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GENERATION IV/NGNP MATERIALS PROJECT
TASK 11: NEW MATERIALS

PART I- SELECTION OF CANDIDATE MATERIALS

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ABSTRACT

The bounding conditions were briefly summarized for the Next Generation Nuclear Plant (NGNP) that is the leading candidate in the Department of Energy Generation IV reactor program. Metallic materials essential to the successful development and proof of concept for the NGNP were identified. The literature bearing on the materials technology for high-temperature gas-cooled reactors was reviewed with emphasis on the needs identified for the NGNP. Several materials were identified for a more thorough study of their databases and behavioral features relative to the requirements ASME Boiler and Pressure Vessel Code, Section III, Division 1, Subsection NH.
A multi-year collaborative effort has been established between the Department of Energy (DOE) and the American Society of Mechanical Engineers (ASME) to address technical issues related to codes and standards applicable to the Generation IV Nuclear Energy Systems Program [1]. To date, twelve tasks have been identified that are managed through the ASME Standards Technology, LLC (ASME ST-LLC) and involve significant industry, university, and independent consultant activities. Task 11 in this effort addresses “New Materials for NH.” What is meant by a “new material” is one that currently is not permitted for construction of Class 1 components under the rules of ASME Section III, Division 1, Subsection NH. The goal of this task is to review the information available for new materials that have been identified as primary or “backup” materials and to scope out the requirements to gain approval for Subsection NH construction [2]. The low-alloy steels and ferritic/martensitic steels that are candidates for the reactor pressure vessel, piping, and similar applications are excluded from this activity, although they will be mentioned at some places in the report. Task 11 has three subtasks or parts. Part I, addressed here, requires the identification of the component temperature-time operating conditions and the selection of candidate materials. Part 2 requires a review of the available data for the new materials, the identification of any gaps in the data, and the development of a basis for a rough estimate of the cost and schedule that would lead to approval of the candidate for Subsection NH construction. Because it is likely that the operating conditions identified in Part I will imply a relative broad range of data requirements, Part II will briefly review the status of the database for a broad range of design-related materials properties. Part III requires a benchmark comparison of the candidate materials with respect to creep-rupture strength at 800°C and 100,000 hours.
IDENTIFICATION OF METALLIC COMPONENTS
AND OPERATIONAL REQUIREMENTS

Recently, the Idaho National Laboratory (INL) has documented the fundamental and operating requirements for the Next Generation Nuclear Plant (NGNP) project [3,4]. The report includes summaries of the reactor types, design power levels, gas outlet temperatures, gas inlet temperatures, and important operating parameters. The bounding conditions are sufficiently broad to encompass three “conceptual design trade studies” reported by three reactor vendor teams [5, 6, 7]. The three concepts include the Westinghouse/Pebble Bed Modular Reactor, the AREVA-Prismatic Design, and the General Atomics-Prismatic Design and are developed from the Gas Turbine-Modular Helium Reactor (GT-MHR) of General Atomics, the ATARES design of AREVA, and the Pebble Bed Modular Reactor (PBMR) Demonstration Pilot Plant of PBMR (pty) Ltd. These concepts differ in details with respect to the power generation and the specific applications such as co-generation, hydrogen production, process heat for oil sand and oil shale recovery, or other usages [8]. However, it is clear that the NGNP Bounding Condition-002 requires that the NGNP shall be capable of power levels up to 600 Mw, and that the plant be designed, licensed, and constructed for the maximum power level specific to the vendor design but in the range of 500 to 600 Mw for an operational lifetime to 500,000 hours. Bounding condition-003, however, requires that the system be capable of design, development, and certification for a lower power rating as a proof of concept.

The proposed reactor gas outlet design temperature falls in the range of 900 to 950°C, but the initial reactor island operating temperature, satisfying Bounding condition-003, may be in the 750 to 800°C range. The “optimum” reactor gas inlet temperature will be reactor specific and will fall in the range of 350 to 500°C with the lower temperature range being the most likely. Design pressure will be in the range of 7 to 9 MPa.

For the purpose of hydrogen production, the NGNP system heat source will include an intermediate loop and intermediate heat exchanger(s) (IHX) that deliver up to 60 Mwth process heat. Assuming a two-stage IHX, the high-temperature IHX will be designed for at least a five-year lifetime and the low-temperature IHX will be designed for at least a 20-year lifetime.

Components that can be replaced during the 60 year operation lifetime requirement are acceptable. It is required that the non-replaceable structural materials resist corrosion and erosion without loss of function.

The system heat source will be capable of load following, which implies thermal or mechanical load cycling of some components.

“NUREG/CR-5973 will be used as a starting point for the identification of codes and standards to be followed during conceptual design.”
In the initial planning stages, the leading materials for metallic components for the NGNP were expected to be similar to those selected for the Gas-Turbine Modular Helium Reactor (GT-MTR) and emphasis was placed on alloy 617, for the very hot components operating up to 950°C, alloy 800H for service to 750°C, and 9Cr-1Mo-V steel for the reactor pressure vessel, cross vessel, and intermediate heat exchanger vessels [9, 10]. However, designs that provided cooling to the vessels allowed lower design temperatures and alternative pressure vessel steels were considered for the GT-MTR. These included SA-533B Type B Class 1 and SA-508 Grade 3 Class 1, which were proposed for the Modular High Temperature Gas Cooled Reactor (MHTR), 2 1/4Cr-1Mo-V steel, and 3Cr-1Mo-V steel, which were developed for large process vessels in the petroleum and petrochemical industries [9, 11]. The proposed hot metallic structural components included the core barrel within the reactor pressure vessel, other reactor internal support structures covered by ASME Code Case NB, sleeves, hot ducting, and perhaps a bellows within the cross vessel, and the intermediate heat exchangers. More details regarding the performance requirements of materials for the NGNP were provided in 2004, when a materials research and development plan was identified for the Very High Temperature Reactor (VHTR) [12]. The components, expected service conditions, and primary candidate materials were tabulated in the plan for the “hot components” and are shown in Table 1. Although the limiting temperatures were not reported under the conditions column, the types of loading were provided along with several choices for primary materials. The prime materials listed in Table 1 were similar to those considered for the GT-MTR, namely alloy 617, alloy 800H, and alloy X/XR. Of these, alloy 617 and alloy X/XR were considered to be new materials. A few other “new” materials appeared, namely alloy 230, 316FR stainless steel, alloy 120, and 347H stainless steel. Almost without comment, a listing of 27 potential candidate materials was provided in the plan. That listing is shown in Table 2. In contrast to Table 1, the potential candidate materials were not linked to specific components.

In 2007, the results of a survey of operating conditions and material selections were published by Westinghouse LLC in connection with an activity to update Section III Code Case N-201 (CC N-201). The responders included PBMR, AREVA, and General Atomics. The materials mentioned for the high-temperature structural components included 316H, 321H, 347H, 316FR, and alloy 800H [13]. Alloy 617 and alloy X were not identified. It was recognized, however, that the focus of the update of CC N-201 was to bring the code case into harmony with III-NH, so two of the materials of concern were those currently in III-NH. Also, at the time of the survey, the specific requirements of the IHXs were not clearly identified.

The IHXs were critical components in the NGNP, and a number of options were considered in the preconceptual designs. Several concepts were under consideration for the GT-MHR and included tube and shell, plate-fin, and prime-surface configurations [9, 10, 12]. Stainless steel plate-fin heat exchangers for lower temperatures were analyzed by Jiang, et al. [14] and Chen et al. [15] while Sharma, Choi, and Kang examined the behavior of alloy 617 foils that could be used in compact heat exchangers operating to 900°C [16]. Pra et al. examined the potential of compact heat exchangers developed for recuperators in Brayton cycle applications [17]. Conceivably, these designs would be
useful for the low-temperature IHX. Kim, et al. looked at the use of the compact heat exchangers fabricated from alloy X for temperatures to 950°C [18]. A review of intermediate heat exchangers was undertaken by AREVA NP Inc in 2008 [19]. The emphasis by AREVA was on four concepts: Tubular Helical Coil, Plate-Stamped, Plate-Fin, and Plate-Machined. Although operation up to 1000°C was considered, the recommendation was to limit gas temperature to 900°C. The alloys that were considered included alloy 617, alloy 230, alloy X, alloy XR, and alloy 800H. Alloy 617 was the leading candidate.

Table 1. Candidate Materials Listed for Intermediate and High-Temperature Components for the Very High Temperature NGNP

<table>
<thead>
<tr>
<th>Component</th>
<th>Loading</th>
<th>Prime Materials</th>
</tr>
</thead>
<tbody>
<tr>
<td>SCS Tube</td>
<td>Thermal Stress LCF/HCF</td>
<td>316FR, Alloy 800H</td>
</tr>
<tr>
<td>Core Barrel</td>
<td>Core Weight</td>
<td>Alloy 800H, 316FR</td>
</tr>
<tr>
<td>Core Support Floor</td>
<td>Own Weight</td>
<td>Alloy 800H, 316FR</td>
</tr>
<tr>
<td>IHX Indirect</td>
<td>Thermal Transients</td>
<td>Alloy 617, 230, HR120, X, XR</td>
</tr>
<tr>
<td>Hydrogen HX</td>
<td>7 MPa, Cycles</td>
<td>Alloy 617, 230, HR120, X, XR</td>
</tr>
<tr>
<td>Hot Duct</td>
<td>Own Weight</td>
<td>Alloy 800H, 316FR</td>
</tr>
<tr>
<td>Bellows</td>
<td>Fatigue</td>
<td>Alloy 800H, 316FR</td>
</tr>
<tr>
<td>Helium Circulator</td>
<td>Fatigue, Creep Fatigue</td>
<td>316FR, Alloy 800H</td>
</tr>
<tr>
<td>Primary to Secondary Piping</td>
<td>7 MPa</td>
<td>Alloy 617, 230, HR120, X, XR</td>
</tr>
<tr>
<td>Recuperator</td>
<td></td>
<td>347 SS, 316FR</td>
</tr>
</tbody>
</table>
Table 2. Potential Candidate Materials for Intermediate and High-Temperature Metallic Components in the VHTR Concept of the NGNP Reactor

<table>
<thead>
<tr>
<th>Nominal Composition</th>
<th>UNS No</th>
<th>Name</th>
<th>Maximum Temp of Data</th>
<th>Code Status</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni-16Cr-3Fe-4.5Al-Y</td>
<td></td>
<td>Haynes 214</td>
<td>1040</td>
<td></td>
</tr>
<tr>
<td>63Ni-25Cr-9.5Fe-2.1Al</td>
<td>N06025</td>
<td>VDM 602CA</td>
<td>1200</td>
<td></td>
</tr>
<tr>
<td>Ni-25Cr-20Co-Cb-Ti-Al</td>
<td></td>
<td>Inconel 740</td>
<td>825</td>
<td></td>
</tr>
<tr>
<td>60Ni-22Cr-9M03.5Cb</td>
<td>N06625</td>
<td>Inconel 625</td>
<td>900</td>
<td>I, VIII</td>
</tr>
<tr>
<td>53Ni-22Cr-14W-Co-Fe-Mo</td>
<td>N0230</td>
<td>Haynes 230</td>
<td>1100</td>
<td>I, VIII</td>
</tr>
<tr>
<td>Ni-22Cr-9Mo-18Fe</td>
<td>N06002</td>
<td>Hastelloy X</td>
<td>1000</td>
<td>I, VIII</td>
</tr>
<tr>
<td>Ni-22Cr-9Mo-18Fe</td>
<td></td>
<td>Hastelloy XR</td>
<td>1000</td>
<td></td>
</tr>
<tr>
<td>46Ni-27Cr-23Fe-2.75Si</td>
<td>N06095</td>
<td>Nicrofer 45</td>
<td></td>
<td></td>
</tr>
<tr>
<td>45Ni-22Cr-12Co-9Mo</td>
<td>N06617</td>
<td>Inconel 617</td>
<td>1100</td>
<td>I, VIII</td>
</tr>
<tr>
<td>Ni-23Cr-6W</td>
<td></td>
<td>Inconel 618</td>
<td>1000</td>
<td></td>
</tr>
<tr>
<td>Ni-33Fe-25Cr</td>
<td>N08120</td>
<td>HR 120</td>
<td>930</td>
<td>I, VIII</td>
</tr>
<tr>
<td>35Ni-19Cr-1 1/4Si</td>
<td>N08330</td>
<td>RA 330</td>
<td></td>
<td>I, VIII</td>
</tr>
<tr>
<td>33Ni-42Fe-21Cr</td>
<td>N08810</td>
<td>Incoloy 800H</td>
<td>1100</td>
<td>I, VIII, NH</td>
</tr>
<tr>
<td>33Ni-42Fe-21Cr</td>
<td>N08811</td>
<td>Alloy 800HT</td>
<td>1100</td>
<td>I, VIII</td>
</tr>
<tr>
<td>21Ni-30Fe-22Cr-18Co-3Mo-3W</td>
<td>R30566</td>
<td>Haynes 556</td>
<td>1040</td>
<td>I, VIII</td>
</tr>
<tr>
<td>18Cr-8Ni</td>
<td>S30409</td>
<td>304H SS</td>
<td>870</td>
<td>I, VIII, NH</td>
</tr>
<tr>
<td>16Cr-12Ni-2Mo</td>
<td>S31609</td>
<td>316H SS</td>
<td>870</td>
<td>I, VIII, NH</td>
</tr>
<tr>
<td>16Cr-12Ni-2Mo</td>
<td></td>
<td>316FR</td>
<td>700</td>
<td></td>
</tr>
<tr>
<td>18Cr-10Ni-Cb</td>
<td>S34709</td>
<td>347H SS</td>
<td>870</td>
<td>I, VIII</td>
</tr>
<tr>
<td>18Cr-10Ni-Cb</td>
<td></td>
<td>347HFG SS</td>
<td>760</td>
<td>I</td>
</tr>
<tr>
<td>18Cr-9Ni-3Cu-Cb-N</td>
<td></td>
<td>Super 304</td>
<td>1000</td>
<td>I</td>
</tr>
<tr>
<td>15Cr-15Ni-6Mn-Cb-Mo-V</td>
<td>S21500</td>
<td>Esshete 1250</td>
<td>900</td>
<td></td>
</tr>
<tr>
<td>20Cr-25Ni-Cb-N</td>
<td></td>
<td>NF 709</td>
<td>1000</td>
<td>I</td>
</tr>
<tr>
<td>23Cr-11.5Ni-N-B-Ce</td>
<td></td>
<td>NAR-AH-4</td>
<td>1000</td>
<td></td>
</tr>
<tr>
<td>Ni-20Cr-Al-Ti-Y2O3</td>
<td>N07754</td>
<td>Inconel MA754</td>
<td>1100</td>
<td></td>
</tr>
<tr>
<td>Ni-30Cr-Al-Ti-Y2O3</td>
<td></td>
<td>Inconel MA754</td>
<td>1100</td>
<td></td>
</tr>
<tr>
<td>Fe-20Cr-4.5Al-Y2O3</td>
<td>S67956</td>
<td>Incoloy MA956</td>
<td>1100</td>
<td></td>
</tr>
</tbody>
</table>

In summary, the fundamental and operating requirements issued by INL for the NGNP will result in designs that will produce hot gas at 800°C or less and utilize two heat exchangers, one high-temperature and one low temperature. These designs will satisfy bounding condition 003. Quite likely, the load conditions and material selections will be similar to those identified in Table 1. However, the need to satisfy bounding condition 002 leaves open the possibility that alternative materials will continue to be of interest, should the primary materials (alloy 617 and alloy 800H) prove to be inadequate.
The behavioral features of metallic materials approved for the design and construction of components under the rules of ASME III-NH are outlined by Jetter [20]. Required knowledge of the behavioral features is linked to the seven structural failure modes that should be precluded: (1) ductile rupture from short-term loading; (2) creep-rupture from long-term loading; (3) creep-fatigue failure; (4) gross distortion due to incremental collapse and ratcheting; (5) loss of function due to excessive deformation; (6) buckling due to short-term loading; and (7) creep-buckling due to long-term loading. Given consideration of these failure modes, it is apparent that the properties required for the acceptance of a new material into ASME III-NH will include tensile, creep, stress-rupture, fatigue, and creep-fatigue interaction properties for the specific products (specifications) that will be used for construction. Because ASME III-NH requires consideration of the load-time-histogram for all load-controlled conditions, the stress allowables are time-dependent and must be applicable to the duration of the expected components life. This life can be up to 500,000 hours for NGNP components. Similarly, the criteria governing strain and deformation-controlled loading must be applicable for the duration of service. Further, ASME III-NH requires consideration of aging effects on short term properties. Fabrication and welding issues exist that must be resolved in the database supplied for code acceptance. Although environmental effects are not addressed in ASME III-NH, resistance to corrosion for the design lifetime is specifically mentioned in the NGNP bounding conditions issued by INL. Currently, environmental effects that should be considered are summarized in Appendix W of Section III, for service below the creep range, and Appendix A of Section II-D, which includes some high-temperature degradation mechanisms. The data requirements for III-NH and the status of candidate materials will be covered in more detail in Part II of Task 11. However, the German Code KTA 3221 provides an example of data requirements for alloy 800HT to 750°C and alloy 617 to 1000°C [21, 22], and a draft ASME III-NH Code Case for alloy 617 has been described by Corum and Blass [23]. Specific data requirements for alloy 617 and alloy 230 have been presented by Ren and Swindeman [24].
A BRIEF REVIEW OF DEVELOPMENT OF THE PRIMARY AND ALTERNATIVE ALLOYS CONSIDERED FOR STRUCTURAL HTGR COMPONENTS

The materials technology for the NGNP reactor is the beneficiary of decades of research on materials proposed for construction of components for high-temperature gas-cooled reactors.

