Development of Creep-Fatigue Design Methodology for Base Metal and Weldments of G91

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Development of Creep-Fatigue Design Methodology for Base Metal and Weldments of G91

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ABSTRACT

This report provides an update on research activity at the Argonne National Laboratory (ANL) in the area of high temperature materials design methodology for advanced ferritic-martensitic steels for advanced Small Module Reactor (aSMR) Concepts. This is a Level 2 report delivered in FY15 (M2AT-15AN1602013) under the work package AT-15AN160201 “Materials Design Technology for Advanced SMR Concepts” performed by ANL for the Department of Energy, Office of Nuclear Energy.

The overall objective of the Materials for Advanced SMR Concepts is to address key long-term design needs for applications of advanced materials. Advanced materials are a critical element in the development of advanced SMR concepts. High temperature materials design technology must be developed to realize the benefits of advanced materials in advanced reactor designs. Among a number of technical issues relevant to materials performance criteria and high temperature design methodology, creep, creep-fatigue, weldment design methodologies, and long-term thermal aging effects are of the highest priority. The effort in this project has focused on improvement of current ASME creep-fatigue design rules and development of a mechanism-based creep-fatigue design method for ferritic-martensitic steels, G91 and G92. This report presents the development of a fundamentally new creep-fatigue life model for G91 base metal and improvements of understanding of creep-fatigue interactions and creep-fatigue design rules for G91 weldments.

A major advance has been made in understanding and modeling cyclic softening and stress relaxation behavior, and creep-fatigue life prediction for G91 and G92 steels. We established a fundamentally new creep-fatigue model that is explicitly applicable to cyclically softened materials, so-called the Cyclic Deformation Life Model (CDLM). We found striking similarities between the cyclic softening behavior and creep behavior. Major findings are:

- The cyclic stress continues to decrease with increasing number of cycles, so-called cyclic softening. The cyclic softening curve can be characterized by three stages, resembling a creep curve: in the primary stage, the cyclic strain rate increases with the number of cycles; in the secondary stage, the cyclic strain rate remains constant; in the tertiary stage, the cyclic strain rate increases again with the number of cycles. The tertiary stage is presumed to be dominated by crack propagation. The cyclic softening curve can be described by a logarithmic equation, namely, the Cyclic Softening Model.

- The secondary cyclic strain rate is inversely proportional to the crack initiation life, in the form similar to the Monkman-Grant relationship for creep. The crack initiation life was defined as the number of cycles at the onset of the tertiary stage of the cyclic softening curve. We named it as the “Cyclic Deformation Life Model (CDLM)”. The model implies that cyclic softening can be regarded as time-dependent plastic deformation under cyclic loading and temperature. The model is applicable to G91 and Opt. G92 tested under various fatigue and creep-fatigue loading conditions and at temperatures of 550-600°C. The constant, referred to as cyclic ductility, is independent of the loading conditions, test temperature and metallurgical factors. In comparison to the Monkman-Grant relationship, the cyclic ductility constant is two orders of magnitude smaller than the creep ductility constant in the Monkman-Grant equation for G91 steel.
Further analysis of the fatigue and creep-fatigue data revealed that the secondary cyclic strain rate follows a power-law relationship with the applied total strain. The power-law exponent is sensitive to the test temperature, but insensitive to the waveform of cyclic loading.

The Cyclic Deformation Life Model correlates the steady-state cyclic strain rate with cycles to crack initiation, and provides a new crack-initiation-based creep-fatigue life prediction methodology that is critical for high-temperature structural designs. The model represents a critical strain criterion, implying that failure will occur when the accumulated creep-fatigue damage reaches a critical level that is manifested as the failure that can be predicted from the steady-state cyclic deformation rate and the cycles/time to failure. As in the creep mechanism models, it is promising that creep-fatigue mechanisms can be established and integrated into the life prediction model, which is the focus of future work.

In addition, an ASME Code action has been made collaboratively between DOE Office of Nuclear Energy and JAEA to incorporate the stress relaxation model developed under this project into the ASME creep-fatigue design procedure. The creep-fatigue tests of the G91 weldment specimens have been initiated, and the effort will continue in FY2016.
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1 Introduction

Economic competitiveness, safety and reliability are key elements in the development of advanced reactor technologies. Advanced materials allow compact and simpler design of structural components and have the potential to reduce construction and operational costs for advanced reactors. High temperature-, corrosion- and irradiation-resistant alloys are important for flexible, reliable and safe designs of reactor components. The materials design technology is an enabling tool to implement new materials in the design of advanced reactors.

Advanced fast reactors will operate at higher temperatures than the current light water reactors (LWRs). The design of high-temperature structural components must consider time-dependent effects and damage mechanisms, such as long-term thermal aging, creep, creep-fatigue, creep ratcheting, and environmental effects. These issues have not been dealt with in LWR designs and operation. Due to the temperature dependence of material properties, extrapolation of short-term, high-temperature accelerated test data to a long-design life and relatively low-service temperatures becomes extremely difficult and often with large uncertainties. Nuclear components that operate at elevated temperatures must be designed in accordance with the American Society of Mechanical Engineers (ASME) Section III Division 5 (previously Subsection NH – Class 1 Components in Elevated Temperature Service), which provides high temperature design rules for structural components in nuclear service. Materials that are allowed for use in high temperature structural designs in nuclear components must be qualified and approved by Division 5, and so far only five materials have been qualified, namely, Type 304 and 316 austenitic stainless steels, 2.25Cr-1Mo steel, modified 9Cr-1Mo (G91) steel, and Alloy 800H. For any new materials to be included in Division 5, an adequate database must be established and associated high temperature materials design rules must be developed to account for unique high temperature damage modes in each material. High temperature materials design methodology is unquestionably a critical technology that can accelerate the qualification of new materials and realize the benefits of these materials in advanced reactors.

