CREEP RUPTURE MECHANISMS IN NOTCHED SPECIMENS OF RENE 95

by

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S.B. Materials Science and Engineering Massachusetts Institute of Technology (1983)

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Abstract

The creep rupture behavior of René 95 was studied in air at 704°C (1300°F). Smooth bar, blunt-notch \( (k_n=1.6) \), and sharp-notch \( (k_n=4.2) \) specimens from a hot isostatically pressed (HIP) turbine disk and from an extruded and isothermally forged (EIF) turbine disk were tested. Creep strains were measured in smooth bar specimens with the direct current potential drop (DCPD) technique, as well as with an extensometer and LVDT. The DCPD technique was used to measure creep strain and damage accumulation in blunt-notch specimens, and to detect crack initiation and growth in sharp-notch specimens.

Smooth and sharp-notch specimens failed after the initiation of surface cracks, indicating the importance of environment to creep rupture in René 95. Initiation occurred much sooner in sharp-notch specimens, resulting in notch weakening for the acute notches. In blunt-notch specimens, surface stresses relax sufficiently to prevent the atmosphere-assisted initiation of surface cracks, resulting in notch strengthening for the less severe notch geometry. Intergranular creep cavities nucleate, grow, and coalesce, leading to failure of the blunt-notch specimens. The notch rupture behavior of René 95 is shown to depend upon time-dependent stress distributions and upon the environment.

There was little or no effect of processing history on rupture life. In all cases, HIP specimens failed along prior particle boundaries (PPBs), while EIF specimens failed on grain boundaries. The creep strain rates in blunt-notch specimens were slightly higher in EIF René 95. This difference is attributed to a larger volume fraction of fine \( \gamma' \) in HIP material, impeding dislocation motion, and to grain boundary sliding in the EIF material. The HIP material cavitated on the PPBs, while the EIF material cavitated on grain boundaries, which are much more closely spaced. The EIF blunt-notch specimens therefore accumulated much more cavitation and exhibited greater changes in DCPD feedback than their HIP counterparts. The HIP sharp-notch specimens showed longer initiation times but faster creep crack propagation than the EIF sharp-notch bars; the result is only a slight increase in HIP sharp-notch rupture lives over those of EIF sharp-notch specimens. The difference in crack initiation is related to the difference in carbide distribution between the HIP and EIF processes. The continuity of PPB carbides in the HIP alloy and the GBS-induced crack tip blunting in the EIF material accounts for the faster crack growth rates observed in HIP sharp-notch specimens.

The potential drop technique was shown, through a simple computer simulation of creep deformation, to be an effective method to monitor creep strain in smooth and blunt-notch specimens. PD procedures could accurately determine crack initiation in sharp-notch specimens. The DCPD technique can be used to monitor cavitation in EIF blunt-notch specimens after a reasonable amount of damage has accumulated, while the small amount of damage in HIP blunt-notch bars could not be resolved. DCPD measurements could conceivably be used to measure sharp-notch crack lengths, but nonsymmetrical crack formation eliminated this possibility.

Thesis Supervisor: Regis M. Pelloux
Professor of Metallurgical Engineering
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to Cynthia
1. Introduction

Since the development of the jet engine before and during World War II, there has been a great emphasis on the development of alloys to help produce lighter, more powerful, and more reliable engines. Requirements for these alloys include high strength, to withstand centrifugal forces in rapidly rotating components and to save weight by making smaller structural components; capability for high temperatures, for combustion chambers and turbine blades and vanes; and resistance to corrosion caused by impurities in fuel and intake air.

Nickel base superalloys have high yield and tensile strengths and good corrosion resistance. They can be used at temperatures corresponding to large fractions of their melting temperatures. To achieve these properties, the superalloys have become very complex, with many alloying elements and finely engineered microstructures. Various powder metallurgy (P/M) and controlled solidification processing techniques have been developed to eliminate casting segregation and other problems attributed to high alloy content. High strength superalloys are prone to problems such as low ductility, notch sensitivity, and low and high cycle fatigue.

René 95 is an advanced superalloy designed for high strength, moderate temperature service in turbine disks. René 95 is produced by powder metallurgy techniques, and consolidated by several means, including hot isostatic pressing (HIP) and extrusion and isothermal forging (EIF). The high temperature deformation and cracking resistance of René 95 are strongly dependent on microstructures determined by processing history.

The direct current potential drop (DCPD) technique has been used extensively to monitor phenomena which produce changes in the resistivity of the material under study. These phenomena may include deformation, phase transformations and precipitation, and cavity formation. In general, potential drop is most widely used to measure creep and fatigue crack growth.

This study has been undertaken for two purposes. The first was to investigate the effect of processing history, namely, the HIP and EIF P/M consolidation techniques, on creep rupture properties of René 95, on smooth bars and on sharp- and blunt-notched specimens. The second purpose was to evaluate the usefulness of the DCPD technique to monitor accumulation of creep damage and deformation in a creep brittle alloy.
2. Nickel Base Superalloys.

Nickel base superalloys are used at higher homologous temperatures (i.e., fractions of their melting temperatures) than any other engineering alloys. Metallurgically perhaps the most complicated alloys, with many elemental ingredients and many phases, the superalloys are also the most widely studied. Nichrome-type alloys, simple nickel-chromium systems which came into use about 1905, showed improved elevated temperature strength and oxidation resistance over austenitic and ferritic stainless steels. With the discovery around 1930 that additions of titanium and aluminum could greatly improve the strength of the Nichrome alloys, the evolution of the superalloys was underway. Further developments in the understanding and control of alloy chemistry, processing, microstructure, and mechanical and environmental properties of these alloys have led to their widespread use in engineering systems, including air- and land-based gas turbines, fossil fuel power plants and nuclear reactors, chemical reaction vessels and equipment, coal gasification plants, and numerous other applications requiring long-term structural and environmental stability at elevated temperatures.

Pure nickel does not have exceptional strength or corrosion resistance, but it has a great capacity for the addition of alloying elements which improve these properties. The nickel matrix is strengthened by additions of various transition metals in solid solution. Chromium present in superalloys lends excellent oxidation resistance through the formation of an adherent oxide film on exposed surfaces. Aluminum and titanium react with nickel to form an intermetallic compound, the coherent $\gamma'$ phase, which greatly strengthens the alloy by impeding dislocation motion. Carbides form at grain boundaries, where they limit grain growth and grain boundary sliding.

Deleterious effects can arise due to extensive alloying in nickel base systems. Large castings are subject to gross segregation during solidification, resulting in microstructural and compositional inhomogeneities, local variations in properties, and incipient melting. In addition, superalloys are prone to oxide or sulfide inclusions, or the precipitation of plates or needles of hard, brittle, topologically close packed (TCP) phases. Through careful control of chemistry, solidification, and thermomechanical processing of the alloys, these detrimental effects can be avoided, and the benefits of extensive alloying are maximized.

2.1. Superalloy Chemistry and Microstructure.

The great majority of nickel base superalloys are characterized by a duplex $\gamma$-$\gamma'$ microstructure. $\gamma$, or austenite, is a face-centered cubic (fcc) matrix made up primarily of nickel. $\gamma'$, an ordered intermetallic compound with nominal composition Ni$_3$Al, precipitates throughout the $\gamma$ matrix, strengthening the alloy by impeding the flow of dislocations.
Table 2.1 groups the elemental ingredients according to their location within the microstructure of the alloy, *i.e.*, those which partition to the γ matrix and to the γ' precipitates, and those which segregate to the grain boundaries. The table also indicates which elements are carbide formers, and several trace elements are identified [1, 2]. Table 2.2 lists crystal structures, chemical compositions, and typical morphologies of phases commonly found in superalloys.

**Table 2.1. Elements in Nickel Base Superalloys.**

<table>
<thead>
<tr>
<th>Location</th>
<th>Element</th>
<th>Comment</th>
</tr>
</thead>
<tbody>
<tr>
<td>Austenitic</td>
<td>Ni</td>
<td>Basis for γ matrix</td>
</tr>
<tr>
<td>Matrix (γ)</td>
<td>Cr</td>
<td>Forms protective Cr₂O₃ oxide barrier</td>
</tr>
<tr>
<td></td>
<td>Co</td>
<td>Solid solution strength, promotes γ' formation</td>
</tr>
<tr>
<td></td>
<td>Fe, V, Mo, W</td>
<td>Solid solution strength</td>
</tr>
<tr>
<td>Ordered γ'</td>
<td>Al</td>
<td>Combines with Ni to form Ni₃Al</td>
</tr>
<tr>
<td>Precipitate</td>
<td>Ti, Cb, Ta</td>
<td>Substitute for Al in Ni₃Al</td>
</tr>
<tr>
<td>Grain Boundary</td>
<td>Hf, B</td>
<td>Refine GB microstructure, improve ductility</td>
</tr>
<tr>
<td>Zr, Mg</td>
<td></td>
<td>Improve ductility</td>
</tr>
<tr>
<td>C</td>
<td></td>
<td>Form GB carbides: strengthening</td>
</tr>
<tr>
<td>Carbide Forming</td>
<td>MC</td>
<td>Ti, Cb, Ta, Mo, W, Fe, Hf, Zr</td>
</tr>
<tr>
<td>Elements</td>
<td>M₂₃C₆</td>
<td>Ta, Cr, Ti, Al</td>
</tr>
<tr>
<td>Trace or Tramp</td>
<td>Ca, Mg, Ce, La</td>
<td>Beneficial refining aids</td>
</tr>
<tr>
<td>Elements</td>
<td>O, H, N, Ar, He</td>
<td>Residual gases</td>
</tr>
<tr>
<td></td>
<td>S, P, Pb, Sb, As, Se, Ag, Cu, Ti, Te</td>
<td>Miscellaneous impurities</td>
</tr>
</tbody>
</table>

**The Austenite Matrix**

Nickel does not have a high modulus of elasticity or low diffusivity, important considerations for creep design, but certain other characteristics make nickel desirable for high temperature applications. Nickel can be alloyed extensively with a minimum of phase instability. This is attributed to its nearly full 3d electron shell [1]. Superalloys are derived from early nickel-chrome alloys; one example of these is Nichrome V, essentially an
<table>
<thead>
<tr>
<th>Phase</th>
<th>Crystal Structure</th>
<th>Chemistry</th>
<th>Morphology</th>
</tr>
</thead>
<tbody>
<tr>
<td>(\gamma)</td>
<td>fcc</td>
<td>Ni (Cr,Co,Mo,W)</td>
<td>Continuous austenitic matrix.</td>
</tr>
<tr>
<td>(\gamma')</td>
<td>simple cubic (ordered derivative of fcc)</td>
<td>(Ni,X)(_3) (Al,Y) (X=\text{Co,Fe}; \ Y=\text{Ti,Cb,Ta})</td>
<td>Intragranular cuboidal or spheroidal precipitates, blocky particles or continuous film on GB, rafted intragranular layers; precipitation hardening phase.</td>
</tr>
<tr>
<td>MC</td>
<td>fcc (NaCl)</td>
<td>M=Ti,Ta,Cb,Hf,etc.</td>
<td>Large, coarse, blocky particles, oriented randomly.</td>
</tr>
<tr>
<td>(M_{23}C_6)</td>
<td>complex cubic</td>
<td>M=Ta,Cr,Ti,Al</td>
<td>Semi-continuous to discrete GB precipitates.</td>
</tr>
<tr>
<td>(M_6C)</td>
<td>complex cubic</td>
<td>M=Mo,W</td>
<td>See (M_{23}C_6).</td>
</tr>
<tr>
<td>(M_{3}B_2)</td>
<td>tetragonal</td>
<td>M=Mo,Ni,Cr,Fe,etc.</td>
<td>Discrete GB particles.</td>
</tr>
<tr>
<td>(\sigma)</td>
<td>tetragonal</td>
<td>(Cr,Mo)(_x) (Ni,Co)(_y)</td>
<td>Brittle acicular or Widmanstätten intragranular ppts.</td>
</tr>
<tr>
<td>(\mu)</td>
<td>hexagonal</td>
<td>(Co,Mo)(_7) (Cr,W)(_6)</td>
<td>See (\sigma).</td>
</tr>
<tr>
<td>Laves</td>
<td>hexagonal AB(_2)</td>
<td>(FeCrMn)(_2) (MoTiCb)</td>
<td>See (\sigma).</td>
</tr>
<tr>
<td>(\eta)</td>
<td>hexagonal</td>
<td>Ni(_3)Ti</td>
<td>See (\sigma).</td>
</tr>
<tr>
<td>(\delta)</td>
<td>orthogonal</td>
<td>Ni(_3)Cb</td>
<td>See (\sigma); also random grain boundary globules.</td>
</tr>
<tr>
<td>(\gamma')</td>
<td>bct</td>
<td>Ni(Cb,X); (X=\text{Al,Ti})</td>
<td>Small disks or platelets: precipitation hardening of Fe-Ni-base alloys.</td>
</tr>
</tbody>
</table>
80% Ni-20% Cr alloy still used today as an electrical resistance heating element [3]. Chromium forms a thin surface film of Cr$_2$O$_3$, an adherent oxide which limits the rates of diffusion of metal and oxygen ions through the film and thereby prevents continuous oxidation of the alloy [1, 3-5]. Chromium typically comprises 10 to 15 % (by weight) of modern superalloys.

Chromium is not a very potent solid solution strengthener in nickel, so other elements must be added. Cobalt is added for strengthening and to improve the fabricability of the alloy. Cobalt also serves to decrease the solubility of aluminum and titanium in the matrix, enhancing precipitation of $\gamma'$ [1, 3, 4, 6-8]. Typically the level of cobalt plus nickel in a Superalloy is about 50%. Cobalt is a politically strategic and very expensive raw material; recent studies [9, 10] suggest that up to half of the cobalt added to nickel base superalloys can be replaced by nickel, without affecting forgeability, rupture life, ductility, creep rates, and tensile strength.

Solid solution strengthening is provided by additions of molybdenum [1, 4-6, 8], tungsten [1, 6, 8], and $\gamma'$-formers vanadium and titanium [1] and columbium [11]. Solid solution strengthening is attributed (a) to the size misfit of solute atoms, which hinders the glide of dislocations; (b) to a decrease in stacking fault energy, resulting in dislocations becoming more extended, and making cross-slip more difficult; and (c) to the introduction of slow-diffusing atoms, such as molybdenum and tungsten [1], which interfere with diffusion of more mobile species and reduce diffusion-controlled creep rates.

Aluminum is a potent solid solution strengthening agent in nickel base superalloys [1], and it forms a more protective oxidation barrier than Cr$_2$O$_3$ [1, 12]. The most important role played by aluminum, however, is to harden superalloys through formation of $\gamma'$ precipitates.

**Gamma Prime**

The greatest contribution to the elevated temperature strength of most nickel base superalloys is due to hardening by the interaction of dislocations with $\gamma'$ precipitates. The degree of strengthening increases as the volume fraction of $\gamma'$ is increased. $\gamma'$, or Ni$_3$Al, has an L1$_2$ ordered structure, typical of many A$_3$B-type ordered intermetallic compounds.

In general, if elements A and B have the same crystal structure, they will form a complete series of A-B solid solution alloys if their atomic radii differ by less than about 15%. Greater atomic size differences will result in limited solubility fields of one element in the other. As the difference in the electronegativities of A and B increases, the tendency to form intermetallic compounds increases [13, 14].
The \( \text{A}_3\text{B} \) (or \( \text{L}_1_2 \)) crystal structure is illustrated in Figure 2.1: the \( \text{B} \) atoms are located on the corners of a simple cubic (SC) unit cell, and the \( \text{A} \) atoms are found on the cube faces. Because the atoms occupy the same locations in the unit cell as in the fcc \( \gamma \), the structure is often referred to as an “ordered fcc superlattice.” The ordered structure is constructed by stacking \( \{111\} \) planes of the type depicted in Figure 2.2 in an \( abcabc \) pattern, such that each \( \text{B} \) atom is surrounded by 12 nearest neighbor \( \text{A} \) atoms and no \( \text{B} \) atoms. To satisfy geometrical constraints, the ratio of the atomic radius of element \( \text{B} \) to that of element \( \text{A} \) must lie between 1.0 and 2.4. The classic example of the \( \text{A}_3\text{B} \) ordered fcc structure is \( \text{Cu}_3\text{Au} \); other examples include \( \text{Cu}_3\text{Pt} \) and \( \text{Cu}_3\text{Pd} \), and \( \text{Ni}_3\text{Fe} \), \( \text{Ni}_3\text{Mn} \), and \( \text{Ni}_3(\text{Mn},\text{Fe}) \). These order-disorder transformations occur coherently, with little or no lattice distortion, *i.e.*, the lattice parameters of the two phases are nearly equal [13-15].

![Austenitic (\( \gamma \)) Face Centered Cubic and \( \gamma' \) Simple Cubic ("Ordered FCC")](image)

**Figure 2.1.** Austenitic FCC Lattice and \( \gamma' \) SC Lattice (Ordered FCC Superlattice).

The major constituents of superalloys are ranked in Table 2.3 according to atomic radius, electronegativity, and group (column) of the periodic table. Elements which replace nickel in \( \text{Ni}_3\text{Al} \) are located toward the top of the lists, while those replacing aluminum are found near the bottom. According to the rules for \( \text{A}_3\text{B} \) formation, the smaller, more electronegative elements (*i.e.*, those like nickel: cobalt, molybdenum, or tungsten) are likely to substitute for nickel in \( \gamma' \), while the larger, more electropositive elements (titanium, columbium, or tantalum) should replace aluminum [7, 8]. More than two-thirds of the aluminum in \( \gamma' \) can be substituted for without causing \( \gamma' \) instability [16, 17]. Depending upon the alloy composition, chromium, iron, or vanadium may substitute for either nickel or
aluminum. It is further noted that A atoms come from groups V, VI, and VIII, while B atoms are from groups III, IV, and V [1].

![Figure 2.2. Close Packed (111) Plane in L12 Structure of Ni3Al.](image)

Table 2.3. Ranking of Elements for \(\gamma'\) Formation.

<table>
<thead>
<tr>
<th>Atomic Radius</th>
<th>Electronegativity</th>
<th>Group</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni 1.24 Å</td>
<td>Ni 1.8</td>
<td>Ni VIIIB</td>
</tr>
<tr>
<td>Co 1.25</td>
<td>Co 1.8</td>
<td>Co VIIIB</td>
</tr>
<tr>
<td>Fe 1.26</td>
<td>Fe 1.8</td>
<td>Fe VIIIB</td>
</tr>
<tr>
<td>Cr 1.30</td>
<td>Mo 1.8</td>
<td>Mo VIB</td>
</tr>
<tr>
<td>V 1.34</td>
<td>W 1.7</td>
<td>W VIB</td>
</tr>
<tr>
<td>Mo 1.39</td>
<td>Cr 1.6</td>
<td>Cr VIB</td>
</tr>
<tr>
<td>W 1.41</td>
<td>V 1.6</td>
<td>V VB</td>
</tr>
<tr>
<td>Al 1.43</td>
<td>Cb 1.6</td>
<td>Cb VB</td>
</tr>
<tr>
<td>Cb 1.46</td>
<td>Ta 1.5</td>
<td>Ta VB</td>
</tr>
<tr>
<td>Ti 1.47</td>
<td>Ti 1.5</td>
<td>Ti IVB</td>
</tr>
<tr>
<td>Ta 1.49</td>
<td>Al 1.5</td>
<td>Al IIIA</td>
</tr>
</tbody>
</table>

Figure 2.3 illustrates the shapes of the \(\gamma'\) phase fields in several Ni-Al-X ternary systems [1]. The \(\gamma'\) phase fields for ternaries in which X=Ti, Cb, or V extend toward a hypothetical Ni\(_3\)X compound, evidence that these elements substitute for aluminum in Ni\(_3\)Al. Conversely, the \(\gamma'\) phase field of the Ni-Al-Co phase diagram reaches in the direction of the (nonexistent) Co\(_3\)Al, indicating the substitution of cobalt for nickel. The intermediate orientations of the Ni\(_3\)Al phase region for ternary diagrams containing molybdenum, chromium, or iron suggest that these elements could replace either nickel or aluminum in \(\gamma'\), depending upon the concentrations of other substitutional elements. The \(\gamma'\) formula cited is usually a variation of (Ni,Co)\(_3\)(Al,Ti) or (Ni,Co,Fe)\(_3\)(Al,Ti,Cb).
Addition of $\gamma'$ forming elements is a critical balancing act. These elements increase the $\gamma'$ volume fraction and therefore the strength of the alloy, but precipitation of too much $\gamma'$ decreases the amount of nickel remaining in the austenite matrix, increasing the tendency to precipitate detrimental TCP and $\eta$ phases [18]. Superalloys typically contain up to 8% each by weight of $\gamma'$ formers aluminum and titanium, and smaller amounts of columbium. If the titanium content exceeds about 4%, there is a tendency to form $\delta$-Ni$_3$Cb [1, 6, 7]. $\gamma'$ nucleates coherently from the supersaturated austenite matrix due to the spontaneous development of periodic concentration gradients along $\langle100\rangle$ crystallographic directions [16]. This uniform homogeneous nucleation of Ni$_3$Al is made possible by low $\gamma$-$\gamma'$ misfit strains. The lattice parameter mismatch ranges from 0 to 1.5%, although in most cases it is much less than 1% [1, 3, 4, 6-8]; the lattice parameter of $\gamma'$ is generally slightly larger than that of $\gamma$ [19, 20].

The major strengthening mechanism in $\gamma$-$\gamma'$ alloys is due to formation of antiphase boundaries (APBs) in $\gamma'$ [1, 19, 21]. Figure 2.4 shows burgers vectors for the austenitic matrix and for ordered $\gamma'$. The burgers vector for the ordered superlattice is twice as long as that for the fcc lattice, so in order to shear the $\gamma'$ completely, i.e., to return one layer of atoms
to the same position relative to the next layer, two perfect fcc dislocations are required. The first dislocation will cause slip in the $\gamma'$, but it will leave behind an APB, a planar region across which equivalent superlattice positions are not conserved during slip. The passage of a second dislocation will erase the antiphase boundary. The APB energy in $\gamma'$ is very large, due to disruption of nearest neighbor relationships. This large APB energy hardens the alloy by preventing the first dislocation from penetrating deeply into the $\gamma'$ precipitates. A second dislocation will actually facilitate shear to some degree. The repulsive force between the two dislocations will push the first dislocation through the $\gamma'$ particles, and the APB created by the first dislocation will attract the second dislocation [19, 21].

![FCC Burgers Vector](image)

**Figure 2.4.** Burgers Vectors in FCC Lattice and in Simple Cubic (Ordered FCC) Superlattice.

An important consequence of APB strengthening is that the yield strength of the alloy increases with increasing temperature, up to a maximum of 700-800°C [1, 19, 22-25]. This dependence of strength on temperature is explained as follows. Because APB energy is less on {100} cube planes than on {111} octahedral planes, segments of the leading dislocation are subject to thermally-activated cube cross-slip. Sections of the lead dislocation which remain in the octahedral slip system are pinned where they join the cross-slipped dislocation lengths. With increasing temperature, more dislocation segments cross-slip, and shorter lengths of dislocation remain on {111} planes; larger stresses are required to bow out these shorter dislocations. Therefore, the strength of the alloy increases with temperature up to the point where most of the dislocation line length experiences cross-slip [19, 23, 24]. The inherent ductility of $\gamma'$ precipitates make them more effective strengthening agents than carbide particles or oxide dispersoids, which may crack and lead to local fracture [1].

At low temperatures, the $\gamma$-$\gamma'$ mismatch contributes some additional strengthening, due to interactions between the elastic strain fields of the dislocations and the precipitates...
[1, 3]. On the other hand, the temperature capability of a superalloy will be increased by up to 200°C if mismatch strains are decreased [6]. For $T/T_m>0.6$, γ' particles tend to ripen, making it easier for dislocations to bypass the precipitates [1, 3]. The ripening rate can be slowed by decreasing the $\gamma$-$\gamma'$ interfacial energy, e.g., by adjusting the Al:Ti ratio, or by varying the amounts of molybdenum, cobalt, or tungsten in the alloy. The ripening rate can also be controlled by slowing diffusion in the austenite lattice and by decreasing the stability of γ' in the matrix; both of these changes are made through additions of chromium, columbium, molybdenum, cobalt, and tungsten. Increasing the volume fraction of γ' also will retard the rate of precipitate coarsening [1]. Some single crystal superalloys develop oriented γ-γ' structures during exposure to an applied stress (e.g., rafts perpendicular to the applied stress axis) [5, 6, 7, 26]. During particle ripening, as fine γ' dissolves, larger γ' particles are more likely to be cut by dislocations into smaller particles, which also dissolve. Eventually, only large aligned γ' structures remain in the γ matrix [26]. Development of such a rafted structure could conceivably strengthen the alloy; dislocations are compelled to cut through these precipitates, because there is no path in the γ by which they can bypass the γ' [27]. Rafting has reportedly improved rupture lives by a factor of four over an unrafted microstructure. In this case, mismatch strains would be increased by altering the alloy chemistry, to promote formation of the desired γ' morphology [3, 5-7, 24, 27].