In the U. S., both developmental and commercial alloys were evaluated over a span of thirty years [25-43]. The alloys were exposed to testing in various HTGR helium environments at temperatures of 800°C (1272°F) and higher. Alloy 800H, was introduced into the HTGR research programs in the early 1970s and gained experience in the Fort St Vrain reactor. Research was undertaken to qualify both alloy 800 and alloy 800H in the nuclear Code Case 1331-5. Alloy X was included in the research programs but only alloy 800H was incorporated into the nuclear code case [25-32]. Additional research was undertaken on alloy X, alloy 617, and other materials in the late 1970s and 1980s [26-43]. Alloys included alloy 556, alloy 802, alloy 618, IN-519, and alloy 214. Several of these were included in Table 2. Sufficient research on alloy 800H was undertaken by the 1990s to support the SIII-NH stress allowables for alloy 800H and research on alloy 617 permitted the development a draft code case for alloy 617 [23].

The high-temperature gas-cooled reactor research programs in Japan started in 1969 and included work on commercial alloys as well as efforts to develop new alloys with specific performance characteristics. Commercial alloys included some of those listed in Table 2 (alloy 800H, alloy X, and alloy 625). In the 1970s, an effort was undertaken to qualify alloy 617 and alloy X for use in a nuclear steel making project [44-46]. The aim was to develop a high-temperature construction code similar to ASME Code Case 1592 but extended to higher temperatures [46]. A number of tungsten-bearing developmental alloys and alternative “code” materials were investigated in Japan for HTGR service over the next decade [44, 47-59]. These developmental alloys had designations such as NSC-1, SZ, KSN, 113MA, and R4286. In the end, alloy XR became the focus for much of the effort to develop the high temperature construction code [60-62]. Modifications of alloy XR, such as alloy XR2, were undertaken [56] but only alloy XR was incorporated into the high temperature design code [61, 62].

Efforts in Europe centered on alloy 800H and alloy 617, but a number of other promising materials were evaluated in helium atmospheres [63-82]. Alternate alloys included several of those listed in Table 2. Schubert and coworkers included alloy 802, alloy IN-519, Manaurite-36X, alloy S, and alloy 86 [70]. Huchtemann looked at a high-tungsten alloy, Thermon 4972, in addition to alloy 617 [76] In the German design code KTA 3221, three grades of the 20Cr-32Ni-Fe alloy were qualified and included alloy 800HT [22]. Also, alloy 617 was covered to 1000°C. More recently, alloy 230 has been under consideration in France [81].
Conformance to the NGNP bounding conditions will require the use of currently approved materials, “new materials,” and promote the modification of existing materials or the development of alternate materials.

**Currently approved materials:** Task 6 of the ASME/DOE Gen-IV Materials Project has undertaken an effort to validate the current design allowables for the five materials currently approved for construction under the rules of ASME III-NH. The databases are being expanded and the extension of the time-dependent allowables is being investigated. These materials include 304H, 316H, 800H, 2 1/4Cr-1Mo steel, and 9Cr-1Mo-V steel. The applicable specifications and restrictions on chemistry are under review.

**Primary candidate materials:** The NGNP bounding requirement that hot non-replaceable components be designed to meet the maximum helium outlet temperature of 850°C or greater presents a significant challenge. The primary candidate material for this usage has been alloy 617. However, alloy 230 remains as a strong contender. Other candidates listed in Table 2, such as alloy X, should be included in the assessment but to a lesser degree of detail. Alloy XR remains as a primary candidate and is evaluated in a separate activity within Task 11.

**Alternate materials:** There is a long list of alternate materials. Some are listed in Table 2. This need for alternate materials is based, in part, on the performance requirements that will be identified for the two heat exchangers. The tube-and-shell concept is proven and experience has been gained with alloy XR in the Japanese HTTR with this type of heat exchanger. The same appears to be true for the German experience with alloy 617 and their version of alloy 800H, although long-time experience is lacking. The compact heat exchangers, however, require thin-section construction and specifications for most high temperature materials require grain sizes which are large relative to the section thicknesses of the compact heat exchangers. Fine-grained materials have been used in recuperators for many years, and some alloys have been specifically designed to produce good high-temperature strength as fine-grained products. Some of these materials will be identified in Part II of Task 11. Included will be modifications of 20Cr-25Ni-Nb stainless steel and 347H stainless steel for service to 700°C.
SUMMARY

The bounding conditions the New Generation Nuclear Plant (NGNP) require the use of metallic materials capable of long-term service at temperatures up to 850°C or higher.

The literature bearing on the materials technology for high-temperature gas-cooled reactors was reviewed with emphasis on the needs identified for the NGNP.

Metallic materials essential to the successful development and proof of concept for the NGNP were identified. Some were judged to be “new” materials relative to those approved for construction in ASME Boiler and Pressure Vessel Code Section III Division 1 Subsection NH.

High temperature alloys of interest were those that have been primary candidate for more than thirty years and include alloy 617, alloy 800H, and alloy X. One relatively new material, namely alloy 230, was added to the list. The relative performance and data needs for these materials will be reviewed in Part II of Task 11.

At least two materials, produced as strip products were added to the list of materials of interest. These were 20Cr-25Ni-Nb stainless steel and 347H stainless steel, which are used in primary surface recuperators.
ACKNOWLEDGMENTS

The authors acknowledge the assistance of Dr. T.-l. (Sam) Sham of the Oak Ridge National Laboratory and Dr. R. I. Jetter for assistance in gathering information on the materials data needs.
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PART II – REVIEW OF AVAILABLE DATA FOR CANDIDATE MATERIALS

and

PART III – ESTIMATE OF STRENGTH CHARACTERISTICS

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APPENDIX I: A LISTING OF APPLICABLE TEST MATRICES FROM THE
NATIONAL LABORATORIES NGNP IHX R&D PLAN WITH SUPPLEMENTARY
TESTS RECOMMENDED FROM THIS EVALUATION
ABSTRACT

Material properties required for the design and construction of components meeting the rules of ASME Section III Subsection NH (ASME III-NH) were reviewed. An overview of the data available for candidate “new materials” for the Next Generation Nuclear Plant (NGNP) was undertaken with respect to meeting the needs for incorporation of the materials into ASME III-NH. These materials included alloy 617, alloy 230, and alloy 556 for service to 800°C and above. For service below 800°C, an “enhanced strength” stainless steel typical of a new group of such steels was included. Although not a new material, alloy 800H was included in the review. The data needs identified in the National Laboratories testing plans for the NGNP were considered. In these plans, emphasis was placed on alloy 617 which is the leading candidate for the high-temperature metallic components in the NGNP for components operating above 800°C. It was found that the plans were very comprehensive and identified the data needs for both incorporation of a new alloy into ASME III-NH and the complementary database needed for the application of the Code. A comparison of the strength of several candidate alloys approved for ASME Section I or Section VIII, Division 1 construction was made and this comparison supported the selection of alloy 617 as the leading candidate on the basis of strength. With respect to compact heat exchangers, some concerns about the behavioral features of these alloys as fine-grained strip products were developed, and some comparisons were made between candidate alloys developed for high-temperature recuperators was made.
INTRODUCTION

This report provides Part II and Part III of Task 11 on the ASME/DOE General IV Reactor Materials Project to address new materials for ASME Section III, Division 1, Subsection NH (ASME III-NH) [1, 2]. What is meant by a “new material” is one that currently is not permitted for construction of Class 1 components under the rules of ASME III-NH. The goal of this task is to review the information available for new materials that have been identified as primary or “backup” materials and to scope out the requirements to gain approval for Subsection NH construction [2]. Part I of this task, reported earlier, identified the component temperature-time operating conditions and the selection of candidate materials based largely on the information supplied or referenced in a summary report of bounding requirements for the NGNP Demonstration Plant [2, 3]. Part II requires a review of the available data for these new materials, the identification of any gaps in the data, and the development of a basis for a rough estimate of the cost and schedule that would lead to approval of the candidate for ASME III-NH construction. Because it is likely that the operating conditions identified in Part I will imply a relative broad range of data requirements, Part II will briefly review the status of the database for a broad range of design-related materials properties. Part III requires a benchmark comparison of the candidate materials with respect to creep-rupture strength at 800°C and 100,000 hours. However, the bounding conditions for the NGNP require a design that would produce outlet gas at temperatures in the range of 750 to 800°C for times that could extend to 500,000 h (60 years at 95% availability), so the Part III includes consideration of materials that would serve such intent.

The low-alloy steels and ferritic/martensitic steels which are candidates for the reactor pressure vessel, piping, and similar pressure boundary components in the NGNP are excluded from this activity. One leading austenitic alloy candidate for high-temperature components, namely the Japanese-developed alloy XR, has been excluded from this evaluation but is covered in a companion report [4].
NEEDED PROPERTIES FOR CONSTRUCTION OF SECTION III SUBSECTION NH COMPONENTS FOR ELEVATED TEMPERATURE SERVICE

Current requirements for ASME III-NH:

Materials data requirements for ASME III-NH have been outlined by Jetter [5], and those specific to the Very High Temperature Reactor (VHTR) Program have been summarized in a program plan by Corwin et al. [6]. With respect to the leading candidate materials for use in high-temperature metallic components, namely alloy 617, the ASME III-NH requirements were reviewed in general by Ren and Swindeman [7] and provided in detail in the NGNP IHX Materials R&D Plan developed by Wright at the Idaho National Laboratory [8]. These data requirements include:

- Cold work and recovery data to establish cold forming limits provided in NH-4212 and Fig. NH-4212-1;

- Tensile ultimate strength ($S_u$) and yield strength ($S_y$) which extend the ASME Section II-D Table Y-1 and Table U into the time-dependent temperature range and are needed to produced time-independent stress-intensity values ($S_m$) in Table NH-3225-1 and Table I-14.5;

- Tensile reduction factors for aging as in Table NH-3225-x;

- Creep and rupture data for setting time-dependent allowable stress intensity values in Table I-14.3 for $S_{mot}$ and Table I-14.4 for $S_i$;

- Minimum stress-to-rupture values, $S_r$, in Table I-14.6;

- Stress-rupture factors for weldments in Table I-14.10;

- Strain-controlled fatigue data in Fig. T-1420-1x;

- Creep-fatigue damage data in Fig. T-1420-2;

- Tensile curves and creep data for external pressure time-temperature limit curves in Fig. T-1522-x;

- and tensile curves and creep data for isochronous stress-strain curves in Fig. T-1800-x.

In addition to the data specific to ASME III-NH, supplementary data in ASME Section II-D need to be supplied if not currently provided. These data include such items as the maximum allowable stress intensity, $S_o$, in Table I-14.2 which is based on Table 1A or 1B in ASME Section II-D, physical properties such as thermal expansion (Table TE-x),
thermal conductivity (Table TCD-x), thermal diffusivity (Table TCD-x), modulus of
elasticity (Table TM-x), Poisson’s ratio (Table PRD), and density (Table PRD).

Other considerations regarding current and future data needs for ASME III-NH:

Further, there are additional considerations and needed data because material properties
serve two distinct functions in the design and construction of engineering equipment. The
first of these is material selection. The second is engineering design.

Material selection tends to lead the process. Consequently the properties most commonly
considered, particularly in the early planning stages of development, are those which can
be most easily used to rank one material against another. Mechanical design is therefore
often left in the position of making best use of the properties that have been determined
with other objectives in mind.

This has not been a totally unsatisfactory system, at least until relatively recently, because
common sense dictates that material selection criteria should at least approximate the
duty cycle of the component. Fundamental properties, such as Young’s Modulus, which
defines the elastic stiffness, or yield and ultimate strengths, or % elongation as a measure
of the ability to accommodate plastic deformation without cracking, all appear to serve
both purposes reasonably well, and this situation has been fairly accurate up to the
present.

Major changes have occurred in the practice of engineering design in recent years. While
computers have existed as tools for several decades, the power and versatility provided
by the modern generation of hard- and software have completely revised the practice of
mechanics as it applies to engineering design. Approximate methods were once a
necessary requirement to cope with complex problems as recently as a decade ago, due to
limitations of computational power, and these methods generally found the modest range
and type of material properties collected routinely to be adequate fuel for the purpose of
analysis.

This environment has been displaced by another of virtually limitless capability in which
precise models of complex geometries, together with detailed constitutive models of
material behavior can, in principle, be built and analyzed with comparative ease.

The qualification, “in principle”, is derived largely from the fact that, while the analytical
tools are readily available, the requisite material data of a type and degree of detail
necessary to generate reliable quantitative solutions with them is not. Unfortunately, the
properties once considered sufficient to describe material behavior for practical purposes,
while they perhaps continue to be useful parameters for ranking materials, are no longer
the ideal measures for use in the current computational environment, but they continue to
artificially constrain the way structural computations are performed.
As an example of this process, consider the concept “yield stress.” This is no more than a mathematical construct, originally derived from a rather eccentric yielding behavior observed in some plain carbon steels, which was generalized to define the point of transition from linear elastic deformation to irreversible plastic deformation for materials in general. Virtually no practical engineering alloy exhibits a true yield stress. A convenient substitute has long been accepted, consisting of the stress corresponding to a defined offset, usually 0.2%, from the initial linear elastic part of the stress-strain curve. Deformation beyond the “yield stress” is commonly approximated by a linear section with a reduced slope, referred to as “linear strain hardening” or, when the slope is zero, “perfect plasticity”.

These concepts were, and still are crucial to our conceptual understanding of structural behavior. They are also very convenient measures with which to characterize a typical stress/strain curve with a minimum of parameters. However, they are not sufficient to provide the input necessary to drive the kind of detailed structural analysis of which designers are capable today, and for which there is increasing pressure to implement for reasons of both improved economy and safety.

Many modern – and mature – structures operate under cyclic conditions in strain ranges at, or just below the “yield stress”. According to the bilinear idealization, the plastic strain at a stress just below yield, is zero, whereas we know it, in reality, to be 0.2%. In terms of some damage mechanisms, this is the difference between zero and significant damage.

As long as computational limitations dictated the need for simple representations of stress/strain behavior such as the bilinear curve, both yield stress and strain hardening modulus remained useful tools for the analyst. However, with a typical modern FE program, it is no more difficult to enter the actual stress/strain curve than to enter a bilinear idealization, and it is a whole lot more accurate. Due to the smooth transition it provides from linear elasticity to plastic yield there may actually be computational advantages to using the “real thing”, because of improved convergence of the iterative routine invariably used to solve problems of this nature.

However, if the only information available to the analyst is the elastic modulus, the “yield stress” and possibly, but not always, the strain hardening modulus, these are the parameters he will be constrained to work with, thus deliberately reducing the precision of his model in order to accommodate a data filtering step which artificially removed some of the information, the true shape of the curve, of which he could have made some good use.

One can argue that, from the viewpoint of the structural analyst living in today’s computational environment, a more accurate and more meaningful description of the stress/strain curve would be something like the Osgood-Ramberg formulation,

\[ \varepsilon_p = \kappa \sigma^n \]  

(1)
where \( \varepsilon_p \) is the inelastic strain, \( \sigma \) is stress, and \( \kappa \) and \( n \) are constants. Not only is this equation more accurate over a larger strain range than the bilinear curve, it still only requires two parameters to define it fully. Furthermore, no extra work is involved in fitting the equation to the data since, in both cases, the complete stress/strain curve is required. It is just a question of what one does with it to characterize the curve.

When considering elevated temperature applications, where rate effects in the tensile test become important, the artificial definition of a “yield stress” by an arbitrary offset can produce wide variations in the perceived “material property” which may be merely a small variation in the shape of the stress/strain curve, whereas a fit of the Osgood-Ramberg type, should be expected to provide greater consistency, because the whole curve is used in determining the two constants, \( n \) and \( \kappa \).

Similar criticisms can be leveled at other venerable measures of material behavior, such as “ultimate tensile strength”, and “percent elongation”. These are both well established parameters for characterization of materials and it may come as a surprise to discover that neither of them has any direct value as the input to a detailed inelastic analysis.

Certainly, by some ingenious manipulation, it is possible to work one’s way back from the yield and ultimate tensile strengths of a material to infer its probable underlying true stress/strain relationship – providing you either know, or can deduce the strain corresponding to the UTS (about 60% on average of the % elongation in a standard proportioned tensile test bar by the way).

Except as a mainly qualitative measure of material ductility however, how much value is the % elongation? It may well have had value in the early days of material science when nobody had any better idea of how to characterize plasticity, especially when there was no accompanying science of solid mechanics to make use of the property even if it were measured correctly. The % elongation is an empirical measure of plastic deformation which is comprised in part of uniform plastic strain over most of the length of a specimen of standardized proportions, together with an additional increment of deformation due to a highly localized instability know as a “neck”, the failure of which is the result of void formation in the neck accelerated by hydrostatic stress build up caused by the concave formation of the neck itself. This is a very complicated situation and requires a considerable amount of advanced FE modeling to reproduce it theoretically.