During the licensing review of the Clinch River Breeder Reactor Plant (CRBRP) and the Power Reactor Innovative Small Module (PRISM) project by the Nuclear Regulatory Commission (NRC), a number of technical issues were identified, including creep-fatigue damage, weldment safety evaluation, notch weakening effect, environmental effects, etc. Many of these issues remain unresolved, and require extensive experimental and analytical effort for a successful resolution. A group of researchers at Argonne National Laboratory (ANL) and Oak Ridge National Laboratory (ORNL) conducted detailed, in-depth assessment of materials’ qualification and licensing needs, identified a list of key technical issues, and recommended a viable approach to resolve these issues and the R&D priority [1,2]. Among them, creep-fatigue interaction and long-term thermal aging-induced mechanical degradation are two of the most critical issues that must be addressed for structural applications of advanced materials in sodium-cooled fast reactors.

Creep-fatigue is the most damaging and least understood mechanism in high temperature structural designs. Creep is a time-dependent strain development under a constant load, while fatigue is an accumulated damage process under cyclic loading. Creep-fatigue interaction is a complex dynamic process involving combined effects of creep, fatigue, and environment on the accumulated damage. The process depends on a number of mechanical and metallurgical factors including test temperature, strain rate, hold time, types of hold, environment, thermo-mechanical...
development, microstructure, etc. There is extensive evidence that the combined effect of creep and fatigue is much greater than their individual contributions, and this synergistic interaction is much less understood than individual events of creep and fatigue.

The creep-fatigue damage in high temperature structural design has been evaluated using an empirical approach. Current ASME creep-fatigue design procedures are based on simple damage accumulation rules and basic material properties (e.g. yield stress, fatigue and creep life curves, etc.). The creep-fatigue life is evaluated using a linear summation of fractions of cyclic damage and creep damage in an uncoupled manner. The interaction of creep and fatigue is accounted for empirically by the total damage, and is represented graphically by the creep-fatigue interaction diagram. The design rules provide a simple solution to the design of structural components subjected to complex temperature, strain and environment conditions, but inevitably involve excessive conservatisms and are limited in their generality. Such an engineering approach is effective only if it is supported by a robust database that is representative of the operating conditions and when it does not involve significant data extrapolation. When the extrapolation is beyond the database where the empirical constants are derived, the reliability and accuracy of life prediction are highly questionable. The reliability of the life-prediction methods would be greatly enhanced if they were based on an understanding of the mechanisms underlying the damage processes. There is however, a large knowledge gap between a physical understanding and the engineering rules.

The empirical nature of the design methodology leads to a number of issues when applying the current creep-fatigue design procedures to ferritic-martensitic steel G91. The ASME creep-fatigue design rule was established primarily for austenitic stainless steels. Austenitic stainless steels behave drastically different from ferritic-martensitic steels under creep-fatigue loading. G91 is subjected to considerable cyclic softening during strain-controlled cyclic loading [3,4]; in contrast, austenitic stainless steel, Types 304 and 316, cyclically harden [5]. Cyclic softening can have a profound effect on creep, fatigue, creep-fatigue, fracture, and high-temperature tensile performance. The advantage of the initial high strength of G91 steel can be lost even after a small number of loading cycles due to dramatic changes of initial microstructure. Depending on the strain range and strain rate, this stress may drop to nearly one-half of its starting value. The unstable microstructure of G91 steel results in unsaturated creep-fatigue behavior, which makes extrapolation of short-term test data to long-term service performance extremely difficult and unreliable as illustrated in Fig. 1-1. The creep-fatigue design method developed for cyclically hardened materials such as austenitic stainless steels with relatively stable microstructure cannot be directly applied to cyclically softened G91 steel with continuously evolved microstructure. The very different behavior of these two classes of alloys implies the operation of different mechanisms that cannot be captured by an empirical approach.

Lack of an appropriate creep-fatigue model applicable to G91 steel is a central issue in the current ASME creep-fatigue design methodology. A multi-pronged approach is employed in Division 5 to ascertain that the design life determined from the creep-fatigue criterion is adequately conservative. The over-conservatism of the G91 creep-fatigue design procedure has been addressed in a number of studies [6-9]. Continued efforts have been made to alleviate the excessive conservatism issue. Previously, the creep-fatigue damage envelope for G91 had the interception point of (0.1, 0.01), which results in overly conservative creep-fatigue design, especially for the creep load. The recently-developed ASME Code Case, CC N-812 offers an
alternative creep-fatigue damage envelope with an interception of (0.3, 0.3) when creep damage is evaluated in accordance with T-1433(a) Step 5(2) to satisfy the requirements of T-1431(d) and only the elastic analysis is involved. While this alternative procedure provides a certain level of relief of conservatism and an intermediate solution, conservatism remains large and needs to be further reduced. Three fundamental issues of the current creep-fatigue evaluation method for G91 must be addressed: (1) lack of proper evaluation methods for cyclic softening and stress relaxation behavior of G91, (2) empirical correlations have no physical basis, and is unreliable for extrapolation, and (3) environmental effects, e.g. sodium and irradiation exposures, are completely ignored.

Life prediction of G91 welded components subject to creep-fatigue loading is another critical design concern. The creep-fatigue database of G91 weldments is much smaller than that for the base metal, and a separate creep-fatigue procedure for welds is lacking. The weldment creep-fatigue design procedure in ASME Section III Division 5 was established based on G91 base metal creep-fatigue data, and the adequacy of the creep-fatigue design procedure for weldments needs to be verified by sufficient weldment creep-fatigue data. G91 weldments are susceptible to Type IV cracking, a premature failure in the heat affected zone (HAZ) of welded joints. Type IV cracking occurs because of the gradients of microstructure developed in the HAZ, and it is localized in the fine grained and inter-critical region of the HAZ, the so-called creep-weak region. Current creep-fatigue design rules do not explicitly account for Type IV cracking in G91 weldment under creep-fatigue loading.
With the overarching goal of developing mechanism-based predictive models and ultimately developing a robust design methodology for creep-fatigue in advanced structural alloys, our research effort has been focusing on understanding and modeling cyclic softening and stress relaxation behavior, and developing a new creep-fatigue life model that is explicitly applicable to cyclically softened G91 and G92 ferritic-martensitic steels. We have successfully developed a cyclic softening model and a stress relaxation model that can accurately describe the cyclic softening curves and the hold-time stress relaxation curves for G91 and G92. An ASME Code action is being taken to incorporate these models into the current creep-fatigue design procedure. More importantly, we have recently developed a fundamentally new creep-fatigue life prediction model that captures the cyclic softening behavior of G91 and G92 steels. We discovered that the cyclic softening behavior of G91 and G92 steels under fatigue or creep-fatigue loading resembles the creep deformation behavior, and the creep-fatigue lifetime can be estimated using a comparable life prediction model for creep.