Incorporating molybdenum into γ' increases the γ' lattice parameter (and therefore the γ-γ' mismatch), the γ' solvus temperature, and the volume fraction of γ' in the alloy. Columbium increases the γ' solvus and the γ' volume fraction. Cobalt, chromium, and iron additions also increase the volume fraction of γ'. This increase in γ' content does not necessarily strengthen the alloy, however, because iron and chromium additions are known to decrease the strength of the γ matrix [1].

Typical superalloys contain up to 50% γ' (by volume). Advanced investment-cast alloys, in which forging operations are unnecessary, and powder metallurgy alloys, which are inherently more forgeable than cast and wrought ingots with the same chemistry, may contain up to 60% γ'. In large castings, such alloying results in gross macrosegregation, but in small cast blades, segregation is much less severe [5].

Many superalloys contain a bimodal γ' size distribution. Relatively large precipitates form and grow during aging; these are responsible for creep strength of the alloy. A smaller volume fraction of small γ' particles continue to nucleate as the alloy is cooled following the aging treatment, but their growth is limited by the decreasing temperature. This “cooling” γ' results in high tensile strengths at lower temperatures [1, 4].
Intragranular $\gamma'$ precipitates can have spheroidal or cuboidal morphologies, depending on $\gamma$-$\gamma'$ mismatch strains [1, 8, 20, 22]. Spherical $\gamma'$ forms when the coherency strain is below about 0.2%, while mismatches ranging between 0.5 and 1.0% result in cuboidal precipitates. $\gamma'$ plates tend to form at lattice misfits greater than about 1.25% (usually in single crystals only) [1, 22]. In heavily alloyed investment-cast alloys, there is a tendency to form $\gamma'$-$\gamma'$ eutectic nodules during solidification, which do not completely dissolve during subsequent solution treatment [1, 7]. In addition to the intragranular $\gamma'$ morphologies described above, some alloys form a continuous film of $\gamma'$ covering grain boundaries and grain boundary $\text{M}_{23}\text{C}_6$ carbides. $\gamma'$ and $\text{M}_{23}\text{C}_6$ are products of decomposition of MC carbides [1, 7, 28].

**Carbides**

Several varieties of carbides are found in superalloys. For the most part, these carbides are located on grain boundaries. Carbides prevent grain growth during exposure to high temperatures (heat treatments and operating conditions) by pinning grain boundaries. In addition, carbides interrupt the smoothness of the grain boundaries, often in conjunction with intergranular $\gamma'$, to prevent grain boundary sliding, and therefore increase the creep strength of the alloy. Elements forming carbides include columbium, tantalum, titanium, chromium, molybdenum, tungsten, and vanadium [1, 4-8, 28, 29]. Because the ratio of the atomic radius of carbon to the atomic radii of chromium, iron, cobalt, and nickel exceeds 0.59, crystal structures of these interstitial compounds are very complex [13, 14].

During solidification, primary MC carbides form at grain boundaries, within grains, and in interdendritic regions, without any degree of orientation between carbides and the austenitic matrix [1, 16]. During solidification and heat treatment of superalloy powders, MC carbides tend to form at powder particle surfaces; this results in poor adhesion between particles during compaction, and in brittle prior particle boundaries (PPBs) in consolidated powder products [3, 18, 30-32]. These deleterious effects can be remedied by controlling carbon content and solidification parameters and by decreasing the temperature of hot isostatic pressing (HIP) of the powder [3, 30, 31]. MC carbides, which have a face centered cubic (NaCl) crystal structure [8, 14], can be blocky [1, 4, 8], random, coarse, or cubic in shape [1, 8].

Increasing the columbium and tantalum content in the alloy tends to stabilize the MC carbides, that is, retard transformation of MC into the lower carbides $\text{M}_{23}\text{C}_6$ and $\text{M}_6\text{C}$. Molybdenum, tungsten, and iron stabilize MC to a lesser extent. Conversely, increased chromium content promotes decomposition of MC. Group III, IV, V, and VI transition
metals are the only elements to form monocarbides [14]. Typical MC compositions include TiC, TaC, CbC, VC, MoC, (Ti,Cb)C, and (Ti,Mo)C [1].

$M_{23}C_6$ forms at elevated temperatures by decomposition of MC carbides or by precipitation from the matrix, generally at grain boundaries [1, 7, 16]. The morphology of $M_{23}C_6$ can be controlled by heat treatment; its shape can range from discrete particles to nearly continuous grain boundary platelets [7, 8, 28]. Often these grain boundary carbides are engulfed in a thin layer of $\gamma'$, another product of MC decomposition. $M_{23}C_6$ carbides strengthen grain boundaries by retarding grain boundary sliding, although the carbide particles may eventually fracture or decohere [1, 7]. The $\gamma'$ sheath serves to slow propagation of such a void or crack into the bulk of the grain [1], and the $\gamma'$-$M_{23}C_6$ structure acts as a dispersion-strengthened sheet [7].

Nucleation of cellular $M_{23}C_6$ at grain boundaries can be severely detrimental to the high temperature mechanical properties of an alloy, but through careful control of chemistry and heat treatment, cellular growth of $M_{23}C_6$ can be avoided [1, 7]. For example, the presence of grain boundary $\gamma'$ will prevent formation of cellular $M_{23}C_6$ [1].

During elevated temperature deformation, large blocky $M_{23}C_6$ carbides grow on grain boundaries which support a tensile stress. These carbides grow at the expense of carbides on unstressed boundaries and on boundaries in compression. Tensile boundaries with a greater number of these carbides exhibit lower creep rates than boundaries with fewer carbides, although cavities eventually form at the grain boundary carbides, leading to fracture [28].

$M_{23}C_6$ carbides have a complex cubic crystal structure. Large chromium compositions favor the formation of $M_{23}C_6$, and titanium, aluminum, and tantalum further stabilize this carbide. A typical formula for this compound is $Cr_{21}(W,Mo)_2C_6$; nickel, cobalt, and iron can substitute for chromium [1, 8].

Another carbide product from MC degradation is $M_6C$. Like $M_{23}C_6$, the $M_6C$ carbide is heat-treatable [7], and often the two carbides are found together on grain boundaries [1]. $M_6C$ occurs as blocky grain boundary particles, and sometimes as acicular intragranular precipitates [8, 28].

$M_6C$ has a complex cubic crystal structure. High concentrations of refractory metals (e.g., greater than 6 to 8% molybdenum or tungsten) replace chromium in other carbides and induce formation of this compound [1, 16]. Examples of $M_6C$ are $(Ni,Co)_3Mo_3C$, $(Ni,Co)_2W_4C$ [1], and $Fe_3Cb_3C$ [8]. In fact, $M_6C$ covers a range of metal-to-carbon ratios between $M_3C$ and $M_{13}C$. $M_6C$ is more beneficial to grain boundary strengthening than is $M_{23}C_6$, due to its greater stability at higher temperatures [1].
Another blocky grain boundary carbide is Cr$_7$C$_3$ [1, 8, 16]. Cr$_7$C$_3$ is found in alloys with only small amounts of alloying additions, and not in most nickel base superalloys. Cobalt, molybdenum, tungsten, and columbium additions will prevent formation of this compound [1, 9].

Several solid state reactions have been formulated to describe decomposition of MC into the lesser carbides. The first of these, which operates from 760°C to 980°C, is given by [1, 3, 7, 26]:

$$MC + \gamma \rightarrow M_{23}C_6 + \gamma'$$  \hspace{1cm} (2.1a)

Although it cannot formally be balanced, the equation can be rewritten to locate relevant elements [1, 16]:

$$(Ti,Mo)C + (Ni,Cr,Al,Ti)\gamma \rightarrow Cr_{21}Mo_2C_6 + Ni_3(Al,Ti)$$  \hspace{1cm} (2.1b)

According to this relationship, chromium in solid solution in the austenite substitutes for titanium in the carbide, converting MC to $M_{23}C_6$; this displaced titanium combines with aluminum and nickel from the $\gamma$ to form more $\gamma'$ [1, 10]. Conversely, nucleation of $\gamma'$ and $M_{23}C_6$ depletes the austenite of carbon and titanium, leading to further decomposition of TiC [16]. Although the decomposing MC particle may be located within a grain, $\gamma'$ and $M_{23}C_6$ will form on grain boundaries, where they can nucleate heterogeneously and where increased diffusion results in faster growth rates [26].

Another MC decomposition reaction, occurring between 815° and 980°C, is [1, 3]:

$$MC + \gamma \rightarrow M_6C + \gamma'$$  \hspace{1cm} (2.2a)

or, including the appropriate elements [1, 16]:

$$(Ti,Mo)C + (Ni,Co,Al,Ti)\gamma \rightarrow Mo_3(Ni,Co)_3C + Ni_3(Al,Ti)$$  \hspace{1cm} (2.2b)

Again, titanium is displaced from the MC carbide, in this case by cobalt and nickel, to form $M_6C$; titanium combines with aluminum and nickel in the austenite to form $\gamma'$. The $\gamma'$ product of both MC decomposition equations accounts for the thin film of $\gamma'$ located around intragranular and grain boundary carbides, and along the grain boundaries themselves [1].

Equations 2.1 and 2.2 can be combined to describe the equilibrium between $M_{23}C_6$ and $M_6C$:

$$M_6C + M' \rightleftharpoons M_{23}C_6 + M''$$  \hspace{1cm} (2.3a)

or, alternatively:

$$M_3(Ni,Co)_3C + Cr_{\gamma} \rightleftharpoons Cr_{21}Mo_2C_6 + (Ni,Co,Mo)_{\gamma}$$  \hspace{1cm} (2.3b)
This relationship shows that chromium favors the formation of $M_{23}C_6$, while cobalt, molybdenum, and nickel promote $M_6C$. Depending on the specific alloy composition, this last transformation has been observed to proceed in either direction [1].

**Superalloy Grain Boundaries**

Grain boundary carbides and $\gamma'$ particles are important during heat treatments: they pin grain boundaries, preventing excessive grain growth and preventing grain boundary sliding, an important low-stress, high-temperature mechanism of creep deformation. If carbides form as a continuous film along the boundaries rather than as discrete particles, the grain boundary is embrittled, as the carbide layer acts as an easy path for crack propagation [3, 4, 29].

A common grain boundary microstructure consists of discrete $M_{23}C_6$ or $M_6C$ carbides, with a thin $\gamma'$ film engloving these carbides and lining the grain boundaries. This structure results from degeneration of MC carbides into the metal-rich carbides and $\gamma'$, and it behaves as a dispersion-strengthened, ductile, and creep-resistant layer at the grain boundaries [1, 7]. A further benefit of the $\gamma'$-enshrouded carbide structure is that, in case the carbide fractures or breaks away from the film, the $\gamma'$ is tough enough to delay growth of the crack into the bulk of the grain [1].

Several elements segregate to grain boundaries of nickel base superalloys, including magnesium, zirconium, hafnium, boron, and carbon. These elements exhibit little solid solubility in the $\gamma$ matrix, due to large variations between their atomic radii and that of nickel (on the order of 20 to 30%), but they fit into vacancies along grain boundaries or form grain boundary compounds [1, 4-7, 29, 33].

Hafnium additions have been shown to increase creep strength, rupture life, creep ductility, and resistance to cracking of superalloys [3-7, 34, 35]. Examination of grain boundary microstructures of hafnium-containing alloys shows that intergranular script-like (film-like) carbides are more discrete in nature and that there is more blocky $\gamma'$ along grain boundaries [3, 29, 35]. The boundary itself has changed from planar to wavy; this wavy appearance is due to straight segments of grain boundary connecting unevenly distributed carbides. Formation of such tortuous boundaries is enhanced by furnace cooling (rather than air cooling) from the solution temperature to below the $\gamma'$-solvus or the carbide solvus. This slower cooling rate ensures precipitation of grain boundary particles that distort the boundaries [3]. The wavy grain boundaries are rougher, to resist grain boundary sliding, and longer, to increase grain boundary diffusion distances and retard diffusional creep. Conversion of script carbides to blocky particles removes the continuous, brittle path for easy
crack propagation [3, 4, 29]. The use of hafnium as a grain boundary strengthener has obviated the need for a continuous grain boundary film of γ' [7].

Boron and zirconium transform script MC carbides into blockier particles, with a corresponding improvement in ductility [29, 365]. Boron also is credited with improving the grain boundary cohesive strength, while zirconium reacts with any residual sulfur to form inert Zr$_2$S particles [29]. Magnesium refines grain boundary M$_6$C particles and promotes precipitation of blocky or globular rather than cellular δ-Ni$_3$Cb, without affecting the morphology of the γ' or γ" hardening phases. Magnesium also improves microstructural stability for long time exposure to high temperature [33].

Another benefit of hafnium is the preferred precipitation of HfC from the melt, rather than TiC. Hafnium carbide in the bulk alloy does not transform into other carbides (eg., MC layers at prior particle boundaries in powder metallurgy alloys, or grain boundary carbide films) [3, 35], and titanium is freed for its role as γ'-former [3]. Addition of too much hafnium or other grain boundary segregants can be detrimental: γ' may overage excessively, there may be cellular nucleation of γ' at grain boundaries, or the incipient melting point may be too severely depressed [3, 29].

Because of their odd sizes, grain boundary segregants hafnium, boron, zirconium, and magnesium fit into vacant lattice positions on the grain boundaries. Once these vacancies are no longer “vacant,” diffusion along grain boundaries is inhibited [1, 29, 34]. The decrease in diffusion lowers the rate of creep due to grain boundary diffusive flow [1], and reduces the rate of penetration of oxygen (or other atmospheric embrittling species) into the material through grain boundaries [3, 34].

Miscellaneous Phases

Due to the complicated alloy chemistry of the superalloys, there are many phases that may form in addition to beneficial γ' and carbide precipitates. These miscellaneous phases may be beneficial or benign, but more frequently they have a detrimental influence on the properties of the alloy. The presence of these phases is typically accompanied by decreases in strength, ductility, rupture life, and corrosion resistance.

The major class of deleterious phases in superalloys are topologically close-packed (TCP) phases. The crystal structures of these compounds are characterized by close-packed atomic planes, widely separated by layers of larger atoms. Being close-packed in only one plane, the structures are referred to as topologically close-packed (TCP). Phases such as γ or γ', close-packed in all directions, are called geometrically close-packed (GCP) [37, 38]. Hard, brittle TCP particles form slowly in an alloy, not during heat treatment, but after long periods of exposure to service conditions. This delayed appearance makes it difficult to
predict the long-term behavior of susceptible alloys [16]. TCP compounds, including σ, μ, and Laves phases, occur predominantly as long, thin plates or needles [1, 7, 8, 16, 38]; the Laves phase can also form as short rods [7]. Widmanstätten plates of TCP phases often stretch from one grain boundary to another; the faces of these plates (parallel to the close-packed TCP planes) are coherent with the octahedral \( (i.e., \{111\} \) close-packed) planes of the \( \gamma \) matrix, while the edges of the thin precipitates are incoherent with austenite [1, 16].

The σ phase has a complex ordered tetragonal crystal structure [14]. The chemical composition of σ can vary between \( \text{A}_7\text{B} \) and \( \text{AB}_7 \), although the typical composition is close to AB; element A comes from Groups IIIB through VIB of the periodic table, commonly chromium and molybdenum, while element B is from Groups VIIB and VIIIB, generally nickel and cobalt [1, 8, 16, 37, 38]. μ has a hexagonal lattice structure, with a chemical structure similar to that of σ, where molybdenum and cobalt dominate; its composition is often given by a variation of \((\text{Co},\text{Ni})_7(\text{Cr},\text{W},\text{Mo})_6\) [1, 8, 38, 39]. The Laves phase has a hexagonal \( \text{AB}_2 \) (or \( \text{MgZn}_2 \)) structure [14, 36, 38], in which hexagonal planar arrays of large A atoms are separated by smaller B atoms [14]. The chemical formula for Laves is typically \((\text{Fe},\text{Cr},\text{Mn},\text{Co},\text{Si})_2(\text{Mo},\text{W},\text{Ti},\text{Cb})\) [8, 37, 40]. σ and μ are electron compounds, \( i.e., \) their bonding is determined by the ratio of electrons to atoms in the structure [14]. σ is comprised of atoms of similar size, while μ is made up of slightly different sized atoms. Laves, on the other hand, is not an electron compound, but a “size” compound; it is composed of atoms with large size differences, and the bonding is due to size-related factors [14, 37, 38].

σ is crystallographically related to \( \text{M}_{23}\text{C}_6 \) and μ is similarly related to \( \text{M}_6\text{C} \); in both cases, the TCP compound has a crystal structure like that of the corresponding carbide, from which carbon atoms have been removed and the positions of the metal atoms have been slightly adjusted to relieve any lattice distortions [1, 38]. Therefore, TCP phases can nucleate coherently on their corresponding carbide [1, 8, 38], although growth seems to proceed by precipitation from the matrix [38]. The presence of molybdenum will result in formation of μ rather than σ, just as molybdenum promotes precipitation of \( \text{M}_6\text{C} \) rather than \( \text{M}_{23}\text{C}_6 \) [1]. In general, alloys which tend to form \( \text{M}_{23}\text{C}_6 \) will be prone to σ formation, while alloys forming \( \text{M}_6\text{C} \) will be susceptible to μ precipitation [38]. Growth of Laves phase is favored by increased amounts of titanium, columbium, tungsten, and molybdenum, and by decreases in nickel, iron, and chromium [40].

There are two ways in which TCP precipitates can embrittle an alloy. Because they form hard, brittle plates, they serve as easy sites for crack nucleation (by fracture of the TCP particle itself) and as paths for crack extension (along the particle-matrix interface) [1, 7, 8,
Additionally, large amounts of refractory elements segregate to the TCP phases; depletion of these elements from the $\gamma$ matrix reduces solid solution strengthening and coherency hardening from the $\gamma$-$\gamma'$ lattice parameter mismatch [38].

Another group of potentially harmful phases have a nominal $\text{Ni}_3\text{X}$ chemical formula, with crystal structures different from the ordered lattice of $\gamma'$-$\text{Ni}_3\text{Al}$. These GCP phases include $\eta$-$\text{Ni}_3\text{Ti}$, $\delta$-$\text{Ni}_3\text{Cb}$, and $\gamma'$-$\text{Ni}_3\text{Cb}$.

$\text{Ni}_3\text{Ti}$, or $\eta$, has a hexagonal close-packed (hcp or $\text{DO}_{24}$) crystal structure [4, 8, 17, 41]. $\eta$ can precipitate in a cellular manner from grain boundaries or as semicoherent intragranular plates (similar to TCP plates); either form is detrimental to strength and ductility of an alloy [1, 7, 8, 41]. $\eta$ occurs by transformation of $\gamma'$ [17, 41], particularly of metastable $\text{Ni}_3\text{Ti}$-$\gamma'$, which forms in alloys with little or no aluminum [17]. Formation of $\eta$ is promoted by a high ratio of titanium to aluminum in the alloy [1, 17]. Tungsten and aluminum retard transformation of $\gamma'$ to $\eta$ [1, 17, 18], while grain boundary boron prevents nucleation of cellular $\eta$ [36].

$\delta$-$\text{Ni}_3\text{Cb}$ has an orthorhombic crystal structure [1, 4, 8, 16, 41]. $\delta$ is found as semi-coherent intracrystalline plates (like TCP and $\eta$ plates) and as random grain boundary globules [8, 36, 41]. $\delta$ has a tendency to form if the amount of columbium in an alloy exceeds about 4% [1, 6].

$\text{Ni}_3\text{Cb}$ also forms as $\gamma'$, which has an ordered body centered tetragonal (bct or $\text{DO}_{22}$) crystal structure [1, 8, 36, 41, 42]. Aluminum and titanium can substitute into $\gamma'$, resulting in $\text{Ni}_3(\text{Cb,Al,Ti})$. $\gamma''$ is very rare except in iron-nickel base superalloys, in which it precipitates as small disks or platelets [1, 36, 41, 42] and provides $\gamma$-$\gamma'$ lattice coherency strain hardening [41, 42].

In alloys with additions of boron, there may be a tendency to form borides. Borides have a tetragonal crystal structure [8] and a chemical formula of $\text{M}_3\text{B}_2$, where the metal atom may be molybdenum or nickel [1, 8, 16], chromium or titanium [1, 16], cobalt [1], or tantalum, columbium, iron, or vanadium [8]. Borides typically have a blocky grain boundary morphology. Grain boundary borides provide no strengthening, but merely act as sources of boron atoms [3, 8, 29, 34].

Typically an alloy will tend to form TCP compounds if the ratio of chromium plus refractory elements to cobalt plus nickel exceeds a critical value [37]. A computer model has been constructed [16], based upon solid state electron theory, to determine whether a given alloy composition will be prone to TCP formation. This model, called PHACOMP, works by subtracting from the total alloy composition the amounts of each element tied up in formation of carbides, borides, and $\gamma'$ (i.e., all phases other than $\gamma$). The remaining
ingredients give the chemical composition of the austenite matrix, from which TCP phases precipitate. For this $\gamma$ composition, PHACOMP determines the number of electron holes $N_\gamma$ (essentially the number of electrons needed to fill the d-electron shell), and determines a weighted average of $N_\gamma$ for the matrix. If this $N_\gamma$ value exceeds some critical value, the alloy may be prone to TCP phase precipitation [16, 38].

PHACOMP works reasonably well in predicting which alloys are prone to TCP precipitation. It is difficult to determine the actual electron hole number for each element [38, 43], and these values may be composition-dependent [38]. A more recent version of the technique makes predictions based on the actual d-electron levels in the $\gamma$ matrix; this model is able to predict more accurately the $\gamma$-TCP phase boundaries for several alloy systems [43].

2.2 Processing of Nickel Base Superalloys.

The processing history of nickel base superalloys is very important to achieving the optimum behavior under the in-service conditions experienced by superalloy components. Figure 2.5 shows a flow chart of several processing procedures, from melt to final product [3, 44]. In a modern gas turbine engine, investment casting is used to produce high-temperature blades and vanes, forgings made from cast ingots or from powder are used as turbine disks, compressor blades, and other structural components, and rolled sheet is formed into combustion liners [4].

Early superalloys were produced from massive cast ingots, which then were rolled or forged into shape. Melting and alloying under a vacuum, by such processes as vacuum induction melting (VIM), vacuum arc remelting (VAR), and electroslag remelting (ESR), result in cleaner alloys, without embrittlement due to oxide stringers, sulfides, and nitrides. These processes provide better control over casting, solidification, and segregation [18, 45, 46]. Cast and wrought (C&W) procedures are still in use to produce disks, compressor blades, and sheet for various applications. C&W components can exhibit high strength, but conventional ingot metallurgy (I/M) processing can lead to problems in superalloys with large amounts of alloying additions. The large size of the ingots limits the rate at which the solidifying alloy can give off heat. The resulting slow cooling rates lead to large grain sizes (up to 6 mm) and to chemical and microstructural segregation across the ingot, for example, large interdendritic carbides and an uneven $\gamma'$ dispersion [3, 18, 47]. Segregation on such a large scale results in inhomogeneities in ductility, strength, corrosion and fatigue resistance, and other properties. These inhomogeneities become more severe as the alloy composition becomes more complicated [3-5]. The use of high temperature soaks to homogenize the chemistry of an ingot is impractical due to the long times necessary to achieve a reasonable degree of homogeneity [45, 47, 48].
Figure 2.5. Superalloy Processing Flow Chart.
Increased alloying results in decreased fabricability for several reasons. The incipient melting point of the alloy is decreased, while the $\gamma'$-solvus is increased, limiting the hot-working temperature range. In addition, the volume fraction of $\gamma'$ in the alloy, and the degree of solid solution strengthening, are increased. These conditions lead to increasingly poor forgeability as the alloy content is increased. A related problem is increased wear of forging dies [6, 45, 48, 49].