As long as % elongation remains a semi-qualitative indication of adequate ductility for practical purposes, there is no strong incentive to look for an alternative. However, times have changed, structural analysis has expanded in scope to include as one of its objectives, the evaluation for the potential of cracking or rupture in complex geometries and, for this purpose, a more rational measure of ductility is required. This issue is particularly acute in elevated temperature applications, where it is feared that ductility becomes more critical as time and temperature increase.
Summary of materials properties needs for modern design analysis:

Transition should be made from presentation of data in graphical form to parametric equation format to facilitate use in design computations.

Wherever possible material data should be provided in as transparent format as possible, i.e. with the minimum of data post processing. The reason for this is that design data needs are expanding in scope and it is becoming increasingly difficult to anticipate analytical needs. The result of post processing into preconceived formats is often to remove essential material characteristics which might otherwise be useful in structural assessment.

One specific post processing practice to which serious consideration should be given to its elimination is the practice of publishing “lower bound” or “minimum” material properties with no clear indication of the process used in setting the bounds. For preference in conducting complex structural analyses it is often preferable to carry out the computation with mean or nominal data and consider the effects of property variation as a separate entity. The need for this transparency stems both from the increasing requirements on design to include trend and/or Probabilistic Risk Analysis (PRA) evaluations, and the responsibility placed on the designer, by the ASME Code (see Foreword to BPV Code, all sections) to ensure that any computations are accurate and appropriate.

**Time Independent Tensile Data**

- **Stress/strain Curve** - Osgood-Ramberg or equivalent representation of full tensile stress/strain curve

- **Nominal Ductility Measure** - Strain at peak of engineering stress/strain curve (at the UTS) – in austenitic alloys, this is called the uniform strain and defines ductility at onset of unstable necking.

- **Rate dependency** - At elevated temperature (creep range) – provide stress/strain data for a range of strain rates, e.g. a minimum of 3 rates spaced at factors of 100 on rate.

- **Multiaxial Ductility Criteria (Notched Bar Testing)** – This should be collected on a routine basis for the purpose of establishing multiaxial ductility criteria and trends.

- **Cyclic stress/strain data** – It is well known that cyclic stress/strain behavior can differ radically from the monotonic stress/strain curve. It has long been known that the cyclic curve is the appropriate input for carrying out fatigue calculations, for instance. Depending on the material and its initial condition hardening or softening can occur under cyclic conditions. Since it is suspected that short term strength is related in some way to long term creep resistance, it is important to know cyclic properties as well as the more customary monotonic ones, in order to
make an accurate forecast of structural performance. While the acquisition of cyclic stress/strain data is recognized as being a tedious and possibly costly operation, some effort needs to be made to fill this void that exists for most materials. The Masing model approximation model frequently used in ground vehicle assessments, and once used in Nuclear Code Case N-47, provides some direction on how to provide approximate information, but if it is to be used, work needs to be done to incorporate the effects of cyclic softening and hardening in a quantitative manner.

**Time-Dependent Tensile Data**

*Creep Deformation I* - Reporting of creep deformation should include, at minimum, both the minimum creep rate (mcr) and the total primary creep accumulated in reaching the mcr. Preferably the latter should be provided as the intercept on the strain axis at time zero formed by back extrapolating the secondary portion of the curve.

*Creep Deformation II (Tertiary Creep)* – Some representation of the complete creep curve, up to rupture, should be provided. A parametric representation of the secondary and tertiary creep phases, such as is captured by the Omega model developed by the Materials Properties Council would be an acceptable procedure. At minimum the mcr, time-to-rupture, and the ratio of strain-at-rupture to the Monkman-Grant strain (referred to as the “λ” factor) would be sufficient to construct a useable working material model for design purposes.

*Creep Deformation III (Isochronous Stress/strain curves)* - Since design analysis covers a range of conditions, there is merit in providing material data in alternative formats that best suit the application. One such is the cross plotting of creep data into isochronous stress/strain curves. This could either be done at source, or published as such directly in the Code or, if a parametric form of representation is adopted, the cross plotting can be done as needed by the analyst according to a set of rules provided in the Code. The latter approach is utilized, for instance in API 579. The advantage of direct isochronous curve presentation is that this format can provide total strain v. stress, thereby eliminating potential errors incurred by factorizing material data first into elastic, plastic and creep strain components, and then recombining them during analysis. Issues of rate dependency that emerge in some materials, such as alloy 617 at very high temperature, can be effectively bypassed by this procedure.

*Creep Deformation IV (Relaxation)* – Creep relaxation features prominently in many critical component assessment procedures, notably creep/fatigue damage calculations. The only analytical tool available to deal with this problem is indirect computation of the relaxation during hold periods from constant stress
creep data. While this process is correct in principle, insufficient knowledge of the constitutive behavior underlying creep exists to enable this relaxation calculation to be carried out consistently and accurately in all circumstances. Therefore, there is a case for offering, as part of the material database, some independently generated creep relaxation data, to be used either directly, or to provide a benchmark for the validation of computational estimates of relaxation. A segment of the existing code in all Sections addressing elevated temperature operation, where relaxation data would be valuable, if not essential, to accurate prediction, is high temperature bolting, especially as it applies to flanged joints.

Failure Criteria – Failure in a tensile creep test, the most common method of compiling creep data, can occur in a variety of ways. The most readily perceived, but not the only ones, are ductile failure by time-dependent extension and necking leading to what amounts to a ductile instability, strain induced softening leading to tertiary creep without observable internal “creep damage”, and true damage due to void initiation and growth. Currently ASME criteria lump all aspects of this failure event into a single allowable stress. Due to the complex nature of stress redistribution in redundant structures, creep failures by these various mechanisms do not translate similarly to the complex multiaxial environment seen by a component in service. Therefore, creep failure criteria obtained from simple tensile tests should be reported separately, since their application in component design may well lead to radically different answers.

BENCHMARK COMPARISON OF CANDIDATE MATERIALS WITH RESPECT TO CREEP-RUPTURE STRENGTH

Part III of Task 11, which calls for a comparison of candidate materials with respect to the 100,000-hour strength at 800°C, is covered here. If one limits the temperature of the outlet gas to the temperature range of 750 to 800°C, the number of candidate alloys greatly increases. Only a few have been selected for the comparison required in Part III and these are included in Figure 1, which plots the ASME Section II-D Table 1A and 1B stress allowables against temperature for the range of 725 to 825°C. At 800°C, alloy 230 is slightly stronger than alloy 617, followed closely by alloy 625. Alloy 556 is significantly weaker but maintains a strength advantage over alloy 120 and alloy X. The NF-709 represents a class of enhanced strength stainless steel tubing alloys and is approved for Section I construction in Code Case 2582. It has been rolled to sheet and foil and has undergone testing to 750°C in those product forms. Alloy 811 (800HT) and alloy 810 (800H) are significantly weaker than the other candidates. Alloy XR is not included in the comparison nor is 316H stainless steel or 316L(N) stainless steel. Likewise, there are several other alloys intended for service around 800°C that have been excluded for one reason or another but the work undertaken on the candidates produced many of the data needed to establish these materials for high temperature service.
Figure 1. Comparison of the strength based on 100,000 hours for alloys intended for service at temperatures around 800°C.

REVIEW OF AVAILABLE DATA FOR NEW MATERIALS RELATIVE TO THE NEEDS FOR INCORPORATION INTO ASME III-NH

In Part I of Task 11, a review was undertaken of the research and development (R&D) programs directed toward high-temperature metallic materials for gas-cooled reactor components [2]. In Part II, data produced in these and other programs will be identified relative to the needed data identified above.

Cold work effects:

With respect to cold work and recovery data to construct permissible time/temperature service conditions for cold worked materials, there are data for most of the candidate alloys. An extensive research undertaking was reported by Sessions [9] in 1976, which partly formed the basis for the alloy 800H curve in ASME III-NH Figure 4212-1. Bassford and Hosier reported the effect of cold work in both alloy 800H and 617 in 1984 [10]. Annealing temperatures ranged from 760 to 1200°C, but annealing times were short (60 minutes maximum). Klarstrom [11] examined the recrystallization characteristics of alloy X, alloy 230, and alloy 625 after cold working from 10 to 50%. Again, the annealing temperatures were very high (above 1000°C) and times were short. However, the work is of value is establishing the optimum conditions to produce a target
grain size. More recently, the recrystallization behavior of alloy 230 and a restricted chemistry version of alloy 617 (CCA 617) has been explored by Shingledecker [12] and Mohn and Stanko [13] in connection with materials for use in the ultrasupercritical (USC) boiler. Mohn and Stanko introduced strain levels of 15, 20, 30, and 40% in heavy wall tubing. Alloy 230, for example, was exposed at temperatures ranging from 816 to 899°C and for times up to 1000 hours. A Larson-Miller parametric (LMP) approach was taken to correlate the data related to the initiation of recrystallization to strain, temperature, and time. Thus:

$$\varepsilon_{Rx} = a_0 \text{LMP}^2 + a_1 \text{LMP} + a_3,$$  \hspace{1cm} (2)

and

$$\text{LMP} = T_k \left(C + \log t\right),$$

Where: \(\varepsilon_{Rx}\) is the on-set of recrystallization strain, \(T_k\) is temperature in Kelvin, \(t\) is time in hours,

\[
a_0 = 7.3308 \\
a_1 = -374.52 \\
a_3 = 4797.9 \\
C = 19.7.
\]

In the work of Shingledecker, some indications of recrystallization was found after 8,000 h at 775°C in the 33% cold worked region but there was no evidence of recrystallization in the tube that was cold worked 20% in forming the bend.

**Tensile properties:**

Tensile data are required when a data package is submitted to ASME to obtain approval for a Code Case covering elevated temperature construction. Generally, the yield and ultimate strengths for at least three commercial heats are included. Often, the tensile databases are quite large. This is the case for most of the candidate materials mentioned above. The tensile and yield data are used to develop trend curves which anchor the Y-1 and U tables in ASME Section II-D once the minimum yield and ultimate strengths are specified. Trend curves, Y-1 values, and U values exist for all of the candidate alloys of interest. However, in ASME II-D the values do not extend to temperatures where the yield and ultimate strength become somewhat rate dependent. Moreover, the specific strain rate of testing may vary from one data set to another. And further, the Y-1 and U values do not represent minimum or average values in any statistical sense. Nevertheless, the \(S_y\) and \(S_u\) values provided by ASME III-NH in Tables NH-3225-1 and Table I-14.5 are intended to be consistent with the Y-1 and U values at lower temperatures (say, up to 538°C) but are extended to the maximum temperature covered by the ASME III-NH. In Figures 2 and 3, typical yield and ultimate strength data are plotted against temperature and it may be seen that data for most materials extend to 980°C. For these materials, the trend curves may have to be recovered from the files held by ASME or MPC Inc. Such an effort was undertaken for alloy 800H, limited to 760°C in the figures, but with Y-1 and U values available to 900°C [14].
Figure 2. Typical yield strength versus temperature for several candidate alloys.

Figure 3. Typical ultimate strength versus temperature for several candidate alloys.
Tensile reduction factors:

The tensile and yield strength reduction factors provided in ASME III-NH Tables NH-NH-3225-2, NH-3225-3A, and NH-3225-3B are based on estimates of the aging effects spanning the temperature-time coverage of the Code. Such data were produced for the five materials currently covered by ASME III-NH and for several of the candidate alloys considered here. Typically, aging studies included the development of Temperature-Time-Precipitation (TTP) diagrams, hardness studies, toughness studies (Charpy-V impact energy), and tensile properties. The TTP diagrams have been developed for alloy 625 [15], alloy 617 [16, 17], alloy 800H [18] and NF-709 [19] for times to beyond 10,000 hours, but few studies have been extended beyond 100,000 hours. These TTP diagrams have been modified many times over to accommodate compositional and thermomechanical processing factors. An example of a revised TTP diagram for alloy 617 is shown in Figure 4 [17]. Studies of the effect of aging on hardness, Charpy-V impact energy, and tensile properties of candidate alloys are far too numerous to be covered here. However, review papers and reports are available [20-26]. Depending on the material and aging conditions, both increases and decreases in yield strength have been reported and aging times approaching 100,000 hours have been achieved. Alloys in which Ni$_3$Al or Ni$_3$Nb precipitates form generally show an increase in strength for a time but overaging or re-solution of precipitates results in loss of strength. Further, massive Laves phase is known to form in some tungsten-bearing alloys after long times and the appearance of this phase results in the loss of solid-solution strengthening and considerable embrittlement. A more detailed study of the aging effects will be needed to produce a phenomenological model that can be used to predict tensile reduction factors for times to 500,000 or 600,000 hours. With the power of modern computational thermodynamics, it may be possible to predict the very long-time stable microstructures and from this knowledge infer the stabilized strength values.

Figure 4. Temperature-Time-Precipitation (TTP) diagram for alloy 617 by Wu et al. [17]. Long-time aging performed at ORNL by McCoy [21].
Creep and stress-rupture:

Just as for tensile properties, creep and stress-rupture data packages are required for the assignment of stress allowables for ASME Code Cases and incorporation of allowables into ASME Section II-D when temperature extends into the creep range. However, none of the candidate alloys mentioned above, with the exception of alloy 800H, are permitted for construction under the rules of ASME III-NH. All are permitted for construction under the rules of ASME Section I, Section VIII, Division 1, or both. The criteria for setting the stresses for the latter two sections, specified in ASME Section II-D Appendix 1, differ from the criteria in ASME III-NH specified in NH-3221 which require consideration of 1% total strain creep data and tertiary creep data for setting the allowable in Table I-14.3 and Table I-14.4. The status of the data for alloy 800H relative to the 1% and tertiary creep data, needed to extend coverage in ASME III-NH to 500,000 or 600,000 hours and for temperatures to at least 850°C, has been reviewed several times and recommendations for further testing have been made [7, 27, 28]. For the most part, a modest experimental effort is needed [7] along with a more vigorous evaluation of existing data. With respect to alloy 617, a draft Code Case for alloy 617 was prepared some years back and this Code Case was based on a creep model developed from a limited database [29]. Somewhat encouraging is the fact that sufficient creep data exist to permit both alloy 800HT (a restricted chemistry version of alloy 800H) and alloy 617 to be introduced into the German high-temperature Code Case KTA 3221.1 [30, 31].

With respect to stress-rupture, the databases for several candidate alloys were reviewed by ASME Section II a few years ago and the stress allowables based on average rupture strength in ASME Section II-D were revised to incorporate a factor, $f_{avg}$, on the average strength to produce rupture in 100,000 hours that was equal to or less than 0.67. The $f_{avg}$ was developed from the work of Marriott [32] and Prager [33] and was applied to the average rupture strength and not to the minimum strength and only for stress allowable values at 815°C (1500°F) and above.

Generally, the stress-rupture behavior of the candidate alloys is well-represented by a temperature-time parameter introduced by Larson and Miller (LMP):

$$\text{LMP} = T (C + \log tr)$$  \hfill (3)

Where $T$ is absolute temperature, $tr$ is rupture life, and $C$ is a parametric constant. The LMP is a function of stress alone which permits a plot of stress (or log stress) versus LMP to collapse the stress-rupture data and be used for purposes of extrapolating times within the limits of the stress function. Plots of stress versus the LMP are shown for several candidate alloys in Figure 5 through Figure 10. The values for the parametric constants, $C$, have been taken from Prager [33] and represent correlations for data at the upper temperature range of data. The arrows in the figures for the nickel-base alloys represent the LMP values for 600,000 hours at 800 and 900°C. It is clear that there is a dearth of data on which to base stress allowables at 600,000 hours and 900°C, especially in view of the fact that the ASME III-NH criterion is based on minimum strength, not the average strength, and such data are needed for Table I-14.10.
Figure 5. Stress versus the Larson-Miller parameter for rupture of alloy 617. (Arrows show the Larson-Miller parameter values at 600,000 hours.)

Figure 6. Stress versus the Larson-Miller parameter for rupture of alloy 230.
Figure 7. Stress versus the Larson-Miller parameter for rupture of alloy X.