Effort has also been initiated to develop experimental data of G91 weldments and an improved creep-fatigue design methodology.
2 Experimental Procedure

2.1 G91 and Opt. G92 Base Metals

Two ferritic-martensitic steels were investigated in this study, namely, Grade 91 Heat 30176 and optimized Grade 92 Heat G92-3. G91 was developed for intermediate heat exchanger and steam generator applications for LMFBRs in the United States. There has been a continued increase in the use of G91 in sodium-cooled fast reactors worldwide. Optimized G92 was recently developed by the Oak Ridge National Laboratory (ORNL). The chemistry of optimized G92 was tightly controlled within the American Society Testing Materials (ASTM) specifications, while the microstructure was optimized to provide significantly better tensile and creep properties than conventional G91 and G92 steels. Optimized G92 was down-selected for further development by Advanced Reactor Concepts Advanced Materials Program in 2012 [10].

The chemical compositions of G91-H30176 and opt. G92-3 are given in Table 2-1. The chemical compositions of G91 steel were taken from the reference [11]. It is noted that the content of silicon in the G91 is below the ASTM specifications. The chemical composition of optimized G92 steel is within the ASTM Grade 92 specification with constrained concentration for a few elements [10].

Grade 91-H30176 was normalized at 1050°C for 1 h, air cooled, and tempered at 760°C for 2 h, and air cooled. Opt. G92-3 was hot-forged and/or hot-rolled at 1170°C, followed by normalization at 1080°C for 1 h, air cooling, and tempering at 750°C for 2h, air cooling.

Figure 2-1 shows the microstructure of G91-H30176 and opt. G92-3 in the normalized and tempered (N&T) condition. Both alloys have tempered martensite structure, M\textsubscript{23}C\textsubscript{6} carbides decorated along martensite boundaries and prior γ grain boundaries, and MX carbonitrides distributed within martensite laths/subgrains. Subgrains were well developed in the as-received G91, while in optimized G92 fine martensite lath structure prevailed. Carbides and carbonitrides density was considerably higher in optimized G92 than in G91.

The choice of investigating two ferritic-martensitic alloys (G91-H30176 and opt. G92-3) rather than a single material has proved to be remarkably valuable in creep-fatigue modeling. The fundamentally same class of alloys with different strengths revealed the deformation and failure mechanisms that are not readily understandable in one alloy.

2.2 G91 Weldment

Four welded plates of G91-H30176 steel were made by Specialty Welding & Machining, LLC. Harrison, TN. The welding procedure conforms to the ASME B&PV Section IX requirements. ANL supplied eight G91 H30176 plates in normalized and tempered condition with dimensions of 14” L x 3” W x 1” T. The welded plate has dimensions of 14” L x 6” W x 1” T. The welding process was gas tungsten arc welding (GTAW). The weld wire was ER90S-B9 procured by the vendor with the certificate of analysis. Post-weld heat treatment was conducted at 760°C for 2 hours and air cooling. Documentation including the material certification of conformance for the weld wire, weld groove drawing, Welding Procedure Specification (WPS) form, Procedure Qualification Record (PQR) form, radiograph test reports, and bend and tensile test reports can be found in reference [12].
Table 2.1. Chemical composition (in wt%) of G91-H30176 and opt. G92-3 steels.

<table>
<thead>
<tr>
<th>Alloy</th>
<th>Fe</th>
<th>C</th>
<th>Cr</th>
<th>Mn</th>
<th>Mo</th>
<th>N</th>
<th>Nb</th>
<th>Ni</th>
<th>Si</th>
<th>Ti</th>
<th>V</th>
<th>W</th>
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</thead>
<tbody>
<tr>
<td>Opt.G92-3</td>
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<td>0.05</td>
<td>9</td>
<td>0.5</td>
<td>0.5</td>
<td>0.05</td>
<td>0.1</td>
<td>0.1</td>
<td>0.2</td>
<td>-</td>
<td>0.2</td>
<td>1.9</td>
</tr>
<tr>
<td>G91-H30176</td>
<td>Bal</td>
<td>0.08</td>
<td>8.6</td>
<td>0.37</td>
<td>0.89</td>
<td>0.055</td>
<td>0.07</td>
<td>0.09</td>
<td>0.11</td>
<td>0.004</td>
<td>0.21</td>
<td>&lt;0.01</td>
</tr>
</tbody>
</table>

Figure 2-1. Microstructure of (a) N&T G91-H30176 and (b) N&T opt. G92-3.

2.3 Specimens

ASTM standard-size round specimens were used in creep-fatigue tests. The geometry and dimensions of creep-fatigue specimens for G91 and G92 base metals are shown in Figs. 2-2(a) and (b), respectively. The G91-H30176 base metal specimens have a gauge diameter of 0.156 in, a gauge length of 0.5 in, and a total length of 4.5 in. The opt. G92-3 base metal specimens have a gauge diameter of 0.160 in, a gauge length of 0.64 in, and a total length of 4.0 in. The gauge length is along the rolling direction.

ASTM standard-size round creep-fatigue specimens were fabricated from the G91 welded plates. The geometry and dimensions of creep-fatigue weldment specimens are shown in Fig. 2-3. The gauge section of the creep-fatigue weldment specimen includes weld, heat affected zone, and base metal, as illustrated in Fig. 2-4.

Gauge sections of creep-fatigue specimens were polished longitudinally with 1-µm diamond paste to remove surface scratches and oxide layer, if any, before creep-fatigue testing.
Figure 2-2. Geometry and dimensions of creep-fatigue specimens for (a) G91-H30176 and (b) opt. G92-3.
Figure 2-3. Schematic drawing of G91 creep-fatigue weldment specimen.

Figure 2-4. Fabrication drawing of G91 creep-fatigue weldment specimens.
2.4 Creep-Fatigue Tests

Creep-fatigue tests were performed on closed-loop servo-hydraulic test frames at ANL. Specimen heating was achieved by a three-zone split furnace. The test temperature was controlled and monitored by two Type K thermocouples wired at the top and bottom grip sections of the specimen. A high temperature extensometer was used to control/measure the total axial strain. All creep-fatigue tests were carried out in air. The creep-fatigue test systems are shown in Fig. 2-5.