To minimize the detrimental effects of casting of large ingots, and to increase material utilization (i.e., to reduce waste) during fabrication, several advances have been made in superalloy processing. Investment casting techniques can be used to achieve complicated shapes precisely, with a minimum of final machining and unutilized material. In addition, internal hollows with complicated geometries, which channel cooling air from the compressor stage of the engine, can easily be incorporated into blade design [18, 46, 47]. Powder metallurgy (P/M) processing also yields near net shape components, reducing machining costs and material waste, while increasing the fabricability of superalloys. In addition to near net shape capabilities, investment casting and powder metallurgy can lead to components with increased $\gamma'$ content (greater than 60% by volume), greater solid solution strengthening, and novel microstructures, and therefore to components with superior properties [18, 48].

**Investment Casting**

Investment casting is the application of the ancient lost wax technique to production of superalloy turbine blades. The technique involves injection molding wax or plastic blade molds and connecting these molds to each other and to guide inlets and runners for molten metal, using wax or plastic guide inlet molds. To provide cooling hollows in the cast blades, ceramic cores can be inserted into these plastic molds. These cores must be able to withstand the high temperature of the molten metal, and readily removable from the final casting. Fused quartz tubes or rods are typically used because silica is easily leached away, while the blade is unattacked by acid. This plastic turbine blade tree is repeatedly dipped into a clay slurry and dried, in order to form a ceramic shell to contain the molten metal. The assembly is then baked, to harden this ceramic shell and to melt and pour out the plastic [46].

The conventional process for casting blades is to pour in molten metal, allowing it to freeze inward from the ceramic mold. This technique leads to reproducible results without the need for extensive process control. The microstructure consists of fine equiaxed grains and very little segregation. The fine grain size results in good fatigue and tensile properties at up to moderately elevated temperatures [48]. The fitting together of adjacent dendritic grains results in undulating or interlocking grain boundaries, while conventional cast and wrought alloys are recrystallized after forming, resulting in smooth, planar boundaries. The contorted
boundaries in investment cast alloys prevent grain boundary sliding and decrease the effective rate of grain boundary diffusion by increasing the diffusion path length [50]. On the other hand, the small grains promote diffusional creep, and the large number of grain boundaries aligned normal to the axis of the blades (i.e., normal to the tensile stress) provide many sites for nucleation of creep cavitation, leading to intergranular fracture [48].

Investment cast blades can be directionally solidified (DS) to eliminate transverse grain boundaries. The mold is constructed with a chill plate at its base and is placed in a furnace when the molten metal is poured. The chill plate and the slow withdrawal of the mold assembly from the furnace result in a steep axial thermal gradient, which induces columnar grain growth away from the chill plate. Growth of the crystals occurs within a few degrees of the <001> direction; since this direction has a low elastic modulus, thermal stresses resulting from temperature gradients along the length of the blade are lower than in a conventionally cast blade, resulting in improved thermal fatigue resistance. The <001> direction also has a greater resistance to creep deformation [18, 35, 47, 48].

The as-cast DS microstructure consists of coarse, columnar grains, providing better resistance against diffusional creep than fine grains, few or no transverse grain boundaries on which creep damage is prone to accumulate, and fewer grain boundaries in general, and therefore less chemical segregation and fewer sites for intergranular corrosive attack [18, 35, 48]. Addition of hafnium to DS alloys strengthens these axial grain boundaries, improving the transverse ductility of the alloy [4, 5, 18]. Due to the high alloy content of the DS material and to very slow solidification rates, these blades experience microstructural segregation. Interdendritic MC carbides and γ-γ' eutectic nodules form on solidification [5, 48], and the distribution of γ' precipitation during heat treatment will be uneven [1]. Because of the small size of the castings, however, the degree of segregation is much less than that found in large conventionally cast ingots, and the chemistry can be partially homogenized during typical solution treatments [1].

Directionally solidified turbine blades and vanes have been in use in advanced military engines since the mid-1960s and in commercial engines since about 1970 [5]. Because of the more complicated mold design and the more involved solidification process (i.e., mold withdrawal), the cost of DS components is up to four times that of their conventionally investment cast components [48]. Through automated production facilities, DS blades can be produced with consistent properties [48], and their thermal fatigue life is improved by a factor of about eight, rupture life by about two, and rupture ductility by about four [5].

The next logical step in casting of turbine blades is removal of all grain boundaries to produce single crystal (SX) blades. This can be accomplished by inserting a helical segment
(or "pigtail") into the mold between the chill plate and the blade form. As the metal freezes, the helix blocks growth of all but one crystal; the remainder of the component has the crystallographic orientation of this remaining grain. Removal of all grain boundaries eliminates sites for segregation and corrosive attack. In addition, elements added to strengthen grain boundaries (hafnium, zirconium, boron, and carbon) are no longer needed; their removal increases the incipient melting point of the alloys, and allows higher solution treatment temperatures and more complete chemical and microstructural homogenization [5, 18, 48].

SX blades are somewhat costlier than DS blades, due to the addition of a helical crystal selector to the mold and to slower solidification. This difference in cost is more pronounced if selection of the transverse crystallographic orientation is desired, because it requires placement of a properly oriented seed crystal into the base of the mold [48]. Due to improvement in rupture life, however, the lifetime-normalized cost of SX blades is less than that of DS blades, which in turn is less than that of conventionally cast blades [18].

A further modification of the directional solidification process is casting of DS eutectic components. This results in an aligned composite structure of strong intermetallic or carbide whiskers or fibers embedded in a more ductile matrix [5-7, 48]. Strengthening fibers, which grow axially between columnar grains of the matrix, can be $\text{M}_7\text{C}_3$ or $\text{M}_3\text{C}_2$ carbides [5], TaC or $\text{M}_{23}\text{C}_6$ carbides, or $\delta$-Ni$_3$Cb intermetallic [7, 47]. These eutectic phases do not exhibit coarsening during exposure to high temperatures, although they may be unstable in thermal cycling [47]. The alloy matrix may be a $\gamma$-$\gamma'$ alloy, as in the $\gamma$-$\gamma'$-$\delta$ DS eutectic [7, 47], or it may be a more simple iron-base alloy with additions of chromium, carbon, and aluminum. A simpler alloy would have a higher incipient melting temperature and lower density, contain lesser amounts of geopolitically strategic elements, and therefore would cost less than nickel base alloys [5]. The lower cost of directionally solidified eutectic components would be offset somewhat by the much slower solidification rates required to grow oriented eutectic fibers [47].

Tailoring of solidification parameters has improved the properties of integrally cast turbine wheels. The small cross-sections of the blades results in fast cooling rates and a fine grain structure, while the large thermal mass of the hub cools much more slowly, producing coarse grains. The trend in blade processing is toward large-grained (and even single crystal) blades, however, to improve creep resistance. Disks are produced with a fine grain structure to improve strength and fatigue resistance. A program of controlled rotation and oscillation during solidification of the molten metal can lead to the desired grain structures. During the initial stages of solidification, the wheel is slowly rotated; this rotation disrupts the convective
heat flow within the mold, resulting in steeper thermal gradients and reduced crystal nucleation, and therefore larger-grained blades. When the blades have solidified, the mold is oscillated, increasing nucleation of new grains and decreasing the final grain size in the hub [47].

Oxidation resistance is provided by a surface layer of Cr₂O₃. The oxide layer adheres well to the alloy and retards oxidation due to slow ionic transport through the layer. Chromium has a small hardening effect in the γ matrix of the alloy (in fact, it displaces elements which provide solid solution hardening to a greater degree), and at very high temperatures, Al₂O₃ provides a more diffusion-resistant oxide film. Removal of chromium from the alloy results in a diminished resistance to hot corrosion. Hot corrosion is the acceleration of oxidation due to sodium or sulfur, which form deleterious (often liquid) phases which attack the oxide layer. Aluminum-enhanced coatings such as NiAl or NiCoCrAlY are applied to blades to provide corrosion resistance. These coatings contain sufficient levels of chromium to prevent hot corrosion, so less chromium is needed in the alloy. Coatings also act as thermal barriers: heat flow is reduced through the oxide layer, so blade temperatures are lower than without coatings [4, 7, 12].

**Powder Metallurgy**

Powder metallurgy (P/M) techniques are used to produce turbine disks with properties superior to those of disks produced by conventional cast and wrought techniques. P/M consolidation techniques produce near net shape (NNS) components in fewer steps and with greater material utilization (i.e., less waste), and therefore at lower cost [35, 45, 51-54]. For example, to produce a 5 kg compressor disk by conventional means, a 95 kg ingot must be cast, forged three times, and rough machined, prior to heat treatment and finish machining. To produce the disk by hot isostatic pressing (HIP) plus hot forging, 33 kg of powder are hot isostatically pressed, forged once, then heat treated and finish machined. Only 18 kg of powder is required to make the disk by HIP alone, followed by heat treatment and final machining. Final machining involves the removal of less material in powder processed disks, resulting in further savings over conventionally produced disks [54].

The earliest P/M superalloys had poor consolidation characteristics and subsequent poor mechanical properties. Atomized in air, powder particles formed skins of TiO₂, Al₂O₃, or Cr₂O₃. These oxides interfere with interparticle cohesion during extrusion or pressing, and in the powder are not reducible into their elemental components [31]. Techniques have evolved for producing powder under an inert atmosphere, after melting in vacuum, to keep out air; atomized powders contain less than 100 ppm of oxygen [5, 31].
Because of the small size and corresponding fast cooling rates of the solidifying droplets of metal, there is no macroscopic chemical segregation like that found in large castings; microsegregation is limited to the size of the “microingot” particles (typically 10-300 μm). Because there is less segregation, more elements can be added to the alloy, and the γ' content of the finished product can be increased to as much as 60% [3, 5, 30, 31, 35, 45, 53-56]. Elimination of gross segregation allows higher solutionizing and service temperatures, as variations in the alloy solidus and the γ' and carbide solvus temperatures are minimized [53]. In addition to macrosegregation, powder processing eliminates other casting defects, such as large grains, inclusion stringers, large carbides and carbide clusters, and shrinkage porosity [18, 35, 56].

Solidified powder is filtered to remove any ceramic inclusions from the slag or broken from the atomization nozzle [18]. This powder is collected and enclosed in a mild steel can prior to consolidation. The can of powder must be evacuated completely; any trapped gas will lead to incomplete interparticle bonding and porosity [30, 31, 51, 57]. Porosity and inclusions will have an adverse effect on the forgeability and mechanical properties of the alloy [30, 51].

Powder consolidation can be accomplished by two methods: extrusion and hot isostatic pressing (HIP). Both procedures are carried out at temperatures somewhat below the γ' solvus, to stabilize grain boundary γ' and to prevent excessive grain growth during subsequent heat treatment. To attain 100% of theoretical density, the extrusion ratio must exceed about 9:1; a typical ratio is 10 to 20:1, leading to an extremely fine grain size and grain aspect ratios of 30 to 60:1 [31].

Densification during the HIP process occurs by creep of small particles to fill spaces between larger particles, which undergo little or no deformation [32]. If HIP is carried out at a temperature greater than the γ' solvus, there is a tendency to precipitate a titanium-rich carbide film at prior particle boundaries (PPBs) surrounding the large undeformed powder particles during consolidation [3, 18, 31, 32, 52, 54, 56]. This MC carbide film consumes carbon from the bulk of the particle, preventing precipitation of grain boundary strengthening carbides, and providing an easy crack propagation path along the brittle “eggshells” of the PPBs [31, 45]. To eliminate PPB embrittlement, most P/M superalloys contain less carbon than their I/M counterparts, and they contain greater amounts of refractory elements to stabilize primary MC carbides. Powder-processed components are HIP at a temperature below the γ' solvus of the alloy; this temperature must remain high enough to ensure recrystallization throughout the component [3, 30, 31, 54, 56, 56].
A forging operation can be carried out following HIP or extrusion to modify further the microstructure of the component. Due to the extremely fine grain size and the great degree of homogeneity resulting from powder consolidation techniques, conventionally unworkable alloys can be forged superplastically. Superplastic deformation is performed isothermally at slow strain rates under relatively low applied forces; a much larger amount of deformation is attained with less die wear than in conventionally forged I/M parts [5, 31, 35, 42, 45, 53-55, 58]. In addition, forging breaks up and disperses ceramic inclusions and PPB carbides and closes up voids and cavities [18, 45, 52, 54].

The improved homogeneity of P/M superalloys results in rejection of fewer parts, and a larger percentage of rejected parts are removed earlier in the manufacturing process. Properties of alloys produced from powder are also more uniform than those of conventionally cast and wrought alloys. The P/M processes result in near net shape (NNS) components; less material is required to produce a part of a given size, so less material is wasted. For these reasons, it is less costly to produce disks from powder than from a cast ingot. Forged P/M disks need little machining after final forging; HIP disks require more machining, but the forging step and the need for expensive forging dies are eliminated [3, 5, 18, 30, 31, 47, 52, 57].

P/M procedures can be modified to achieve further improvements in superalloys. RSP (rapid solidification processing) techniques can produce faster cooling rates in powder than conventional atomization, with a further reduction in segregation, greater chemical and microstructural homogeneity, and an increased incipient melting point [3, 5, 53, 54]. Diffusion bonded wafer turbine blades can be produced from RSP powder through the following procedure. Powder is canned and compacted by extrusion or hot isostatic pressing, and superplastically rolled into thin sheet form. Patterns which will form cooling passages in the blade are etched into these sheets (or wafers). Many wafers are clamped together into a stack, and are diffusion bonded together under a vacuum. The resulting block is directionally recrystallized by zone annealing to produce a coarse, columnar grain structure. Finally, the part is electrochemically machined into blade form [5, 54].

Superalloy powders can be mechanically alloyed with small inert oxide powders, such as ThO<sub>2</sub> or Y<sub>2</sub>O<sub>3</sub>. These insoluble particles provide stable dispersion hardening at greater T/T<sub>M</sub> than is attainable with precipitates such as γ'. Oxide dispersion strengthened (ODS) superalloys need less alloying for high temperature strength, but a γ-γ' structure is retained to provide strength at intermediate temperatures [3, 6, 7, 18]. The higher incipient melting temperature resulting from the lower alloy content of ODS alloys, and their greater strength at high temperatures, enables their use as a turbine blade material [18, 47]. One
possible drawback to their use is that directional recrystallization results in preferred orientation of the [110] crystal axis along the length of the blade; this leads to inferior thermal fatigue and creep deformation resistance [47].

**Thermal and Thermomechanical Processing of Superalloys**

Superalloy heat treatment has developed from a black art into an application of current knowledge of precipitation kinetics and phase stability. The initial step of heat treatment is a solution treatment of the alloy. The solution temperature is determined by three temperatures specific to the alloy composition: the solidus, or incipient melting temperature, the $\gamma'$ solvus, and the carbide solvus temperature. The solidus is higher than the $\gamma'$ solvus; both of these temperatures decrease with increased alloying additions. The carbide solvus is the lowest of the three characteristic temperatures; the carbide solvus increases with increased levels of carbon and carbide-forming elements.

The objective of solution treating is to homogenize the microstructure, dissolving the primary $\gamma'$ and MC carbides, and preparing the $\gamma$ matrix for uniform $\gamma'$ precipitation [1]. These goals are achieved by solutionizing as close to the solidus as possible. Solutionizing disk alloys at such a high temperature is undesirable, however, because it results in excessive grain growth. In practice, solution treatments are carried out slightly below the $\gamma'$ solvus; a small amount of primary $\gamma'$ is retained, pinning grain boundaries to prevent grain growth [3, 32]. The solution temperature must be greater than the carbide solvus to prevent formation of a brittle grain boundary carbide film [3, 10]. Blade alloys are solution treated above the $\gamma'$ solvus to obtain a larger grain size and to increase homogeneity [9]. Solution temperatures range between about 1040° and 1230°C.

Consolidation of P/M alloys is done between the $\gamma'$ and carbide solvus temperatures to maintain a fine grain size [3, 32]. Investment cast components are solution treated for longer periods in order to homogenize their segregated dendritic structure [1]. Single crystal alloys, with no grain boundaries to embrittle, small carbon contents, and very high incipient melting points, can be solutionized at temperatures 100°C higher than standard solution temperatures; this results in a large degree of homogenization of the component [5, 18, 48].

Often the solution treatment consists of two solution anneals. The primary solution dissolves most of the $\gamma'$ which formed upon cooling after solidification. Air cooling from this anneal produces very fine $\gamma'$ precipitates. The secondary solution anneal, carried out at a lower temperature than the primary anneal, dissolves most of the $\gamma'$ which precipitated during cooling from the primary treatment. The remaining $\gamma'$ precipitates grow, and more fine $\gamma'$ forms during air cooling. $M_6C$ carbides may form during solution treating, although $M_{23}C_6$ will not, because it is less stable at high temperatures [1].
Aging treatments are designed to precipitate the optimum dispersion of $\gamma'$ based on the intended use of the component. High temperature creep properties (i.e., rupture life and creep rate) are improved by formation of relatively large $\gamma'$ precipitates, accomplished at high aging temperatures, in the range of 840° to 1100°C. Conversely, tensile strength at low and intermediate temperatures results from finer $\gamma'$, which forms at lower solution temperatures (e.g., about 760°C) [1].

As with solutionizing, aging is often a two-step procedure. Primary $\gamma'$ aging is carried out at a higher temperature to produce large $\gamma'$, resulting in good creep properties. To develop good tensile properties, a secondary aging treatment at a lower temperature follows the primary age [1].

Thermomechanical processing (TMP) can be used to improve the mechanical properties of conventional I/M and advanced P/M superalloys [53, 56, 59, 60]. TMP consists of warm working of the alloy followed by an appropriate heat treatment. Warm working results in a relatively coarse grain structure, strengthened by a fine dislocation cell network anchored in place by small intragranular $\gamma'$ precipitates. The material must be rolled or forged to attain the proper warm worked structure; HIP does not provide the necessary shear deformation to create the dislocation substructure. Because warm working is carried out slightly below the $\gamma'$ solvus temperature, these $\gamma'$ precipitates are present to prevent the dislocation network from annealing out; blocky intergranular $\gamma'$ prevents grain boundary migration, so the warm worked microstructure will not recrystallize. If the alloy were hot worked, that is, deformed above the $\gamma'$ solvus, there would be no $\gamma'$ particles to prevent recovery of the dislocation network or recrystallization of the warm worked grains [52, 54, 56, 58, 60].

Following warm working, the alloy can be recrystallized at a temperature below its $\gamma'$ solvus. Strain-free grains nucleate at boundaries between warm worked grains by subgrain coalescence, but their growth is limited by coarse (1-2 $\mu$m) $\gamma'$ particles, which block grain boundary migration. Nucleation continues on boundaries between recrystallized grains and warm worked grains; no nucleation occurs at PPBs or within warm worked grains. The resulting microstructure has a very fine (2 to 5 $\mu$m) grain size [54, 56, 58, 60].

The properties of thermomechanically processed alloys can be optimized by combining the above microstructures. Because recrystallization proceeds into warm worked grains from their grain boundaries, a partial anneal will produce a "necklace" structure of warm worked grain centers surrounded by fine (2 to 5 $\mu$m) recrystallized grains. As in the fully recrystallized microstructure, coarse $\gamma'$ limits growth of new grains in the necklace by preventing excessive grain boundary movement. The degree of recrystallization increases as
either the anneal time or the amount of deformation during warm working is increased. This necklace is stronger than a fully recrystallized structure, due to retention of warm working, and it has fewer grain boundaries on which to nucleate cavities. The fine recrystallized grains form a tortuous crack path around large warm worked grains; these large grains resist transgranular crack propagation. A turbine disk with a necklace microstructure will be stronger than a recrystallized disk, and more ductile than a fully warm worked disk [54, 56, 58, 60, 61].
3. High Temperature Deformation.


Several parametric methods have been empirically developed to predict lifetimes of high temperature components. Figure 3.1 illustrates lifetime vs. temperature lines for different stress levels assumed by several of these methods. All methods predict shorter lives with increasing stress or temperature, but the exact dependence of lifetime on these conditions differs between methods. Using these parametric life prediction methods for design purposes involves constructing a master plot of stress vs. a polynomial fit of some parameter P, which is defined in terms of temperature and rupture life. A brief description of some parametric methods is given below.

The Larson-Miller parameter is the most widely used parametric life prediction tool [6, 62-69], based on the following dependence of rupture life $t_R$ on temperature:

$$\log t_R = \log t_a + \frac{P}{T} \quad (3.1a)$$

In this equation, $T$ is the temperature, in convenient units, $t_a$ is a constant that represents the value of $t$ at which the lines intersect (in this case, also the t-axis intercept), and $P$ is a function of stress. For design purposes, stress is plotted vs. $P$, given as:

$$P_{LM} = T (\log t_R - \log t_a) \quad (3.1b)$$

The Goldhoff-Sherby approach [64, 65] generalizes that of Larson and Miller by setting the intersection point of the isostress lines away from the t-axis to the point corresponding to $t_a$ and $T_a$. Rupture life is modeled by:

$$\log t_R = \log t_a + P \left[ \frac{1}{T} - \frac{1}{T_a} \right] \quad (3.2a)$$

where the stress-dependent variable $P$ is given by:

$$P_{GS} = \frac{\log t_R - \log t_a}{\frac{1}{T} - \frac{1}{T_a}} \quad (3.2b)$$

When $1/T_a = 0$, $P_{GS} = P_{LM}$, and the two approaches are identical.

In the Orr-Sherby-Dorn approach [6, 62, 64-70], the $\log t$ vs. $1/T$ lines are parallel, according to:

$$\log t_R = P + \frac{B}{T} \quad (3.3a)$$
Figure 3.1. Assumed Dependence of Rupture Life on Temperature for Several Parametric Life Prediction Models ($\sigma_3 > \sigma_2 > \sigma_1$).
The corresponding stress-dependent parameter is:

\[ P_{\text{OSD}} = \log t_R - \frac{B}{T} \tag{3.3b} \]

In contrast to the above models, the Manson-Haferd and the Manson-Succop parametric methods propose linear relationships between \( \log t \) and \( T \) rather than \( 1/T \). The Manson-Haferd model \([63-66]\) is based upon:

\[ \log t_R = \log t_a + P \left[ T - T_a \right] \tag{3.4a} \]

The Manson-Haferd parameter is defined by:

\[ P_{\text{MH}} = \frac{\log t_R - \log t_a}{T - T_a} \tag{3.4b} \]

As in the Goldhoff-Sherby treatment, the lines intercept at the point \((T_a, \log t_a)\); the parameters differ in the temperature dependence of the denominator. The \( \log t \) vs. \( T \) lines on which the Manson-Succop model is based \([64, 67]\) are parallel, like the lines in the Orr-Sherby-Dorn model; rupture life is described by:

\[ \log t_R = P - BT \tag{3.5a} \]

and the Manson-Succop parameter is:

\[ P_{\text{MS}} = \log t_R + BT \tag{3.5b} \]

A recent addition to the catalog of parametric rupture life correlations is that of White and LeMay \([65]\). This approach is based on the rupture life dependence assumed by Goldhoff and Sherby in Equation 3.2a; their stress function is the inverse of the Goldhoff-Sherby parameter:

\[ P_{\text{WL}} = \frac{1}{T} \cdot \frac{1}{T_a} \tag{3.6} \]

A “master curve” which encompasses some of these models defines a parameter \( P(\sigma) \) as follows \([70]\):

\[ P(\sigma) = \frac{(\sigma^q \log t_R) - \log t_a}{(T - T_a)^r} \tag{3.7} \]

When \( q=0 \) and \( r=1 \), for example, this leads to the Manson-Haferd parameter (Equation 3.4b), while for \( q=0, r=-1 \), and \( T_a=0 \), this reduces to the Larson-Miller parameter (Equation 3.1b).
An approach called the minimum commitment method (MCM) is based on a more complicated relationship between stress, temperature, and rupture life [62, 64, 66]:

\[
\log t_R = \frac{P(\log \sigma) - F(T)}{1 + A F(T)}
\]  

(3.8a)

Here, \(P\) is a function of \(\log \sigma\), \(F\) is a function of temperature, and \(A\) is a variable that depends upon the condition of the material. Proposed functions for \(P\) and \(F\) include [66]:

\[
P = A_1 + B_1 \log \sigma + \frac{C_1}{\sigma} \quad \text{for } \sigma \geq \sigma_a
\]  

(3.8b)

\[
P = A_2 + B_2 \log \sigma + C_2 \sigma \quad \text{for } \sigma \leq \sigma_a
\]  

\[
F = R_1 (T - T_a) + R_2 \left( \frac{1}{T} - \frac{1}{T_a} \right)
\]  

(3.8c)

where \(T_a\) and \(\sigma_a\) are arbitrarily chosen near the midrange of temperature and stress values, and \(A_i, B_i, C_i,\) and \(R_i\) are constants. The parameter \(A\) ranges between about -0.2 and 0.2; a value of \(A\) greater than zero indicates that the alloy undergoes a strengthening reaction with time (e.g., precipitation or work hardening); if \(A\) is negative, the alloy is weakening with time (due to recovery of dislocation structure or overaging); \(A=0\) indicates a stable microstructure [64]. The stress-dependent parameter in the minimum commitment method is given by [62, 64, 66]:

\[
P(\log \sigma) = \log t_R + (1 + A \log t_R) F(T)
\]  

(3.8d)

Monkman and Grant [71] proposed the following relationship between minimum creep rate (mcr) and rupture life:

\[
(mcr)^\alpha t_R = C_{MG}
\]  

(3.9a)

For a wide range of alloys, the value of \(\alpha\) ranges between 0.75 and 1.0, with an average of about 0.85 [20, 69, 71, 72]. Because \(\alpha\) is close to unity, the Monkman-Grant relationship is often presented simply as:

\[
(mcr) t_R = C_{MG}
\]  

(3.9b)

The Monkman-Grant constant, \(C_{MG}\), typically ranges from 0.01 to 0.3 [69, 72]; it is often presented as a multiple of the rupture ductility, \(\varepsilon_R\), modifying Equation 3.9b as follows:

\[
\frac{(mcr)^\alpha t_R}{\varepsilon_R} = C'
\]  

(3.9c)

or in logarithmic form,
\[
\log \frac{t_R}{\varepsilon_R} + \alpha \log (\text{mcr}) = C'' \tag{3.9d}
\]

Several of the preceding models can be rationalized through an Arrhenius rate equation approach [6, 62, 65, 68]. Strain rate can be related to stress and temperature by:

\[
\frac{d\varepsilon}{dt} = A \sigma^n \exp \left(-\frac{Q_{\text{app}}}{RT}\right) \tag{3.10}
\]

where \(Q_{\text{app}}\) is an apparent activation energy for creep. Combining Equation 3.10 with the Monkman-Grant relationship (Equation 3.9b) and taking logarithms, this general relationship follows:

\[
\log t_R = \frac{C}{T} - D \tag{3.11a}
\]

If \(C\), but not \(D\), is assumed to be a function of stress,

\[
C(\sigma) = T (D + \log t_R) \tag{3.11b}
\]

which is the Larson-Miller parameter (Equation 3.1b). If, on the other hand, \(D\) is the stress-dependent parameter,

\[
D(\sigma) = \frac{C}{T} \log t_R \tag{3.11c}
\]

This is the Orr-Sherby-Dorn parameter (Equation 3.3b).