Figure 8. Stress versus the Larson-Miller parameter for rupture of alloy 556.
Figure 9. Stress versus the Larson-Miller parameter for rupture of NF 709

Figure 10. Stress versus the Larson-Miller parameter for rupture of alloy 800H.
Tensile stress-strain curves:

Although the databases for tensile properties and stress-rupture of the candidate alloys are quite substantial, tensile stress-strain curves and creep curves for the candidate alloys are not abundant. Tensile curves to a few percent tensile strain are needed for the construction of external pressure charts in ASME II-D and for external pressure time-temperature limit curves in ASME III-NH Figure T-1522-x. They are also needed for construction of the “hot tensile curves” which anchor the isochronous stress-strain curves in ASME III-NH Figure T-1800. Such curves need to be consistent with the Y-1 or $S_y$ values, after consideration of whether minimum or average values are needed. Good records of the original hot tensile curves are available in the Code subcommittee minutes for the five materials currently approved for ASME III-NH, and the development of the curves for alloy 800H was reviewed recently [14]. There are concerns in collecting and representing the curves for the nickel-base alloys. The first concern is the specification of the strain-rate for testing at and above 700°C. Nickel-base alloys often exhibit an increase in yield strength near this temperature due to strain-aging and the phenomenon tends to be strain-rate dependent. Such behavior creates a problem, since the “rules” for establishing Y-1 or $S_y$ values do not permit an increase in the values with increasing temperature. The standard strain-rate for tensile testing is around 0.005/minute, which is sufficiently slow to produce strain-aging effects at the temperatures of interest. A second concern with respect to the strain-rate is the possibility of creep effects. The separation of time-independent plasticity and time-dependent creep is difficult at the standard testing strain-rate above 700°C, hence the tensile flow curve may manifest a creep effect. In the draft code case for alloy 617, a unified constitutive model was used to represent the stress-strain behavior [29]. This model produced a strain-rate dependent yield curve that captured to some extent an inelastic flow phenomenon in which a peak yield stress occurred at a low strain and the subsequent flow stress dropped to a lower flow stress that was maintained at a level that depended on the imposed strain rate. The behavior is typical of several of the nickel-base alloys. See Figure 11 for curves representing alloy 230 at 871°C and above [34]. At 760°C (not shown) a typical yield curve is observed with no yield point phenomenon. The presence or absence of the yield point phenomenon depends on the thermo-mechanical history. In the German code, however, a power law function identical to equation (1) above was used to represent the stress versus plastic strain behavior of alloy 617 for all temperatures [30]. No strain-rate effect was included. Of course, it is possible to represent the high strain-rate behavior by a power law and capture the decreasing flow stress with decreasing strain-rate by combining the tensile hardening with a creep function. With such an approach, it is necessary to define the strain-rate that has no time-dependent effect. The available databases of the candidate alloys are too meager to establish this condition, so some tensile testing to establish the strain rate effect on the flow stress is needed.
Creep strain versus time data:

Creep data in various forms are available for all of the candidate nickel-base alloys. Vendor data for alloy 230, alloy 556, and alloy X have been made available in the form of minimum creep rate (mcr) and times to specific strains up to a few percent. Copies of hand plotted curves have been available for Code committee work for alloy 230, alloy X, and alloy 800H. Papers written to describe various aspects of creep in the candidate alloys are too numerous to be covered here in any detail, however, papers and reports that develop creep models are of value. A summary of sources for alloy 617 creep data was provided by Wright in the National Laboratories NGNP IHX Materials R&D Plan [8], and a review of creep models for alloy 617 provided several sources that could be used to re-generate creep curves for this alloy [35].

Some features of creep in the nickel-base alloys proposed for construction of nuclear components were recently discussed [4]. What emerged from this review was that there were several issues that needed to be resolved and additional testing could be of value in this respect. Two issues are derived from the criteria for setting allowable stresses: one based on the stress to produce 1% total strain in a specific time and the other based on the 80% of the minimum stress to initiate tertiary creep in a specific time.

For the conditions where most creep data are available, 1% strain in nickel-base alloys may comprise some primary creep which depends on a number of metallurgical factors and is highly variable from one heat (or heat treated condition) to the next. Alloy 617 exhibits such sensitivity, as illustrated in Figure 12 for the creep of two heats at 800°C and 100 MPa. Heat A shows a more-or-less classical behavior with the primary creep
rate slowly diminishing with time while heat B shows tertiary creep with creep rate increasing throughout the life. Alloy X and alloy 800H have shown similar creep variability. The current ASME III-NH criterion based on 1% creep uses the average strength rather than the minimum strength, so the penalty on stress is not as severe as when it is based on minimum strength for 1% creep. Nonetheless, a large variability complicates the development of creep models needed from producing isochronous stress-strain curves in the low strain region. Thermo-mechanical studies may offer insight with respect to methods to minimize the variability of primary creep behavior. A simple stabilizing anneal may provide the answer [35].

Tertiary creep in alloy 800H has been a concern since the material was first proposed for use in gas-cooled reactors in the early 1970s. As mentioned above, consideration of tertiary creep in materials behavior modeling for component analysis is the essential feature of the Omega approach utilized in API 579 [36]. The Omega function is only one of several models for continuum damage mechanics (CDM) that capture tertiary creep behavior, but it has been shown to represent ferritic steels quite well. The appearance of some primary creep in solid-solution nickel-base alloys such as alloy 230 (Figure 13), alloy 556 (Figure 14), and in alloy 617 (Heat A in Figure 12) and alloy 800H after a stabilizing heat treatments suggest that consideration of the primary creep component is needed as an add-on to the Omega model. In this respect, other representations of the full creep curve, such as those of Evans and Wilshire [37] and Dyson [38], may be better suited to accommodate the creep behavior of the nickel-base alloys.

![Figure 12. Two creep curves for alloy 617 showing the variability in primary creep.](image)
Figure 13. Creep curves for alloy 230 at 871°C.

Figure 14. Creep curves for alloy 556 at three temperatures.
Relaxation data:

Relaxation has been examined for most of the candidate nickel-base alloys. Typically, relaxation “runs” last for 100 hours, or so, and in some studies repetitive runs are employed to examine the high-temperature creep hardening or softening behavior. Of course, relaxation can be calculated from a creep model with some assumption as to the creep-hardening mechanism and this has been demonstrated for some materials. Examples of materials included in such relaxation studies are alloy 617 [39], alloy X [40], alloy 556 [41], and alloy 800H [32]. Generally, test results revealed satisfactory prediction of behavior, with a tendency for some hardening for repetitive relaxation runs at lower temperature and stabilized relaxation rates at higher temperatures. The long-time relaxation data produced by NIMS for alloy 800H [42] was of interest because of a slight increase in stress at long time was consistently observed. Possibly, this increase in stress reflects a precipitation or re-solution process that produces a slight volume contraction.

Most long-time relaxation data represent monotonic deformation, and the calculation of the relaxation from secondary stresses produced in a component from cyclic operation that experiences cyclic plastic strains is another matter. It is unlikely that there will be an uninterrupted period of service in a component that would result in relaxation over tens of thousands of hours. It is possible, however, that cycling, when it occurs, could cause a relatively brief change in the elastic stress and a return to the pre-interruption relaxation condition after the transient. This situation would be more likely at “lower” temperatures where thermal recovery during unloading is not rapid. In ASME III-NH, a procedure is outlined that estimates relaxation from the isochronous curves which are based on monotonic deformation. The need to validate the procedure is important. Although creep-fatigue testing with hold times at the strain limits provide some data on the effect of cyclic strain on relaxation rates, the hold times are generally short, say 10 hours or less, and the validation of the prediction of relaxation from usage of the isochronous curves less certain. Figure 15 shows the third run of a relaxation test on alloy 556 at 871°C for 90 hours. (The first two runs were shorter and reflected the influence of a more rapid primary creep component which may be exhausted with repetitive testing.) Early relaxation is rapid at 871°C and in only 100 hours the stress decreases to a value that corresponds to a creep rate near 10^{-5}/h. Included in Figure 15 is a “damage” curve which corresponds to the estimated creep damage fraction during the relaxation hold.

Figure 16 shows the first cycle of a C-F test at 871°C and 0.62% strain range with a hold-time period at the tensile limit. The starting stress at the tensile strain limit will depend on the strain rate when the temperature is in the range where the flow stress is strain-rate dependent. In this case where the loading strain-rate was 0.001/s, the starting stress was high, relaxation from the flow stress was rapid, and values quickly approached the long-time allowable stress. The data are compared to calculation from a model based on a simple Ramberg-Osgood law for plasticity and a simple Norton-Bailey law for creep which accommodates creep during slow loading rates and the relaxation during the hold period [41]. The model does a good job for the first loading but the predictability with continued cycling will depend on the validity of the hardening rules. The relaxation
behavior above 800°C may be more complicated for alloy 617 than alloy 230, alloy X, or alloy 556 because the microstructure of alloy 617 appears to be somewhat dependent on the stress direction with grain boundary precipitates orienting on grain boundaries parallel to the applied stress [43]. Further, the issue of diffusional creep should be considered when estimating relaxation at very low stresses and high temperature.

Figure 15. Relaxation behavior for alloy 556 near 871°C.

Figure 16. Start of a 0.5-hour relaxation-hold C-F test on alloy 556 at 871°C and 0.62% strain range.
Strain-controlled fatigue data:

Wright summarized the continuous cycling fatigue data available for alloy 617 and alloy 800H [8]. The database covers a sufficient range of temperatures and strain ranges to characterize the alloys and develop the curves in ASME III-NH Figure T-1420-x. In additions to alloy 617 [27, 44, 45], low-cycle fatigue data are also available for alloy 230 [46], alloy X [45], and alloy 556 [47], to cite a few. A comparison of typical strain-range versus fatigue life (LCF curves) for the nickel-base alloys at 871°C is shown in Figure 17. In all cases the data were produced using a triangular strain-time cycle with strain rates in the range of 0.001 to 0.004/s. The 0.004/s rate is considered to be high enough to avoid creep effects at temperatures below 800°C and is recommended in ASME III-NH. A higher rate above 800°C may be needed to avoid rate effects. The comparison shows that alloy 230 data are at the high side of the LCF scatter and alloy 617 are at the low side of the scatter. The databases for alloy 617 and alloy X include testing of aged materials and testing in helium.

Not all of the literature papers and reports provide cyclic hardening/softening data. What is needed is representation of the stress range versus cycles at constant strain range. Also, the cyclic stress-strain hardening curve (or stress amplitude versus strain amplitude) is needed. Comparison of the cyclic hardening curves for 5% of life, half life, and the cyclically stabilized condition with the hot-tensile curve that anchors the isochronous curves is needed to determine the factor, k, used in ASME III-NH Appendix T for fatigue damage analysis.

Figure 17. Comparison of continuous cycling low-cycle fatigue data for some nickel base alloys.
There are some issues with respect to the choice of the reference condition. The fatigue curves in Appendix T are developed from a fit to the raw data using the minimum life developed from a factor of two on strain range or a factor of twenty on fatigue life. This restriction amounts to 5% or less of life, yet the cyclic hardening curve are based on either half-life data or “stabilization” of hardening. At the temperatures of interest for NGNP, however, stabilization in the nickel-base and similar alloys comes early in life. As an example, data for alloy 556 are shown in Figure 18 [47]. These hardening curves are typical of the solid-solution nickel-base alloys, as well. Here, it may be seen that the hardening that occurs beyond 5% of life depends on the temperature and the strain range. Lower temperatures and intermediate strain ranges tend to produce a lower rate of hardening. More information is needed in regard to the cyclic hardening characteristics of alloy 617 if the isochronous curves are to be used to anchor the starting stresses for creep-fatigue damage calculations.

![Figure 18. Typical stress versus cycles behavior for alloy 556.](image)

With respect to alloy 800H, the database at Petten is available and includes information on continuous cycling fatigue with limited information of the stabilized hysteresis loops and stress ranges. However, substantial data are available about the continuous cycling behavior of alloy 800H and the creep-fatigue behavior. Utilization of these data has produced the material representations currently in ASME III-NH. As discussed in the next section, data at 850°C exist that will be of value in extending the temperature limit above 760°C

**Creep-fatigue interaction:**

There is consensus that, at the low-temperature end of the use range for nickel-base alloys, the life is dominated by strain-fatigue damage while, at the high-temperature end of the use range, the life is dominated by creep damage. The problem of creep-fatigue (C-F) interaction covers the range at high temperatures where, under rapid cycling, strain-
fatigue cracks initiate at the surface, at defects, or at some metallurgical feature and propagate in a transgranular mode while, under relatively low frequency, multiple cracks develop on grain boundaries by a creep (or oxidation) mechanism and propagate in a mixed or intergranular mode. The most common creep-fatigue test that forms the basis for the evaluation of C-F damage uses the relaxation-hold cycle similar to Figure 16 in which the strain-fatigue damage develops during the rapid cycle and creep damage is accumulated during the relaxation hold as shown in Figure 15 [48]. Although this type of testing dominates the C-F testing field, other methods, such as slow-fast cycle (used for the frequency separation model [49]), stress hold periods at the maximum or minimum stress amplitude (needed for strain-range partitioning [50]), and thermal-mechanical fatigue with hold time, have produced some C-F data for nickel-base alloys and have provided insight into the understanding of C-F damage. A modest database is available that includes these methods for the nickel-base alloys. However, the C-F interaction damage diagram in ASME III-NH Figure T-1420 is largely based on damage calculation from the relaxation hold cycle.

Presented in a simplified form, the fatigue damage in any cycle is given by:

$$\delta_f = 1/N_f$$  \hspace{1cm} (4)

where $\delta_f$ is an incremental damage fraction and $N_f$ is the cycles to failure at that strain range and temperature, and the creep damage at any increment of time is given by:

$$\delta_c = \delta t_c/t_r$$  \hspace{1cm} (5)

where $\delta_c$ is an incremental creep damage fraction, $\delta t_c$ is the time at any stress, and $t_r$ is the time to failure at that stress and temperature. For the relaxation hold cycle, $\delta_c$ must be summed over all times in the relaxation hold period to obtain the creep damage per cycle, $\Sigma \delta_c$. Often, a single cycle corresponding to the half-life is chosen to be representative of the whole test. Then, $N\delta_f$ becomes the fatigue damage, $D_f$, and $N\Sigma \delta_c$ becomes the creep damage, $D_c$. These damage terms must be summed to a value less than or equal to the total allowed damage $D$. Thus:

$$D_f + D_c \leq D$$  \hspace{1cm} (6)

This damage equation defines the interaction diagram and the problem is to produce the data that quantifies $D$. Ideally, $D$ should be unity, but experimental work shows that $D$ is less than one and varies with the ratio of $D_c/D_f$. The results differ greatly from one material to another, from one analysis method to another, and from one temperature to another. Alloy 617 generally exhibits low values for $D$ when $D_f$ and $D_c$ are similar. Higher $D$ values have been found for alloy 230 and alloy 556 than for alloy 617. For alloy 800H, Shill found that damage (in alloy 800) summed to around unity [51] while Kaae’s work showed that alloy 800H summed to near 0.25 when $D_f$ and $D_c$ were similar [52]. Huddleston also calculated low values for alloy 800H when his calculations were based on Kaae’s data [53]. The three damage interaction curves are shown in Figure 19. As mentioned above, these damage interaction curves do not consider the possibility that
the degree of interaction between creep and fatigue could be temperature dependent and do not consider environmental effects. Even ignoring these two important factors, there exists a need for carefully designed testing on alloy 617 or its alternate to resolve the issue of inconsistency in the estimation of C-F damage. It may be possible to resolve the issues for alloy 800H if the database can be recovered in sufficient detail.

Figure 19. Damage interaction diagram for alloy 800 and alloy 800H determined from three analyses.

There are other issues with respect to the creep-fatigue interaction studies that should be considered in estimating the needs for data. Clearly, a well-coordinated testing program that includes continuous cycling fatigue, creep-fatigue, and creep-rupture for the same heat is needed. It must be recognized that the most likely conditions for the NGNP will involve long periods between cycles. Depending on the temperature, microstructural recovery is likely during these periods and this recovery will influence the subsequent cyclic hardening, so a few long hold-time C-F tests are needed. Much of the C-F testing with long hold times has been performed at strain ranges greater than 0.6%. The very poor hardening characteristics of the solid-solution nickel-base alloys at and above 800°C often results in dimensional instability in fatigue specimens. Uniform gage specimens tend to form a neck in the middle when the extensometer is axially placed and necks on either side of the middle when the extensometer is placed on the diameter. Creep buckling of specimens, especially tubular configurations, is a problem, so care must be taken in the development of the test methods.

Some consideration should be given to specifying test conditions that could help in the assessment of other models for C-F. These tests might include cyclic creep testing to establish a cyclic creep-rupture curve, low strain rate testing to establish the strain-rate ductility relation for the ductility exhaustion C-F models, tests to establish the true-stress creep-rupture and true-stress ductility relation to correct for the shortened rupture life associated with the increase in engineering stress in constant-load creep-rupture testing, and testing to provide data for the assessment of CDM fatigue models.
Data from a few thermal-mechanical tests are available for some of the nickel-base alloys [54], and Horton and Hollander performed exploratory thermal-stress fatigue tests with hold times up to one hour on alloy 800 and found good performance [55]. A few tests of this character are needed to assess the ability to model the mechanical behavior and predict the C-F damage. Such tests are identified in the National Laboratories NGNP IHX Materials R&D Plan [8] for alloy 617.

Multiaxial stress and strain:

ASME III-NH provides guidelines for addressing multiaxiality for both time-independent and time-dependent loading situations. Yield, flow, and hardening rules have been generalized and guidelines for cyclic multiaxial analysis problems have been provided in a report by Corum and co-workers [56]. Some consideration of multiaxiality issues was undertaken when Corum and Blass developed the draft Code Case for alloy 617 [29], but the rules were largely drawn from the experience with alloys currently approved for construction in ASME III-NH. Based on German experimental work on alloy 617 [57, 58, 59] and alloy 800H [58], there is no reason to expect that the treatment of multiaxiality for the NGNP materials will be significantly different; however, new work is needed to validate the rules and obtain material-specific values for the parametric constants in the applicable formulations. As mentioned above, the testing program should include notched-bar tests designed to assess the applicability of the multiaxial stress-rupture criteria. Emphasis on the 1/1 biaxial stress state would be of value for stress-rupture testing. Multiaxial testing is included in the NGNP IHX Materials R&D Plan [8].

