of type 310 stainless steel to 871°C (1600°F) (Prager, 1992). Type 310 stainless steel, for example, is used as liner piping in the Tidd plant PFBC. Newer alloys with improved corrosion resistance have also been proposed for service above 816°C (1500°F) and could serve a similar function. These include a silicon-aluminum modified 18-8 stainless steel identified as RA85H® (Kelly, 1989), a high-strength Fe-Ni-Cr-Nb-N alloy identified as HR-120® (Myers, 1991 and Ernst and Lai, 1992) and a high-strength cobalt-bearing alloy identified as HR-160® (Haynes, 1990). Experience is being gained in the use of these materials, and the data needed to produce allowable design stresses are being accumulated. Figs. 3a and 3b compare typical alloys on the basis of creep strength at 10,000 h reported by alloys vendors. Fig. 3a includes the alloys approved for pressure boundary applications in the BPV Code, while Fig. 3b compares the alternate structural alloys. It may be seen that a wide range of strength levels exists, and at 982°C (1800°F) the difference in strength between the weakest material (253MA stainless steel) and the strongest (alloy 617) is a factor of two. The developmental alloys HR-120 (essentially a nickel-based alloy) and HR-160 (containing cobalt) have excellent strengths at 982°C (1800°F), while both RA333® and RA85H have much improved strength relative to type 310 stainless steel. RA333 and type 310 stainless steel have been used for components in the PFBC hot-gas cleanup vessel at Grimethrope pilot plant (Stringer, et al., 1991) and are used in the hot-gas cleanup vessel at the Tidd plant, where strength, fabricability, and corrosion resistance are needed (Manjoine and Filstrup, 1992). Much remains to be learned about fabrication technology before the full potential of the newer alloys can be exploited, and issues related to fabrication technology are discussed later.

DESIGN CRITERIA ISSUES

Most of the experience in the design and operation of components for long-time service above 816°C (1500°F) has been gained in the petroleum and petrochemical industries. Frequent inspection, repair, and replacement of components is common, and special procedures are used to estimate remaining life. The American Petroleum Institute (API) Recommended Practice (RP) 530 is an example of the design philosophy for heater tubes. Similar to the ASME piping codes B31.1 and B31.3 and BPV Code, specific alloys are approved, and temperature limits are set for each. In contrast to the ASME piping codes and the BPV Sect. I and Sect. VIII, however, data are provided in RP-530 that identify the stresses corresponding to design lives of 20, 40, 60, and 100 thousand hours. Damage and residual life assessments are commonly performed on serviced components using RP-530. ASME Sect. I and Sect VIII only address new construction, one allowable stress is provided for each material as a function of temperature, and there is no time limit. Also, RP-530
provides information on corrosion allowances and guidelines for accommodating creep-fatigue, whereas the ASME Sect. I and Sect. VIII codes provide no information on fatigue and creep-fatigue for service above 427°C (800°F), and rules for expansion joints, needed in PFBC piping, have only recently been introduced into the ASME BPV code (Becht, 1989).

The complexity of establishing rules for construction at very high temperatures is discussed in connection with the ASME Sect. III Code Case N-47 by Corum and Blass (1991). Here, alloy 617 is being considered for use to 982°C (1800°F) in modular high-temperature gas-cooled reactors. Lacking long-time experience, the draft code case for alloy 617 requires design for finite life, consideration of strain limits, consideration of creep-fatigue damage interaction, and the construction of isochronous stress versus strain curves for use in analysis for times extending to 100,000 h. Data for most structural materials of interest to second-generation combined-cycle systems are inadequate to develop a code case comparable to N-47, and, even if a comparable design methodology could be developed, the expense of using such a procedure is prohibitive for many applications. Nevertheless, awareness of the design margin that comes from a design-by-analysis code case is needed for reliable design of second-generation combined-cycle structural components.