The slopes of the \(\log t_R\) vs. \(1/T\) lines at constant stress can be interpreted as a creep activation energy. In the case of stress activated creep processes, the activation energy \(Q\) decreases with increasing stress; the Larson-Miller or Goldhoff-Sherby parameters will give the best fit of the experimental data, and the microstructural parameter \(A\) used in the minimum commitment method is negative. Conversely, if \(Q\) increases with stress, the Manson-Haferd or White-LeMay methods will give a better fit, and the minimum commitment parameter \(A\) is positive. If the lines are parallel, \(Q\) is constant, and the Orr-Sherby-Dorn or Manson-Succop models provide the best fit; in the minimum commitment method, \(A=0\) [65].

Despite the apparent scientific basis for the predictive models described above, all were initially based upon empirical observations of creep behavior. Many of the constants are defined arbitrarily [62, 70], with a vague or nonexistent physical basis for their use [62, 65]. Often the data correlates equally well with the simpler relationships as with the more complicated models [67]. The methods work well enough when used for interpolation of data, but it could be dangerous to use these parameters to extrapolate behavior beyond about three times the longest life measured in the laboratory [65]. Actual behavior can deviate from
predictions because of instabilities such as overaging or oxidation. Because different creep mechanisms are treated together, the models are not able to account for their differing contributions to creep under different conditions [6, 62, 70, 73, 74]. These parametric methods also cannot account for changes in stress or temperature during service [62].

3.2. Creep Strain Rate Equations.

Figure 3.2. Creep Curve Schematic. Primary, Secondary, and Tertiary Creep Regimes Are Illustrated. $\varepsilon_o$ - Initial Elastic Plus Plastic Strain. $\varepsilon_s$ - Steady State Creep Strain. $\varepsilon_R$ - Total Rupture Strain. $mcr$ - Minimum Creep Rate.

A typical curve of creep strain vs. time is shown in Figure 3.2. Before the formulation of creep deformation mechanisms, many empirical relationships were developed in an attempt to model strain accumulation in elevated temperature service components. These equations for strain typically had a term for instantaneous (elastic plus plastic) strain $\varepsilon_o$, a linear term in $t$ (time) to describe steady state creep rate ($d\varepsilon/dt)_{ss}$ (also called the minimum creep rate, or $mcr$), and various fractional powers or exponential functions of time to fit primary creep. Some examples of these equations follow [6, 70, 75, 76]:

$$\varepsilon = \varepsilon_o + \beta t^{1/3} + \left[ \frac{d\varepsilon}{dt} \right]_{ss} t$$  \hspace{1cm} (3.12a)

$$\varepsilon = a_1 + a_2 t^{1/3} + a_3 t^{2/3} + a_4 t$$ \hspace{1cm} (3.12b)
\[ \varepsilon = \varepsilon_0 + \varepsilon_p [1 - \exp (-\gamma_p t)] + \dot{\varepsilon}_{ss} t \quad (3.12c) \]

where $\beta$, $\varepsilon_p$, $\gamma_p$, and $a_1$ through $a_4$ are all arbitrarily defined constants. In an attempt to refine these models by including tertiary creep strains, often another exponential or power function of $t$ was included [6, 69, 70]:

\[ \varepsilon = \varepsilon_0 + b_1 t^{1/3} + b_2 t + b_3 t^3 \quad (3.13a) \]

\[ \varepsilon = \varepsilon_0 + \varepsilon_p [1 - \exp (-\gamma_p t)] + \dot{\varepsilon}_{ss} t + \varepsilon_T \exp[\gamma_T (t - \tau)] \quad (3.13b) \]

where $b_1$, $b_2$, $b_3$, $\varepsilon_T$, and $\gamma_T$ are curve-fitting constants. Some models proposed a balance between primary and tertiary creep, where secondary creep is merely the inflection point of the creep curve [6, 70]:

\[ \varepsilon = \varepsilon_0 + c_1 t^{1/2} + c_2 t^{2} \quad (3.14a) \]

\[ \varepsilon = \theta_1 [1 - \exp (-\theta_2 t)] + \theta_3 [\exp (\theta_4 t) - 1] \quad (3.14b) \]

Again, $c_i$, $\delta_i$, and $\theta_i$ are constants.

It should be emphasized that although these equations may fit the creep curves quite well, they should be considered purely empirical fittings of the creep curve. Most fitting parameters are arbitrarily defined, non-physical constants [6, 70]. The variable $t$ (time) is not an intrinsic material property [77], and many of the formulas, upon differentiation with respect to $t$, predict infinite creep rates at $t=0$ [6, 70]. Unless actual deformation mechanisms are characterized, there are dangers in using the above equations to extrapolate very far beyond the conditions under which the curves were fit. There can be increases in creep rate due to contributions of different creep mechanisms, or due to various mechanical or microstructural instabilities, including necking, particle overaging or precipitation of an embrittling phase, grain growth, oxidation, cavitation, or crack nucleation and growth [6, 24, 70].

Creep strain rates can be defined in terms of stress, temperature, and strain or time as follows [70]:

\[ \frac{d\varepsilon}{dt} = f_1(\sigma, T, t) \quad \text{or} \quad \frac{d\varepsilon}{dt} = f_2(\sigma, T, \varepsilon) \quad (3.15a) \]

Assuming that the functions $f_1$ and $f_2$ can be separated into functions of the individual variables $\sigma$, $T$, and $t$ or $\varepsilon$, these strain rate equations can be rewritten as:

\[ \frac{d\varepsilon}{dt} = u_1(\sigma) v_1(T) w_1(t) \quad \text{or} \quad \frac{d\varepsilon}{dt} = u_2(\sigma) v_2(T) w_2(\varepsilon) \quad (3.15b) \]
Typically, $u(a)$ is given by an exponential, hyperbolic, or power function of $a$, and $v(T)$ is an Arrhenius rate function; the form of $w_1(t)$ or $w_2(\varepsilon)$ is chosen to try to incorporate curvature due to primary and tertiary creep into the creep curve, and often they resemble the empirical fits described above. Some proposed forms for these functions are listed below:

$$u(a) = \begin{cases} \frac{(\sigma/E)^n}{\exp(\sigma/E)} \\ \sinh(\sigma/E) \\ \left[\sinh(\sigma/E)\right]^n \end{cases}$$

$$v(T) = \exp\left(-\frac{Q}{kT}\right)$$

$$w(t) = a + b t^{-2/3} + c t^2 + \frac{\theta_1 \theta_2 \exp(-\theta_2 t) + \theta_3 \theta_4 \exp(\theta_4 t)}{(1 - \omega)^r}$$

The function of strain has been related to transient and steady state dislocation densities [65, 75] or to an increase in applied stress due to reduction of load carrying area [63, 78-80].

A generalized equation for creep strain rate is given by:

$$\frac{d\varepsilon}{dt} = A \mu b \left(\frac{b}{d}\right)^n \frac{\sigma}{\mu} D_o \exp\left(-\frac{Q}{kT}\right)$$

where $\mu$ is the shear modulus, $b$ is the length of the burgers vector, $d$ is the grain size, and $D_o$ is the pre-exponential diffusion constant. $Q$ is the activation energy for diffusion: $Q$ can be $Q_L$ for lattice diffusion, $Q_{GB}$ for grain boundary diffusion, $Q_p$ for dislocation core (pipe) diffusion, or $Q_1$ for solute interdiffusion [6, 75, 81]. The diffusion coefficient $D$ for the appropriate diffusion process is defined by the Arrhenius equation:

$$D = D_o \exp\left(-\frac{Q}{kT}\right)$$

The creep rate is therefore proportional to the diffusion coefficient.

At low stresses, the stress exponent is unity; this condition is referred to as Newtonian viscosity, and can result from Nabarro-Herring creep (lattice diffusional flow), Coble creep (grain boundary diffusional flow), Harper-Dorn creep (a poorly understood intracrystalline dislocation mechanism), or grain boundary sliding [6, 75, 76, 81]. Power law creep occurs at intermediate stresses, where the stress exponent can range from 3 to 7 or higher. At high stresses, power law breakdown is observed; the creep rate varies exponentially with stress [6, 24, 76, 81, 82]:

$$\frac{d\varepsilon}{dt} = A \mu b \left(\frac{b}{d}\right)^n \frac{\sigma}{\mu} D_o \exp\left(-\frac{Q}{kT}\right)$$

where $\mu$ is the shear modulus, $b$ is the length of the burgers vector, $d$ is the grain size, and $D_o$ is the pre-exponential diffusion constant. $Q$ is the activation energy for diffusion: $Q$ can be $Q_L$ for lattice diffusion, $Q_{GB}$ for grain boundary diffusion, $Q_p$ for dislocation core (pipe) diffusion, or $Q_1$ for solute interdiffusion [6, 75, 81]. The diffusion coefficient $D$ for the appropriate diffusion process is defined by the Arrhenius equation:

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$$\frac{d\varepsilon}{dt} = A \mu b \left(\frac{b}{d}\right)^n \frac{\sigma}{\mu} D_o \exp\left(-\frac{Q}{kT}\right)$$
This increase in strain rate at very high stresses can be explained by a greater dislocation density in the material. Dislocation core diffusion is greatly enhanced by the increased dislocation density. In addition, increased numbers of jogs are created on these dislocations when they cut each other. Vacancies are produced by dragging these jogs through the material, resulting in a greater than equilibrium vacancy concentration [76]. A hyperbolic function of stress can be used to fit both power law creep and power law breakdown [6, 24, 73, 75, 82]:

\[
\frac{de}{dt} = C \exp (B \sigma)
\]  

(3.19)

At low stresses this unified creep law reduces to the power law equation, while at high stresses, it models the exponential creep rate.

Diffusional Creep and Grain Boundary Sliding

At low stresses, the dominant mechanism of creep involves stress-induced diffusional flow. Because of the applied stress, some grain boundaries are stressed in tension, and others are in compression. Under the influence of the resulting gradients in stress, there is a net flux of vacancies away from tensile boundaries toward compressive boundaries. The accumulated flux of vacancies results in a macroscopic creep strain rate. This creep rate is proportional to the applied stress, making diffusional creep a special case of power law creep, where the stress exponent is unity [6, 21, 24, 62, 69, 75, 76].

The atoms can diffuse through the bulk of the grains or along grain boundaries. When this transport is through the grains, Nabarro-Herring creep models strain rate as:

\[
\frac{dc}{dt} = C \frac{\sigma \Omega}{kT} \frac{D_L}{d^2}
\]

(3.21a)

where \(C\) is the equilibrium vacancy concentration, \(\Omega\) is the atomic volume, \(d\) is the grain size, and \(D_L\) is the lattice diffusion coefficient. Coble creep models strain rate due to grain boundary diffusion as:

\[
\frac{dc}{dt} = C \frac{\sigma \Omega}{kT} \frac{\pi \delta D_{GB}}{d^3}
\]

(3.21b)

Here, \(D_{GB}\) is the grain boundary diffusion coefficient and \(\delta\) is the effective grain boundary thickness. A combined formula for diffusional creep is [6, 62, 69]:

\[
\frac{de}{dt} = [C_1 \sinh (B_1 \sigma)]^n
\]

(3.20)
\[
\frac{d\varepsilon}{dt} = C \frac{\sigma \Omega}{kT} \frac{D_L}{d^2} \left[ 1 + \frac{\pi \delta}{d} \frac{D_{GB}}{D_L} \right], \text{ or}
\]

\[
\frac{d\varepsilon}{dt} = C \frac{\sigma \Omega}{kT} \frac{D_L}{d^2} + \frac{\pi \delta D_{GB}}{d^3}
\]

(3.22)

In addition to the tensile and compressive grain boundaries described above, many are subjected to a shear stress; this causes two adjacent grains to slide along their shared boundary. If the boundary is planar, there is no resistance to grain boundary sliding (GBS); therefore sliding does not limit the strain rate, but occurs to accommodate deformation of the grains by any other mechanism [75, 83]. Most real grain boundaries are not flat, but twisted or curved. Such irregularities along a boundary give rise to large normal stresses across the boundary after very small amounts of sliding; the macroscopic creep rate may be controlled by this grain boundary friction. The sliding rate of two grains across a boundary is given by [69, 72, 83]:

\[
\frac{dU}{dt} = \frac{\tau_a}{\eta_{GB}}
\]

(3.23)

where \(\tau_a\) is the applied shear stress on the grain boundary, and \(\eta_{GB}\) is the shear viscosity of the grain boundary.

In order for steady state GBS to proceed, some mechanism must be present to relieve normal stresses across the grain boundaries. Elastic accomodation of sliding results in high reverse stresses that oppose any further sliding along the boundary. A mechanism of diffusional transport of matter from compressive to tensile boundary regions will accomodate normal stresses and allow sliding to continue. Assuming a sinusoidal grain boundary, the sliding rate can be expressed as [75, 83]:

\[
\frac{dU}{dt} = \frac{8 \tau_a \Omega \lambda}{\pi kT h^2} D_L \left[ 1 + \frac{\pi \delta}{\lambda} \frac{D_{GB}}{D_L} \right], \text{ or}
\]

\[
\frac{dU}{dt} = \frac{8 \tau_a \Omega}{\pi kT h^2} \left[ \lambda D_L + \pi \delta D_{GB} \right]
\]

(3.24)

where \(\lambda\) and \(h\) are the wavelength and amplitude of the sine function approximating the shape of the grain boundary, and the other terms have been defined above. \(\lambda\) and \(h\) can be related to grain size, and \(\tau_a\) is proportional to the externally applied stress \(\sigma\). The shear strain rate is
equal to the sliding rate divided by the grain size; this shear strain rate is proportional to the macroscopic longitudinal strain rate, given by:

$$\frac{de}{dt} = C \frac{\sigma \Omega}{kT} \frac{1}{d^3} \left[ D_L + \frac{\delta D_{GB}}{d} \right]$$  \hspace{1cm} (3.25)

This is identical to Equation 3.22 above, which describes strain accumulation by diffusional flow by the Nabarro-Herring or Coble creep mechanisms. Because diffusional flow and grain boundary sliding models produce identical strain rate equations, the two processes are closely coupled: both mechanisms must occur simultaneously in order to deform the material [21, 27, 83].

The grain boundary sliding mechanism can be adapted to describe the effect of intergranular particles on sliding resistance. The sliding rate of a boundary containing an array of discrete impermeable particles, with size $a$ and spacing $\lambda$, is given by:

$$\frac{dU}{dt} = 1.6 \frac{\tau a \Omega \lambda^2}{kT a^3} D_L \left[ 1 + 5 \frac{\delta D_{GB}}{D_L a} \right], \text{ or}$$

$$\frac{dU}{dt} = 1.6 \frac{\tau a \Omega \lambda^2}{kT a^3} \left[ \frac{D_L}{a^3} + 5 \frac{\delta D_{GB}}{a^4} \right]$$  \hspace{1cm} (3.26)

From these equations it is seen that, for a given volume fraction, large particles are more effective than small particles at slowing the sliding rate. It has been concluded, however, that the grain size, rather than grain boundary particle size, determines macroscopic creep rates [83]. Particles have the added effect of restricting the ability of the grain boundary to emit and absorb vacancies and otherwise interfere with microscopic grain boundary sliding; this interference leads to a threshold stress for sliding. Particles also have been observed to distort grain boundaries in an alloy (e.g., borides and carbides in nickel base superalloys), decreasing the wavelength $\lambda$ of grain boundary fluctuations, and thereby decreasing the rate of sliding [83].

The activation energy for Coble creep ($Q_{GB}$) is less than that for Nabarro-Herring creep ($Q_L$); therefore, Coble creep is more important at relatively lower temperatures. In Coble creep, the strain rate exhibits a $d^{-3}$ dependence on grain size, while in Nabarro-Herring creep, the strain rate varies as $d^{-2}$. Due to the difference in grain size dependency, Coble creep dominates in finer grained materials, while Nabarro-Herring creep is more important as grain size increases [6, 81, 83]. Grain boundaries are necessary to act as vacancy sources and sinks, and Coble creep requires boundaries along which to transport vacancies and
atoms; therefore, neither mechanism contributes to the creep of large-grained investment castings or single crystal blades [6, 81].

In addition to elongation of grains in the tensile direction, direct evidence of diffusional creep can be found in alloys containing precipitates or dispersoids. As vacancies flow from tensile to compressive boundaries, the uniform distribution of particles is disrupted. Because atoms leave regions around compressive boundaries, particles near these boundaries become more closely spaced; eventually these particles become packed along compressive boundaries. At the same time, migrating atoms plate out onto tensile boundaries, with a resulting particle free zone on either side of the boundaries. Often this behavior is interpreted as denudation of the precipitate phase near tensile grain boundaries and coarsening at compressive boundaries. The formation of a precipitate free zone results in localized weakening and eventual failure at grain boundaries oriented perpendicular to the maximum tensile stress axis [21, 27].

Power-Law Creep

Nearly all creep deformation in engineering alloys can be described by power law creep. The basic equation for power law creep is:

\[
\frac{de}{dt} = A \frac{\mu b}{kT} \left( \frac{\alpha}{\mu} \right)^p Q^n D_0 \exp\left( -\frac{Q}{kT} \right)
\]

(3.17)

The grain size exponent \( p \) is non-zero only for the diffusional creep mechanisms described above, for which the stress exponent is unity. Intragranular dislocation motion depends not on grain size but on smaller microstructural features (for example, dislocation density, subgrain size, or particle size and spacing), some of which which are dependent upon stress; their effects are incorporated into the stress exponent \( n \) [6, 81]. If two competing mechanisms are controlling creep strain, the mechanism with the larger stress exponent will be dominant at higher stresses. For example, the exponent for diffusional creep is one, while that for dislocation mechanisms is at least three; diffusional creep is dominant at lower stresses, dislocation glide or climb controls higher stress creep [6, 27].

Table 3.1 [75, 76, 81] gives values of stress exponent \( n \) and pre-exponential factor \( A \) from Equation 3.17 for pure metals and two classes of solid solution alloys. Dislocation motion in these classes of materials occurs by two mechanisms: glide on the slip plane between obstacles, and thermally activated climb over obstacles [27, 75, 76, 81]. In pure metals, there is little resistance to glide, so climb around obstacles controls the strain rate. Dislocation climb models predict a typical stress exponent of about 4.5; according to Table 3.1, pure metals have an exponent of 5 [6, 24, 75]. Class II (Class M) alloys also
deform according to $\sigma^5$ at high temperatures. Strain rates are controlled by dislocation climb; the activation energy is the same as that for lattice self diffusion [27, 75, 76, 81].

In Class I (or Class A) alloys, high temperature strain rates depend upon $\sigma^3$. The transition to Class I behavior is enhanced by large atomic size misfits, large stacking fault energies, and large solute concentrations. Size misfit increases drag on dislocations gliding on the slip plane, and drag increases with solute concentration; large stacking fault energies do not allow dislocations to become extended, so they are able to cross slip around obstacles. The creep rate limiting mechanism in these alloys is not dislocation climb, but rather dislocation glide between obstacles. The stress exponent is three, and the activation energy is that for solute interdiffusion [75, 76, 81]. At lower temperatures, the stress exponents increase to 5 for Class I alloys and to 7 for Class II alloys. The activation energy for either controlling mechanism is that for diffusion along the dislocation cores [27, 75, 76, 81].

In precipitation strengthened alloys, measured values of the stress exponent can range from a low of 5 up to about 40, while oxide dispersion strengthened alloys may exhibit exponents as large as 75. Such a strong stress dependence can be attributed to a large internal (or threshold) stress, below which dislocation motion is very difficult due to the presence of particles, but above which, dislocations move easily around the particles. The apparent activation energy for creep also is very large in particle hardened alloys, taking a value two to three times the activation energy for lattice diffusion [6, 21, 24, 27, 75, 82, 84-88].

**Table 3.1. Power Law Creep Exponents and Constant Factors.**

<table>
<thead>
<tr>
<th>System</th>
<th>Stress Exponent n Range (Typical)</th>
<th>Constant Factor A Range (Typical)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Pure fcc</td>
<td>4.4-5.3</td>
<td>$10^2$-$10^8$</td>
</tr>
<tr>
<td>Pure bcc</td>
<td>4-7</td>
<td>$10^5$-$10^{15}$</td>
</tr>
<tr>
<td>Pure hcp</td>
<td>4-6</td>
<td>$10^3$-$10^8$</td>
</tr>
<tr>
<td>Class I (A)</td>
<td>3-4</td>
<td>$10^{-2}$-$10^4$</td>
</tr>
<tr>
<td>Class II (M)</td>
<td>4.5-6</td>
<td>$10^5$-$10^9$</td>
</tr>
</tbody>
</table>

**Hardening Mechanisms**

In nickel base superalloys, the nickel matrix is strengthened by solid solution additions of cobalt, chromium, molybdenum, tungsten, iron, and other elements. These elements segregate to dislocations, reducing mobility due to solute drag. Before dislocations
can move, they must break away from the solute atoms. Solute atoms may reduce diffusivity in the alloy, slowing diffusion assisted climb around obstacles in the slip plane. These elements decrease stacking fault energy, enabling dislocations to become extended, and therefore making cross slip, climb, and recovery more difficult. Extended dislocations also make very poor point defect sources and sinks [6, 21, 27]. Creep strain rates are reported to depend on stacking fault energy raised to a power of about 3 in fcc alloys, and the stress exponent increases with decreasing stacking fault energy [27, 75, 76].

Three types of particles can strengthen nickel base superalloys: coherent precipitates (i.e., \( \gamma' \)), incoherent precipitates (carbides), and inert dispersoids (insoluble oxide particles in oxide dispersion strengthened systems). The strain hardening exponent in particle strengthened alloys ranges from 5 to 40, reaching as high as 75 in ODS alloys [6, 21, 24]. The creep activation energy is two to three times the activation energy for lattice diffusion. Such high values for these parameters are attributed to interactions between dislocations and particles [24].