Stress-rupture factors for weldments:

Data for the candidate NGNP materials supporting the implementation of ASME III-NH rules for welds and weldments are meager. Tensile test data for weldments developed by the base metal alloy vendors are available [60, 61, 62]. Also, short-time stress-rupture data from the same sources are available. Long-time stress-rupture, fatigue, C-F, and aging effects on strength and ductility for most of the candidate alloys are scarce. Although ASME III-NH limits strain in welds to 0.5%, there are few creep models available for calculating strains in weld metal or weldments. Isochronous curves, $S_0$, and $S_w$ for deposited weld metals would be helpful. ASME III-NH supplies stress-rupture reduction factors for weldments in Table I-14.10. For austenitic alloys, these factors were derived from the ratio of the average rupture strength for weld metal to the average rupture strength for base metal. However, simple tests of cross-welds and longitudinal-welds in plates and cylinders were performed to validate the conservative aspects of the stress-rupture factors [63]. Clearly, there is a need to develop such factors for candidate nickel-base NGNP materials if weldments are to experience pressure-boundary or similar load-bearing service conditions.
Recently, the situation for alloy 800H welds and weldments was reviewed in some detail [64]. Here, it was found that the stress-rupture factors for weldments in ASME III-NH Table I-14.10 were supported by the database to 730°C but higher temperature and longer time stress-rupture testing was needed to provide the database for extending coverage of alloy 800H weldments to higher temperatures. It was suggested that the 21-33 filler metal be included in the testing programs since it appeared to be a better match for the strength of the base metal than the materials currently specified in ASME III-NH, namely ERNiCr-3.

The candidate nickel-base alloys have good weldability. The filler metals, alloy 617 and the alloy 117 electrode, have been used to join alloy 617 and many high-temperature alloys that lack matching filler metals, so the experience base is significant for this filler metal [65]. It is true that long-time, high-temperature stress-rupture data are lacking. Some available weldment data sources were identified in a review by Ren and Swindeman [66]. Allen summarized work in Europe to study failures that have been encountered in alloy 617 weldments and has discussed efforts to modify composition to improve creep-rupture performance [67, 68]. Also, weldments are under investigation in the EPRI/DOE USC boiler materials program using a restricted composition version of alloy 617 base metal that should produce additional information on tensile and creep-rupture for temperature to 800°C [69]. Wright summarized the status for alloy 617 weldments [8] and designed a testing program to address the needs for the NGNP. The authors have gone well beyond the needs for stress reduction factors and have included fatigue, C-F, and aging effect studies for this material.

The weldment performance of alloy 230 has been reported by a number of researchers [70-72] but an extensive consistent database is lacking. To a lesser extent, weldment behavior in alloy 556 [60] and alloy X [73] have been studied. Similar to alloy 617, alloy 556 is used to join alloys that have no matching filler metal and its performance has been satisfactory in many high temperature applications.

**Fine-grained strip products for compact heat exchangers:**

Several intermediate heat exchanger designs have been proposed for the NGNP and these have been described by Wright [8] and in the AREVA report dealing with Task 7 of the ASME/DOE Gen IV Materials Project [74]. Most likely, the tube-and-shell design could be covered by ASME III-NH once the candidate material is incorporated into a high-temperature Code Case. Even here, as shown by Nickel and coworkers [75], the creep characteristics of tubes may differ significantly from other products tested to develop the Code Case. The compact heat exchanger materials, however, present an entirely new problem because they will require the use of strip products. The width, thickness, grain size, and heat treating requirements are not known at this time. In their studies, Séran and coworkers studied the creep of ribbon materials including alloy 617, alloy 230, and alloy X [76]. The ribbons were 1 to 2 mm thick and materials were of a medium grain size (50 to 100μ). These ribbons, tested in bending, were expected to behave like thicker products. At the very low stresses, however, they crept with a linear stress dependency, although some “primary” creep occurred. A Norton-Bailey relation was used to represent
the creep behavior. For purposes of discussion of data needs, it will be assumed that the strip product will be a “foil” less than 0.5-mm thick and that a grain size of ASTM No. 8, or finer, will be required. Although experience in design, fabrication, and operation of some types of compact heat exchangers have been reported, very little information is available on the high-temperature mechanical properties of such foil materials. A paper by Sharma, Choi, and Knag that discussed the creep-rupture of fine-grained alloy 617 foil revealed that the strength was greatly degraded [77].

A source of information about creep of foil materials is available from work on a DOE CRADA with Solar Turbines [78] to develop materials for primary surface recuperators for service to 730°C. This activity primarily examined stainless steels [79], but in a follow-on effort other classes of materials were included and higher temperatures were targeted. At first, these were alloy 120, alloy 230, and alloy 625 [80]. Later, alloy 214, PM 2000, and 256MA were added [80]. In the stainless steel grouping, the material that emerged was a modified 20Cr-25Ni-Nb steel. Similar to NF-709, but with somewhat modified chemistry and heat treatment, the AL20/25Ni+Nb™ steel foil exhibited good strength relative to other solid-solution strengthened foil materials, as shown in Figure 20. The advantage of the 20Cr-25Ni-Nb steel base composition is its long history of successful operation as a thin-wall cladding alloy in the British AGR reactors. The air, steam, and CO₂ corrosion resistance have been satisfactory.

![Figure 20. Stress versus Larson-Miller parameter for 0.08 to 0.13-mm foils.](image)

For strength above 730°C, alloy 625 and alloy 214 show promise. As a coarse grained material, alloy 625 is approved for construction in ASME Section VIII Division 1 to 871°C. In the fine-grained condition as a foil, alloy 625 has seen usage in primary surface recuperators. For pressure boundary service, alloy 625 is of concern due to the embrittlement produced by the Ni₃Cb precipitate which forms during service. In the foil
form, however, the embrittlement may be less severe, since foils exposed in 4500 hours creep at 730°C have been bent 120° without breaking. Alloy 214 enjoys excellent oxidation resistance and is thought to have good corrosion resistance to molten salts. In thick sections it is difficult to weld but its fabrication characteristics as a foil are not known. It is clear that with either material a substantial experimental program would be needed to meet the data required for the compact IHX of the NGNP. The R&D materials program for the NGNP has identified testing of an alloy 617 foil product [8].

Figure 21. Comparison of creep curves for alloy 625 and alloy 214 foils at 800°C.
OVERVIEW OF THE ESTIMATES FOR DATA NEEDS

Material property requirements for the construction of high-temperature components to the explicit rules of ASME III-NH are but a subset of the properties needed for the design, construction, and licensing of the NGNP. A more complete plan is provided by Wright who has developed the NGNP IHX Materials Research and Development Plan [8]. Wright covers environmental effects, radiation effects, and issues such as fracture toughness, creep-crack growth, and fatigue-crack growth, and these data requirements comprise a significant portion of the testing requirements in the National Laboratories testing plan for the NGNP materials.

Although the overall needs for approval of a material in ASME III-NH can be identified, the specific needs are dependent upon the availability of data for the candidate material. It is clear from the scope of the requirements that a very short list of candidate material is available.

Alloy 800H is one material that appears on nearly all lists of materials and any needs to complete the data requirements for alloy 800H should have high priority. Certainly, the issue of the weldment behavior, the possibility of diffusional creep, and the avoidance of relaxation cracking are three issues that need to be addressed from an experimental viewpoint.

A significant database exists for alloy X, but recognizing the alloy XR is under consideration the database for alloy X serves largely as useful background information. The database and data needs for alloy XR are provided by Suzuki and Asayama [4], so they are not covered here.

Several nickel-base alloys that are competitive with alloy XR have been included in the survey of alloys and some their behavioral features have been presented. Of these, the prime candidate in the National Laboratories plan is alloy 617. Alloy 617 also appears to be the first choice of the Koreans for their VHTR [81]. There are several sound reasons for its selection which include an extensive database, superior corrosion resistance in a carburizing environment, good creep-rupture strength, and retention of ductility after long-time, high-temperature exposure in a carburizing environment. Its drawbacks appear to be poor creep-fatigue resistance relative to some other candidates, “nonclassical” creep behavior for some products or thermal-mechanical treatments, the lack of a very long-time creep-rupture database, and uncertainty with respect to weldment behavior. Experimental work is needed to address these issues.

Suggested testing of alloy 800H to support ASME III-NH:

Two heats of alloy 800H could be used for testing. One could be a plate product purchased for the NPR-HTGR and characterized with respect to tensile, creep-rupture, and weldment behavior to 750°C [82]. The other should be a tubing heat if it is determined the tube-and-shell heat exchanger is the leading candidate.
Table 800H-1. **Effect of a stabilization heat treatment on tensile properties and primary creep behavior of alloy 800H:**

Several austenitic alloys, including alloy 800H, alloy 617, and the creep-strength enhanced 20Cr-25Ni-Nb steels are susceptible to “relaxation cracking” in the service temperature range of 575 to 700°C (750°C for the case of alloy 617). Over 40 failures in process industry vessels have been identified [83]. A second “stabilization heat treatment” at 980°C after the solution heat treatment is recommended to avoid relaxation cracking in alloy 800H. If introduced in ASME III-NH, the stabilization heat treatment could have an effect on the yield and ultimate tensile strengths, so the extent of the effect should be established by performing a few tensile tests over the range of temperatures covered by $S_U$ and $S_Y$ (in Table NH-3225 and Table I-14.5, respectively, in ASME III-NH). See Table 800H-1A in Appendix I. The stabilization treatment could alter the character of primary creep, as well, so a few “short-time” tests to examine the change in primary creep would be of value. See Table 800H-1B in Appendix I.

Table 800H-2. **Strain-rate effects on the strength and ductility of alloy 800H:**

The need to examine strain rate effects on the strength and ductility of alloy 800H was identified in Task 1 of the ASME/DOE Gen IV Materials program [37]. The testing matrix (Table A28) and cost are covered in the National Laboratories NGNP IHX Materials R&D Plan [8]. See Table 800H-2 in Appendix I.

Table 800H-3. **Diffusional creep in alloy 800H:**

The issue of diffusional creep was addressed in a report under the ASME/DOE Gen IV Materials Program Task I [64]. A program to examine the issue has been undertaken under the support of the DOE University Research and is briefly discussed by Wright [8]. The testing program (Table A29) and cost are covered in the National Laboratories NGNP IHX Materials R&D Plan [8]. See Table 800H-3 in Appendix I.

Table 800H-4. **Multiaxial creep-rupture in alloy 800H:**

A fairly significant database exists regarding the multiaxial creep-rupture of alloy 800H. However, some supplementary testing would be of value to more clearly establish the parametric constants in the criteria that have been proposed in ASME III-NH and would be applicable. The testing program (Table A31) and cost are covered in the National Laboratories NGNP IHX Materials R&D Plan [8]. However, supplementary notched bar testing are considered to be of value and validating the expectations based on the multiaxial testing of tubes. These supplementary tests are identified in Table 800H-4 in Appendix I.
Table 800H-5. *Filler metals and weldments for alloy 800H:*

The extension of ASME III-NH to higher temperatures and longer time is dependent on the resolution of the issues related to weldment performance. The Weld Strength Factors (WSFs) currently provided in ASME III-NH for alloy 800H are not adequately supported by the available database for temperatures in excess of 730°C [61]. It is clear that experience in the chemical process industry with alloy 800H welded construction has been good when the composition of the filler metals and fabrication processes are well-controlled. To establish this good experience on a more explicit basis, the National Laboratories NGNP IHX Materials R&D Plan recognizes the need to supplement the existing creep-rupture database with additional testing of deposited weld metal compositions and cross-welds in alloy 800H [8]. The testing program (Table A27) and cost are covered in the National Laboratories NGNP IHX Materials R&D Plan. Some additional testing at 800°C is recommended and these tests are provided in Table 800H-5 in Appendix I.

**Suggested testing for alloy 617:**

On the basis that alloy 617 will be selected as the primary candidate among the nickel base alloys, the National Laboratories NGNP IHX Materials R&D Plan developed and very comprehensive testing plan [8]. This was based on a careful review of the available databases, and, as mention above, included much more than the restricted list of data needs for an ASME III-NH Code Case.

Table 617-1. *Effect of a stabilization heat treatment on tensile properties and primary creep behavior of alloy 617:*

As mentioned above, alloy 617 is known to be susceptible to “relaxation cracking” in the service temperature range of 550 to 750°C, and failures in process industry vessels have been identified [78]. A second “stabilization heat treatment” in the range of 900 to 980°C for 3 hours and air cooled after the solution heat treatment is recommended to avoid relaxation cracking in alloy 617. If introduced in ASME III-NH, the stabilization heat treatment could have an effect on the yield and ultimate tensile strengths, so the extent of the effect should be established by performing a few tensile tests over the range of temperatures covered by $S_U$ and $S_Y$. See Table 617-1A. The stabilization treatment could alter the character of primary creep, as well, so a few “short-time” tests to examine the change in primary creep would be of value. See Table 617-1B in Appendix I.
Table 617-2. *The effect of temperature and strain rate on the tensile stress-strain curves for alloy 617*:

The trend curves that form the basis for $S_t$ and $S_u$ for alloy 617 are available from the archives of ASME and MPC and cover room temperature to 815°C. However, the tensile stress-strain curves have not been recovered. To satisfy needs and characterize new material for the subsequent testing program, the NGNP IHX Materials R&D Plan calls for tensile tests on two different product forms. The testing program (Table A1) and cost are covered in the National Laboratories NGNP IHX Materials R&D Plan [8]. A study of the effect of strain rate on the tensile flow of alloy 617 is included in the testing plan to assist in the development of a unified constitutive model (UCM) for deformation (Table A5), but these will be interrupted tests and supplementary testing is needed to examine the influence of strain rate on the uniform elongation, tensile strength, and tensile ductility to assist in the understanding of the temperature limits for single values of $S_u$ and the ductility exhaustion model for C-F evaluations. Recommended tests are provided in Table 617-2 in Appendix I.

Table 617-3. *Diffusional creep in alloy 617*:

The issue of diffusional creep was addressed in a report under the ASME/DOE Gen IV Materials Program Task [63] and the issue with respect to alloy 617 was addressed by Séran et al. [76] and Duty and McGreevy [84]. A program to examine the issue has been undertaken under the support of the DOE University Research and is briefly discussed by Wright [8]. The work is linked to the issue of grain size which is related, in turn, to the need for a strip or foil product for the compact heat exchanger. The testing program (Table A24 and Table A25) and cost are covered in the National Laboratories NGNP IHX Materials R&D Plan [8]. See Table 617-3 in Appendix I.

Table 617-4. *Multiaxial creep-rupture in alloy 617*:

A fairly significant database exists regarding the multiaxial creep-rupture of alloy 617. However, the National Laboratories NGNP IHX Materials R&D Plan links multiaxial testing to the development of the Unified Constitutive Model (UCM) and a very extensive the testing program was identified for this task. (Tables A5 to A14). Included in the UCM testing plans are relaxation tests with some lasting to 20,000 hours. Tube burst tests and multiaxial creep-rupture tests are included in Tables A15 and A21, respectively, and these are designed to provide information to determine the parametric coefficients in the multiaxial creep-rupture criterion. However, supplementary notched bar testing are considered to be of value in assessing the applicability of the criterion when stress gradients are present. These supplementary tests are identified in Table 617-4 in Appendix I.
Table 617-5. *Filler metals and weldments for alloy 617:*

The development of the ASME III-NH Code Case for alloy 617 requires the specification of Weld Strength Factors for the temperature range to be covered by the Code Case to higher temperatures and longer time is dependent on the resolution of the issues related to weldment performance. It is not clear whether these factors will be temperature and time dependent. The National Laboratories NGNP IHX Materials R&D Plan recognizes the need to supplement the existing creep-rupture database for alloy 617 with additional creep-rupture testing of deposited weld metal compositions and cross-welds in alloy 617 (Table A2). Both gas-tungsten arc (GTA) and shielded-metal-arc (SMA) processes are included, and testing times are expected to reach 60,000 hours and could even exceed 100,000 hours. The testing plan goes well beyond the requirements for the current ASME III-NH rules and includes fatigue and C-F testing (Table A3), and a very large testing program on the effect of aging on fracture toughness (Table 4). The testing program and cost covered in the National Laboratories NGNP IHX Materials R&D Plan for Table A2 are provided in Table 617-5 in Appendix I.

Table 617-6. *Long-time creep-rupture testing:*

Chandra, et al. show no rupture data for alloy 617 extending beyond 42,000 hours and no testing without failures beyond 70,000 hours [85], whereas the intent for the NGNP is to design for 500,000 hours (60 years at 95% availability). In addition to planning creep tests to support the UCM effort (Table A13a), the National Laboratories NGNP IHX Materials R&D Plan includes some long-time creep rupture tests, some extending to beyond 140,000 hours (Table A17). These tests are essential to the estimation of the stress allowables for the design lifetime of the NGNP and approach the times currently available in the test database for alloy 800H. The testing program and cost in the National Laboratories NGNP IHX Materials R&D Plan for Table A2 are provided in Table 617-6 in Appendix I.