Creep-fatigue tests conducted in this study were specifically designed to support the development of creep-fatigue mechanistic models and improvement of ASME Code design rules. The model-oriented creep-fatigue tests require flexibility in selecting creep-fatigue loading conditions, e.g. hold-time cycling with a hold time applied either in tension, or compression, or at both tension and compression, symmetric continuous cycling (fast-fast (FF), slow-slow (SS)), or asymmetric cycling (slow-fast (SF) and fast-slow (FS)). These tests are typically conducted in the strain-controlled mode, as schematically shown in Fig. 2-6(a). During a strain-controlled test, creep-fatigue damage may be introduced by applying a hold time at peak strains, either in tension, or in compression, or in combined tension and compression, or by applying a slower strain rate either in tension, or in compression such as asymmetric cycling (slow-fast, fast-slow, etc.), or in both tension and compression (slow-slow).

To better understand the interaction of creep and fatigue damage, it is necessary to conduct sequential creep-fatigue tests, as schematically shown in Fig. 2-6(b). In sequential loading tests, a two-step loading is applied with creep followed by fatigue (referred as “Creep + Fatigue”), or fatigue followed by creep (referred as “Fatigue + Creep”). In the sequential creep-fatigue test, pure fatigue and pure creep are studied in sequence, and thus the contribution of each process in the collective creep-fatigue effect can be evaluated. These sequential creep-fatigue tests will allow us understand how fatigue impacts creep, and creep impacts fatigue as well as how fatigue and creep interact to accelerate the material damage.

To investigate the effect of an extended hold time on stress relaxation behavior, creep-fatigue tests were conducted by applying various lengths of hold time up to 60,000 s (17 h), as schematically shown in Fig. 2-6(c).

The Instron WaveMatrix™ Dynamic and Fatigue Materials Testing Software provide great flexibility in applying various loading conditions, and have been used for all creep-fatigue tests in this study.

2.5 Microstructural Characterization

The microstructure of as-received and creep-fatigue-tested specimens was characterized by optical microscopy (OM), scanning electron microscopy (SEM), and transmission electron microscopy (TEM). Fractography was performed on fractured specimens by SEM. A broken half of the creep-fatigue-tested specimen was sectioned along the specimen gauge for examination of cracking behavior by OM and SEM. TEM discs were made from the gauge section of the creep-fatigue-tested specimen perpendicular to the loading axis for microstructure characterization by TEM.
Figure 2-5. High temperature creep-fatigue test systems at ANL.
(a) Strain-controlled creep-fatigue interaction tests

(b) Sequential creep-fatigue tests

(c) A hold-time creep-fatigue test with varied length/type of hold times

Figure 2-6. Waveforms used in creep-fatigue tests.
3 Experimental Results and Creep-Fatigue Modeling

3.1 Cyclic Softening Curves of G91 and G92 Base Metals

One of the key issues in evaluation of creep-fatigue damage in high-strength ferritic-martensitic steels, G91 and G92 is their cyclic softening behavior, which can cause continuous degradation of tensile and creep properties during service. G91 and G92 steels cyclically soften under creep-fatigue loading, in contrast to conventional austenitic stainless steel, Types 304 and 316, which exhibit pronounced cyclic hardening. The concepts of cyclic hardening and cyclic softening are illustrated in Figs. 3-1 and 3-2, respectively [13]. Under the constant total strain amplitude, a cyclically hardened material shows a continued increase of the peak stress as a function of cycles followed by a saturation of peak stress, while a cyclically softened material shows a continued decrease of the peak stress and therefore a continued increase of plastic strain accumulation with increasing cycles. Figure 3-3 shows the tensile and compressive peak stress profiles for G91-H30176 fatigue tested with the total strain amplitude of 0.5% and creep-fatigue tested under a combined tension and compression hold (TCH) of 60 s with the total strain amplitude of 0.25% at 550°C. Cyclic softening is evident in G91-H30176 tested at 550°C under both pure fatigue and creep-fatigue loading conditions. A similar softening behavior was observed in opt. G92-3, as shown in Fig. 3-4 where the peak stress profiles are shown for opt. G92 creep-fatigue tested under a tension hold (TH) of 600 s with the total strain amplitude of 0.5% at 600°C.

We have found that the cyclic softening of G91 and G92 steels is characterized by a multiple-stage behavior, and can be described by the following equation, the so-called Cyclic Softening Model [14-16]:

\[
\frac{\sigma(N)}{\sigma(N=1)} = 1 - A \cdot \ln(kN + 1) - BN
\]

where \(\sigma(N)\) and \(\sigma(N=1)\) are the peak stress at cycle N and cycle 1, respectively, and A, k and B are the material constants. As shown in Figs. 3-3 and 3-4, an excellent agreement was achieved between modeling and experiment for all the testing conditions. The Cyclic Softening Model has been applied to various cyclic loading conditions, including test temperature, strain amplitude, waveform, hold type, and hold time.

Figure 3-1. Cyclic hardening (a) constant strain control, (b) stress response, (c) cyclic stress-strain response.
The plastic strain developed in each cycle can be calculated using the following equation:

\[ \frac{\Delta e}{2} = 0.25\% \text{, TCH 60s} \]

\[ \frac{\Delta e}{2} = 0.5\% \text{, CC} \]

\[ \frac{\Delta e}{2} = 0.5\% \text{, TH 600s} \]

\[ \frac{\Delta e}{2} = 0.5\% \text{, TH 600} \]
where $\Delta \varepsilon_e$, $\Delta \varepsilon_p$, and $\Delta \varepsilon_r$ are the total strain range, elastic strain range and plastic strain range, respectively, $\Delta \sigma$ is the total stress range, $N$ the number of cycles, and $E$ the Young’s modulus. Combining Eqs. (3-1) and (3-2), we can obtain:

$$\frac{d(\Delta \varepsilon_p(N))}{dN} = -\frac{d(\Delta \varepsilon_e(N))}{dN} = -\frac{1}{E} \frac{d(\Delta \sigma(N))}{dN} = \frac{\Delta \sigma_0}{E} \left( \frac{Ak}{kN + 1} + B \right)$$

(3-3)