Depending on the size, spacing, and mechanical properties of the particles, there are several mechanisms by which an alloy is strengthened by their presence [21, 27, 89]. Passage of dislocations through small coherent precipitates provides resistance to plastic flow. As particles become larger, stronger, and less coherent, dislocations bypass them by bowing around them, leaving dislocation loops behind, or by cross-slip onto a different slip plane. Due to their low stacking fault energies, nickel base alloys are not subject to cross slip; dislocations become widely extended, with large stacking fault areas between partial dislocations, making it nearly impossible for a dislocation to initiate slip on another plane. Cutting, bowing, and cross-slip mechanisms operate at low temperatures, and at high temperatures they are accelerated through thermal activation. An additional mechanism operating at high temperatures is diffusion assisted climb of dislocations over the particles.

The main resistance to particle cutting by dislocations in Superalloys is due to formation of an antiphase boundary (APB) by the passage of the first dislocation. The APB is erased by the second dislocation. An approximate relationship for the increase in strength \( \Delta \tau \) due to APB formation is [89]:

\[
\Delta \tau = \frac{2f \Gamma_{\text{APB}}}{b}
\]

where \( b \) is the burgers vector, \( f \) is the volume fraction of the particles, and \( \Gamma_{\text{APB}} \) is the antiphase boundary energy. A more exact formula for the interaction of a single dislocation with a particle is:
\[ \Delta \tau = \frac{\Gamma_{\text{APB}}}{2b} \left[ \frac{8\Gamma_{\text{APB}} f r}{\pi G b^2} \right]^{1/2} \]  

(3.28)

where \( r \) is the radius of the particle and \( G \) is its shear modulus \([19, 21]\). For paired dislocations traveling through the alloy, the strengthening is decreased as follows because the stress field of the second dislocation forces the first dislocation through the particle, and passage of the second dislocation heals the antiphase boundary \([19, 21, 24]\):

\[ \Delta \tau = \frac{\Gamma_{\text{APB}}}{2b} \left[ \left( \frac{8\Gamma_{\text{APB}} f r}{\pi G b^2} \right)^{1/2} - f \right] \]  

(3.29)

This analysis assumes that the first dislocation penetrates the particle. The analysis is more complicated if the first dislocation bows around the particle rather than cutting it. A loop is left around the particle; the second dislocation pushes the loop through the particle, cuts the particle itself, and follows the first dislocation. The loop left by the first dislocation is annihilated once it has cleared the precipitate \([19, 21, 24]\). Very large \( \gamma' \) particles are more difficult to cut, and therefore can support multiple loops before penetration occurs \([24]\).

The lattice parameter misfit between matrix and particle gives rise to another resistance to particle shearing \([19, 21, 89]\). The strain field surrounding a particle exerts a force on a dislocation, giving rise to the following increase in strength:

\[ \Delta \tau = AG \sqrt{\frac{rf}{b}} \varepsilon^{3/2} \]  

(3.30)

where \( \varepsilon \) is the coherency or lattice misfit strain and \( A \) is a constant. A mismatch in the elastic moduli of the precipitate and matrix also produces a small increase in strength. In superalloys, increments in strength due to lattice parameter and modulus differences are much less than the strengthening caused by APB formation.

As precipitates increase in size, and as their APB energy or lattice mismatch strain increases, the particles become impenetrable; they behave like incoherent dispersoids. Dislocations are forced by an applied stress to bow out between particles; when the dislocation arc between two particles becomes semicircular, it breaks free, leaving a dislocation loop around each particle. The strengthening from this mechanism is approximated by the Orowan relationship \([19, 21, 89, 90]\):

\[ \Delta \tau = \frac{Gb}{\lambda} \]  

(3.31)

where \( \lambda \), the particle spacing, is related to particle size and volume fraction by \([21]\):
The exact definition of an average spacing $\lambda$ depends upon the assumed distribution of particle sizes and spacings [21]. For large volume fractions, particles no longer act as point strengtheners, and it is necessary to define an effective spacing, equal to the center-to-center spacing minus the particle diameter [21, 24]. Additionally, as loops accumulate around a particle, its effective radius increases with each successive loop [19, 21, 89, 90].

The particle strengthening behavior for an alloy with a constant volume fraction of second phase is shown schematically in Figure 3.3 [19, 27, 89]. For small particle sizes, dislocations cut the particles, and the degree of strengthening from APB formation or coherency hardening increases with $r^{1/2}$ (see Equations 3.28, 3.29, and 3.30). At larger sizes, dislocation bowing results in strengthening proportional to $r^{-1}$ (see Equations 3.31 and 3.32). The result of these two competing phenomena is a maximum hardness at an intermediate particle size; the goal of heat treatment is to produce a large volume fraction of precipitates having this size. The time dependent mechanical properties of precipitation hardened alloys in high temperature service is explained by the above behavior. As aging occurs, smaller particles dissolve and larger particles grow; the average particle size increases, while the volume fraction of the second phase does not change. The alloy shows an early increase in strength, followed by a steady weakening with time.

\[
\lambda = 1.25 \, r \sqrt{\frac{2\pi}{3f}}
\]  

(3.32)

Internal Stress

For many engineering alloys, measured values of the strain hardening exponent can become quite large, and the apparent creep activation energy exceeds the activation energy of the operating diffusion mechanism. This effect is particularly noticeable in high-strength,
creep-resistant, particle-hardened alloys, such as γ-γ' nickel base superalloys and ODS alloys. To account for this phenomenon, a stress σ₀, variously referred to as an internal stress, friction stress, resisting stress, or threshold stress, modifies the power law creep rate equations as follows [3, 20, 21, 24, 27, 69, 75, 82, 84-88]:

\[
\frac{de}{dt} = A \left[ \frac{(\sigma - \sigma_0)}{E} \right]^{n_0} \exp \left( \frac{Q_0}{kT} \right)
\]

In this equation, \(n_0\) and \(Q_0\) are the “true” stress exponent and creep activation energy. The effective stress is \(\sigma - \sigma_0\), which determines the size of the dislocation network in the creeping material according to:

\[
\sigma = \sigma_0 + \alpha \mu b \sqrt{\rho}
\]

where \(\mu\) is the shear modulus, \(b\) is the Burgers vector, and \(\rho\) is the dislocation density. Recovery involves the growth of the dislocation network, and \(\rho\) is proportional to \((\sigma - \sigma_0)^3\); hardening involves refinement of this network, and is proportional to \((\sigma - \sigma_0)^{-1}\). Steady state creep occurs when the recovery rate is balanced by the hardening rate, so the steady state creep rate is proportional to \(r/h\) [20, 69, 75, 76, 84-88]; therefore, \(n_0 = 4\). The apparent stress exponent \(n\) and activation energy \(Q\) increase as the internal stress increases, although the values for \(n_0\) and \(Q_0\) remain constant. The fact that these parameters are constant, and consistent between alloy systems with different strengthening mechanisms, has led investigators to conclude that creep is controlled by similar processes in many alloy systems, and that a universal creep rate equation incorporating internal stress can be developed to describe high temperature deformation [20, 24, 84-88].

The internal stress (or “back” stress) is attributed to the Peierls lattice resistance, solute drag by solid solution atoms, the creation of antiphase boundaries or surface steps in strengthening particles [24, 84], or back motion of curved dislocations and bowed subgrain boundaries [75, 85]. The magnitude of the back stress in particle hardened systems has been equated to the Orowan stress or the stress necessary to shear γ' precipitates [85, 91]. For a constant γ' volume fraction, as particle size and spacing increase (e.g., during ripening at elevated temperatures), dislocation bowing becomes easier, so the Orowan stress decreases; the friction stress has also been observed to decrease [91].

**Deformation Mechanism Maps**

It is possible to construct diagrams, such as Figure 3.4, which indicate the operating creep mechanism for given loading conditions [27, 62, 69, 72, 92]. Within each region on the map, one mechanism will produce the greatest creep rate for the given normalized stress.
and homologous temperature. The boundary between two regions is the locus of points for which the models of the two creep mechanisms predict equal creep rates. For example, the vertical line in the lower half of Figure 3.4 represents equal creep rates by bulk diffusive flow (Nabarro-Herring creep) and by grain boundary diffusive flow (Coble creep). According to this deformation map, bulk diffusion dominates over boundary and dislocation core diffusion at higher temperatures. Power law creep dominates over diffusional flow as stress increases; at much higher stresses, power law breakdown and even plastic dislocation glide may occur. As grains become finer, the diffusional flow region becomes larger, and the boundary diffusion regime reaches into that of bulk diffusion. These predictions are in agreement with observations noted earlier in this chapter.

![Deformation Mechanism Map](image)

Figure 3.4. Representative Deformation Mechanism Map. $\mu$, Shear Modulus. DF, Diffusional Flow. PL, Power Law Creep: High Temperature by Lattice Diffusion, Low Temperature by Dislocation Core Diffusion. Dotted Lines Indicate Behavior of Smaller Grained Component. Shaded Area Represents Regime of Typical Turbine Operation.

Creep mechanism maps can be altered to show strain rate, rupture life, or grain size on either axis, depending on the desired application of the maps. In addition, maps can be constructed to show regions of predominance of fracture or cavity growth mechanisms, rather than creep mechanisms.
Tertiary Creep

As time elapses and creep strain accumulates, an acceleration of creep rate is observed. This increase in creep rate can be due to reduction in cross-sectional area, with a resulting increase in stress and strain rate. There may also be a microstructural deterioration, which results in softening of the alloy. For example, dislocation pinning precipitates may coarsen or dissolve. As particles become larger and more widely spaced, the stress necessary to bow dislocations out between them is decreased. Growth of particles may also deplete the matrix in certain solid solution strengthening elements. Embrittling phases, such as TCP phases in superalloys, may precipitate, promoting microcrack formation throughout the alloy. The dislocation substructure may undergo recovery, reducing interactions between dislocations, and reducing the incidence of dislocation cutting. Grain growth and recrystallization produce new, undeformed grains with less resistance to dislocation motion. Finally, nucleation and growth of intergranular cavities occurs, resulting in a smaller effective load carrying area, and therefore higher stresses and creep rates. As cavities grow together and form cracks, the strain rate continues to increase, up to fracture of the material. All these effects, reduction in area due to strain, microstructural degradation, and cavity and crack formation, are combined under the general term “creep damage” [6, 24, 26, 28, 62, 93-95].

3.3. Continuum Damage Mechanics.

Norton’s law can be used to derive an expression for the rupture life of a creeping specimen. Norton’s equation for power law creep can be given as simply:

\[
\frac{de}{dt} = B \sigma^n
\]  

(3.35)

As strain accumulates, the cross sectional area \(A\) decreases, and the actual net section stress on the specimen increases with strain according to:

\[
\sigma = \frac{P}{A} = \frac{P}{A_i} \frac{A_i}{A} = \frac{\sigma_i}{A_i} A_i \frac{L_i}{L} = \sigma_i \exp(\epsilon)
\]  

(3.36)

where \(\sigma_i\) is the initial applied stress, \(A_i\) and \(A\) are the initial and current cross sectional areas, and \(L_i\) and \(L\) are the initial and current gauge lengths of the specimen. Substituting Equation 3.36 into Equation 3.35 gives:

\[
\frac{de}{dt} = B \sigma^n \exp(ne)
\]

(3.37)

where \(\sigma\) now denotes the initial stress. Integrating from \(\epsilon=0\) to \(\epsilon_R\) and \(t=0\) to \(t_R\), one obtains:

\[
\epsilon_R = -\frac{1}{n} \ln(1 - nB\sigma^n t_R)
\]

(3.38)
for strain at failure, and

\[ t_R = \frac{1 - \exp(-n\varepsilon_R)}{n \sigma_0^n} \]  

(3.39)

for rupture life. Given large values of \( n \) or \( \varepsilon_R \), Equation 3.39 reduces to:

\[ t_R = \frac{1}{n \sigma_0^n} \]  

(3.40)

These equations for rupture life and ductility apply to ductile (viscous) fracture due to creep strain accumulation [63, 79, 96-98].

Most creep-resistant alloys used in high temperature applications do not exhibit great amounts of ductility at fracture. In an attempt to model creep behavior of such an alloy, a scalar damage parameter \( \omega \) has been introduced. This parameter, the basis of continuum damage mechanics (CDM), is essentially a measure of the reduced ability of the material to support the applied load, can be defined in terms of cross-sectional area as:

\[ \omega = \ln \frac{A}{A_R} \]  

(3.41)

As time elapses, \( \omega \) increases, and the effective load carrying area \( A_R \) decreases; the stress on this area increases according to [96]:

\[ \sigma = \sigma_i \frac{A_i}{A_R} = \sigma_i \frac{A_i}{A} \frac{A}{A_R} = \sigma_i \exp(\varepsilon + \omega) \]  

(3.42)

For a brittle material, \( \omega \) is small and \( \varepsilon \) can be neglected, so Equation 3.42 reduces to:

\[ \sigma = \frac{\sigma_i}{1 - \omega} \]  

(3.43)

The quantity \((1 - \omega)\) is often interpreted as the reduction in load-carrying area by nucleation and growth of voids in the material. Norton’s Law can be rewritten to incorporate the damage parameter, and a similar relationship can be defined for the development of damage with time [63, 78, 79, 80, 93, 96-109]:

\[ \frac{d\omega}{dt} = C \left( \frac{\sigma}{1 - \omega} \right)^k \]  

(3.44a)

\[ \frac{d\varepsilon}{dt} = B \left( \frac{\sigma}{1 - \omega} \right)^n \]  

(3.44b)
where \( \sigma \) denotes the initial stress, or more generally:

\[
\frac{d\omega}{dt} = \frac{C \sigma^k}{(1 - \omega)^r} \tag{3.45a}
\]

\[
\frac{d\varepsilon}{dt} = \frac{B \sigma^n}{(1 - \omega)^q} \tag{3.45b}
\]

Equation 2.45a is integrated from \( \omega = 0 \) to 1 to obtain the rupture life \( t_R \):

\[
t_R = \frac{1}{(r + 1) C \sigma^k} \tag{3.46}
\]

Figure 3.5 shows the brittle and ductile extremes in rupture life dependence on stress [63, 79, 97]. At high stress levels, the material fails by ductile rupture, with lifetime dependent upon \( \sigma^n \), while at low stresses, brittle failure modes result in a \( \sigma^k \) dependence on life. Typically \( n > k \), as shown in the figure. Damage and strain at time \( t \) are given by:

\[
\omega = 1 - \left[ 1 - (r + 1) C \sigma^k t \right]^{1 \over r + 1} = 1 - \left( 1 - {t \over t_R} \right)^{1 \over r + 1} \tag{3.47}
\]

\[
\varepsilon = B \sigma^n t_R \left[ \frac{1}{r + 1 - q} \right] \left[ 1 - \left( 1 - {t \over t_R} \right)^{1 \over r + 1} \right], \text{ or}
\]

\[
\varepsilon = \varepsilon_R \left[ 1 - \left( 1 - {t \over t_R} \right)^{1 \over \lambda} \right] \tag{3.48}
\]

The damage tolerance parameter \( \lambda \) is defined by:

\[
\lambda = \frac{r + 1}{r + 1 - q} = \frac{\varepsilon_R}{\varepsilon_S} = \frac{\varepsilon_R}{t_R(mcr)} \tag{3.49}
\]

and rupture strain \( \varepsilon_R \) by:

\[
\varepsilon_R = \lambda \varepsilon_S = \lambda t_R(mcr) = \frac{\lambda B \sigma^n}{(r + 1) C \sigma^k} = \frac{1}{(r + 1 - q) C} \sigma^{n-k} \tag{3.50}
\]

The damage tolerance parameter \( \lambda \), the ratio of rupture strain \( \varepsilon_R \) to steady state creep strain \( \varepsilon_S = t_R(mcr) \), is a measure of the tertiary creep strain which the material can accommodate, as illustrated in Figure 3.6 [63, 79, 80, 93, 96-102]. According to Equation 3.50, strain at
Figure 3.5. Stress Dependence of Rupture Life in Ductile and Brittle Regimes.

Figure 3.6. Form of a Typical Creep Curve, Defining Primary, Secondary, and Tertiary Creep Regions, Minimum Creep Rate, and Steady State and Rupture Strains.

Rupture increases as stress increases because $n > k$. Figure 3.7 illustrates the stress dependences of strain rate, rupture life, and ductility [63].

Damage is presented in terms of time in Equation 3.47. This quantity can also be defined as a function of strain:

$$\omega = 1 - \left( 1 - \frac{\varepsilon}{\varepsilon_R} \right)^{\frac{1}{r+1}} = 1 - \left( 1 - \frac{\varepsilon}{\varepsilon_R} \right)^{\frac{1}{r + 1 - q}}$$  \hspace{1cm} (3.51)
Rates of accumulation of strain and of damage can be defined in terms of damage, as in Equations 3.45a and 3.45b, in terms of fractional life [63, 79, 80, 93, 96-102]:

\[
\frac{d\omega}{dt} = C\sigma^k \left(1 - \frac{t}{t_R}\right)^\frac{-r}{r+1}
\]

(3.52a)

\[
\frac{d\epsilon}{dt} = B\sigma^n \left(1 - \frac{t}{t_R}\right)^\frac{-q}{r+1}
\]

(3.52b)

or in terms of strain:

\[
\frac{d\omega}{dt} = C\sigma^k \left(1 - \frac{\epsilon}{\epsilon_R}\right)^\frac{-\lambda r}{r+1} = C\sigma^k \left(1 - \frac{\epsilon}{\epsilon_R}\right)^\frac{-r}{r+1 - q}
\]

(3.53a)

\[
\frac{d\epsilon}{dt} = B\sigma^n \left(1 - \frac{\epsilon}{\epsilon_R}\right)^\frac{-\lambda q}{r+1} = B\sigma^n \left(1 - \frac{\epsilon}{\epsilon_R}\right)^\frac{-q}{r+1 - q}
\]

(3.53b)

Although these relationships are derived arbitrarily, the latter two equations make more sense from a physical standpoint: time is not a material property, and it is reasonable to expect that damage is somehow dependent upon strain. In fact, for a case of strain-controlled damage, the stress exponents \(n\) and \(k\) are equal; according to Equation 3.50, \(\epsilon_R\) and \(\epsilon_S\) are independent of stress [99]. The material satisfies the Monkman-Grant relationship (Equation 3.9b):
The continuum damage mechanics treatment can be modified to account for multiaxial states of stress [93, 97, 98, 103, 104, 110]. First, the equivalent stress $\sigma_e$ and strain $\varepsilon_e$ must be defined:

\[
\sigma_e = \left[ \frac{3}{2} S_{ij} S_{ij} \right]^{1/2} = \frac{1}{\sqrt{2}} \left[ (\sigma_1 - \sigma_2)^2 + (\sigma_2 - \sigma_3)^2 + (\sigma_3 - \sigma_1)^2 \right]^{1/2} \quad (3.55a)
\]

\[
\varepsilon_e = \frac{2\sigma_e}{3S_{ij}} = \frac{\sqrt{2}}{3} \left[ (\varepsilon_1 - \varepsilon_2)^2 + (\varepsilon_2 - \varepsilon_3)^2 + (\varepsilon_3 - \varepsilon_1)^2 \right]^{1/2} \quad (3.55b)
\]

where $\sigma_1$, $\sigma_2$, $\sigma_3$ and $\varepsilon_1$, $\varepsilon_2$, $\varepsilon_3$ are the principal stresses and strains, and $\varepsilon_{ij}$ and $S_{ij}$ refer to the strain tensor and the deviatoric stress tensor. Equation 3.34 is modified to give the power law relationship:

\[
\frac{d\varepsilon_{ij}}{dt} = \frac{3}{2} B\sigma_e^{n-1} S_{ij} \quad (3.56)
\]

The contribution of damage to strain rate is included as in Equation 3.45b [82, 93, 97, 98, 104-106, 110]:

\[
\frac{d\varepsilon_{ij}}{dt} = \frac{3}{2} B\sigma_e^{n-1} S_{ij} (1 - \omega)^{-q} \quad , \text{or} \quad (3.57a)
\]

\[
\frac{d\varepsilon_{ij}}{dt} = \frac{3}{2} B\sigma_e^{n-1} S_{ij} \omega^{-q} \quad , \text{where} \quad (3.57b)
\]

\[
\sigma = \frac{1}{\sqrt{2}} \left[ \left( \frac{\sigma_1}{1 - \omega} - \sigma_2 \right)^2 + (\sigma_2 - \sigma_3)^2 + \left( \frac{\sigma_3}{1 - \omega} - \sigma_1 \right)^2 \right]^{1/2} \quad (3.58)
\]

Although stress and strain are given as tensor quantities, due to the complexity of the mathematical treatment, damage is presented as a scalar; Equation 3.58 is constructed so that damage affects stress in a tensor fashion.

The form of the damage rate law (Equation 3.45a for uniaxial stress) depends on whether the material accumulates damage according to the maximum principal stress, the maximum equivalent stress, or by some mixed behavior [62, 80, 93, 98, 99, 106, 109-111].
If damage accumulation depends on the maximum principal stress $\sigma_1$, the damage law is given by:

$$ \frac{d\omega}{dt} = \frac{C \sigma_1^k}{(1 - \omega)^r} $$

(3.59)

If damage depends instead upon the equivalent stress, the damage accumulation rate equation takes the form:

$$ \frac{d\omega}{dt} = \frac{C\sigma_e^k}{(1 - \omega)^r} \quad \text{or} \quad \frac{d\omega}{dt} = C\sigma_e^k \omega $$

(3.60)

When damage accumulation exhibits a mixed stress dependence, the damage rate law incorporates a representative rupture stress $\sigma_R$ as follows:

$$ \frac{d\omega}{dt} = \frac{C \sigma_R^k}{(1 - \omega)^r} $$

(3.61)

This rupture stress is defined as:

$$ \sigma_R = \alpha \sigma_1 + \beta \sigma_e + \gamma \sigma_h, \quad \text{where} \quad \alpha + \beta + \gamma = 1 $$

(3.62a)

In most cases, $\sigma_h$, the hydrostatic stress, is neglected, and the representative stress is simply:

$$ \sigma_R = \alpha \sigma_1 + (1 - \alpha) \sigma_e $$

(3.62b)

The representative stress can also be defined as:

$$ \sigma_R = (\sigma_1 \sigma_e)^{\frac{k_x}{k}} $$

(3.62c)

A final version of the damage law incorporates the damage dependence into the representative rupture stress:

$$ \frac{d\omega}{dt} = C\sigma_R^k \quad \text{where} \quad \sigma_R = \frac{\alpha \sigma_1}{(1 - \omega)} + (1 - \alpha) \sigma_e $$

(3.63)

Some early attempts to fit the damage accumulation law, Equation 3.44a or 3.45a, to observed physical behavior involved treating cavities as small penny-shaped cracks on grain boundaries [eg., 63]. For grain boundary cavities with radius $r$ and spacing $\lambda$, damage may be defined as the area fraction of grain boundary which has cavitated, or:

$$ \omega = \left( \frac{r}{\lambda} \right)^2 $$

(3.64)
Differentiating this equation with respect to time yields:

\[ \frac{d\omega}{dt} = \frac{2}{\lambda} \left( \frac{r}{\lambda} \right) \frac{dr}{dt} \quad \text{or} \quad \frac{\kappa}{2} \sqrt{\omega} \frac{d\omega}{dt} = \frac{dr}{dt} \quad (3.65) \]

where \( \frac{dr}{dt} \) is the creep crack growth rate, measured on fracture mechanics specimens. This approach is flawed because the stress field around a large macroscopic crack is very different from that near a small crack-like void. This dissimilarity in stress field results in different growth rates for the respective flaws. In addition, different mechanisms may be controlling growth rates of vastly different sized flaws, and the above treatment does not account for micromechanisms of creep cracking.


![Figure 3.8. Formation of Wedge-Shaped Voids due to Grain Boundary Sliding and Round Cavities due to Stress Perpendicular to the Boundary.](image)

Cavity Nucleation

The existence of cavities in creeping materials has long been known. Figure 3.8 illustrates two early classifications of cavities. One type of cavity forms at triple point junctions, where three grain boundaries meet. Grain boundary sliding in opposite directions results in a large tensile stress across the boundary oriented normal to the applied tensile stress. The two grains are forced apart, creating a W-type void (W for wedge). W-type cavities are associated with smooth boundaries, which promote grain boundary sliding. R-type (R for round) cavities also form on boundaries oriented perpendicular to the tensile axis; these cavities are often associated with grain boundary particles and with serrated boundaries, both of which retard grain boundary sliding. W-type voids form at low temperatures under large applied stresses, while R-type voids are found at higher
temperatures and lower stresses; typical turbine operating conditions promote the formation of round cavities in superalloys. Closer observations of W-type cavities have shown them to result from the preferential coalescence of R-type cavities in the stress concentration of the grain boundary triple point. Microcracks formed in this manner propagate by the formation of small grain boundary cavities in the stress field ahead of the crack tip, and by the linking of these cavities with the main crack [26, 69].