Table 617-7. *Fatigue and creep-fatigue testing:*

A large number of fatigue and C-F tests are identified in the National Laboratories NGNP IHX Materials R&D Plan. These tests are listed in Table A19, for fatigue, Table A20 for C-F, Table A23 for an examination of the saturation of C-F damage with extended hold time, and Table A16 for a Simplified Model Test (SMT). All C-F tests in Table A20 involve the relaxation hold cycle, as illustrated in Figure 16 of this report, except that the tests are performed at a $\varepsilon_{\min}/\varepsilon_{\max}$ (R) = -1.0 instead of R = 0, as shown in Figure 16. A single temperature of 950°C is listed. The maximum hold time is chosen to be 600 minutes (10 hours). The C-F tests, designed to look at damage saturation at 1000°C, are likewise relaxation hold cycles in tension and the hold times range from zero to
600 minutes. The SMT tests are planned for 950°C with relaxation holds to 600 minutes. The C-F testing plans are consistent with the data base used to develop the C-F damage diagrams for ASME III-NH but do not include any of the alternative methods to study C-F damage that were mentioned earlier in this report. At the temperatures designated for testing, it is unlikely that mean stresses, developed at low strain ranges when the cycling is at \( R=0 \), can be maintained, so such testing to look at “shakedown” will not be necessary. Also, at temperatures of 950°C, or so, the primary stresses are likely to be so low that the strain-range partitioning tests would have to be performed at very low stresses in the \( \varepsilon_{cp} \) cycle. Strain-rate effects in the cycle are included in the testing plans (Table A19) but all the C-F testing is at a ramp rate of \( 10^{-3}/s \). For consistency, a few tests at a ramp rate typical of the rate used for the tensile curves would be of value. Also, a few tests at a ramp rate of \( 10^{-2}/s \) may be of value for the assessment of the creep damage component in the cycle. To some extent, a successful UCM could be used to make such assessments. Some thermal cycle tests are contained within the testing plans to develop the UCM. It is not clear that these tests will be taken to failure. A few thermal-mechanical fatigue tests would be of value. The testing program and cost covered in the National Laboratories NGNP IHX Materials R&D Plan for Table A2 are provided in Table 617-7 in Appendix I.

SUMMARY

The materials properties required for the development of a Code Case for the construction of metallic components for the NGNP under the rules of ASME III-NH were reviewed.

The available databases for candidates relative to these needs were reviewed.

A benchmark comparison was made for the candidate materials on the basis of the current allowable stresses for ASME Section II-D Tables 1A and 1B. The leading candidates on the basis of strength were found to be alloy 617, alloy 230, and alloy 625. The choice of alloy 617 as the primary choice for the National Laboratories NGNP IHX Materials R&D Plan was found to be based on a relatively mature database, superior corrosion resistance in the anticipated operating environment, and the acceptable retention of properties after long-time thermal exposure.

The National Laboratories NGNP IHX Materials R&D Plan was reviewed relative to the data needs for the ASME III-NH Code Case and a subset of the plan was identified for this purpose. A few supplementary tests were recommended to the plan.
ACKNOWLEDGMENTS

This work was supported by ASME ST-LLC and managed by J. Ramirez with the assistance of Anthony Amato. The authors acknowledge the oversight and guidance provided by R. I. Jetter, consultant, and Sam Sham of the Oak Ridge National Laboratory. Also, Weiju Ren of the Oak Ridge National Laboratory was helpful in collecting reports and papers dealing with the materials properties.
REFERENCES


3. Anon., Summary of Bounding Requirements for the NGNP Demonstration Plant F&ORs, INL/EXT-08-14395, Idaho National Laboratory, Idaho Falls, ID (June, 2008).


34. D. Klarstrom, Haynes International, data provided for inclusion into *Alloys Digest*


APPENDIX I. TESTS AND ESTIMATED COSTS

Table 800H-1A. Tensile tests on stabilized alloy 800H:

Material: Solution annealed plus 980°C for 3 hours

<table>
<thead>
<tr>
<th>Specimen No</th>
<th>Temperature °C</th>
<th>Cross Head Rate</th>
</tr>
</thead>
<tbody>
<tr>
<td>800H-1A1</td>
<td>23</td>
<td>0.005/min</td>
</tr>
<tr>
<td>800H-1A2</td>
<td>600</td>
<td>0.005/min</td>
</tr>
<tr>
<td>800H-1A3</td>
<td>700</td>
<td>0.005/min</td>
</tr>
<tr>
<td>800H-1A4</td>
<td>800</td>
<td>0.005/min</td>
</tr>
</tbody>
</table>

Estimated cost for testing: $2K

Table 800H-1B. Creep tests on stabilized alloy 800H to check primary creep:

Material: Solution annealed plus 980°C for 3 hours

<table>
<thead>
<tr>
<th>Specimen No.</th>
<th>Temperature °C</th>
<th>Stress</th>
<th>Expected Time</th>
</tr>
</thead>
<tbody>
<tr>
<td>800H-1B1</td>
<td>600</td>
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<td>1000</td>
</tr>
<tr>
<td>800H-1B2</td>
<td>650</td>
<td>TBD</td>
<td>1000</td>
</tr>
<tr>
<td>800H-1B3</td>
<td>700</td>
<td>TBD</td>
<td>1000</td>
</tr>
<tr>
<td>800H-1B4</td>
<td>750</td>
<td>TBD</td>
<td>1000</td>
</tr>
</tbody>
</table>

Estimated cost for testing: $5K

Table 800H-2: Strain-rate effects on the strength and ductility of alloy 800H:

These tests are identical to those listed in the National Laboratories NGNP IHX Materials R&D Plan Table A-28.

Estimated cost for testing: $13K

Table 800H-3. Diffusional creep in alloy 800H:

These tests are identical to those listed in the National Laboratories NGNP IHX Materials R&D Plan Table A29.

Estimated cost of testing: $17K

Table 800H-4. Multiaxial creep rupture in alloy 800H:

The testing is identical to the National Laboratories NGNP IHX Materials R&D Plan Table A31

The estimated cost of testing: $126K
Supplementary multiaxial stress testing of alloy 800H:

Material: 800H bar or plate product machined to round notched bars

<table>
<thead>
<tr>
<th>Specimen No.</th>
<th>Notch Type*</th>
<th>Kt Value*</th>
<th>Temperature (°C)</th>
<th>Stress</th>
<th>Expected Life (h)</th>
</tr>
</thead>
<tbody>
<tr>
<td>800H-4-1</td>
<td>TBD</td>
<td>TBD</td>
<td>760</td>
<td>TBD</td>
<td>1000</td>
</tr>
<tr>
<td>800H-4-2</td>
<td>TBD</td>
<td>TBD</td>
<td>760</td>
<td>TBD</td>
<td>3000</td>
</tr>
<tr>
<td>800H-4-3</td>
<td>TBD</td>
<td>TBD</td>
<td>760</td>
<td>TBD</td>
<td>10000</td>
</tr>
<tr>
<td>800H-4-4</td>
<td>TBD</td>
<td>TBD</td>
<td>760</td>
<td>TBD</td>
<td>30000</td>
</tr>
<tr>
<td>800H-4-5</td>
<td>TBD</td>
<td>TBD</td>
<td>850</td>
<td>TBD</td>
<td>1000</td>
</tr>
<tr>
<td>800H-4-6</td>
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<td>850</td>
<td>TBD</td>
<td>3000</td>
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<tr>
<td>800H-4-7</td>
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<td>TBD</td>
<td>850</td>
<td>TBD</td>
<td>10000</td>
</tr>
<tr>
<td>800H-4-8</td>
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<td>TBD</td>
<td>850</td>
<td>TBD</td>
<td>30000</td>
</tr>
</tbody>
</table>


Estimated cost for testing $90K

Table 800H-5. Filler metals and weldments for stress factors for alloy 800H:

The recommended testing is identical to the National Laboratories NGNP IHX R&D Materials Plan Table A27.

The estimated cost of testing: $475K

Supplementary creep-rupture testing of filler metals and weldments for alloy 800H:

<table>
<thead>
<tr>
<th>Specimen No.</th>
<th>Filler</th>
<th>Process</th>
<th>Type of Specimen</th>
<th>Temperature (°C)</th>
<th>Stress</th>
<th>Expected Life (h)</th>
</tr>
</thead>
<tbody>
<tr>
<td>800H-5-w1</td>
<td>82</td>
<td>GTA</td>
<td>Cross</td>
<td>750</td>
<td>TBD</td>
<td>3000</td>
</tr>
<tr>
<td>800H-5-w2</td>
<td>82</td>
<td>GTA</td>
<td>Cross</td>
<td>750</td>
<td>TBD</td>
<td>10000</td>
</tr>
<tr>
<td>800H-5-w3</td>
<td>82</td>
<td>GTA</td>
<td>Cross</td>
<td>750</td>
<td>TBD</td>
<td>30000</td>
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<tr>
<td>800H-5-w4</td>
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<td>GTA</td>
<td>Cross</td>
<td>850</td>
<td>TBD</td>
<td>3000</td>
</tr>
<tr>
<td>800H-5-w5</td>
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<td>GTA</td>
<td>Cross</td>
<td>850</td>
<td>TBD</td>
<td>10000</td>
</tr>
<tr>
<td>800H-5-w6</td>
<td>82</td>
<td>GTA</td>
<td>Cross</td>
<td>850</td>
<td>TBD</td>
<td>30000</td>
</tr>
<tr>
<td>800H-5-w7</td>
<td>82</td>
<td>GTA</td>
<td>Weld</td>
<td>750</td>
<td>TBD</td>
<td>3000</td>
</tr>
<tr>
<td>800H-5-w8</td>
<td>82</td>
<td>GTA</td>
<td>Weld</td>
<td>750</td>
<td>TBD</td>
<td>10000</td>
</tr>
<tr>
<td>800H-5-w9</td>
<td>82</td>
<td>GTA</td>
<td>Weld</td>
<td>750</td>
<td>TBD</td>
<td>30000</td>
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<tr>
<td>800H-5-w10</td>
<td>82</td>
<td>GTA</td>
<td>Weld</td>
<td>850</td>
<td>TBD</td>
<td>3000</td>
</tr>
<tr>
<td>800H-5-w11</td>
<td>82</td>
<td>GTA</td>
<td>Weld</td>
<td>850</td>
<td>TBD</td>
<td>10000</td>
</tr>
<tr>
<td>800H-5-w12</td>
<td>82</td>
<td>GTA</td>
<td>Weld</td>
<td>850</td>
<td>TBD</td>
<td>30000</td>
</tr>
</tbody>
</table>

The estimated cost of testing: $172K
Table 617-1A. Tensile tests on stabilized alloy 617:

Material: Solution annealed plus 980°C for 3 hours

<table>
<thead>
<tr>
<th>Specimen No</th>
<th>Temperature °C</th>
<th>Cross Head Rate</th>
</tr>
</thead>
<tbody>
<tr>
<td>617-1A1</td>
<td>23</td>
<td>0.005/min</td>
</tr>
<tr>
<td>617-1A2</td>
<td>700</td>
<td>0.005/min</td>
</tr>
<tr>
<td>617-1A3</td>
<td>800</td>
<td>0.005/min</td>
</tr>
<tr>
<td>617-1A4</td>
<td>900</td>
<td>0.005/min</td>
</tr>
</tbody>
</table>

Estimated cost for testing: $2K

Table 617-1B. Creep tests on stabilized alloy 617 to check primary creep:

Material: Solution annealed plus 980°C for 3 hours

<table>
<thead>
<tr>
<th>Specimen No.</th>
<th>Temperature °C</th>
<th>Stress</th>
<th>Expected Time</th>
</tr>
</thead>
<tbody>
<tr>
<td>617-1B1</td>
<td>750</td>
<td>TBD</td>
<td>1000</td>
</tr>
<tr>
<td>617-1B2</td>
<td>800</td>
<td>TBD</td>
<td>1000</td>
</tr>
<tr>
<td>617-1B3</td>
<td>850</td>
<td>TBD</td>
<td>1000</td>
</tr>
<tr>
<td>617-1B4</td>
<td>900</td>
<td>TBD</td>
<td>1000</td>
</tr>
</tbody>
</table>

Estimated cost for testing: $5K

Table 617-2. Effect of temperature and strain rate on the tensile stress-strain curves for alloy 617:

The tests are identical to the National Laboratories NGNP IHX Materials R&D Plan Table A1.

Estimated cost for testing: $20K

Supplemental testing to check strain-rate dependence of ultimate strength & ductility

<table>
<thead>
<tr>
<th>Specimen No.</th>
<th>Temperature °C</th>
<th>Crosshead Rate 1/min</th>
</tr>
</thead>
<tbody>
<tr>
<td>617-2-1</td>
<td>750</td>
<td>0.5</td>
</tr>
<tr>
<td>617-2-2</td>
<td>750</td>
<td>0.05</td>
</tr>
<tr>
<td>617-2-3</td>
<td>750</td>
<td>0.005</td>
</tr>
<tr>
<td>617-2-4</td>
<td>750</td>
<td>0.0005</td>
</tr>
<tr>
<td>617-2-5</td>
<td>800</td>
<td>0.5</td>
</tr>
<tr>
<td>617-2-6</td>
<td>800</td>
<td>0.05</td>
</tr>
<tr>
<td>617-2-7</td>
<td>800</td>
<td>0.005</td>
</tr>
<tr>
<td>617-2-8</td>
<td>800</td>
<td>0.0005</td>
</tr>
<tr>
<td>617-2-9</td>
<td>850</td>
<td>0.5</td>
</tr>
</tbody>
</table>
Estimated cost of testing: $16K

Table 617-3. Grain size and diffusional effects in creep of alloy 617:

These tests are identical to those listed in the National Laboratories NGNP IHX R&D Materials Plan Table A24 and A25.

Estimated cost of testing: $460K

Table 617-4. Multiaxial creep rupture in alloy 617:

The majority of testing is blended into the tables that focus on the development of the UCM and are included in the National Laboratories NGNP IHX Materials R&D Plan Tables A5 to Table A14 and the tube-burst and multiaxial creep tests in Table A15 and A21.

The estimated cost of testing: $520K

Supplementary multiaxial stress testing of alloy 617

<table>
<thead>
<tr>
<th>Specimen No.</th>
<th>Notch Type*</th>
<th>Kt Value*</th>
<th>Temperature (°C)</th>
<th>Stress</th>
<th>Expected Life (h)</th>
</tr>
</thead>
<tbody>
<tr>
<td>800H-4-1</td>
<td>TBD</td>
<td>TBD</td>
<td>760</td>
<td>TBD</td>
<td>1000</td>
</tr>
<tr>
<td>800H-4-2</td>
<td>TBD</td>
<td>TBD</td>
<td>TBD</td>
<td>TBD</td>
<td>3000</td>
</tr>
<tr>
<td>800H-4-3</td>
<td>TBD</td>
<td>TBD</td>
<td>TBD</td>
<td>TBD</td>
<td>10000</td>
</tr>
<tr>
<td>800H-4-4</td>
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<td>TBD</td>
<td>TBD</td>
<td>TBD</td>
<td>30000</td>
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<tr>
<td>800H-4-5</td>
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<td>850</td>
<td>TBD</td>
<td>1000</td>
</tr>
<tr>
<td>800H-4-6</td>
<td>TBD</td>
<td>TBD</td>
<td>850</td>
<td>TBD</td>
<td>3000</td>
</tr>
<tr>
<td>800H-4-7</td>
<td>TBD</td>
<td>TBD</td>
<td>850</td>
<td>TBD</td>
<td>10000</td>
</tr>
<tr>
<td>800H-4-8</td>
<td>TBD</td>
<td>TBD</td>
<td>850</td>
<td>TBD</td>
<td>30000</td>
</tr>
</tbody>
</table>


Estimated cost for testing $90K
Table 617-5. Filler metals and weldments for alloy 617:

The recommended testing is identical to the National Laboratories NGNP IHX R&D Materials Plan Table A2 and Table A3.

The estimated cost of testing: $2400K for A2 and $300K for A3

Table 617-6. Long-time creep-rupture testing:

The recommended testing is identical to the National Laboratories NGNP IHX R&D Materials Plan Table A17 and A13a.

The estimated cost of testing: $800K for A17 and $420K for A13a.

Table 617-7. Fatigue and creep-fatigue testing:

The recommended testing is identical to the National Laboratories NGNP IHX R&D Materials Plan Table A19, A20, A23.

The estimated cost of testing: $70K for A19 (Tests in helium not included.), $80K for A20 (Tests in helium not included.), and $200K for A23.

Note on cost estimates

Elevated temperature tensile tests are priced at $500 each.

No distinction is made between rupture testing and creep testing. The estimate for creep-rupture testing is $1/h plus a $200 set-up charge for tests lasting less than 100 hours.

Tube-burst and multiaxial creep tests performed in creep frames are priced at $2/h.

Multiaxial tests, fatigue test, and creep-fatigue tests are priced at $10/h.

The cost of specimen preparation and materials are not included. Calibrations and record keeping in conformance with NQA-1 and other standards are included.