Figure 3-5 shows the tensile peak stress vs. cycles and corresponding plastic strain as a function of cycles for optimized G92 fatigue tested at 550°C with the total strain amplitude of 0.5%. It can be seen that there are striking similarities between the cyclic strain – cycle curve and the creep strain – time curve. Similar to a classical creep curve, the cyclic strain – cycle curve can be divided into three characteristic stages: primary (stage I), secondary (stage II), and tertiary (stage III). As shown in Fig. 3-5(b), in the primary stage the strain increases monotonically with cycles at a reduced rate; in the secondary stage the plastic strain shows a linear dependence of cycle, i.e. a constant plastic strain rate, $d\varepsilon_p/dN$; in the tertiary stage the plastic strain develops rapidly, leading to the final failure. The primary and secondary stages of the cyclic strain curve can be well described by Eq. (3-3). It is apparent that cyclic softening of G91 and G92 steels is essentially a creep process. While creep is the time-dependent deformation under the constant load, cyclic softening is the time-dependent deformation under the cyclic load. Both processes proceed in three stages of primary, secondary, and tertiary deformation.

3.2 New Creep-Fatigue Interaction Model – Cyclic Deformation Life Equation

One of the most important findings in this study is that cyclic softening of G91 and G92 steels is a time-dependent deformation process under the cyclic load, resembling creep, a time-dependent
deformation under the constant load. In the secondary stage of the cyclic softening curve, the plastic strain follows a linear relationship with the number of cycles, therefore a constant cyclic strain rate. We define this constant strain rate as the steady-state cyclic softening rate, \( \frac{d\varepsilon_{ss}}{dN} \). At the onset of the tertiary stage, the cyclic strain rate increases rapidly, leading to unstable failure of a material. We define the number of cycles at the onset of the tertiary stage as the crack initiation life in cycles, \( N_i \), as shown in Fig. 3-5(b). Figure 3-6 plots the steady-state cyclic softening rate as a function of the crack initiation life, for G91 and G92 tested under various fatigue/creep-fatigue loading conditions at 550-600°C. It was found that the steady-state cyclic softening rate follows a power law relationship with the crack initiation life and it is independent of waveforms, types and lengths of hold times, and this relationship can be described as:

\[
\frac{d\varepsilon_{ss}}{dN} \cdot (N_i)^q = K
\]  

(3-4)

where \( q \) and \( K \) are the constants. Fitting of experimental data in Fig. 3-6 to Eq. (3-4) gives \( q = 0.84 \), and \( K = 9 \times 10^{-4} \).

The data shown in Fig. 3-6 are re-plotted in Fig. 3-7 in terms of the cyclic softening rate, \( \frac{d\varepsilon_{ss}}{dt} \) vs. the crack initiation life in seconds, \( t_i \):

\[
\frac{d\varepsilon_{ss}}{dt} = \frac{1}{\Delta t} \cdot \frac{d\varepsilon_{ss}}{dN}
\]

\[
t_i = \Delta t \cdot N_i
\]

(3-5)

where \( \Delta t \) is the period of one cycle. It was found that the steady-state cyclic softening rate can be described by:

\[
\frac{d\varepsilon_{ss}}{dt} \cdot t_i = C'
\]

(3-6)

where \( C' \) is the constant. Fitting of experimental data in Fig. 3-7 to Eq. (3-6) gives \( C' = 2.45 \times 10^{-4} \). We call this new function the Cyclic Deformation Life Model (CDLM).

Recall that in a creep rupture test, the steady-state creep strain rate, \( \frac{d\varepsilon_{ss}}{dt} \) follows a similar relationship with the rupture life, \( t_r \), the so-called the Monkman-Grant relationship:

\[
\frac{d\varepsilon_{ss}}{dt} \cdot t_f = C
\]

(3-7)

where \( C \) is the constant. It is apparent that both creep and creep-fatigue are governed by similar rate-dependent deformation processes, with the former occurring under a static constant load and the later occurring under a dynamic cyclic load. Figure 3-8 compares the cyclic strain rate vs. crack initiation life curve for G91 and G92 tested under fatigue/creep-fatigue loading with various waveforms and hold times at 550-600°C, and the creep strain rate as a function of rupture life for G91 creep-tested at 550-600°C. Both sets of data can be described by the same function (Eqs. (3-6) and (3-7)), namely, the lifetime is inversely proportional to the steady-state strain rate, but the values of the constants, \( C \) and \( C' \) vary by two orders of magnitude. The constant, \( C \) is often called the Monkman-Grant “ductility, which is the contribution of secondary creep strain to the total creep failure strain. The constant, \( C' \) in the CDLM represents the contribution of secondary cyclic
strain to the total fatigue/creep-fatigue strain for crack initiation, and is significantly smaller than the Monkman-Grant “ductility”.

Figure 3-6. Cyclic strain rate as a function of the initiation life (in cycle) for G91 and Opt. G92 tested under fatigue/creep-fatigue loading with various waveforms and hold times at 550-600°C.

Figure 3-7. Cyclic strain rate as a function of the initiation life (in seconds) for G91 and Opt. G92 tested under fatigue/creep-fatigue loading with various waveforms and hold times at 550-600°C.
Figures 3-8 and 3-9 (a) and (b) plot the cyclic strain rate, $\frac{d\varepsilon}{dN}$ and $\frac{d\varepsilon}{dt}$ as a function of the applied total strain amplitude, respectively, for G91 and opt. G92 tested at 550-600°C under various fatigue/creep-fatigue loading conditions. The $\frac{d\varepsilon}{dt}$ varies much more widely than the $\frac{d\varepsilon}{dN}$, primarily due to the large variation in the applied hold time. It appears that the cyclic strain rate follows a power-law relationship with the applied total strain. The power-law exponent is sensitive to the test temperature, showing stronger dependence of the cyclic strain rate on the applied total strain at 550°C than at 600°C, but less sensitive to the fatigue/creep-fatigue waveforms.