It can be shown that both grain boundary particles and sliding are necessary to nucleate cavities on the boundaries [69, 72]. Grain boundary sliding results in stress concentrations at triple points and at grain boundary particles which impede sliding. A detailed comparison of triple points and intergranular particles in the presence of grain boundary sliding [72] shows that stress concentrations at triple points are insufficient to cause cavity nucleation, while the much larger stresses at grain boundary particles are high enough to form cavities at the particle-matrix interfaces. These stresses may also be sufficient to fracture particles, nucleating cracks or cavities from the fractures. Figure 3.9 shows how the combination of grain boundary sliding with an applied compressive stress can lead to tensile stresses across a grain boundary aligned parallel to the compressive axis [69].

![Figure 3.9](image)

Cavity nucleation is a process which occurs continuously during exposure to the appropriate conditions. The number of cavities has been found to vary directly with strain [26, 69, 72, 93, 112]; the rate of cavity nucleation can be related directly to strain rate [69]. In the presence of a multiaxial state of stress, the stress component controlling cavity nucleation on a given grain boundary is the tensile stress perpendicular to that boundary [69].
Cavity Growth

Several models have been developed to predict the growth rate of cavities in a material based on various rate limiting processes. Growth of small cavities at low stresses is controlled by diffusion of vacancies into the cavities and of atoms from cavities to sinks, such as adjacent grain boundaries. Figure 3.10 illustrates diffusion-controlled cavity shapes. For small cavity sizes, surface diffusion is fast enough to maintain the equilibrium shape of the cavity: two spherical sections (or spherical "caps") with constant radii of curvature. Because surface diffusion occurs rapidly, cavity growth is limited by the rate of diffusion through the grain boundary. As cavities grow, it becomes more difficult for surface diffusion to maintain their shape; the effective cavity spacing decreases because the cavities are larger, and because more cavities have nucleated. Atoms are carried away from the void by grain boundary diffusion faster than surface diffusion can equilibrate the cavity shape. The cavity has become extended or crack-like, and its growth is now limited by surface diffusion. Higher stresses favor crack-like cavity growth over equilibrium cavity growth [69, 72, 78]. For larger cavities exposed to even higher stress levels, growth is controlled by power law creep of the material remaining between cavities [69].

Typically, different mechanisms operate at different stages of cavity growth, with diffusional transport, first by grain boundary diffusion, then by surface diffusion, limiting early cavity growth rates, and power law creep controlling cavity growth later in life. Two or more mechanisms may operate together, enhancing growth rates; alternatively, two mechanisms may impede each other, constraining cavity growth [69].

(a)

(b)

Figure 3.10. Shapes of Grain Boundary Cavities Corresponding to (a) Grain Boundary Diffusion Controlled Growth, and (b) Surface Diffusion Controlled Growth.

Grain Boundary Diffusion Limited Cavity Growth

The growth rate of an intergranular cavity can be calculated based on an atomic flux balance on the surface of the cavity and along the grain boundary away from the cavity, and related to the strain rate. Strain accumulates by plating out of atoms onto the grain boundary, forcing apart the two grains. If growth of equilibrium shaped cavities of radius \( r \) and spacing
\( \lambda \) is limited by grain boundary diffusion, the rate of growth has been formulated as [69, 113, 114]:

\[
\frac{dr}{dt} = \frac{D_{GB} \delta_{GB} \Omega}{kT h(\psi) r^2} \frac{[\sigma - (1 - \omega)2(\Gamma_S/r) \sin \psi]}{[\ln(1/\omega) - (3-\omega)(1-\omega)/2]} \tag{3.66}
\]

where \( D_{GB}, \delta_{GB}, \Omega, k, \) and \( T \) are defined in previous sections, \( \Gamma_S \) is the energy of a free surface, \( \omega \) is the area fraction of cavitated boundary, defined in Equation 3.64, \( \psi \) is half of the dihedral angle at the cavity-boundary vertex, and \( h(\psi) \) is the following geometrical function:

\[
h(\psi) = \frac{1 + \cos \psi}{2} \frac{-\cos \psi}{\sin \psi} \tag{3.67}
\]

The value of \( h(\psi) \) is typically about 0.65. Equation 3.66 has been derived in similar forms elsewhere in the literature [72, 78, 93, 112, 115].

The cavity growth rate can be combined with Equation 3.65 to obtain damage and strain accumulation rates of the form [78, 115]:

\[
\frac{d \omega}{dt} = \left( \frac{d \epsilon}{dt} \right)_{\text{min}} \left[ \frac{d}{2 \lambda \sqrt{\omega}} \frac{\ln(1/\omega)}{\ln(1/\omega)} \right] \tag{3.68a}
\]

\[
\frac{d \epsilon}{dt} = \left( \frac{d \epsilon}{dt} \right)_{\text{min}} \left[ \frac{\ln(1/\omega)}{\ln(1/\omega)} \right] \tag{3.68b}
\]

where \( (d\epsilon/dt)_{\text{min}} \) is the minimum creep rate and \( \omega_i \) is the initial cavitated area. For values of \( \omega \) near unity, \( i.e., \) near full cavitation, Equations 3.68a and 3.68b reduce to the CDM form. Diffusional cavity growth is relevant at much smaller \( \omega \), however, where CDM predictions are not valid. For instance, CDM predicts an increasing \( d\omega/dt \) for small \( \omega \), while cavity growth models and experimental observations indicate that \( d\omega/dt \) is actually decreasing [78, 115].

**Surface Diffusion Limited Cavity Growth**

Growth of crack-like cavities can be modelled as above. The growth rate of cavities with radius \( r \) and spacing \( \lambda \) is given by [69]:

\[
\frac{dr}{dt} = \frac{\Omega \delta_S D_S \Gamma_S}{8kT} \left[ \frac{\lambda^2 \sigma}{(\lambda^2 - r^2) \Gamma_S \sin(\psi/2)} \right]^3 \tag{3.69}
\]
\( \Gamma_{GB} \) is the grain boundary free energy; other symbols have been defined. Similar forms have been presented elsewhere [72, 78, 112, 114]. The formulas for damage and strain accumulation are [78]:

\[
\frac{d\omega}{dt} = \left( \frac{de}{dt} \right)_{\text{min}} \frac{d\sigma}{4\Gamma_s} \left( \frac{\omega}{\omega_i} \right)^{1/2} \left( \frac{1 - \omega_i}{1 - \omega} \right)^3
\]

(3.70a)

\[
\frac{de}{dt} = \left( \frac{de}{dt} \right)_{\text{min}} \left( \frac{\omega}{\omega_i} \right)^{1/2} \left( \frac{1 - \omega_i}{1 - \omega} \right)^3
\]

(3.70b)

These simplify to the CDM forms for \( \omega \) close to unity (full cavitation), but diffusive cavity growth is important for smaller cavities and growth rates, as described above for grain boundary diffusion growth control [78].

**Power Law Creep Controlled Cavity Growth**

The rates of damage and creep strain accumulation have been formulated for cavity growth controlled by power law creep of the uncavitated boundary as [78, 115]:

\[
\frac{d\omega}{dt} = \left( \frac{de}{dt} \right)_{\text{min}} \left[ \frac{1}{(1-\omega)^n} - (1-\omega) \right]
\]

(3.71a)

\[
\frac{de}{dt} = \left( \frac{de}{dt} \right)_{\text{min}} \left[ 1 + \frac{2r_i}{d} \left( \frac{1}{(1-\omega)^n} - 1 \right) \right]
\]

(3.71b)

For large values of \( \omega \), these equations are equivalent to Equations 3.44a and 3.44b, while at smaller values of \( \omega \), Equation 3.71a predicts slower damage accumulation than does Equation 3.44a. For \( \omega=0 \), CDM predicts a finite rate of damage accumulation, whereas the above model predicts no damage accumulation. This makes sense from a physical standpoint: nonexistent cavities do not grow [78, 115].

**Coupled Cavity Growth Mechanisms**

Figure 3.11 shows an isolated cavity growing by diffusional creep; the surrounding grains are undergoing power law creep. Atoms flow out of the cavity, and plate onto the boundary close to the cavity. As matter builds up, the stresses near the cavity are relaxed, and the driving force for diffusional creep is decreased. Because the grain boundary is being wedged open by the material from the cavity, the stresses further from the cavity are increased, and power law creep of the grains proceeds at a greater rate. Because of this accelerated creep, stresses again build up close to the cavity. This is an example of two
mechanisms operating together to produce cavity growth rates greater than those of either mechanism alone [69, 78].

A different situation is depicted in Figure 3.12. A grain boundary containing a distribution of cavities is shown; few other boundaries are cavitating. The stresses in the grains on either side of the cavitating boundary have been relaxed, so the cavity growth rate decreases. The surrounding grains creep faster because the load is redistributed onto the noncavitating regions, but cavities on the boundary only grow fast enough to accommodate power law creep in the surrounding material. This cavitation rate is slower than when all boundaries are cavitated. Such behavior occurs for small, widely spaced cavities under conditions of slow diffusion and slow creep rates in the bulk of the material; the cavity growth rate is given by [69, 72]:

\[
\frac{dr}{dt} = \frac{\alpha d}{4 \omega h(\psi)} \frac{de}{dt}
\]  

(3.72)

\(\alpha\) is a constant, equal to about 0.9. The time to coalesce grain boundary cavities into a single facet crack is obtained by integrating Equation 3.72 [72]:

\[
t_{\text{facet}} = \left( \frac{4 h(\psi)}{3 \alpha} \right) \frac{\lambda}{d} \left[ 1 - \omega_1 \right] \frac{1}{de/dt}
\]

(3.73)
Similar relationships have been derived which incorporate the inverse dependence upon strain rate but not on grain size [20]. If the time to rupture is taken as $t_{\text{facet}}$ (a reasonable assumption [20, 69, 72]), then Equation 3.73 satisfies the Monkman-Grant relationship.

Figure 3.12. Coupling of Diffusional Cavity Growth and Power Law Creep Resulting in Constrained Cavity Growth.
4. René 95 Research Program.

4.1. René 95.

René 95 is used widely to make compressor and turbine disks and other components in advanced military and commercial jet aircraft engines. One of the strongest nickel base superalloys, René 95 has relatively low ductility. This alloy has very high yield and ultimate tensile strengths and good resistance to creep strain accumulation. René 95 is notch sensitive and is characterized by low tensile and creep rupture elongations. The alloy also exhibits very fast creep crack growth rates in the presence of an air environment. Rupture life is effectively limited by the time required to initiate a small crack [3, 25, 52, 60, 116-120].

René 95 Chemistry

Table 4.1 gives typical compositions for conventional ingot metallurgy and powder metallurgy René 95. The high content of refractory elements (cobalt, molybdenum, and tungsten) leads to substantial solid solution hardening. There are also large amounts of γ’-forming elements (aluminum, titanium, and columbium): the γ’ content of René 95 approaches 50% by volume. Grain boundaries are strengthened through additions of carbon, zirconium, and boron. Such extensive alloying leads to segregation, inhomogeneities in structure and properties, and localized incipient melting [25, 61, 118].

<table>
<thead>
<tr>
<th>Element</th>
<th>Minimum</th>
<th>Maximum</th>
<th>Conventional</th>
<th>Powder Met.</th>
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</thead>
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<td>0.15</td>
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</tr>
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<td>3.7</td>
<td>3.5</td>
<td>3.6</td>
</tr>
<tr>
<td>Al</td>
<td>3.3</td>
<td>3.7</td>
<td>3.5</td>
<td>3.6</td>
</tr>
<tr>
<td>Ti</td>
<td>2.3</td>
<td>2.7</td>
<td>2.5</td>
<td>2.6</td>
</tr>
<tr>
<td>Cb</td>
<td>3.3</td>
<td>3.7</td>
<td>3.5</td>
<td>3.6</td>
</tr>
<tr>
<td>Zr</td>
<td>0.03</td>
<td>0.07</td>
<td>0.005</td>
<td>0.053</td>
</tr>
<tr>
<td>B</td>
<td>0.006</td>
<td>0.015</td>
<td>0.010</td>
<td>0.010</td>
</tr>
</tbody>
</table>

Compositions stated in weight per cent.
Processing of René 95

Originally designed as a high strength cast-and-wrought alloy for turbine disk applications, René 95 is no longer produced practically by conventional I/M processes [18]. The high alloying content results in severe macrosegregation and poor forgeability in large cast ingots. To alleviate these problems and to improve material utilization, René 95 components are produced through P/M techniques [3, 18, 52, 116, 121]. Compared to its wrought condition, P/M René 95 exhibits somewhat higher ultimate tensile strength and lower yield strength, greater tensile and creep rupture ductilities, and much greater rupture life [3].

René 95 is generally consolidated from powder via HIP or extrusion below the γ' solvus temperature, which is about 1150°C (2100°F). Consolidation is carried out below this temperature to retain primary γ' precipitates, which prevent excessive grain growth by pinning grain boundaries. The fine-grained structure results in good tensile properties and enables the preform to be hot worked without cracking. The first lots of HIP René 95 were susceptible to embrittlement at prior particle boundaries and to fracture at ceramic inclusions (from atomization nozzles or slag). The carbon and chromium contents of P/M René 95 have been decreased to suppress precipitation of carbides on the PPBs during the HIP process. Consolidation may also be accomplished by extrusion, or consolidation may be followed by a forging operation, such as isothermal forging ("isoforging"). This will break up any inclusions, and disperse the broken inclusions and PPBs throughout the microstructure of the alloy. If a necklace grain structure is desired, the billet is cold worked or hot forged, because neither isothermal forging nor HIP provide the necessary large-grained warm-worked microstructure.

Standard heat treatment includes a recrystallization anneal below the γ' solvus. The recrystallized HIP microstructure is very uniform, comprised of fine grains (ASTM 9-11; 8-11 μm); recrystallization of forged P/M material produces a somewhat finer grain size. There is an even distribution of micron-sized titanium and columbium-rich MC carbides for HIP and isoforged René 95; M₆C and M₂₃C₆ carbides are rare. The HIP alloy contains large (1-2 μm) blocky primary γ' along the recrystallized grain boundaries, intermediate (0.5-1 μm) primary γ', and fine (0.1 μm) spherical intracrystalline γ'. Isoforged René 95 typically contains 0.2-0.3 μm primary cuboidal γ', and fine (0.1 μm) spheroidal transgranular γ' [3, 25, 58, 61, 117, 121, 122].

4.2. Potential Drop Data Acquisition.

The potential drop (PD) technique has been used in many different studies to monitor crack length during creep crack growth or fatigue crack growth testing. The
A study has been made of the use of the potential drop technique to monitor general mechanical behavior [123]. Changes in geometry due to elastic, plastic, or creep deformation can be followed through changes in resistance of a specimen. In addition, potential drop can indicate the development of cavitation, cracking, oxidation, precipitation, or other phenomena which affect the effective resistivity of the material.

Vasatis [109, 124-126] used the PD technique to monitor the notch rupture behavior of Inconel X-750. He showed that potential drop can be used to measure deformation in notched and unnotched specimens of this ductile alloy. He also measured the time to initiate creep cracks in sharply notched specimens. When the cracks grow radially inward from the entire circumference of the notch root toward the center of the specimens, the potential drop can be used to calculate crack lengths and creep crack growth rates.

4.3. Research Objectives.

There were two main goals in this research project. The first goal was to study the creep rupture behavior of René 95. The second goal was to evaluate the usefulness of the potential drop measurement technique to the study of such a brittle alloy as René 95.

Turbine disks are subjected to fatigue loading, and failure mechanisms often include low cycle fatigue or fatigue crack propagation. The loading history of a disk can be simplified into an up ramp, an extended hold at a constant stress, and a down ramp, simulating engine start-up, level flight, and engine shut-down. The behavior of material in the vicinity of a stress concentration in a disk (e.g., a bolt hole or a fir tree blade attachment) can be investigated by testing notched specimens at a constant load. For notch sensitive alloys, the useful life of a disk can be modeled by the initiation life of a notched specimen.

The first goal of the present study was to study the effects of processing history and notch acuity on the creep rupture behavior of René 95. One lot of René 95 had been consolidated by hot isostatic pressing (HIP), the other by extrusion plus isothermal forging (EIF). The goal was to characterize the mechanisms of accumulation of creep deformation and damage in René 95.

The second goal of this research project was to assess the ability of the potential drop method to monitor the creep rupture behavior of an alloy such as René 95, which does not accumulate large amounts of deformation or intergranular cavitation. It was desired to evaluate the sensitivity of potential drop to such small amounts of deformation, cavitation, and microcracking, and to judge the ability of the technique to measure initiation of small cracks and to follow rapid crack propagation rates.
5. Experimental Materials and Procedures.

5.1. René 95.

The specimens tested in this study were machined from two René 95 turbine disks, processed by different powder metallurgy (P/M) techniques from an initial powder size of -150 mesh (106 μm). The initial powders had nominally the same chemical compositions. One disk had been consolidated by hot isostatically pressing (HIP); the other disk had been thermomechanically processed by extrusion and isothermal forging (EIF). The disks had been heat treated according to the schedules given in Table 5.1. The resulting grain size of the HIP material is ASTM 9-11 (8-11 μm), while that of the EIF material is ASTM 12-13 (4-6 μm) [127].

<table>
<thead>
<tr>
<th>Consolidation Method:</th>
<th>HIP</th>
<th>EIF</th>
</tr>
</thead>
<tbody>
<tr>
<td>Solution Treatment (below γ’ solvus)</td>
<td>1129°C (2065°F)/1 hr</td>
<td>1107°C (2025°F)/1 hr</td>
</tr>
<tr>
<td></td>
<td>538°C (1000°F) Salt Quench</td>
<td>Room Temperature Oil Quench</td>
</tr>
<tr>
<td>Aging Treatment</td>
<td>760°C (1400°F)/8 hr/AC</td>
<td>760°C (1400°F)/8 hr/AC</td>
</tr>
<tr>
<td>Grain Size</td>
<td>ASTM 9-11 (8-16 μm)</td>
<td>ASTM 12-13 (4-6 μm)</td>
</tr>
</tbody>
</table>

Microstructures of the experimental alloy are depicted in Figures 5.1 through 5.4. Optical micrographs of HIP René 95 are shown in Figure 5.1 at magnifications of 160× and 640×. Outlines of the larger powder particles are visible; the large particle in the lower middle of Figure 5.1a is about 95 μm in diameter. Primary interdendritic γ’ is evident within the powder particles, and some of the powder particles are outlined by γ’ precipitates. Figures 5.2a and 5.2b are optical photographs of HIP and EIF René 95, taken at 320×. In contrast to the interdendritic primary γ’ and powder particle morphology of the HIP material, the EIF alloy shows a much more uniform distribution of γ’ with no other microstructural features. SEM micrographs at 1000× and 5000× are presented for the HIP and EIF treatments in Figures 5.3 and 5.4 respectively.

Both powder consolidation treatments have produced a duplex distribution of coarse primary and fine intracrystalline γ’. The primary γ’ in the HIP material is divided between large (1-4 μm) irregular globules which seem to have precipitated randomly throughout the alloy and intermediate (0.2-1 μm) globular particles which have precipitated interdendritically. The EIF René 95 contains large (2-5 μm) angular or blocky particles and medium-sized (0.5-1 μm) rounded or globular particles. The primary γ’ is slightly larger in
the EIF alloy than in the HIP alloy. Both aging treatments resulted in a dispersion of fine (0.05-0.1 \(\mu\)m) \(\gamma'\) spheroids, which precipitated transgranularly throughout the alloy. The HIP alloy is solution treated at a somewhat higher temperature than the EIF alloy; therefore, the EIF material retains a larger volume fraction of the coarse and intermediate primary \(\gamma'\), while the HIP material contains a denser distribution of fine aging \(\gamma'\).

5.2. Specimen Geometry.

The three axially symmetric specimen geometries used in this study are shown in Figure 5.5. In addition to a straight cylindrical (smooth bar) specimen with no elastic stress concentration \((k_t=1.0)\), two notch geometries were used. The blunt notch (or U-notch) is essentially a semicircular notch (i.e., notch depth equal to notch root radius) with tapered shoulders and a moderate elastic stress concentration \((k_t=1.6)\). The sharp notch (V-notch) specimen has the British Standard Notch geometry and a severe elastic stress concentration \((k_t=4.2)\) [128-130]. Each notched specimen has two identical notches, so that when one notch breaks, the remaining notch has reached only a fraction of its lifetime. The unfailed notch may be examined to observe the state of creep damage present before failure. Analysis of potential drop feedback can reveal this fraction of “expended” life; alternatively, extrapolation of potential drop data will result in an estimation for the lifetime of the unfailed notch.

5.3. Experimental Conditions.

All tests were carried out at 704°C (1300°F) in laboratory air. Net section stresses in smooth bar tests ranged from 100 to 127 ksi, with rupture lives between 6 and 140 hr. U-notch specimens were tested in the range of 100-160 ksi (690-1100 MPa), resulting in rupture lives of 12-300 hr. V-notch specimens were subjected to loads of 85-130 ksi (580-900 MPa) with rupture times up to 70 hr; several V-notch specimens were loaded to 50-80 ksi (345-550 MPa) with no failure or crack initiation in 250 hr, at which point each test was suspended.

5.4. Creep Testing Apparatus.

Tests were performed on a 20,000 lb capacity ATS Model 2410 Lever Arm Tester with a 20:1 lever ratio. The tester uses double knife edge couplings to minimize bending moments on the specimen, and has manual weight elevator and drawhead. An ATS Model 2961 Split Round Test Furnace, with three 1000 W heating zones and a maximum temperature of 1200°C (2200°F) is controlled to within ±2°C of the desired temperature using an ATS Model 2935 three-zone single controller Furnace Temperature Control System.
Specimen grips are machined of Mar M-246, and pull rods are Inconel 713L. Elongation of smooth bar specimens is monitored using an ATS Model 4115 Axial Retort Extensometer, which attaches directly onto the specimens; a Pickering Series 7300 Linear Variable Displacement Transducer (LVDT) in an ATS housing is attached to the extensometer outside the furnace to measure the specimen elongation. Two Omega type K chromel-alumel thermocouples, with ice point junction compensators, are used to monitor the temperature at either end of the specimen.

5.5. Automated Data Collection System.

Creep tests are monitored by the direct current (DC) potential drop (PD) technique. A constant current is passed along the specimen, while the potential across a region of interest (e.g., a notch or a crack) is monitored. The potential drop signal is a function of the resistivity of the material, the shape of the region being monitored, and the presence of discontinuities (such as cracks or cavities) in the region. Resistivity is affected by fluctuations in temperature and by microstructural stability of the alloy; precipitation as well as changes in dislocation structure can change the measured potential in a specimen. A reference probe can be used to follow potential differences due to temperature fluctuations and to stress-independent phase transformations. The corrected potential then gives a measure of creep strain and damage accumulation in the specimen.

A schematic of the direct current potential drop testing system, described by Vasatis [109, 124-126], is presented in Figure 5.6. Current input wires of commercially pure nickel are spot welded to opposite shoulders of the specimen. A DC power supply (Hewlett-Packard HP 6259B or Sorensen SRL 10-25) applies a constant current (10 A for notched specimens, 5 to 6 A for smooth bar specimens) across the specimen through these wires.

Potential drop probes are made by spot-welding pairs of chromel thermocouple wires to the specimen at appropriate locations. One pair of wires is attached across each notch; the wires are welded as close as possible to the edges of the notch (where it meets the straight sections), and diametrically opposite one another, in order to average out any asymmetries in the distribution of damage or cracking. In some tests, two PD probes were welded at each notch. Another pair of wires is attached across a straight (i.e., non-straining) section of the specimen to serve as a reference probe, which is used to account for any fluctuations in applied current or resistivity (i.e., temperature). The 10 A applied current produces typical initial PD measurements of 2.5-3.0 mV for a U-Notch, 1.0-1.3 mV for a V-Notch, and 0.7-1.0 mV for a reference probe.

The potential drop probes and thermocouples, and the LVDT in smooth bar tests, are connected via a shielded telephone line to a Hewlett-Packard HP 3488A Switching and
Control Unit (or Multiprogrammer), which is in turn wired into an HP 3478A Digital Voltmeter (DVM). These two input/output (I/O) devices are connected to, and controlled by, an HP85A Personal Desktop Computer. Finally, an HP 9121 Dual Disk Drive completes the system; data is stored on 3.5 inch microfloppy disks. It should be noted here that the computerized system does not control the furnace, but only monitors the temperature of the specimen.