The cost of scientific personnel, project management, overhead and the like are not included.
DOE/ASME Generation IV/NGNP Materials Project

A report on Task 11 submitted to ASME ST-LLC
Updated Final Report

Operating Condition Allowable Stress Values

Updated: July 25, 2009
Draft: March 31, 2009

Kazuhiko Suzuki and Tai Asayama
Japan Atomic Energy Agency
Executive summary

This report describes the results of the Task 11 investigation as part of the DOE/ASME Materials Project that was based on a contract between ASME Standards Technology, LLC (ASME ST-LLC) and the Japan Atomic Energy Agency (JAEA). Task 11 concerns the selection of candidate materials and the corresponding operating time and temperature conditions, review of the available data and the provision of ROM cost and schedule estimates, and the estimation of strength characteristics.

When selecting candidate materials their resistance to environmental degradation caused by exposure to a HTGR helium atmosphere is a key factor. Improving the resistance of commercially available Nickel base super alloys to corrosive oxidation in low oxidizing potential atmospheres such as HTGR-He were discussed, with reference to, for example, the improved Hastelloy X resulting in Hastelloy XR. With regard to the operating temperature the required primary helium coolant temperature in the SI process (or IS process) was identified as being 950 deg C at the reactor outlet.

Review of the available information on Hastelloy XR and Inconel 617 as candidate materials was made, and several critical issues discussed. Information on Inconel 617 is from Japanese project. Those issues were identified with help from the author’s experiences in developing the HTTR high temperature structural design guide. Some R&D to obtain approval for Subsection NH construction was then pointed out.

With estimating the strength characteristics the design creep rupture strength was identified as being 14MPa for Hastelloy XR, even in the HTGR-He atmosphere. The OSDP (Orr-Sherby-Dorn Parameter) method was applied to Hastelloy XR as an extrapolation technique to gain creep rupture strength values, primarily because of scarcity of data on the longer rupture life region.
Preface

This report describes the results of the Task 11 investigation as part of the DOE/ASME Materials Project that was based on a contract between ASME Standards Technology, LLC (ASME ST-LLC) and the Japan Atomic Energy Agency (JAEA). Task 11 concerns the selection of candidate materials and the necessary corresponding operating time and temperature conditions, review of the available data and the provision of ROM cost and schedule estimates, and estimation of the strength characteristics.
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Part I: Selection of candidate materials and corresponding time and temperature operating conditions

I.1 Considerations in selecting the candidate materials

In selecting the candidate materials for use in the high temperature parts of the HTGR cooling system components whose inlet temperature of the primary helium is higher than 850-900 deg C the following points need to be paid special attention.

(1) Corrosive oxidation

As stated in ASME B&PV Code Section III, Division 1, Subsection NB, NB-2160, it is the responsibility of the Owner to ensure a material is selected that suits the conditions stated in the Design Specifications (NCA-3250), with specific attention needing to be given to the effect of being in service upon the properties of the material. With selecting the candidate materials for use in the HTGR coolant (helium gas) environment a key issue is the corrosive oxidation that occurs in the helium atmosphere of the HTGR at an operating temperature of around 900 deg C.

Nickel-base super alloys are one of the candidate materials. An important consideration is the different mechanism of corrosion through oxidation that occurs in an HTGR helium atmosphere from that of in air [1]. When conventional nickel-base super alloys such as Inconel 617 and Hastelloy X are placed in strong oxidizing atmospheres the solid-solution-intensifying elements contained therein get almost totally oxidized to form spinel oxides. Spinel oxides, which are dense and strongly adhere to the alloy surface after having formed there, serve to inhibit the oxidation of the alloy thereafter, which is why these conventional super alloys exhibit excellent corrosion resistance at high temperatures in strong oxidizing atmospheres. However, spinel oxide films barely form in low oxidizing potential atmospheres that include, for example, vacuums or the hot helium gas as in an HTGR that uses helium as the cooling medium. A conventional nickel-base super alloy being exposed to the above-mentioned low oxidizing potential atmosphere results in the slightly formed oxide film on the surface easily spalling off or the selective oxidation of the grain boundaries and internal oxidation. This then results in conventional super alloys being incapable of sufficiently withstanding corrosion at high temperatures.

JAERI (Japan Atomic Energy Research Institute, presently JAEA) in collaboration with Mitsubishi Kinzoku Kabushiki Kaisha (Mitsubishi Metal Corp., currently Mitsubishi Material Corp.) developed Hastelloy XR, based on Hastelloy X, after taking into consideration the above-mentioned different corrosion mechanism.

If the conventional nickel-base super alloys do not get improved because of this point of view of needing to exhibit excellent corrosion resistance at high temperatures in strong oxidizing atmospheres they will include the possibility of having reduced high temperature strength after long-term exposure to HTGR coolant at high temperatures.
Appendix A, the U.S. Hastelloy XR Patent, provides details on the mechanism of corrosion resistance in strongly oxidizing atmospheres, the different corrosion mechanism of an HTGR coolant atmosphere, and details on how to improve corrosion resistance in an HTGR coolant atmosphere so as to meet the creep strength requirements of the high temperature parts of an HTTR IHX. The chemical composition of Hastelloy XR is given in Table 1 and compared with that of Hastelloy X.

<table>
<thead>
<tr>
<th></th>
<th>C</th>
<th>Mn</th>
<th>Si</th>
<th>P</th>
<th>S</th>
<th>Cr</th>
<th>Co</th>
<th>Mo</th>
<th>W</th>
<th>Fe</th>
<th>Ni</th>
<th>B</th>
<th>Al</th>
<th>Ti</th>
<th>Cu</th>
</tr>
</thead>
<tbody>
<tr>
<td><strong>Hastelloy XR</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
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<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
</tr>
<tr>
<td>Max</td>
<td>0.15</td>
<td><strong>1.00</strong></td>
<td>0.040</td>
<td>0.030</td>
<td>23.0</td>
<td>2.50</td>
<td>10.0</td>
<td>1.00</td>
<td>20.0</td>
<td>Bal.</td>
<td>0.010</td>
<td><strong>0.05</strong></td>
<td>0.03</td>
<td>0.50</td>
<td></td>
</tr>
<tr>
<td>Min</td>
<td>0.05</td>
<td><strong>0.75</strong></td>
<td>0.25</td>
<td>-</td>
<td>-</td>
<td>20.5</td>
<td>-</td>
<td>8.0</td>
<td>0.20</td>
<td>17.0</td>
<td>Bal.</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
<tr>
<td><strong>Hastelloy X</strong></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
<td></td>
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<td></td>
</tr>
<tr>
<td>Max</td>
<td>0.15</td>
<td><strong>1.00</strong></td>
<td>0.040</td>
<td>0.030</td>
<td>23.0</td>
<td>2.50</td>
<td>10.0</td>
<td>1.00</td>
<td>20.0</td>
<td>Bal.</td>
<td>0.010</td>
<td><strong>0.50</strong></td>
<td>0.15</td>
<td>0.50</td>
<td></td>
</tr>
<tr>
<td>Min</td>
<td>0.05</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>20.5</td>
<td><strong>0.50</strong></td>
<td>8.0</td>
<td>0.20</td>
<td>17.0</td>
<td>Bal.</td>
<td>-</td>
<td>-</td>
<td>-</td>
<td>-</td>
</tr>
</tbody>
</table>

Naturally any reduction in high temperature strength due to corrosive oxidation in a low oxidizing potential atmosphere strongly depends upon the temperature of the coolant. If the coolant temperature at the reactor outlet is below 850 deg C, that reduction will not be overly crucial.

The reduction in creep strength and other properties of Inconel 617 are discussed in Part II and compared with those of Hastelloy XR.

(2) **High temperature strength**

The material to be selected will need to have the high temperature strength that satisfies the design requirements of the hot parts of the cooling system components.

(3) **Low temperature strength and other properties**

Significant stress and therefore creep damage are typically expected when the temperature of metal parts is high. However, deformation-controlled stress is usually fully relaxed within several tens of minutes or less than one hour due to very significant creep with hot parts, and therefore significant residual stress can be expected after a plant has shut down. If the mechanical strength is as high at low temperature as expected there will be no critical issues relating to low temperature strength. And as discussed in Part II thermal ageing at the high temperatures of 850 deg C and above will cause a significant reduction in the fatigue strength and fracture toughness, including impact energy, at low temperatures of Hastelloy XR as well as conventional nickel-base super alloys. This issue is not considered critical in selecting the candidate materials but should be kept in mind.
(4) Workability and weldability

If Hastelloy XR or Inconel 617 is used fabrication and welding of the high temperature parts of a helical coil type IHX have been demonstrated to be feasible through the following construction experience, even in Japan.
1) HTTR for Hastelloy XR
2) IHX of 1.5MW thermal output in ERANS project for Inconel 617

It should be noted here that several techniques need to be developed in order to construct the helical coil type IHX given below:

a) Welding techniques where the best welding conditions and appropriate welding materials can be selected in order to improve the ductility of welded joints and to avoid any separation of the weld bonds of Hastelloy XR (and probably the other nickel-base super alloys). The separation cannot be detected through tensile testing but only through fatigue testing.

b) ECT technique for use in welding the joints of the coils as the length of a coil will be 30m or more and tubes of around 5m in length will need to be joined together.

1.2 Operating time and temperature conditions

All the cycle configuration recommendations in the pre-conceptual design report (INL/EXT-07-12967) [2] include a hydrogen production system. Of the two hydrogen production process recommendations HTE (High Temperature Electrolysis) was selected as the initial target, with SI (or IS) being the longer-term target.

JAEA has continued to research pre-conceptual designs for an IS hydrogen production system by utilizing nuclear-generated heat in addition to other related R&D activities. The operating conditions of the main components in the IS system to which nuclear-generated heat is supplied via secondary helium are given in Table 2 [3].

This reveals then that a reactor outlet temperature of 950 deg C is, therefore, required to meet the operating requirements on an IS system.

The issue of selecting the appropriate material also exists with the SO$_3$ decomposer because of the large differential pressure at the high temperature of about 900 deg C.
Table 2 Operating conditions of the main components in an HTGR-IS hydrogen production system [3]

<table>
<thead>
<tr>
<th>IS system component</th>
<th>Inlet helium temperature in deg C</th>
<th>Medium, outlet temp. and pressure of secondary side</th>
<th>Medium and reaction</th>
<th>Outlet temp. in deg C</th>
<th>Pressure in MPa</th>
</tr>
</thead>
<tbody>
<tr>
<td>SO₃ decomposer</td>
<td>About 900</td>
<td></td>
<td>SO₃(g) → SO₂(g) + 1/2O₂(g)</td>
<td>About 850</td>
<td>1.2</td>
</tr>
<tr>
<td>H₂SO₄ decomposer</td>
<td>About 610</td>
<td></td>
<td>H₂SO₃(l) → SO₃(g) + H₂O (gas)</td>
<td>About 470</td>
<td>1.2</td>
</tr>
<tr>
<td>Co regenerated HI decomposer</td>
<td>About 730</td>
<td></td>
<td>2HI → H₂(g) + I₂ (g)</td>
<td>About 600</td>
<td>1.2</td>
</tr>
<tr>
<td>Purifier and concentrate</td>
<td>About 490</td>
<td></td>
<td>-</td>
<td>About 210</td>
<td>1.2</td>
</tr>
</tbody>
</table>
Part II: Review of available Japanese data on Hastelloy XR and Inconel 617 and required R&D

II.1 Review of Japanese information on Hastelloy XR and Inconel 617

Major data on the mechanical properties of Hastelloy XR is available in Reference 4 and the Year 2007 Task 5 report. The essentials on developing an HTTR high temperature structural design guide, in which the design criteria and allowable limits with Hastelloy XR are given, are presented in Reference 5. Essentials of mechanical properties of Inconel 617, especially on creep-fatigue interaction and its environmental effect, were summarized in Reference 6. Other available information [7, 8] is also referred to. In Japan, material tests on Inconel 617 were mainly conducted as part of ERANS project (ERANS project: project by the Engineering Research Association on Nuclear Steelmaking under the sponsorship of Ministry of International Trade and Industry), and major raw experimental data have been classified.

The present report discusses the critical issues with the mechanical properties of the candidate materials for use in NGNP with help from the author’s experience with gaining the approval for a HTTR high temperature structural design guide from the Japanese regulatory authorities. The critical issues include the following:

1) Environmental effects, and in particular the effect of corrosion on the mechanical properties
2) Unique tensile stress-strain curves due to dynamic recrystallization
3) Very significant creep
4) Creep analysis method

Thermal ageing is also considered to be a critical issue.

II.1.1 Environmental effects

As discussed in Part I to be suitable for use in the high temperature parts of HTGR cooling system components conventional nickel-base super alloys will need to be improved so as to have sufficient corrosion resistance in low oxidizing potential atmospheres. Corrosive oxidation in the HTGR-helium gas environment at the very high temperature of 1000 deg C significantly reduces the creep rupture strength of Inconel 617, as revealed in Figure 1. The high temperature structural design guideline of ERANS project addresses the creep rupture strength reduction factor for very high temperatures ranging to 1000 deg C. Raw data on the strength reduction of Inconel 617, which were obtained for an HTGR-He simulated helium gas with the helium purity of 99.9995%, are expressed as a function of decarburization. However, Sakai et al. [8] pointed out the following for Inconel 617 in a different HTGR-He simulated helium gas which was denoted as “He-2”:

a) Inconel 617 was carburized in the temperature range from 900 deg C to 970 deg C, but was decarburized above 990 deg C.
b) In the temperature range of carburization, the carburized layer was observed at near-surface region, and Cr-Mo carbides precipitated along the grain boundaries and twin boundaries. In the temperature range of decarburization, the carbides along the grain boundaries disappeared, and this led to the formation of altered layer.

c) In contrast to the results that the oxide films formed at high temperature in air were dense and protective, the films formed in the helium environment were very porous and less protective. Further, the chromium content in the oxide films along the grain boundaries beneath the surface decreased with the increase of test temperature. This could lead to make the films much less protective.

Other research shows that even if Inconel 617 is carburized at the temperatures below 950 deg C, the creep rupture strength is reduced in an HTGR-He-simulated helium gas. These confused findings suggest careful selection of impurities compositions in the helium gas for Inconel 617.

Impurities in both of 99.995% helium (or 99.99% helium) used in ERANS project and JAERI-B type helium were controlled or added to simulate the low oxidizing potential atmosphere of HTGR coolant. The detail of JAERI-B type helium was given in Task 5 report dated September 30, 2007, as shown below;

<table>
<thead>
<tr>
<th>Impurities</th>
<th>Unit (Pa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>H₂</td>
<td>20-21</td>
</tr>
<tr>
<td>H₂O</td>
<td>0.08-0.12</td>
</tr>
<tr>
<td>CO</td>
<td>10-11</td>
</tr>
<tr>
<td>CO₂</td>
<td>0.2-0.3</td>
</tr>
<tr>
<td>CH₄</td>
<td>0.5-0.6</td>
</tr>
</tbody>
</table>

The detail of 99.995% He was not identified in the present work.

On Inconel 617, it should be noted here that a reduction of creep rupture strength by decarburization at 1000 deg C was explained in terms of a diminishing of carbides contributing to depress the action of grain boundaries [7]. Its detail is as follows;

Migrating grain boundaries caused a dissolution of carbides, resulting in carbide free zones behind them and in recrystallized grains. Notable grain boundary migration and recrystallization were observed preferentially on longitudinal grain boundaries in the specimen strained by more than 10% (tertiary stage) and they were considered to affect the creep strength in those specimens. Decarburized specimens being free of grain boundary carbides exhibited significant grain boundary migration, and this fact implied a suppression of grain boundary migration by grain boundary carbides.

Environmental effects on the fatigue and cyclic properties of Inconel 617 and Hastelloy XR are given in Figures 2 to 4.

As revealed by Figure 1 at the very high air temperatures of 950 and 1000 deg C Inconel 617
has the advantage of 50 deg C with regard to creep rupture strength over Hastelloy XR. However, the reduction in creep rupture strength of Inconel 617 due to decarburization in an HTGR-He environment results in a comparable creep rupture strength of Hastelloy XR with Inconel 617 in an HTGR-He environment, which is conceptually illustrated in Figure 5.

Air induces more of an adverse effect than HTGR helium gas on the fatigue strength of Inconel 617 and Hastelloy XR while cyclic softening is more pronounced in an HTGR-He environment.
Fig. 1 Environmental effects on the creep rupture strength of Inconel 617 and Hastelloy XR at HTGR temperatures [6]
Fig. 2 Environmental effect and test specimen configuration dependence on low cycle fatigue strength within a fast-fast type waveform for Inconel 617 at 1000 deg C [6]
Fig. 3 Environmental effect on low cycle fatigue strength under fast-fast type waveform for Hastelloy XR at 950 deg C [6]

Fig. 4 Environmental effect on the cyclic stress-strain relationship of Inconel 617 and Hastelloy XR at HTGR temperatures [6]
II.1.2 Unique tensile stress-strain relationship due to dynamic recrystallization

(1) Stress-stain curves with monotonic loading

The tensile engineering stress-engineering strain curves (or nominal stress-nominal strain curves) of Hastelloy XR at various temperatures are given in Figure 6. The tensile testing was conducted according to the JIS tensile testing procedure standard. At temperatures of below 850 deg C, Hastelloy XR is strain-hardened, but at temperatures of 850 deg C and above, unique stress-strain curves can be observed. And at temperatures of 900 deg C and above, in particular, the stress drops after yielding and becomes almost constant (referred to as a “plateau”). Dynamic recrystallization causes this drop in stress after yielding. The term “plateau” means that the creep strain rate in the “plateau” equals that of the extension rate of the tensile testing. The tensile testing at the very high temperature specified in the JIS and other testing standards such as ASTM does not concern “tensile testing” in the acquisition of data on purely plastic properties but rather short-term creep testing, and this is because of the very significant creep which occurs
at that very high temperature.