The design margin provided by the existing ASME criteria is estimated to be at least a factor of 20 in life at temperatures to 816°C (1500°F). The question that must be addressed in the use of existing alloys or the approval of new alloys concerns the adequacy of the current ASME code criteria to assure safety margins of at least 20 for temperatures above 816°C (1500°F). The criteria for setting allowable stresses are identified in Appendix 1 and Appendix 2 of ASME Sect. VIII, and Sect. II, Part D. In the creep range, the allowable stresses are based on the lowest of the following: 100% of the average stress to produce a creep rate of 0.01%/1000 h; 67% of the average stress to cause rupture at the end of 100,000 h; and 80% of the minimum stress to cause rupture at the end of 100,000 h. Usually, the average rupture strength controls the allowable stress at high temperature, and the breaks in the curves plotted in Fig. 2 correspond to the change in the controlling criterion from tensile (time-independent) to stress rupture (time-dependent). In some materials, creep or creep rate is the time-dependent criterion, but such information is not easily obtained. Generally, fewer creep data and creep rate data are available than rupture data. Also, creep data tend to show much more variability than rupture data, hence the analysis of creep data is generally more difficult than analysis of rupture data.

If one assumes that a simple power law relates life, \( t_r \), to stress, \( \sigma \), at any given temperature, (T), then:

\[
t_r = A \sigma^{-n}
\]

where \( A \) and \( n \) are temperature dependent constants. Generally, \( n \) tends to decrease with temperature, being greater than 10 at very low temperature and less than 2 at very high temperature. Typically, \( n \) is in the range 4 to 8. Assuming that the governing stress criterion is 67% of the average strength at 100,000 h \( (\sigma_0) \) and calling this stress \( \sigma_0 \), one can substitute into Equation (1) and obtain:

\[
t_r/\sigma_0 = (\sigma_0/\sigma)^n = (1.5)^n
\]

Thus, the design margin given by the ratio \( t_r/\sigma_0 \) will vary from more than 25 when \( n \) is 8 to 5 when \( n \) is 4. The actual scatter in the rupture life becomes a very important consideration at temperatures where \( n \) is small because of the reduced design margin, yet design codes do not provide much information regarding the scatter in the data. Some information may be gleaned from the RP-530 document which supplies values for \( n \) over a useful temperature range as well as a Larson-Miller parametric curve for average rupture strength and minimum rupture strength. This information is provided in Fig. 4 for alloy 800H. Here, the Larson-Miller Parameter (LMP) is given by:

\[
LMP = (T + 273)(C + \log_{10} t_r) \times 10^{-3}
\]
Fig. 5. Comparison of log stress versus Larson-Miller parameter for alloy 800H data with the API RP-530 curves and the ASME Code allowables.

where T is temperature in Celsius and \( t_r \) is time in h, and the Larson-Miller constant, \( C \), is 15. Included in the figure are allowable stresses for alloy 800H from ASME Sect. VIII. The allowable stresses for this material appear to be less than 67% of the average strength, and the margin on life based on the design stress and the average life at the design stress is approximately a factor of 12. The margin based on minimum life is approximately a factor of 2.5. These margins are smaller than observed at temperatures below 816°C (1500°F). Trends are shown more clearly by an examination of some of the stress-rupture data for alloy 800H relative to the design curves. In Fig. 5 the RP-530 curves and ASME allowable stress trend curves have been redrawn, and data from U.S. and Japanese (NRIM, 1978) sources have been plotted. Examination of the data relative to the curves reveal that the RP-530 curve represents a reasonable minimum, but the data tend to fall below the average curve as the stress decreases and the LMP increases. The smaller margins on life that appear to exist at higher temperatures must cover the effects of many uncertainties that relate to both design and materials related factors. The margins are adequate for alloy 800H, but experience is lacking in the application of new materials in new technology areas. Concerns regarding the effect of materials behavior on design margins for new alloys include the influence of composition, grain size, cold work, environment, and weldments. Examples of these effects are provided in the next section, using alloy 800H as a reference.

Fig. 6. Shape of the creep curve for alloy 800H reported by Nickel, et al. (1986) at 950°C and 30 MPa.