The Monkman-Grant creep equation has been used for the estimation of the creep rupture life in both laboratory tests and for field components [17]. Likewise, the correlation between the cyclic
softening rate and crack initiation life in the CDLM model can provide a fundamentally new approach to creep-fatigue life assessment for G91 and G92 ferritic-martensitic steels in the cyclic softening regime. This relationship implies that strain accumulated per cycle is the macroscopic manifestation of the cumulative creep-fatigue damage, and failure will occur when the damage reaches a critical level. This crack-initiation-based creep-fatigue life prediction methodology can be a useful tool for high temperature structural designs. As in the creep models, it is promising that creep-fatigue mechanisms can be established and integrated into the life prediction model.

3.3 Stress Relaxation Behavior of G91 and G92 Base Metals

When a hold time is introduced at maximum tensile strain, or minimum compressive strain, or combined maximum tensile and minimum compressive strains in a constant total strain controlled test, stress relaxation will occur during the hold period (see Fig. 2-4(a)). Data on stress relaxation during hold time is critical in evaluating material’s inelastic behavior and deformation characteristics and for estimating creep damage under creep-fatigue conditions. Stress relaxation produces an additional amount of plastic strain compared to pure fatigue, accelerating the damage process. The increase in plastic strain can be calculated by the relaxed stress assuming that unloading is elastic:

\[ \Delta \varepsilon_{rp} = \frac{\Delta \sigma_r}{E} \]  

(3-8)

where \( \Delta \varepsilon_{rp} \) is the increase in plastic strain due to relaxation, and \( \Delta \sigma_r \) is the relaxed stress during hold. The changes in plastic strain rate during hold can also be obtained from stress relaxation curves. During hold time, the total strain range is held constant, and the rate of total strain range is zero, i.e.:

\[ \dot{\varepsilon}_t = \dot{\varepsilon}_r + \dot{\varepsilon}_{rp} = 0 \]  

(3-9)

where \( \dot{\varepsilon}_t \), \( \dot{\varepsilon}_r \) and \( \dot{\varepsilon}_{rp} \) are the total strain rate, elastic strain rate and plastic strain rate, respectively. The elastic strain rate can be calculated using Hook’s law, and is defined as:

\[ \dot{\varepsilon}_re = \frac{1}{E} \frac{d\sigma_r}{dt} \]  

(3-10)

where \( \sigma_r \) is instantaneous relaxation stress and \( t \) is time. Hence, the plastic strain rate during hold can be obtained by:

\[ \dot{\varepsilon}_{rp} = -\dot{\varepsilon}_re = -\frac{1}{E} \frac{d\sigma_r}{dt} \]  

(3-11)

Due to the cyclic softening effect in G91 and optimized G92 steels, the stress relaxation response continues to change with the number of cycles over the entire cyclic process. Our previous work has shown that cyclic softening has a strong effect on stress relaxation during hold [14-16]. The initial stress at the beginning of the hold time is reduced significantly with increasing number of cycles, resulting in subsequently lower relaxed stress at a given time during hold at different cycles. However, when the relaxed stress is normalized by the initial stress, the stress relaxation curves are independent of the cycle number. The normalized stress relaxation curve is also independent of the type of hold time (i.e. either tension or compression hold), as shown in Fig. 3-10 for G91 creep-fatigue tested under a tension and compression hold (TCH) of 60 s with
the total strain amplitude of 0.25% at 550°C. The normalized stress relaxation curves over the entire cyclic history may be described by a single model.

![Figure 3-10. Normalized stress relaxation curves for G91 creep-fatigue tested under a tension and compression hold (TCH) of 60 s with total strain amplitude of 0.25% at 550°C, showing independence of hold types (tension or compression) and number of cycles.](image)

We have established a constitutive Stress Relaxation Model for G91 and G92 steels [14-16]. The stress relaxation during hold time was described by a thermally-activated deformation process. The normalized stress relaxation curve is described by the following equation:

\[
\frac{\sigma_r(t)}{\sigma_r(t=0)} = 1 - \frac{RT}{Q_{SR}} \ln(\beta t + 1)
\]  

(3-12)

where \(\sigma_r(t)\) is the relaxed stress at time, \(t\), \(\sigma_r(t=0)\) is the stress at the beginning of the hold \((t = 0)\), \(Q_{SR}\) is the activation energy for the plastic deformation of stress relaxation, \(\beta\) is the constant, \(T\) is the temperature, and \(R\) is the gas constant (8.31 J/mol-K).

Figure 3-11(a) shows an example of a comparison between experimental data and a modeled stress relaxation curve for optimized G92 tested at 600°C with the total strain amplitude of 0.5% and a tension hold time of 600 s. The hold-time stress relaxation data was fitted to Eq. (3-12) to acquire the activation parameters. An excellent fit was achieved for the creep-fatigue test conditions. An important finding was that the activation energy for stress relaxation response was the same as that for the transient stage of cyclic softening for G91 and optimized G92. It is apparent that a single thermally-activated deformation process is operative in both cyclic recovery and stress relaxation during hold.

To investigate whether or not a single model can be applied and the deformation process is changed when the hold time is extended to a much longer term, creep-fatigue tests were conducted by applying various lengths of hold time, as schematically shown in Fig. 2-6(c). The longest hold time applied was 60,000 s (17 h). Figure 3-11(b) shows the stress relaxation data for optimized G92 steels tested with a tension hold time of 60,000 s at the total strain amplitude of 0.5% at 600°C. The stress relaxation data were modeled using Eq. (3-12). A good agreement between the experimental data and modeled curve verified that a single thermally-activated deformation
process is operative during the hold time, and the stress relaxation behavior of a creep-fatigue test can be described by a unique stress relaxation equation.

Figure 3-11. Normalized stress relaxation curves (by the initial stress at t = 0) for optimized G92 at 600°C with the total strain amplitude of 0.5% (a) with a tension hold time of 600 s, and (b) with a tension hold time of 60,000 s.