The DVM measures voltage with a resolution of 0.1 μV. This resolution corresponds to a longitudinal creep strain of approximately 0.03% in a smooth bar or U-notch specimen. In a V-notch specimen, this resolution indicates a 1.4 μm deep axisymmetric crack initiating uniformly around the circumference of the specimen at the notch root [131]. A 1 μV potential increase represents a creep strain of about 0.3% or an axisymmetric crack 6.8 μm deep.

5.6 Creep Test Procedures.

After the specimen is threaded into the grips and the electrical connections have been made, the specimen is connected to the load train. A current is applied and each PD probe is checked. The thermocouples are placed alongside the specimen, the furnace is closed, and its ends are packed with insulation. The furnace is turned on, and the controller brings the temperature up to the desired level, while the computer measures and displays this temperature. The computer alerts the experimenter when the temperature nears the set value, so that precise “fine tuning” is possible.

When the temperature has become stabilized, the computer prepares for the test to begin. It gives the operator a series of cues: a warning that the ice junction compensators on the thermocouples must be activated, a reminder to insert two empty disks into the disk drive, and various questions concerning the particular test being run. When the test is ready to start, the computer directs the operator to apply the load and level the crosshead, and the computer begins taking data.

The data collection routine proceeds as follows. The HP85A orders the DVM to read in turn each active channel on the switching unit. The HP85A records these values (reference and notch potentials and thermocouple readings) and the elapsed time since the start of the test onto a disk in the dual disk drive. When one disk becomes full, the HP85A switches to the other drive and prints a message on its internal printer, telling the operator that the disk must be replaced with an empty one. Potential drop signals are plotted vs. elapsed time on the CRT screen of the HP85A, allowing the operator to monitor the progress of the experiment. A new plot is drawn every 24 hr, and the previous ones are stored on the active disk.
After storing and plotting the data, the computer checks for errors in the software and in the I/O devices. If an error is present, an appropriate message is output on the printer of the HP85A; the test is not terminated, so that data collection may proceed. The computer also checks whether the specimen has failed, and prints an appropriate message when failure occurs. Again, data collection does not terminate, in case a broken PD lead wire has caused a spurious detection of failure. After a 12-second wait, the computer begins a new data collection cycle; during the first several data cycles on each disk, the computer omits this hold time.

Measurements in each data cycle are made within approximately 3 sec, and data cycles are spaced apart by about 30 sec. Each 500 kb microdisk contains room for over 4000 data cycles, or more than 30 hours of data. A more detailed description of the creep rupture testing program, complete with program listing and flow chart, can be found in Appendix A.
Figure 5.1. Optical Micrographs of HIP René 95: (a) 160x, (b) 640x.
Figure 5.2. Optical Micrographs of René 95 at 320x: (a) HIP, (b) EIF.
Figure 5.3. SEM Micrographs of HIP René 95: (a) 1000x, (b) 5000x.
Figure 5.4. SEM Micrographs of EIF René 95: (a) 1000x, (b) 5000x.
Figure 5.5. Axisymmetric Creep Rupture Specimens.
Figure 5.6. Direct Current Potential Drop (DCPD) Automated Data Collection System.
6. Experimental Results.


Creep rupture test results of the three specimen geometries (U-notch, V-notch, and smooth bar) and both processing treatments (hot isostatically pressed or HIP, and extruded and isothermally forged or EIF) of René 95 are summarized in Figure 6.1. Rupture lifetimes for all geometries, and crack initiation times for V-notch specimens, are plotted as a function of applied net section stress; extrapolated lifetimes for unfailed notches of the U-notch specimens are included as an indication of the scatter in creep data. Figure 6.2 magnifies the U-notch and smooth bar data from Figure 6.1 in greater detail. Creep rupture test results are presented in Table 6.1 for U-notch specimens, in Table 6.2 for V-notch specimens, and in Table 6.3 for smooth bar specimens. Extrapolated lifetimes for unfailed U-notches and crack initiation times for V-notches are defined in Section 6.2.

Within a given geometry, HIP and EIF specimens exhibit the same dependence of rupture life on applied stress. U-notch test results fit a power law stress dependence of the form:

\[ t_R = A \sigma^m \]  

(6.1)

Least squares estimates for A and m are given according to the analysis in Appendix D by:

- HIP: \[ t_R = 5.26 \times 10^{15} \sigma^{-6.61} \]  
  (6.2a)
- EIF: \[ t_R = 4.79 \times 10^{13} \sigma^{-5.60} \]  
  (6.2b)
- All: \[ t_R = 2.73 \times 10^{14} \sigma^{-5.98} \]  
  (6.2c)

According to the statistical treatment in Appendix D, there is no significant difference between the individual HIP and EIF stress dependencies. Smooth bar rupture lives do not fit a power law function of stress, but the HIP and EIF specimens exhibit the same rupture life dependence on applied stress. V-notch specimens show an apparent threshold stress of about 580 MPa (85 ksi). Lifetimes for HIP and EIF specimens approach this threshold asymptotically with decreasing stress.

6.2. Potential Drop Behavior.

The extrapolated lifetime for an unfailed U-notch is defined as follows. Potential drop is a measure of strain (reduction of area and elongation) in the region of interest, as well as the amount of damage, such as cavitation or cracking, which reduces the current carrying cross-sectional area of the specimen. Figure 6.3a shows typical DCPD creep curves from the two notches of a single U-notch specimen. Specimen failure occurred at \( t_R(f) \); the failed
notch potential has reached point A. The unfailed notch has only accumulated the potential corresponding to point B; the failed notch had reached the same value of potential (and therefore the same degree of creep strain and damage) at point C. The spent fraction of life for the unfailed notch at $t_{R(f)}$ is therefore the same as the fraction of life that the failed notch had accumulated at point C:

$$\text{Expended Fraction of Life: } \frac{t_c}{t_{R(f)}} = \frac{t_B}{t_{R(u)}} = \frac{t_{R(f)}}{t_{R(u)}}$$ (6.3)

where $t_{R(u)}$ is the extrapolated time to rupture for the unfailed notch. Finally:

$$t_{R(u)} = \frac{t_{R(f)}^2}{t_c}$$ (6.4)

Figure 6.3b illustrates the failed and the extrapolated unfailed notch potentials.

The above extrapolation procedure accounts for the stochastic nature of creep. One notch begins accumulating creep strain and damage slightly faster than the other, due to randomly occurring differences in chemistry, microstructure, temperature, geometry, and other variables. At any given time, this notch contains a greater amount of strain and damage, and it has a greater rate of strain and damage accumulation, than the other notch. At a given value of potential, the failed notch shows a greater potential increase rate than the unfailed notch, illustrating further the conditions which caused it to begin creeping faster. Normalized by the appropriate rupture lives, the potential increase rates become equal, so that the two potential drop curves become virtually coincident when plotted against time/time to rupture. The use of specimens with multiple notches allows examination of creep damage in the unfailed notch before fracture. Using the DCPD extrapolation technique, one can determine the relative fraction of life experienced by the unfailed notch. The PD technique allows the collection of multiple data points from each specimen, effectively running more tests in a shorter time.

Potential drop creep curves for U-notch René 95 specimens are presented in Figures 6.4 through 6.6. Potential is given as $\log(V/V_o)$, where $V$ is the potential, corrected according to procedures outlined in Appendix B, and $V_o$ is the initial potential. The quantity $\log(V/V_o)$ is related to creep strain as described in detail in Appendix D. PD data is plotted vs. time in Figure 6.4 and vs. $t/t_{R}$ in Figure 6.5; part (a) of each figure shows the HIP test results, and part (b) shows the EIF results. Figure 6.6 plots the EIF and HIP creep curves together vs. time/time to rupture.

The PD curves for HIP U-notch specimens are essentially straight lines which follow steady state creep, with only a small acceleration of potential near the end of some
tests. The absence of pronounced tertiary creep indicates very small amounts of strain and cavitation. EIF specimens show a steady state region with a slightly greater slope (i.e., minimum creep rate) than that of the HIP tests. After about half to two-thirds of lifetime in an EIF test, the potential accelerates, indicating increased rates of strain and damage accumulation.

The V-notch crack initiation time is defined according to Figure 6.7. Potential drop signals from the two notches are essentially constant for most of the test. At some point, one notch potential starts to increase, indicating crack formation; this notch eventually fails. The point at which one potential deviates from the other is the time for crack initiation; plotting the difference between the two notch potentials can clarify crack initiation. There is no formulaic definition, such as a prespecified potential difference or slope; initiation is based simply on an arbitrary judgment by the experimenter.

Potential drop curves for V-notch tests are given in Figures 6.8, 6.9, and 6.10. Figure 6.8 shows the increase in potential at both notches during each test, HIP specimens in part (a) and EIF in part (b). Figure 6.9 shows the difference between notch potentials for HIP and EIF specimens in parts (a) and (b), respectively, and Figure 6.10 is a plot of these HIP and EIF potential difference curves together. PD curves from test 24 (low stress HIP test) were omitted because of a large degree of scatter, which obscures data from the other tests. Applied stresses in V-notch tests were low compared to those in smooth bar and U-notch tests, and the region of high stress concentration at the notch root is very small. The accumulated strain and damage within the stress concentration was severe but much more localized than the DCPD technique can resolve, so there was no measurable steady state strain across the notches. In each test, both notches initially showed constant PD signals, followed by crack initiation and rapid creep crack growth in one notch. In only one case was crack initiation detected in the unfailed notch, and the crack was not visible at the notch root upon examination after the test by binocular microscope at 30×.

As the applied stress is decreased and V-notch rupture life increases, the fraction of life spent in crack initiation also increases up to a limiting value of 1. The ratio \( t/t_R \) is plotted in Figure 6.11 vs. stress for tests of HIP and EIF specimens. This ratio is greater in HIP specimens, indicating a faster creep crack growth rate or a smaller critical flaw size (i.e., toughness) than in EIF specimens.

Smooth bar creep curves are given in Figures 6.12 through 6.14. LVDT and DCPD strains for HIP and EIF tests are plotted vs. time in Figures 6.12a and 6.12b. LVDT and DCPD creep curves are plotted vs. \( t/t_R \) in Figures 6.13a and 6.13b, respectively, for HIP specimens and EIF specimens. Figure 6.14 presents LVDT-\( t/t_R \) curves for HIP and
EIF specimens together. The DCPD and LVDT data have been converted into potential drop and longitudinal strains $\varepsilon_{PD}$ and $\varepsilon_L$, defined as:

$$\varepsilon_{PD} = \frac{1}{2} \log_e \left( \frac{V}{V_0} \right) \quad \text{and} \quad \varepsilon_L = \log_e \left( \frac{L}{L_o} \right)$$

where $L$ is the gauge length at time $t$, and $L_o$ is the initial gauge length. It is noted from Figures 6.12 and 6.13 that the potential drop feedback closely matches the strain measured by extensometer and LVDT.

The dependence of minimum creep rate on stress is presented in Figure 6.15 for U-notch and smooth bar specimens; smooth bar results are expanded in Figure 6.16. At a given stress, a smooth bar accumulates creep strain at a greater rate than a U-notch specimen. U-notch test results fall along a line defined by the power law relationship:

$$\text{mcr} = B \sigma^n$$  \hspace{1cm} (6.5)

where the parameters $B$ and $n$ are given by:

- U/HIP: $\text{mcr} = 9.83 \times 10^{-21} \sigma^{6.07}$  \hspace{1cm} (6.6a)
- U(EIF): $\text{mcr} = 9.50 \times 10^{-21} \sigma^{6.12}$  \hspace{1cm} (6.6b)
- U/All: $\text{mcr} = 6.01 \times 10^{-21} \sigma^{6.19}$  \hspace{1cm} (6.6c)

As with rupture life data, there is no statistically significant difference between the HIP and EIF relationships. U-notch tests are fit to the Monkman-Grant relationship:

$$t_R(\text{mcr})^\alpha = C_M$$  \hspace{1cm} (6.7)

where $\alpha$ and $C_M$ are given by:

- U/HIP: $t_R(\text{mcr})^{1.080} = 9.930 \times 10^{-7}$  \hspace{1cm} (6.8a)
- U(EIF): $t_R(\text{mcr})^{0.887} = 3.506 \times 10^{-5}$  \hspace{1cm} (6.8b)
- U/All: $t_R(\text{mcr})^{0.927} = 1.535 \times 10^{-5}$  \hspace{1cm} (6.8c)

Unlike the U-notch test results, smooth bars show different creep results between the HIP and EIF processing histories. The HIP tests closely follow a power-law dependence of minimum creep rate on applied stress; the EIF tests do not conform to such a neat relationship. Smooth bar creep behavior can be summarized by:

- S/HIP: $\text{mcr} = 2.995 \times 10^{-32} \sigma^{11.99}$  \hspace{1cm} (6.9a)
- S/EIF: $\text{mcr} = 7.047 \times 10^{-49} \sigma^{20.02}$  \hspace{1cm} (6.9b)

for minimum creep rate vs. stress, and:
for the Monkman-Grant relationship. EIF tests show greater strains at failure, both reduction in area and increase in potential drop, than HIP specimens.

Strains at rupture are plotted vs. stress in Figures 6.17 and 6.18. Figure 6.17 plots $\varepsilon_{PD}=\ln(V_f/V_o)$ for U-notch and smooth specimens; Figure 6.18 shows reduction in area $RA=2\ln(D_o/D_f)$ for all three geometries. The strains do not vary systematically with stress. EIF material accumulates larger strains than HIP material in U-notch and smooth bar geometries, while HIP V-notch bars show larger reductions in area than EIF V-notch specimens. Strains are an order of magnitude greater in smooth bars than in V-notch specimens, with U-notch specimens lying between these two extremes.

6.3. Fractography.

Fracture surfaces of U-notch specimens are shown for high and low stress tests of HIP material in Figure 6.19 and of EIF material in Figure 6.20. Fracture initiation occurred by cavitation in the interior of the notch in each test, midway between the center of the notch and the notch root surface. The initiation region is located in the lower half of each fractograph and is distinguished by a bright blue oxide layer, in contrast to the dull purple-gray oxide formed on surfaces which failed by fast fracture. The area of crack initiation (the region of internal cavity growth and coalescence) increased as applied stress decreased. EIF specimens showed larger areas of cavitation than HIP specimens at the same stress. The “ballular” appearance of the PPB fracture path was evident on every HIP fracture surface, while EIF fracture surfaces exhibited a torn surface topography (TST). HIP fractures contained no shear lips at low and intermediate stresses, and only small, partial shear lips at the highest stress; EIF specimens had well-defined shear lips at all stress levels, increasing in thickness with increasing stress.

Typical HIP and EIF V-notch fracture surfaces are shown in Figure 6.21. Crack initiation occurred at the notch root surfaces of all tests (at the bottom of each fractograph). Propagation proceeded asymmetrically across the notch, resulting in crescent-shaped cracks. The crescent width, i.e., crack depth, at failure decreased as applied stress increased. EIF specimens had deeper final cracks than HIP specimens. As in the U-notch tests, HIP fractures follow prior particle boundaries, while EIF fractures have a torn appearance. Shear lips are found in all failed notches, increasing in thickness with applied stress. EIF
specimens have more pronounced shear lips than HIP specimens; in any case, shear lips are very small due to the considerable constraint in the V-notch geometry.

Figures 6.22 and 6.23, respectively, show HIP and EIF smooth bar fracture surfaces, tested at high and low stresses. Cracks initiated at the specimen surfaces. HIP specimens are characterized by the ballular PPB fracture appearance, while EIF specimens exhibit the torn surface as seen in the notched specimens. Large shear lips are evident at all stress levels in the EIF material, shear lip thickness increasing with increasing applied stress. The HIP material showed smaller shear lips at only the highest stress; low stress fracture surfaces are free from these features. Secondary cracks were observed in only the high stress EIF specimen, within the shear lip of the main crack.

The PPB-type fracture surface observed in all HIP fractures is shown at greater magnification both optically and by SEM in Figure 6.24. The round features are attributed to spheres ranging in diameter from 75 to 100 μm; the maximum powder particle size was 106 μm. Smaller features on the fracture surface may be correlated to grain size or to the size of the primary γ' precipitates. Figure 6.25 shows the TST-type intergranular fracture seen in EIF specimens. The small features correspond to the size of grain boundary facets or of primary γ' particles. The growth of a thin oxide layer on the fracture surfaces has obscured any evidence of discrete cavities in all specimens.

In addition to the PPB or TST fracture path, several generalizations can be made regarding the creep rupture tests of René 95. Each smooth bar and sharp notch test failed by initiation and growth of a crack at the surface of the specimen, while blunt notch specimens exhibit internal cavity nucleation, growth, and coalescence below the notch surface, eventually leading to fracture. HIP specimens show smaller fracture zone sizes (cavitation areas or crack depths) than EIF specimens, in agreement with fracture mechanics concepts. Shear lips increase in thickness with increasing applied stress in EIF specimens. Shear lips exist only at the highest stresses in HIP specimens, and are smaller than those found on EIF fracture surfaces.

Extensive metallographic examination was performed on cross sections through failed and unfailed notches in HIP and EIF U-notch specimens tested at high and low stresses. Cavitation accumulated to a much larger degree at low stresses than at high stresses, regardless of processing history. The spacing of damage in the axial direction (parallel to the applied stress axis) corresponds to the powder particle size in HIP material and to the grain size in EIF material. Because the EIF grain size is much smaller than the HIP powder size, the density of cavitation in the axial direction, and the total amount of cavitation, was much smaller in HIP material than in EIF material. This difference in cavity density
would account for the great difference in potential drop behavior between the two treatments. In these U-notch cross sections, the greatest density of cavitation is observed partway between the notch root and the center of the notch. Typical regions of cavitation in failed U-notches are shown in Figure 6.26 in HIP specimens and in Figure 6.27 in EIF specimens. No cavitation was observed in any unfailed notches.


<table>
<thead>
<tr>
<th>Specimen Number</th>
<th>Stress</th>
<th>Time to Rupture, hr</th>
<th>Diametral Strain, %</th>
<th>Minimum Creep Rate</th>
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<td>MPa</td>
<td>Failed</td>
<td>Unfailed</td>
<td>Failed</td>
</tr>
<tr>
<td>Hot Isostatically Pressed (HIP)</td>
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<td></td>
<td></td>
<td></td>
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<td>9</td>
<td>1094</td>
<td>12.730</td>
<td>14.488</td>
<td>0.63</td>
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<td>18</td>
<td>1003</td>
<td>30.952</td>
<td>38.552</td>
<td>1.85</td>
</tr>
<tr>
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<td>895</td>
<td>47.160</td>
<td>60.788</td>
<td>1.07</td>
</tr>
<tr>
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<td>84.497</td>
<td>1.38</td>
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<td>688</td>
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<td>378.516</td>
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<table>
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<th>Specimen Number</th>
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<th>Diametral Strain, %</th>
<th>Minimum Creep Rate</th>
</tr>
</thead>
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<td>MPa</td>
<td>Failed</td>
<td>Unfailed</td>
<td>Failed</td>
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<tr>
<td>Extruded and Isothermally Forged (EIF)</td>
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<tr>
<td>1</td>
<td>1125</td>
<td>15.657</td>
<td>18.281</td>
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<tr>
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<td>1114</td>
<td>*</td>
<td>*</td>
<td>------</td>
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<td>12</td>
<td>1105</td>
<td>27.221</td>
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<td>3.36</td>
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<td>5</td>
<td>895</td>
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<td>895</td>
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<td>93.306</td>
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<td>28</td>
<td>688</td>
<td>293.319</td>
<td>311.323</td>
<td>2.78</td>
</tr>
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</table>

* Specimen 10 failed on smooth section of specimen away from notches.

Minimum creep rate is initial slope of LVDT creep curve (s^-1).

"Failed" and "Unfailed" refer to the two notches of each specimen.

Diametral Strain $\varepsilon_d = \ln \left( \frac{D_0}{D} \right)$
Table 6.2. Creep Rupture Data: V-Notch Tests.

**Hot Isostatically Pressed (HIP)**

<table>
<thead>
<tr>
<th>Specimen Number</th>
<th>Stress MPA</th>
<th>Specimen Stress ksi</th>
<th>Time to Rupture, hr Failed</th>
<th>Time to Initiate, hr Failed</th>
<th>Diametral Strain,% Failed</th>
<th>Diametral Strain,% Unfailed</th>
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<tbody>
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<td>16</td>
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<td>99.9</td>
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<td>1.170</td>
<td>0.96</td>
<td>-0.12</td>
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<td>15</td>
<td>619</td>
<td>89.8</td>
<td>43.414</td>
<td>42.546</td>
<td>0.47</td>
<td>0.04</td>
</tr>
<tr>
<td>24</td>
<td>595</td>
<td>86.3</td>
<td>56.818</td>
<td>56.308</td>
<td>0.63</td>
<td>0.11</td>
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</table>

**Extruded and Isothermally Forged (EIF)**

<table>
<thead>
<tr>
<th>Specimen Number</th>
<th>Stress MPA</th>
<th>Specimen Stress ksi</th>
<th>Time to Rupture, hr Failed</th>
<th>Time to Initiate, hr Failed</th>
<th>Diametral Strain,% Failed</th>
<th>Diametral Strain,% Unfailed</th>
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<td>0.171</td>
<td>0.069</td>
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<td>617</td>
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<td>0.474</td>
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<td>0.05</td>
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<tr>
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<tr>
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</table>

* Specimens 7, 8, and 11 did not fail or initiate a crack within the time stated.

"Failed" and "Unfailed" refer to the two notches of each specimen.

Diametral Strain \( \varepsilon_d = \ln \left( \frac{D_o}{D} \right) \)
Table 6.3. Creep Rupture Data: Smooth Bar Tests.

### Hot Isostatically Pressed (HIP)

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Stress MPa</th>
<th>Stress ksi</th>
<th>Time to Rupture, hr</th>
<th>Diametral Strain, %</th>
<th>Minimum Creep Rate</th>
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<td>5.310x10^-7</td>
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<tr>
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<td>772</td>
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<td>87.556</td>
<td>4.11</td>
<td>1.110x10^-7</td>
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<tr>
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<td>691</td>
<td>100.2</td>
<td>60.777</td>
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### Extruded and Isothermally Forged (EIF)

<table>
<thead>
<tr>
<th>Specimen</th>
<th>Stress MPa</th>
<th>Stress ksi</th>
<th>Time to Rupture, hr</th>
<th>Diametral Strain, %</th>
<th>Minimum Creep Rate</th>
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<tr>
<td>26</td>
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<td>5.65</td>
<td>1.851x10^-6</td>
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<tr>
<td>30</td>
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<td>1.670x10^-8</td>
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<td>19</td>
<td>691</td>
<td>100.2</td>
<td>4.075</td>
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<tr>
<td>21</td>
<td>690</td>
<td>100.0</td>
<td>101.554</td>
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</table>

* Specimen 19 failed prematurely at spot weld within gauge length.  
  Specimen 20 failed prematurely at extensometer knife edge along gauge section.  
  Specimen 21 did not creep due to friction in extensometer (later corrected).

Minimum creep rate is initial slope of LVDT creep curve (s^-1).