**FACTORS INFLUENCING CREEP, STRESS RUPTURE, AND FATIGUE**

Alloy 800H often exhibits a creep curve with a large tertiary component at temperatures below 816°C (1500°F) (Booker, 1983). Because of concerns regarding the geometric stability of components fabricated from materials with low tertiary strain thresholds, the ASME CC N-47 includes the time to tertiary creep and strain limits into the criteria for setting allowable stresses in the creep regime. A similar concern exists above 816°C (1500°F), but it must be recognized that materials with low tertiary strain limits have been successfully used for many years at these temperatures. Depending on the composition, processing variables, degree of cold work, grain size, and environment, the tertiary creep characteristics of alloy 800H vary significantly at temperatures above 816°C (1500°F). Nickel, et al. (1986) find that tube products at 950°C (1740°F) and 30 MPa exhibit tertiary creep starting at strains in the range of 1 to 10%, while bar products exhibit tertiary creep from the start of the test. Typical trends that they report are illustrated by the creep curves in Fig. 6. The kinds of differences shown in Fig. 6 are accommodated within the scatter of the stress rupture scatterband but could give rise to difficulties in defining design criteria based on creep rate or creep strain. Cook and Sutton (1978) observe a sigmoidal shaped creep curve for alloy 800H tested in air at 900 and 950°C (1650 and 1742°F). Tertiary creep appears very early in the air test but the creep rate then decelerates to a near constant rate and eventually decreases again when strains exceeded 10%. The trend for one test at 900°C (1650°F) is sketched in Fig. 7. Tests in helium reveal only tertiary character, as shown in Fig. 7, but the
Creep rate is lower in helium than in air. Cook and Sutton attribute the sigmoidal creep behavior in air to a strengthening effect by internal oxidation. Since the design curves for high temperature are based on air testing, some of the design margin may be lost in reducing and inert environments. Environments within the bed of PFBCs are highly variable and difficult to predict. However, compositions are generally defined for the gas stream passing through the piping and gas cleanup vessels. Page et al. (1984) describe the creep of alloy 800H in air and a sulfur-bearing "complex multioxidant gas" (CMG) at 815°C. The equilibrium partial pressures of sulfur and oxygen contents at 815°C and 6.9 MPa are reported to be 1.29x10^-8 and 3.32x10^-19, respectively. Page et al. (1984) find the creep curve to be relatively stable in air but to have a tertiary character in the CMG, as indicated by the sketch in Fig. 8. Creep in CMG is more rapid than creep in air, and the rupture strength is less in CMG than in air, apparently eroding away most of the design margin. However, as stresses are lowered, the rupture lives in CMG tend toward lives observed in air at comparable stresses. Studies by Smolik and Flinn (1985) on alloy 800H indicate that the rupture lives of pressurized tubes in sulfur-containing atmospheres are less than those in air, and the loss of life increases with increasing sulfur partial pressures. Creep ductilities are lower in the sulfur-containing gas. Consistent with the findings of Cook and Sutton (1978), the stress rupture lives of tubes tested in low sulfur-containing gas (<10^-7 atm) tend toward the life of tubes tested in air as stresses decrease.

It is clear that the changing and inconsistent relationships between time to 1% creep, minimum creep rate, and rupture life are issues that are important in the design of components fabricated from alloy 800H, as well as the possible loss in design margins due to reducing, inert, or very aggressive (sulfidizing) environments. Stress-rupture data obtained in "other environments," as well as results from "control tests" in air, are plotted as log stress versus the LMP in Fig. 9 for alloy 800H. The RP-530 and ASME curves are included in the figure for comparison. It may be seen that several data fall short of performance expectations. Highly sulfidizing atmospheres produce data that fall short of the ASME design curve, and one needs to employ large corrosion factors when using the material in an aggressive atmosphere. Even data from some of the tests in helium and control tests in air fall near the ASME design curve, albeit data come from short-time testing programs.