Due to the independence of normalized stress relaxation curves upon cycle and hold types, a unified equation was developed by combining the Cyclic Softening Model (Eq. 3-1) and Stress Relaxation Model (Eq. 3-12) to describe the cycle-dependent stress relaxation, i.e.:

\[
\sigma_r(N, t) = \sigma_{(N=1)} \cdot \left[1 - A \cdot \ln(kN + 1) - BN\right] \cdot \left[1 - \frac{RT}{Q_{SR}} \ln(\beta t + 1)\right] 
\]

(3-13)

where \(\sigma_r(N, t)\) is the relaxed stress at time, \(t\) during hold at Cycle, \(N\), \(\sigma_{(N=1)}\) is the peak stress of the first cycle, \(Q_C\) is the activation energy for the cyclic deformation, and \(k\) and \(\beta\) are constants. As illustrated in Fig. 3-12, this unified model, for the first time, describes the entire history of peak stress and stress relaxation under the constant-strain creep-fatigue loading.
3.4 Microstructural Evolution under Fatigue and Creep-Fatigue Loading

It has been well recognized that microstructure of G91 and G92 ferritic-martenisitic steels changes dramatically after long-term thermal aging and/or under mechanical loading such as creep, fatigue and creep-fatigue. The creep resistance of G91 and G92 steels is significantly reduced due to reduction in dislocation density, subgrain coarsening, precipitate coarsening and dissolution, and solute depletion [18-29]. Similar microstructural changes were observed in fatigue/creep-fatigue tested G91 and G92 specimens. As shown in Fig. 3-13(a), subgrain coarsening was pronounced after the creep-fatigue test of G91 with a tension hold time of 600 s at the total strain amplitude of 0.5% and 600ºC. The initial elongated martensite structure transforms to equiaxed subgrain structure with decreased dislocation density within subgrains. Highly regular dislocation networks formed as shown in Fig. 3-13(b) in optimized G92 tested with a compression hold time of 600 s and the total strain amplitude of 0.5% at 550ºC. The average dislocation density within subgrains was reduced from $9.7 \times 10^{13}$/m² in the as-received condition to $2.5 \times 10^{13}$/m² after a creep-fatigue test at the total strain amplitude of 0.5% and a tension hold time of 600 s at 600ºC in G91 (see Figs. 3-14 (a) and (b)). Reduction in dislocations within subgrains was observed in specimens fatigue/creep-fatigue tested under various temperature and creep-fatigue loading conditions, as shown in Fig. 3-13 (c).
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Figure 3-13. TEM micrographs of (a) G91 tested with a tension hold time of 600 s and total strain amplitude of 0.5% at 600°C, and (b) optimized G92 tested with a compression hold time of 600 s and total strain amplitude of 0.5% at 550°C.

Figure 3-14. Dislocation densities measured by TEM for G91 (a) dislocation density histogram in the normalized and tempered specimen, (b) dislocation density histogram after creep-fatigue test with a tension hold time of 600 s and the total strain amplitude of 0.5% at 600°C, and (c) mean values of dislocation density in as-received and fatigue/creep-fatigue tested specimens.

Figure 3-15 shows TEM micrographs of the microstructure of G91 specimens after a pure fatigue test (total strain amplitude of 0.5% at 550°C) (Fig. 3-15a), a sequential fatigue and creep
test (fatigue at the total strain amplitude of 0.5% for 100 cycles, followed by creep at 314 MPa at 550°C) (Fig. 3-15b), a load-controlled creep-fatigue test (314 MPa, tension hold time of 600 s (TH 600s) at 550°C) (Fig. 3-15c), and a strain-controlled creep-fatigue test (total strain amplitude of 0.5%, tension and compression hold time of 60 s (TCH 60 s) at 550°C) (Fig. 3-15d). When G91 was tested under pure fatigue, subgrain structure changes dramatically. The mean width of the subgrain can be doubled in size, and M23C6 initially decorated the subgrain boundary remained in coarsened subgrains. There was also a strong tendency for formation of highly regular dislocation networks eventually developing into a low-energy boundary and forming equiaxed subgrains. However, only slight microstructural changes occurred after a sequential “fatigue + creep” test. The recovery process was more pronounced under strain-controlled creep-fatigue loading than pure fatigue or the sequential creep-fatigue test. The most dramatic microstructural changes were found in the specimen tested in the load-controlled creep-fatigue test. A number of nano-sized grains were formed, as shown in Fig. (3-15c).

Subgrain coarsening, reduction of dislocation densities and formation of low-energy dislocation structure are important evidence of dynamic recovery at high temperatures. Though microstructural changes are similar under thermal, creep, or fatigue/creep-fatigue loading in G91 and G92 steels, dynamic recovery is more significant under fatigue and creep-fatigue loading than under creep loading or isothermal annealing.

Figure 3-15. TEM micrographs showing the microstructure of G91 specimens (a) tested under pure fatigue loading, (b) in a “fatigue+creep” test, (c) tested under load-controlled creep-fatigue loading (314 MPa, TH 600s) at 550°C, and (d) tested under strain-controlled creep-fatigue loading (0.5%, TCH 60 s).
4 Characterization of G91 Weldment

Micro-hardness measurement and optical metallography were carried out along the gauge of the as-received creep-fatigue weldment specimen for G91-H30176, and the micro-hardness profile and corresponding microstructure in the HAZ are shown in Fig. 4-1. The hardness data in Fig. 4-1(a) are two separate measurements, and solid symbols represent the set of measurements with better statistics. Hardness varies significantly across the weld joint: the weld zone has the highest hardness; the hardness continues to decrease in the HAZ from the weld interface and reaches the minimum adjacent to the base metal. The gradients of microstructure in the HAZ are also evident, and grain size varies from coarse to fine to coarse, consistent with the general microstructural description of the HAZ of G91, i.e. coarse grain region (CGHAZ), fine grain region (FGHAZ), inter-critical region (ICHAZ), and over tempered region [30]. The soft zone where the hardness is the minimum is apparently within the inter-critical region of the HAZ.

Figure 4-2 compares the stress-strain curves of G91 base metal and cross-weld specimens tested at 550°C at a nominal strain rate of 0.001 s⁻¹. The base metal specimen shows a higher tensile strength and significantly better ductility than the cross-weld specimen. Micro-hardness measurement and optical metallography were conducted on the broken tensile piece of the weldment specimen, and the hardness profile and microstructure are given in Figs. 4-2(b) and (c). Necking and final fracture occurred in the FGHAZ/ICHAZ region, the same region as that for Type IV cracking observed in creep tests, as reviewed by Francis et al. [30]. The significant increase in hardness in the region between points A and B is due to heavy localized deformation of the necked region.