Tensile tests at various extension rates were carried out to clarify the dependence of the unique behavior of the stress-strain curves on the extension rate (or strain rate), the results of which are given in Figure 7. The stress-strain curve at the extension rate of 100 %/min is thought to be purely elastic plastic without any occurrence of dynamic recrystallization, and is discussed below.

The extension rate dependences of yield strength (precisely 0.2% proof stress) and tensile strength are given in Figure 8, and the extension rate dependence schematically illustrated in Figure 9. The critical extension rate at which tensile testing is actually “tensile testing” but not short-term creep testing seems to be about 100%/min at the very high temperature of 1000 deg C.

Fig. 6 Tensile stress-strain curves of Hastelloy XR at various temperatures at the strain rate specified in JIS standards [5]
Fig. 7 Tensile stress-strain curves of Hastelloy XR at 950 deg C at various extension rates [5]

![Tensile stress-strain curve diagram]

Test Temperature = 950°C

Fig. 8 Extension rate dependences of yield strength and tensile strength of Hastelloy XR at 800 and 1000 deg C [4]

![Extension rate dependence graph]

0.2% Proof Stress $\sigma_y$ and Ultimate Tensile Strength $\sigma_u$ (MPa) vs. Extension Rate $\dot{\varepsilon}$ (%/min.)

<table>
<thead>
<tr>
<th>Temp. (°C)</th>
<th>$\sigma_y$</th>
<th>$\sigma_u$</th>
</tr>
</thead>
<tbody>
<tr>
<td>800</td>
<td>$\bullet$</td>
<td>$\Delta$</td>
</tr>
<tr>
<td>1000</td>
<td>$\bigcirc$</td>
<td>$\bigtriangleup$</td>
</tr>
</tbody>
</table>
(2) Stress-strain curves with cyclic loading (i.e., cyclic stress-strain curves)

Figure 10 gives the hysteresis loop of Hastelloy XR at 950 deg C and a strain rate of 0.1%/s, namely 6.0%/min. The drop in stress can only be observed in the first cycle and not in any subsequent cycles. Again, the very significant creep causes a “plateau”, making therefore fatigue testing at a strain rate of 0.1%/s at 950 deg C actually cyclic creep testing.
Issues and their resolution as design criteria with unique stress-strain curves

(3) Issues relating to the above-mentioned observations with the stress-strain curves can be categorized as follows:

a) Issues relating to very significant creep
b) Issues relating to the occurrence of dynamic recrystallization

a) The very significant creep issue and its resolution

A reasonable time-independent allowable limit that assures the structural integrity of a plant component against intense earthquakes needs to be set. The strain rate of seismic loading is in the order of 100%/min. As discussed above, at an extension rate of 100%/min, the tensile stress-strain curves consist of purely elastic plastic responses. However, if the tensile testing is conducted at the extension rate of 0.3%/min, as specified in existing technical standards, the observed yield strength and tensile strength would be far below the “true” yield strength and tensile strength.

The HTTR high temperature structural design guide, therefore, specifies the extension rate of tensile testing at high temperatures of 850 deg C and above to be 100%/min. And as a generic rule, time-independent allowable limits for the reasonable prevention of any failure caused by an intense earthquake are needed, thus necessitating that the tensile testing should be conducted in such a way that time-independent stress-strain curve data can be collected from
that testing.

b) The dynamic recrystalization issue and its resolution

As discussed later an inelasticity analysis method (or precisely speaking a stress relaxation analysis method) is needed to predict the structural response to very significant creep. The issue of dynamic recrystalization is a difficulty to formulating the propagation of dynamic recrystalization triggered by inelastic strain, particularly with the first strain cycle in which inelastic strain develops. However, if the time-independent allowable limits for seismic loadings only, discussed in item a) above, are used that difficulty can be avoided because the strain rate of the seismic loading is high, being in the order of 100%/min.

II.1.3 Very significant creep

Very significant creep causes a very fast rate of creep strain, and results in the following observations:

1) With “thermal transient” loading practically no plasticity develops. Only seismic loading can cause plasticity to develop and thereby cause the fatigue damage discussed in item (2) above.

2) The creep damage becomes saturated within a very short period of time, some tens of minutes or an hour at the very high temperatures of 900 deg C and above and at high stress (hereinafter referred to as “early saturation of creep damage”).

3) The time interval of the primary creep regime of a creep curve is short at very high temperatures and high stress

(1) Early saturation of creep damage

Figure 11 gives the reduction in the fatigue life cycles of Inconel 617 at the very high temperature of 1000 deg C in a helium environment as the hold time increases. It should be noted here that in reality it is simply cyclic creep rather than fatigue damage resulted from plastic cycling. The reduction approaches a constant and beyond the cycle time of 300 s (or 5 minutes) is practically saturated. A similar observation can be observed in Figure 12, which concerns Hastelloy XR at 900 deg C. That reduction is not unique to Inconel 617 and Hastelloy XR, with Figure 13 revealing the same type of saturation for SS 304 as the hold time proceeds. The saturation of the reduction in fatigue life cycles (in reality the reduction in strain cycles because there is no fatigue damage at such a very high temperature) is caused by the practically full stress relaxation revealed in Figure 14. That practically full stress relaxation in the strain cycle leads to saturation in the creep damage increments per strain cycle.
Fig. 11 Effect of strain hold time on low-cycle-fatigue lives within the strain hold waveform for Inconel 617 at 1000 deg C in 99.99% helium [6]

Fig. 12 Effects of strain hold time and strain rate on low-cycle-fatigue lives in the so-called creep-fatigue interaction testing for Hastelloy XR in an HTGR-He environment [4], [9]
Fig. 13 Effects of strain hold time and strain rate on low-cycle-fatigue lives in creep-fatigue interaction testing of SS 304 [10]

Fig. 14 Practically full relaxation of Inconel 617 at the high temperature of 900 deg C and comparison of the observed relaxation curve with one estimated using creep data [6]
(2) Primary creep regime of creep curves

Some creep curves at very high temperatures and under high stress appear to have no primary creep regime. However, detailed measurement and analysis at the early stage of those creep curves reveals the existence of a primary creep regime.

Data on the reciprocal time constant of primary creep regime $r$ of Hastelloy XR at the high temperatures of 800 deg C and above is provided in Figure 15. At 1000 deg C and 20 MPa $r \approx 3$ (1/h), making therefore the time constant of the primary creep regime $t_1$ to be about 20min, which seems rather short. However, it should be noted here that at 1000 deg C stress of 20 MPa would also be considered rather high, and also that the time to onset of tertiary creep $t_3$ at 20 MPa is about 10h, as revealed in Figure 16. Moreover, the maximum primary creep strain in the above-mentioned conditions is of the order of 0.1 %.

Data on creep curves is necessary in developing a creep constitutive equation for use in analyzing stress relaxation as well as generating isochronous stress-strain curves, and thus necessitating that the creep deformation measurements be of high accuracy for testing the creep at such high temperatures.

![Fig. 15 Temperature and stress dependences of reciprocal time constant in primary creep regime $r$ for Hastelloy XR in the high temperature region of 800 deg C and above [4]](image-url)
(3) The very significant creep issues and their resolution

a) The need to “continuously” evaluate the transient and steady states

As well known the structural response of a plant component consists of transient response and a steady state or load-hold response. The creep damage and creep strain in the transient and steady states were separately analyzed at elevated temperatures as the initial stress at the beginning of the steady state can be conservatively estimated using the assumption that no significant creep occurs during the transient state (even though in fact the strain rate effect needs to be taken into account).

However, at a very significant creep regime temperature, or very high temperature, deformation-controlled stress is remarkably relaxed, even in the transient. This then leads to the assumption that the initial stress at the beginning of the transient is the same as stress at the end of the corresponding transient without any creep effect being overly conservative. The creep damage and creep strain need to be evaluated using creep effect analysis throughout a plant operation cycle, but in order to do so a stress relaxation analysis method or creep analysis method that can reasonably analyze the relaxation responses will need to be developed.

b) Need for creep damage and creep strain evaluations under cyclic creep conditions but not as a creep-fatigue interaction

At a very high temperature no plastic strain is induced from thermal-transient-induced
loading because deformation-controlled and discontinuity stress relax very fast, while at low temperatures yielding does not occur because the mechanical strength, including fatigue strength, at low temperatures is very high.

Cyclic application of thermal-transient-induced loading, therefore, does not result in any significant fatigue damage and hence no creep-fatigue interaction. In any such cyclic application it would be better to take the cyclic creep effect into consideration in ensuring that reasonable evaluation of the creep damage takes place.

With the accumulated strain limit the membrane strain limit should be derived from the creep ductility but the creep-fatigue interaction ignored.

II.1.4 Creep analysis method: creep constitutive equation and related hardening/flow rules, and a creep analysis-based method of evaluating creep damage and creep strain

A creep analysis method needs to developed for use in the following analyses of the structural integrity assessment of a VHTR plant component

i) Stress relaxation analysis to evaluate the creep damage discussed above

ii) Creep strain analysis, in particular structural discontinuities for use in analyzing the creep deformation localization induced by large differences in stiffness or strength that is well-known as elastic follow-up

In developing such a creep analysis method, in addition to the creation of a creep constitutive equation and related hardening/flow rules, design factors that take into account the several effects causing variation in creep damage and creep strain in plant components will need to be derived, typically by using the variation in creep curves.

The present report discusses the creep constitutive equation and design factors.

(1) Creep constitutive equation

The Garofalo type creep constitutive equation was used to fit the creep curves of Hastelloy XR because of the limited data on the primary creep regime. An example of the creep curve data is given in Figure 17. The secondary creep regime is clearly observed in many creep curves. The data fitting is good, but if the early stage of the primary creep regime (in Figure 17 less than 0.1%) is focused upon a small offset can be observed, which is known to cause poor estimations of stress relaxation behavior. As pointed out when developing the “Monju” elevated temperature design criteria [11] the following Blackburn type creep constitutive equation is more recommendable for use in analyzing creep in NGNP components.

$$\varepsilon_c = \varepsilon_t(1 - e^{-nt}) + \varepsilon_s(1 - e^{-qt}) + \varepsilon_{\min} t$$

(1)

(2) Design factors in developing criteria for evaluating creep damage and creep strain

As well known slower creep strain rates cause higher creep damage but lower creep strain.
An averaged correlation of the creep curves therefore was used in creating a creep constitutive equation or nominal creep constitutive equation. Design factors that take into consideration the variations in creep damage and creep strain primarily caused by those variations in the creep curves when using the nominal creep constitutive equation. About twenty years ago, the author analyzed the variations in creep damage and creep strain caused by variations in the creep curves [12]. The results of analysis resulted in design factors of 20 for creep damage and 2 for creep strain being recommended, with however, the classical strain hardening rule being used in the analysis.

Fig. 17 Creep curve fitting using the Garofalo equation [4]
II.1.5 Thermal ageing effect on low temperature strength

Long-term exposure to the very high temperatures of 850 deg C and above does not cause any significant reduction in mechanical strength at very high temperatures, as revealed in Figure 18, but does cause significant reductions in fatigue strength and toughness at low temperatures, around room temperature for example. Because of this the structural integrity of IHX tube bundles in a stand-by operation against intense earthquake loading (intense seismic loading) need to be evaluated. The gap between the bundles (in particular the tube support devices) and the shell (or liner attached to the internal thermal insulator) is at its maximum in stand-by operations. Intense earthquakes shake the IHX, and thereby also shaking the bundle. That shaking then results in impact loading both to the tube bundle and to the shell.

However, with the HTTR IHX this issue is not considered to be critical because of the following reasons:

1) The stress levels of the high temperature parts of the IHX in both normal operations and with abnormal events are low in comparison with the fatigue strength at room temperature.
2) The impact energy is low in comparison to the toughness (Charpy V-notch values).

This then resulted in the decision that there would be no need to specify any thermal ageing requirements. However, as a generic design criterion for application with NGNP components a specific design rule was recommended.

Thermal ageing effects on other mechanical properties are given for Inconel 617 in Figures 19 to 20. As revealed by Figure 20 no unique behavior in the monotonic stress-strain relationship due to dynamic recrystallization, as discussed above, can be observed after aging at 950 deg C.
(a) Inconel 617 as received and aged at 1000 deg C in air

(b) Hastelloy XR as received and aged at 950 deg C in an HTGR-He environment

Fig. 18 Comparison of low-cycle fatigue strength at a very high temperature as received and aged at the same temperature [6]
Fig. 19 Changes in the range of stress with increasing number of cycles for Inconel 617 [6]
Fig. 20 Monotonic and cyclic stress-strain relationship for Hastelloy XR at 950 deg C in an HTGR-He environment [6]
II.2 Required R&D

A database has been established with regard to Inconel 617 as well as Hastelloy XR. In Japan an Inconel 617 database was developed in the ERANS (Engineering Research Association on Nuclear Steelmaking) project that was sponsored by the Japanese Government MITI. The database includes creep strength reduction factors in an HTGR-He environment and other key correlations for use in developing high temperature design criterion.

However, further R&D will be required to identify the following items or effects on its mechanical properties and to develop a creep analysis method, and the subsequent development of design criteria.

1. Reasonable method of evaluating creep damage under cyclic creep conditions in which even in the transients significant creep damage develops but not as the creep-fatigue interaction.
2. Thermal ageing effects on low temperature fatigue strength and toughness, and then design rules for thermally aged material to prevent thermal fatigue cracking (cracks that propagate at low temperature in particular) and fractures due to impact loading such as in seismic events.
3. Formulation of the early stage in the primary creep regime for use in analyzing the stress relaxation from highly accurately measured creep curves.
5. Reasonable design criteria for thermal-transient-induced loading as time-dependent loading and for seismic loading as time-independent loading under very significant creep conditions.
Data at the high temperatures of 800 deg C and above is given in Figure 21. In addition Figure 22 gives the effect of the helium pressure on creep rupture strength. With Hastelloy XR there is no significant effect from the helium pressure at up to 40 atg (or 4.1 MPa) on the creep rupture strength. It should be noted here that the terminology of “creep fracture” may be better for use than the often used “creep rupture” because a rupture is usually defined as being the full separation of a fractured test specimen.

In both figures a linear relationship between the applied stress and creep rupture (or creep fracture) life on the log-log scale can be observed for the dataset at each test temperature.

The population of data in the shorter rupture life region is high, while that in the longer life region of 10000 h or above is low. The data fitting, therefore, was weighted in favor of the shorter rupture region. When extrapolating the rupture life use of an extrapolation technique incorporating a conservative longer rupture life is recommended, but is not necessarily the best estimation technique.

The maximum creep rupture life of Hastelloy XR is about 27000 h.
III.2 Extrapolation technique

Several types of TTPs (Time Temperature Parameters) are used in extrapolating the creep rupture life with experimental data as well as in interpolating it. The following gives the classical popular types of TTPs:

1) Larson-Miller parameter: \( \text{LMP} = T_k (Y + C), \quad Y = \log_{10} t_R \)

2) Orr-Sherby-Dorn parameter: \( \text{OSDP} = Y - Q/(RT_k) \times \log_{10} e \) \( (2) \)

3) Manson-Succop parameter: \( \text{MSP} = Y - BT_k \)

OSDP was selected as the TTP for application in the HTTR high temperature structural design guide, primarily because it results in more conservative extrapolations than the others, which is desirable due to the limited data on the creep rupture strength of Hastelloy XR in the longer rupture life region, as discussed in III.1.

In selecting the TTP the data scatter of the creep rupture life also needs clarifying. Figure 23 gives the probabilistic distribution of experimental data on Hastelloy XR, with no abnormal distribution being observed for the selected OSDP.
In accordance to that OSDP was selected as the TTP for interpolating and extrapolating the creep rupture life for use in generating a creep rupture strength design. The regression curve of the creep rupture life with Hastelloy XR can be expressed as follows:

\[ \ell \log_{10} \bar{t}_R - \frac{Q}{RT_k} \times \ell \log_{10} \sigma = \sum_{i=0}^{2} \bar{\alpha}_i (\ell \log_{10} \sigma)^i \]

where \( \bar{t}_R \): average creep rupture life in h,
\( \sigma \): stress in MPa,
\( T_k \): absolute temperature in K,
\( \bar{\alpha}_0 = -9.909, \bar{\alpha}_1 = -0.023, \bar{\alpha}_2 = -1.503 \),
and \( Q/R = 44343 \) in K.

Fig. 23 Probability distribution of creep rupture life data for Hastelloy XR [4]
A guideline on extrapolating the time to be “three times the maximum creep rupture life” is specified in Reference [10], resulting therefore in the extrapolation of the creep rupture life of Hastelloy XR being limited to less than or equal to 100,000 h.

III.3 Estimation of creep rupture strength at 800 deg C and 100,000h

Equation (3) results in the creep rupture strength design of Hastelloy XR at 800 deg C and 100,000h being 14 MPa, if a design factor of 20 with the creep rupture life is used.
References