Materials other than alloy 617 and 800H exhibit essentially tertiary creep behavior at temperatures above 816°C (1500°F), but little or no information is available in
the open literature for a systematic study. Vendors often report design curves such as the stress versus the average time to produce a specific strain level or the stress to produce an average creep rate. Such information is available for alloy 556 (Haynes International, 1988), alloy 230 (Haynes International, 1987), and 253MA stainless steel (Kelly, 1983). The design curves represent "disassociated" or "processed" data. Data sets are needed where temperature, stress, time to specific strains, and rupture data are reported for each test. Such full data sets are rarely available in the literature, even for the alloys approved for pressure vessel service, and without such information it is difficult to establish design margins that are based on strains or strain rates. In a few cases, the shape of the creep curves are provided. Page et al (1984) show creep curves for type 310 stainless steel in air and CMG at 815°C. The curves have a tertiary character and indicate no environmental effect. They also show the heat-to-heat variation in the minimum creep rate for both type 310 stainless steel and RA333 tested in CMG. The variation in creep rate is more than an order of magnitude. Kelly (1984) provides data for RA333 tested in air that include several heats and product forms. Sufficient information is available to conclude that RA333 is stable in creep at high stresses and tends toward a tertiary creep character at low stresses. Heat-to-heat variation in minimum creep rate is approximately an order of magnitude for the same stress and temperature. Prager (1992) reports results from a recent testing program designed to expand the data base for types 309 and 310 stainless steels. Testing temperatures are limited to 871°C (1600°F) and below, but results for several heats are provided. Negative creep is reported at the start of some tests, and tertiary creep often begin at strains below 1%. By prior aging, Prager was able to eliminate the negative creep at the start of testing. Grain size is found to be influential, with coarser grain size producing lower creep rates and longer lives at 816 and 871°C.

The issue of cold work in solid solution nickel- and cobalt-based alloys at temperatures above 815°C is addressed by Klarstrom (1991), and the evaluation includes both alloy 230 and alloy 625. Generally, small levels of cold work, say 5%, affect the kinetics of carbide precipitation and produce small improvements in life. Large levels of cold work promote recrystallization, and conditions that cause very fine grain size produce a loss in rupture life. Similar effects are observed by Persson (1978) in alloy 800. He reports stress rupture data after solution treating and cold working 10, 20, and 30%. At 800°F (1472°F), 10% cold work produces approximately 20% strengthening that endures for at least 20,000 h, while 30% cold work produces approximately 20% loss in strength. At 900°C (1650°F), all three levels of cold work produce at least 20% loss in strength at 10,000 h.

The high-temperature fatigue and creep-fatigue behavior of alloy 617 has been examined extensively for the development of ASME Sect. III draft Code Case, and design fatigue curves to 982°C (1800°F) are described by (Corum and Blass, 1992). Low-cycle fatigue curves at temperatures above 816°C (1500°F) have been produced on other alloys of interest for second-generation combined-cycle applications. These include RA333 (Kelly, 1984 and Manjoine, 1991), alloy 556 (Krukemyer, 1991), and alloy 230 (Rothman, 1990). Generally, the high-temperature alloys exhibit similar fatigue curves. Stronger alloys tend to have better fatigue resistance as the cycles increase. Differences may be seen in Fig. 10, which compares the design fatigue curve for alloy 617 at 871°C (1600°F) to fatigue curves for several alloys. All data fall above the design curve for alloy 617. Direct comparisons of one alloy to another, however, are not meaningful since testing frequencies vary from one group of data to another.