The creep-fatigue tests of the G91 weldment specimens have been initiated, and the effort will continue in FY2016.
Figure 4-1 (a) Hardness profile and (b) corresponding microstructure (markers given in (a)) along the gauge of the G91 as-received cross-weld specimen.
Figure 4-2. (a) Stress-strain curves for G91 base metal and cross-weld specimens tested at 550°C at a nominal strain rate of 0.001 s⁻¹, (b) micro-hardness profile and (c) corresponding microstructure along the gauge of the broken tensile piece of the G91 cross-weld specimen.
5 Improvement to ASME Creep-Fatigue Design Methods

A coordinated effort has been initiated to improve the ASME creep-fatigue design methods between the DOE Office of Nuclear Energy and Japan Atomic Energy Agency (JAEA) under the DOE – JAEA bilateral collaboration. An ASME Code action has been made collaboratively between DOE and JAEA to incorporate the stress relaxation model developed under this project into the ASME creep-fatigue design procedure. As presented in Section 3, the cyclic softening behavior of G91 and G92 steels can be described by a constitutive model, i.e.:

$$\frac{\sigma(N)}{\sigma(N=1)} = 1 - A \cdot \ln(kN + 1) - BN$$

(5-1)

where \(\sigma(N)\) and \(\sigma(N=1)\) are the peak stress at cycle N and cycle 1, respectively, and A and k are the material constants, and B is the constant related to the steady-state strain rate. The stress relaxation during hold time can be described by a thermally-activated deformation process. The normalized stress relaxation curve is described by the following equation:

$$\frac{\sigma_r(t)}{\sigma_r(t=0)} = 1 - \frac{RT}{Q_{SR}} \ln(\beta t + 1)$$

(5-2)

where \(\sigma_r(t)\) is the relaxed stress at time, \(t\), \(\sigma_r(t=0)\) is the stress at the beginning of the hold \((t = 0 \text{ s})\), \(Q_{SR}\) is the activation energy for the plastic deformation of stress relaxation, \(\beta\) is the constant, \(T\) is the temperature, and \(R\) is the gas constant \((8.31 \text{ J/mol-K})\). Due to the independence of normalized stress relaxation curves upon cycle and hold types, a unified equation by combining the Cyclic Softening Model and Stress Relaxation Model, was developed to describe the cycle-dependent stress relaxation, i.e.:

$$\sigma_r(N,t) = \sigma_r(N=1) \cdot \left[1 - A \cdot \ln(kN + 1) - BN\right] \cdot \left[1 - \frac{RT}{Q_{SR}} \ln(\beta t + 1)\right]$$

(3)

where \(\sigma_r(N,t)\) is the relaxed stress at time, \(t\) during hold at Cycle, \(N\), \(\sigma_r(N=1)\) is the peak stress of the first cycle, \(Q_C\) is the activation energy for the cyclic deformation, \(k\) and \(\beta\) are constants. We are pursuing the acceptance of this new model by ASME for determination of stress relaxation behavior of G91 steel in the ASME creep-fatigue design procedure.

The fundamentally new creep-fatigue life prediction model developed in this project, namely, the Cyclic Deformation Life Model (CDLM) (see Section 3) is the first model that explicitly account for cyclic softening behavior unique to G91 steel, and has the potential to significantly reduce the over-conservatism in the current G91 creep-fatigue design procedure. While further development is needed to incorporate the creep-fatigue deformation mechanisms in the model, as with the Monkman-Grant creep correlation and creep deformation mechanism models, the creep-fatigue damage and life prediction may be treated in a procedure similar to that used for creep, which is widely accepted by the ASME. The constitutive equation can also be readily incorporated in finite element analysis.
6 Summary and Future Work

A major advance has been made in understanding and modeling cyclic softening and stress relaxation behavior, and creep-fatigue life prediction for G91 and G92 steels. We established a fundamentally new creep-fatigue model that is explicitly applicable to cyclically softened materials, so-called the Cyclic Deformation Life Model (CDLM). We found striking similarities between the cyclic softening behavior and creep behavior. Major findings are:

- The cyclic stress continues to decrease with increasing number of cycles, so-called cyclic softening. The cyclic softening curve can be characterized by three stages, resembling a creep curve: in the primary stage, the cyclic strain rate increases with the number of cycles; in the secondary stage, the cyclic strain rate remains constant; in the tertiary stage, the cyclic strain rate increases again with the number of cycles. The tertiary stage is presumed to be dominated by crack propagation. The cyclic softening curve can be described by a logarithmic equation, namely, the Cyclic Softening Model.

- The secondary cyclic strain rate is inversely proportional to the crack initiation life, in the form similar to the Monkman-Grant relationship for creep. The crack initiation life was defined as the number of cycles at the onset of the tertiary stage of the cyclic softening curve. We named it as the “Cyclic Deformation Life Model (CDLM)”. The model implies that cyclic softening can be regarded as time-dependent plastic deformation under cyclic loading and temperature. The model is applicable to G91 and Opt. G92 tested under various fatigue and creep-fatigue loading conditions and at temperatures of 550-600°C. The constant, referred to as cyclic ductility, is independent of the loading conditions, test temperature and metallurgical factors. In comparison to the Monkman-Grant relationship, the cyclic ductility constant is two orders of magnitude smaller than the creep ductility constant in the Monkman-Grant equation for G91 steel.

- Further analysis of the fatigue and creep-fatigue data revealed that the secondary cyclic strain rate follows a power-law relationship with the applied total strain. The power-law exponent is sensitive to the test temperature, but insensitive to the waveform of cyclic loading.

- The Cyclic Deformation Life Model (CDLM) correlates the steady-state cyclic strain rate with cycles to crack initiation, and provides a new crack-initiation-based creep-fatigue life prediction methodology that is critical for high-temperature structural designs. The model represents a critical strain criterion, implying that failure will occur when the accumulated creep-fatigue damage reaches a critical level that is manifested as failure that can be predicted from the steady-state cyclic deformation rate and the cycles/time to failure. As in the creep mechanism models, it is promising that creep-fatigue mechanisms can be established and integrated into the life prediction model, which is the focus of future work.

In addition, an ASME Code action has been made collaboratively between DOE Office of Nuclear Energy and JAEA to incorporate the stress relaxation model developed under this project into the ASME creep-fatigue design procedure. The creep-fatigue tests of the G91 weldment specimens have been initiated, and the effort will continue in FY2016.
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