**WELDMENTS**

Some hardware and components in the PFBC piping and gas cleanup systems contain structural weldments, and the available rods, electrodes, and filler metals for BPV Code applications are chosen from ASME, Sect. II, Part C. Even when constructed to Code requirements, the integrity of weldments under creep and creep-fatigue loadings conditions remains an issue for high-temperature service. Weldment strength reduction factors linked to design life have been developed for CC-47, but the estimation of reductions factors remains as a major issue to be resolved.
in the development of rules for high-temperature construction codes (Corum and Blass, 1992). For construction with alloy 800H at temperatures above 816°C (1500°F), dissimilar metal welds are recommended when strength is required (INCO, 1986). Materials include welding electrode 117 and filler metal 617. Both of these alloys exhibit strengths well in excess of alloy 800H, since they are matching weld metals for alloy 617 (INCO, 1979). An overmatch in strength causes an uncertainty in weldment performance, since weldment reduction factors are based on the relative creep strength of deposited weld metal to base metal and do not incorporate restraint effects produced by different creep strengths in the dissimilar metals. Filler metals for alloy 556, RA333, and 253MA stainless steel have essentially the same compositions as the base metals, and meager data indicate that the weld metals have similar strengths to base metal (McCoy and King, 1985 and Swindeman, Bolling, and Mallett, 1990). Alloy 556 has been used to join HR-160 and HR-120 but stress rupture data have not been available for these joints. An important issue in the use of weldments at temperatures above 815°C is whether or not to perform solution treatment after welding. It seems likely that residual stresses will be relieved during service, and often cast structures in welds will have better strength than wrought structures. Exploratory studies in RA333 indicate that solution treated weldments have better life and ductility (Swindeman, Bolling, and Mallett, 1990). Little is known about the performance of weldments under stress and PFBC environments, although Natesan and Wang (1991) report data on the corrosion of weld metals in AFBC environments.

**DAMAGE ESTIMATION**

Capability is needed to estimate remaining life in PFBC components after large operational temperature or pressure changes in the duty cycle and thermal transients caused by turbine trips or PFBC instabilities. Excessive creep distortion, thermal fatigue, creep-fatigue interactions, creep-ratcheting, corrosion-fatigue, and erosion-corrosion are damage mechanisms that lead to loss of function. Often, a judgement is made in regard to the predominant failure mode, and an analysis is made on that basis. For example, McFarlane (1991) chose not to consider cycling in evaluating the life of components fabricated from alloy 617 and alloy 800H and operating in oil refinery and similar process applications. The number of cycles was judged to be too low to accumulate much fatigue damage, so only linear damage based on time fractions was used. However, he found it necessary to take into consideration the thermal stresses and their redistribution in order to arrive at a reasonable estimation of life. Gommans, Verheesen, and Heerings (1992) examined several approaches to estimate damage in alloy 800H components at 800°C (1472°F) and chose a creep strain limit of 7%. This approach eliminated the need to consider tertiary creep in damage summation and permitted an estimate of remaining life based on strain rate rather than life fractions. Manjoine and Filstrup (1992), on the other hand, had an application for alloy RA333 in a PFBC vessel. The primary stresses were very low but the thermal stresses were very high. They chose to use a fatigue analysis in which the low-cycle fatigue curve was assumed to be strain rate dependent and damage was summed on a cyclic life-fraction basis. They were able to qualify their design for the intended life by this procedure.

Often, the designer uses linear damage accumulation methods that incorporate time fractions for creep damage and cycle fractions for fatigue damage. When significant creep and fatigue damage is expected, the procedure outlined by Campbell, (1975) is followed. Here, both time fractions for creep damage and life fractions for fatigue damage are summed.

One of the limitations in the use of linear damage based on life and cycle fractions is the lack of understanding concerning continuum damage relationships. A promising alternate approach is Kachanov damage concept. Here, the creep rate, \( \frac{d\varepsilon_c}{dt} \), in a component under constant stress, \( \sigma \), and temperature is given by:

\[
\frac{d\varepsilon_c}{dt} = K \sigma^n / (1 - \omega)^m \tag{4}
\]

\[
\frac{d\omega}{dt} = B \sigma^p / (1-\omega)^q \tag{5}
\]

where \( \omega \) is known as damage and \( B, K, n', m, p, \) and \( q \) are material constants. Solving the differential equations gives an equation for the creep curve of the form:

\[
\varepsilon_c = \lambda \varepsilon_0 \frac{1}{(1 - \omega)} \left[ (1 - (1 - \omega))^{1/\lambda} \right] \tag{6}
\]

\[
\lambda = (1 + p)/(1 + p - q), \tag{7}
\]

\[
\varepsilon_0 = \lambda \varepsilon_0 \frac{1}{(1 - \omega)} \left[ (1 - (1 - \omega))^{1/\lambda} \right] \tag{8}
\]

where \( \varepsilon_0 \) is an integration constant, \( m, p, \) and \( q \) are constants of the same order-of-magnitude as \( n' \), the stress exponent in the Norton law for the minimum creep rate, \( m_{cr} \):

\[
m_{cr} = K \sigma^{n'}. \tag{9}
\]

Typically, \( n' \) is in the range 4 to 8 and is often close to the
The continuous increase in creep rate is consistent with the tertiary character of the creep curve for many of the materials tested at temperature above 816°C (1500°F), especially some of the testing that was performed in inert atmospheres. Hence, the use of the Kachanov approach is attractive for application to very high temperatures. Experimental data are required to quantify the constants given above, and these constants may vary with temperature, stress, environment, and metallurgical factors such as grain size and heat treatment. When \( n \) is close to \( n' \), the constant \( \varepsilon_0 \) may be obtained from the Monkman-Grant (M-G) plot of log rupture life versus log mcr, and this is a measure of the materials tolerance to creep strain. The constant \( \lambda \) is a measure of the materials tolerance to local strain concentration. A high value for \( \lambda \), say 5 or more, may be observed in a creep ductile material and low value, say 1 or 2, would correspond to a creep brittle material.

The M-G plot requires paired values for minimum creep rate and rupture life. Such data are available for alloy 800HT and RA333, and trends are shown in Figs. 11 and 12, respectively. In both figures, it is apparent the scatter of data is large and that rupture life is not inversely proportional to the mcr for alloy 800H and RA333. It follows that Equation (10) is not always applicable for the alloys under the assumption of constant \( \lambda \) and \( \varepsilon_0 \). The slope of the curves suggests a decreasing rupture ductility with increasing test time. The problem may be resolved by introducing an exponent in the M-G correlation to account for a decreasing strain tolerance with decreasing mcr. Another approach is to base the M-G correlation on a
strain limit such as 5%. A strain limit presents no problem in the use of the Kachanov damage summation approach and reduces the influence that recrystallization, internal oxidation, and necking have on the correlating of mcr with the failure criterion. Data that couple temperature, stress, and the time to a specific strain are not readily available in the open literature, but enough information was found to construct M-G plots for a few alloys. Fig. 13 shows the log time to 2% strain versus log mcr for RA333, and data were taken from Kelly (1984). Trends indicate that, to a first approximation, there exists a linear relationship between mcr and the time to 2% creep strain.

SUMMARY AND CONCLUSIONS
Much work is needed to develop a consensus for design methodologies and material performance criteria for pressure vessels and piping in second-generation combined-cycle applications. The various aspects covered in this paper only provide a limited view of the issues and problems that must be addressed, and it is only through shared experiences that meaningful solutions to potential problems will be developed.

Because of the large base of experience and data, alloy 800H is used to identify some of the concerns regarding material behavior. It is shown that the design margins built into construction codes are less than a factor of 20 in life for temperatures above 816°C (1500°F), and that environmental effects may consume a significant portion of life for temperatures above 816°C (1500°F), and that environmental effects may consume a significant portion of life for temperatures above 816°C (1500°F), and that environmental effects may consume a significant portion of life for temperatures above 816°C (1500°F), and that environmental effects may consume a significant portion of life for temperatures above 816°C (1500°F), and that environmental effects may consume a significant portion of life for temperatures above 816°C (1500°F). It is shown that the creep curves for alloy 800H are highly variable and the correlation between the minimum creep rate and rupture life (Monkman-Grant plot) exhibits wide scatter. It is suggested that the use of a strain limit, say 2 or 5%, as a substitute for rupture life may be a useful criterion on which to set design stresses.

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