Diametral Strain \( \varepsilon_d = \ln \left( \frac{D_0}{D} \right) \)
Figure 6.1. René 95 Creep Rupture Results for All Tests: Rupture Life vs. Applied Net Section Stress.
Figure 6.2. René 95 Creep Rupture Results for U-Notch and Smooth Bar Tests: Rupture Life vs. Applied Net Section Stress.
Figure 6.3. DCPD Extrapolation Procedure for U-Notch Specimens. Refer to Text for Explanation.
Figure 6.4. Potential Drop vs. Time for (a) HIP and (b) EIF U-Notch Tests. Filled Symbols: Failed Notch; Open Symbols: Unfailed Notch.
Figure 6.5. Potential Drop vs. Time/Time to Rupture for (a) HIP and (b) EIF U-Notch Tests. Filled Symbols: Failed Notch; Open Symbols: Unfailed Notch.
Figure 6.6. Potential Drop vs. Time/Time to Rupture for U-Notch Tests. Filled Symbols: HIP; Open Symbols: EIF.
Figure 6.7. Definition of Initiation Time for V-Notch Specimens. See Text for Explanation.
Figure 6.8. Potential Drop Increase vs. Time/Time to Rupture for (a) HIP and (b) EIF V-Notch Tests. Filled Symbols: Failed Notch; Open Symbols: Unfailed Notch.
Figure 6.9. Potential Drop Difference Between Notches vs. Time/Time to Rupture for (a) HIP and (b) EIF V-Notch Tests.
Figure 6.10. Potential Drop Difference Between Notches vs. Time/Time to Rupture for V-Notch Specimens. Filled Symbols: HIP; Open Symbols: EIF.
Figure 6.11. Fraction of Rupture Life Spent in Initiation for V-Notch Specimens as a Function of Net Section Stress. Filled Symbols: HIP; Open Symbols: EIF.
Figure 6.12. Strain vs. Time for (a) HIP and (b) EIF Smooth Bar Tests. Filled Symbols: LVDT Strains; Open Symbols: DCPD Strain.
Figure 6.13. Strain vs. Time/Time to Rupture for (a) HIP and (b) EIF Smooth Bar Tests. Filled Symbols: LVDT Strains; Open Symbols: DCPD Strain.
Figure 6.14. Strain (LVDT) vs. Time/Time to Rupture for Smooth Bar Tests. Filled Symbols: HIP; Open Symbols: EIF.
Figure 6.15. Minimum Creep Rates vs. Applied Net Section Stress for U-Notch and Smooth Bar Specimens.
Figure 6.16. Minimum Creep Rates vs. Applied Stress for Smooth Bar Specimens.
Figure 6.17. Strain at Failure (DCPD) vs. Applied Net Section Stress for U-Notch and Smooth Bar Specimens.
Figure 6.18. Reduction in Area vs. Applied Net Section Stress for U-Notch, V-Notch, and Smooth Bar Specimens.
Figure 6.19. HIP U-Notch Fracture Surfaces at 14×: (a) High Stress (Test 9), (b) Low Stress (Test 27).
Figure 6.20. EIF U-Notch Fracture Surfaces at 14x: (a) High Stress (Test 12), (b) Low Stress (Test 28).
Figure 6.21. Typical V-Notch Fracture Surfaces at 12x: (a) HIP (Test 24), (b) EIF (Test 14).
Figure 6.22. HIP Smooth Bar Fracture Surfaces at 13x: (a) High Stress (Test 25), (b) Low Stress (Test 22).
Figure 6.23. EIF Smooth Bar Fracture Surfaces at 13x: (a) High Stress (Test 26), (b) Low Stress (Test 23).
Figure 6.24. Highly Magnified Fracture Surfaces of HIP René 95:
(a) 14× Optical, (b) 200× SEM, and (c) 800× SEM.
Figure 6.25. Highly Magnified SEM Fracture Surfaces of EIF René 95: (a) 500×, (b) 1500×.
Figure 6.26. Typical Cavitated Regions of Metallographic Cross-Section Through Failed HIP U-Notch Specimens: (a) 160×, (b) and (c) 1000×.
Figure 6.27a. Typical Cavitated Region of Metallographic Cross-Section Through Failed HIP U-Notch Specimens at 320x.
Figure 6.27b. Typical Cavitated Region of Metallographic Cross-Section Through Failed EIF U-Notch Specimen at 1000×.
7. Discussion of Results.

7.1. Summary of Results.

The creep rupture test results of René 95 have been presented in Chapter 6, but a brief summary is given below. Smooth bars were tested at stresses ranging from 690 to 880 MPa (100 to 127 ksi), with resulting rupture lives of 6 to 140 hr. Blunt notch (U-notch) specimens were subjected to net section stresses between 690 and 1100 MPa (100 to 160 ksi), leading to failures in 12 to 300 hr. Sharp notch (V-notch) specimens were tested in the range of 580 to 900 MPa (85-130 ksi), with rupture lives up to 70 hr; several V-notch specimens, tested between 350 and 550 MPa (50-80 ksi) did not initiate cracks within 250 hr, and the tests were terminated.

There was no effect of materials processing history on the rupture lives of smooth and U-notch specimens, and the effect on V-notch specimens was small, with hot isostatically pressed (HIP) specimens exhibiting slightly longer lifetimes than extruded and isothermally forged (EIF) specimens. Notch strengthening was found for the U-notch geometry, while severe notch weakening occurred in V-notch specimens. U-notch specimens showed an inverse power law relationship between stress and lifetime; smooth specimens did not follow such a neat relationship. Minimum creep rates were faster in smooth bars than in U-notch specimens. A power law relationship was found between stress and minimum creep rate in the U-notch specimens, with no significant effect of processing history. HIP smooth bar creep rates fit a power law creep rate equation, although the creep rates did not fit a simple stress dependence. At lower stresses, EIF specimens crept more slowly than HIP specimens, while at the highest stress, the EIF material crept faster. V-notch specimens showed very short lifetimes at high stresses, and an apparent threshold stress, below which crack initiation and failure would not occur within the laboratory time scale. The difference in rupture life between the HIP and EIF processed material is due to an increase in crack initiation time in the HIP specimens; on the other hand, the time to propagate a crack to failure in HIP specimens was shorter than in EIF specimens.

While rupture lives showed little or no dependence of processing history, various other measurements showed HIP material to be more brittle than EIF material. EIF U-notch and smooth specimens had a greater reduction in area (diametral strain) than HIP specimens. HIP specimens developed shear lips only at high stresses, while EIF specimens had larger shear lips at all stress levels. EIF notched specimens had larger fracture zones (cavitated regions in U-notch specimens, crack depths in V-notch specimens) than their HIP counterparts. Such a critical flaw size is actually more a measure of toughness than of ductility.
All HIP fractures followed the prior particle boundaries (PPBs) evident in the microstructures shown in Chapter 5. In EIF specimens, cracks propagate along grain boundaries, leading to microscopically rougher fracture surfaces. In cross sections of failed U-notch specimens, HIP material had accumulated small amounts of cavitation and cracking along the PPBs. EIF specimens exhibited much greater amounts of cavitation and microcracking along the grain boundaries and γ-γ' boundaries.

7.2. Notched Specimen Behavior.

The René 95 specimens show notch strengthening, *i.e.*, longer times to rupture in notched specimens than in smooth bars, for low notch acuities (U-notches) but severe weakening for sharper notches (V-notches). Such notch strengthening and weakening can be explained in terms of the time and geometry dependent tensile and equivalent stresses across the notch. The material above and below the notch produces a triaxial state of stress. Tensile stresses arise in the plane of the notch which prevent the material within the notch from contracting freely. The resulting von Mises equivalent stress is lower than the applied net section stress. Because triaxiality restricts plastic flow through this decrease in equivalent stress, the constraint due to the notch suppresses yield of the specimen and increases its load carrying capacity, while it reduces the strain at fracture. The resistance to plastic deformation also decreases the rate of power law creep; these lower creep rate cans result in longer lifetimes in notched specimens [93, 105, 132, 133].

The time dependent stress distribution across an arbitrarily shaped notch is shown in Figure 7.1 [93]. In the figures in this chapter, $a$ is the radius of the notch, and $r$ is the distance from the center of the specimen. Initially, the maximum tensile stress $\sigma_t$ and the equivalent (von Mises) stress $\sigma_e$ are very large at the root of the notch. Because stresses are higher at the notch root, creep strains accumulate there more rapidly than within the notch. As a consequence, stresses relax quickly at the notch root, and the location of the maximum tensile (axial) stress is shifted away from the root of the notch. The steep gradients in the tensile and equivalent stresses also become less severe. At steady state, the equivalent stress has become nearly constant across the notch throat, slightly higher at the surface; the axial stress shows a maximum somewhere between the center of the notch and the notch root. The precise form of the stress distribution across a notch depends on the severity of the notch and on the power law creep parameters of the material. The environment also plays an important role in notch rupture behavior.

The effect of notch acuity on creep rupture is illustrated in Figure 7.2, which plots the initial and steady state stress distributions for the U- and V-notch geometries, for a creep exponent $n=5$ [134-136]. As the notch increases in severity, the maximum tensile and
equivalent stresses are more strongly concentrated at the notch root. The concentration relaxes quickly right at the surface, but the material deeper within the notch, subjected to a much lower stress, creeps very slowly. As a result, it becomes more difficult to redistribute stresses in sharper notches. Crack initiation will occur after the accumulation of a critical amount of strain. While the high stresses in a blunt notch are relaxed before this critical strain is reached, the sharp notch may have accumulated this strain before the stresses have reached their steady state distributions. Thus, the tendency for notch weakening increases as the notch acuity increases [72, 93, 117, 133, 137].

Figure 7.3 shows calculated steady state stress distributions across a notch for low and high values of the creep exponent (n=3 and n=9) [105]. The notch has a moderate stress concentration (k=2.4). As the creep exponent n increases, the material at the notch root

---

**Figure 7.1.** Stress Distributions in a Notched Specimen. Left: Initial Elastic Stresses; Center: Transient Stresses; Right: Steady State Stresses. $\sigma_e$- Equivalent Stress, $\sigma_I$- Maximum Tensile Stress, $\sigma_{nom}$- Nominal (Net Section) Stress.

**Figure 7.2.** Stress Distributions in U-Notch and V-Notch Specimens. $\sigma_i$- Initial Elastic Stress, $\sigma_I$- Maximum Steady State Tensile Stress, $\sigma_e$- Steady State Equivalent Stress. n=5.
creeps more rapidly, shedding its load more effectively onto the material deeper within the notch. At steady state, the maximum in the tensile stress distribution occurs closer to the center of the specimen for a larger n, and the equivalent stress profile becomes flatter. The tensile and equivalent stresses at the surface of the notch decrease with increasing n. A smaller creep exponent can therefore lead to notch weakening due to less efficient stress redistribution across the notch and to higher steady state stresses at the surface of the notch root [93, 105, 117]. The cavity densities shown in Figure 7.3 follow the tensile stress distributions [105]. Because n=6 for the U-notch specimens tested in this study, the shapes of the René 95 stress distributions are intermediate between the two cases shown in Figure 7.3. The maximum in the steady state tensile stress curve, as well as the region of densest cavitation, should lie near r/a=0.5.

The effect of environment on notch rupture behavior is very important. Surface cracks typically initiate on a smooth specimen through the interaction of applied stress and embrittling atmosphere [20, 60, 117]. Several cracks may form, and one will grow to failure

![Figure 7.3](image-url)

*Figure 7.3. Calculated Spatial Distributions of Normalized Tensile and Equivalent Stresses at Steady State and Observed Distributions of Cavity Density (in Arbitrary Units) for n=3 and n=9.*
before significant accumulation of intergranular cavitation below the surface of the specimen. Under conditions of incomplete stress redistribution in a notched specimen, i.e., for low \( n \) or large \( k_t \), a crack will initiate much sooner than in a smooth bar because of the unrelaxed stresses at the surface of the notch. If conditions allow more complete redistribution of stress, crack initiation may not occur at the surface: stresses at the notch root may relax sufficiently before the combination of stress and environment can initiate a crack. The specimen then behaves as if it were tested under inert conditions: creep damage accumulates in the form of intergranular cavitation below the surface of the notch. Crack initiation is delayed or prevented, leading to notch strengthening. If tested in a vacuum, complete stress relaxation is more likely to occur for any notch geometry because the environmental component of cracking is eliminated [139].

Sharply notched specimens may exhibit a threshold behavior: above a certain stress, rupture will occur after a very short time; below this stress, a test will fail only after a very long time, if at all. Above this threshold net section stress, the highly concentrated stresses at the notch root are unable to redistribute before environmentally assisted crack initiation takes place. Below this stress, redistribution is slow, but the magnitude of the notch root stresses are insufficient to initiate a crack even in the presence of an embrittling atmosphere. Overall rates of accumulation of creep strain and damage are very small due to the small magnitudes of the stresses across the notch [72, 93, 117, 133, 137-139].

The creep rupture fracture surfaces and metallographic sections of failed notches from this study illustrate the effects of stress state and environment on the creep behavior of René 95. The stress rupture behavior of the René 95 specimens tested in this study is interpreted in Figure 7.4. This figure shows the initial and steady state tensile stress distributions for each specimen geometry, and the resulting creep damage morphology. The net section stress equals the applied load \( P \) divided by the cross-sectional area \( A \). Because there is no stress concentration in a smooth specimen, the stress remains constant with respect to position and time. The bulk of the specimen creeps steadily, while the combined effects of stress and atmosphere eventually nucleate a crack at the surface of the specimen. This crack grows by creep crack growth mechanisms, that is, by initiation and coalescence of intergranular cracks ahead of the crack tip. This damage is confined to the region of stress singularity at the tip of the crack; fracture occurs before the appearance of cavitation throughout the specimen.

The U-notch geometry shows an initial stress concentration at the root of the notch. This high stress relaxes rapidly due to creep at the notch root; the lower steady state stress at the surface eliminates the interaction of stress and environment. The effective stress in a
U-notch specimen is lower than the net section stress (see Figures 7.2 and 7.3). Because this stress is the driving force for creep strain accumulation, creep rates are lower in U-notch specimens than in smooth bars. Atmosphere-assisted crack initiation is forestalled by the low surface stresses, allowing intergranular creep cavities to nucleate within the notch where the tensile stresses are highest. The creep cavities grow and eventually coalesce to form grain boundary facet cracks; these cracks grow by the coalescence of intergranular cavities in their path or by linking together with other cracks, leading to fracture of the specimen. The decrease in creep rates and the elimination of surface crack initiation lead to notch strengthening in the U-notch geometry.

![Diagram](image)

**Figure 7.4.** Schematic Showing Location and Morphology of Creep Damage in René 95 as a Function of Stress Distribution for the Smooth Bar, U-Notch, and V-Notch Specimen Geometries.

The greatest density of damage observed in sectioned U-notches occurs partway between the notch surface and the center of the specimen; interpolation of the cavity histograms in Figure 7.3 for n=6 indicates that this should be the case. A quantitative analysis of the cavity distributions is beyond the scope of this study, but Figure 7.5 presents data from another research program on P/M HIP René 95 [117]. Tests in that program were performed on blunt notches (κ<sub>c</sub>=1.4) similar to those in this study, at a temperature of 650°C (1202°F) rather than at 704°C (1300°F). The damage shown in Figure 7.5 is more extensive than in this investigation due to the temperature difference, but the relative cavity and crack
distributions are in agreement. Along the midplane of the notch, the densities of cavities and cracks are greatest at $r/a=0.5$ and lowest at the notch surface. These densities drop off rapidly with distance from the midplane of the notch. These damage distributions are consistent with cavitation and cracking observed in sectioned U-notches in the present study.

The stress concentration is much more severe in a V-notch than in a U-notch; the severe constraint at the notch root prevents sufficient redistribution of this high stress. There is an enhanced interaction between environment and stress: crack initiation is observed much earlier than in a smooth bar subjected to the same net section stress. The time to grow a crack to failure in a V-notch specimen is very short for two reasons. First, the effective driving force for crack growth (e.g., stress intensity or $J$-integral) is high because the effective crack length equals the sum of the actual crack length and the notch depth. Second, in an air environment, René 95 exhibits very fast creep crack growth (CCG) [118, 119]. This susceptibility to CCG in air leads to the observation that as rupture life increases, the fraction of life spent in crack propagation decreases. The V-notch specimens exhibit the threshold stress behavior described above: if the applied stress is small enough, even incomplete stress relaxation can prevent cracking. The stresses across the notch throat are so low that the strain and damage accumulation rates are too small to measure.

7.3. Potential Drop Behavior.

A procedure to calculate potential drop feedback based on accumulation of uniform creep strain is detailed in Appendix D [123]. Essentially, a notched specimen is modeled as
a number of thin disks with varying diameters, stacked together to simulate the notch geometry. Strain in each disk is calculated from Norton’s Law for the net section stress on that disk, and the strain and potential drop increments are summed over the entire notch. There is no attempt to account for non-uniform stress distributions across the notch or for compatibility of strain between adjacent slices. Tests can be simulated for a given notch geometry, power law creep relationship, and applied stress. This technique is much less precise than finite element modeling, but it is simpler and produces reasonable results in much less time.

Calculated and experimental potential drop creep curves for smooth specimens are presented in Figure 7.6 for HIP tests and in Figure 7.7 for EIF tests. HIP experimental results follow the predictions closely for most of the rupture lives, deviating only slightly near the ends of the tests. The high and low stress EIF tests follow the predicted curves early in life. Later in life, the high stress PD curve deviates slightly from the predicted curve, while the low stress test shows a large deviation after about half of its lifetime. The intermediate stress EIF test does not follow the predicted curve at all; this test also deviated from the power law creep relationship defined by the high and low stress tests. Because the potential drop and LVDT/extensometer curves for each test do not differ appreciably, any deviation between these experimental PD curves and the predicted behavior is attributed to nonuniform deformation (i.e., necking or tertiary creep acceleration), rather than increased electrical resistance due to cavitation and cracking. In any case, neither measurement technique is expected to detect cracks with any precision. The large potential probe spacing, greater than the gauge length of the specimen, results in poor crack length resolution. The extensometer is insensitive to the minute amount of opening of such small cracks, particularly when averaged over the entire gauge section.

Predicted and measured U-notch potential drop creep curves are given in Figure 7.8 for HIP tests and in Figure 7.9 for EIF tests. HIP tests show good agreement between predicted and observed behavior, indicating that the measured potential drop is attributable to uniform power law creep strain accumulation. The observed potentials in EIF specimens initially follow the calculated curves, but begin to accelerate at some point between 40% and 80% of life. The measured potentials attain final values about two times the values predicted from uniform creep. This deviation is partly due to nonuniform strain; EIF specimens experienced much larger reductions in area than HIP specimens, although the computer predicted only a small difference. Much of the deviation is due to dense cavitation and microcracking which accumulate in EIF U-notch specimens. This cavitation increases the effective electrical resistivity across the notch, thereby increasing the measured potential drop.
The potential drop curves calculated for smooth and U-notch specimens match the measured curves early in life for the EIF tests, and throughout the duration of the HIP tests. This agreement between the predicted and experimental potential drop behavior indicates that the simple computer calculations provide a reasonable description of the accumulation of uniform power law creep strain in a notched specimen. This correlation also indicates that the increase in potential drop signal is due mainly to creep deformation in smooth and U-notch bars for the HIP tests and for the early portions (up to half of the lifetimes) of the EIF tests.

7.4. Modelling Creep Behavior in René 95.

All aspects of the observed creep rupture behavior of René 95 smooth and notched specimens can be explained in terms of the microstructures and properties resulting from the two processing histories studied here and in terms of the specimen geometries. These aspects of creep behavior include the location, type, and accumulation rates of creep damage in the three geometries of HIP and EIF specimens; ductility and toughness as a function of the material condition and geometry; the times to rupture for HIP and EIF notched and unnotched specimens; and the potential drop measurements corresponding to the observed creep behavior.

The effect of processing history on microstructure is summarized as follows. Large powder particles retain their largely undeformed structure in the HIP material, while smaller particles deform to fill in spaces between these larger particles. The heavily deformed material from the small particles subsequently recrystallizes. No evidence of the powder structure remains in the EIF material; all particles are severely deformed during extrusion and forging, and the entire microstructure is recrystallized (refer to Figures 5.1 through 5.4). During the HIP operation, a semicontinuous carbide film precipitates along the prior particle boundaries (PPBs). Because most of the carbon is tied up in the PPB carbide, there is very little grain boundary carbide formation in the HIP material. During treatment of the EIF material, discrete intergranular carbides nucleate throughout the microstructure. The HIP disk is solution treated at a higher temperature than the EIF disk (see Table 5.2); therefore, more primary $\gamma'$ is dissolved in the HIP material, and a larger volume fraction of fine intracrystalline $\gamma'$ can precipitate during the subsequent aging treatment. The EIF disk is oil quenched to room temperature after solutionizing; the fast cooling rate promotes the formation of smooth grain boundaries. The HIP material is quenched in an elevated temperature salt bath; this slower cooling rate allows the precipitation of some $\gamma'$ at the grain boundaries, resulting in contorted boundaries. Therefore, the smooth EIF boundaries are susceptible to grain boundary sliding (GBS), while the contorted HIP boundaries retard sliding.
The observed creep rates in U-notch specimens are slightly higher for EIF material than for HIP material. Because the HIP alloy has a greater volume fraction of fine transgranular $\gamma'$, it has a greater resistance to dislocation motion than does the EIF material. The EIF material has an added creep component due to grain boundary sliding.

As previously sketched in Figure 7.4, the creep damage nucleation site is a function of specimen geometry. In smooth creep specimens, a creep crack initiates at the surface of the specimen under the combined influence of stress and environment. The exact crack nucleation site may be an oxidized and fractured carbide particle at the surface, although details are obscured by oxidation of the specimen. The incidence of crack nucleation is not expected to vary between HIP and EIF material, since the intersection of carbides with the large specimen surface area occurs frequently for either condition. Smooth bar rupture life consists of an initiation life plus a time to propagate the crack to failure. Initiation is probably caused by oxidation and cracking of carbides at the surface of the specimen. Because the DCPD technique is not sensitive enough to monitor crack length in smooth specimens, there is no way to determine what fraction of life was needed to initiate a crack. Comparison with V-notch behavior indicates that propagation was rapid and therefore contributed little to the total rupture life.

V-notch specimens also initiate surface cracks. Crack initiation in V-notch bars occurred by cracking of carbides at the surface of the notch root. Because carbides are closer together in EIF specimens (5 $\mu$m grain size vs. 100 $\mu$m powder particle size), there is a greater probability that a carbide will be located at the point of highest stress at the notch surface. This explains why crack initiation times are somewhat shorter in EIF specimens than in HIP specimens. The faster CCG rates in HIP material partially balances the time difference for crack nucleation; rupture lives are slightly longer for HIP specimens.

Because the stress distribution reaches a maximum well below the notch root, surface crack initiation is prevented in U-notch specimens. Damage accumulates in the form of cavities, which nucleate, grow, and eventually coalesce into macroscopic cracks. The difference in behavior between the HIP and EIF treatments is in the location of these cavities. In HIP specimens, cavities nucleate along the prior particle boundaries, probably at the carbide-austenite phase boundaries. The absence of grain boundary sliding and of intergranular carbides retard formation of intergranular cavitation. These cavities have an effective z-axis spacing (parallel to the applied stress) of about 100 $\mu$m, corresponding to the powder size. Cavities nucleate on EIF grain boundaries, probably at intergranular carbides and at $\gamma-\gamma'$ phase boundaries. Grain boundary void nucleation is promoted in EIF material by
grain boundary sliding. The EIF cavity spacing is about 5 μm in the applied stress direction, corresponding to the grain size.

Creep crack growth (CCG) occurs by nucleation, growth, and coalescence of cavities on boundaries ahead of the tip of the growing crack, regardless of specimen geometry. Therefore, HIP specimens exhibit "ballular" PPB fracture surfaces, while EIF specimens have intergranular fracture surfaces. CCG rates are somewhat higher in HIP specimens. Part of this difference is due to the greater ease in cracking along continuous PPB carbide films compared to crack jumping from one discrete grain boundary carbide particle to the next. In addition, grain boundary sliding in EIF material results in crack tip blunting, less severe crack tip stress fields, and lower crack growth rates. Growth of an existing creep crack is easier than initiation of a crack. At lower stresses, the time to initiate a crack increases more than the time to propagate that crack to failure; therefore, the fraction of life spent in crack initiation increases as the applied stress decreases.

Rupture lives for U-notch specimens essentially are the times needed to initiate creep cavities on the appropriate interfaces (i.e., prior particle boundaries in HIP specimens or grain boundaries in EIF specimens), to grow these cavities until they coalesce into facet cracks, and to grow these facet cracks together into a macroscopic crack of critical size. The distribution (size and spacing) of this cavitation is unknown for the specimens tested in this study. Fracture surfaces are obscured by an oxide layer, and metallographic preparation of U-notch cross sections has resolved only coalesced microcracks. It is not unreasonable to assume, however, that the areal distributions of cavitation are similar on prior particle boundaries in HIP specimens and on grain boundaries in EIF specimens.

The time to coalesce interfacial cavities into a facet (PPB or grain boundary) crack is the time needed to grow a cavity from its initial size to the size equal to the cavity spacing. Given similar initial and final cavity sizes in HIP and EIF material, the coalescence time is not dependent on processing history. The spacing of cracks in the axial direction (parallel to the applied stress) does depend on processing history. In HIP specimens, cracks are spaced apart by the powder particles (up to 100 μm), while in EIF specimens, cracks can be as close as a single grain (about 5 μm). These coalesced cracks grow until they reach a critical size; crack growth should occupy only a small fraction of total specimen life.

EIF René 95 exhibits greater ductility and toughness than HIP material, in part because of the sliding of EIF grain boundaries. This GBS leads to crack tip blunting and reduction of crack tip stresses, as described above. For a given applied stress, the EIF material can sustain a longer crack, or a greater amount of cavitation, than the HIP material. Grain boundary sliding also contributes to the strain measured in EIF specimens at fracture.
The direct current potential drop (DCPD) feedback measures deformation in smooth specimens of René 95. Section 7.3 describes the correlation between DCPD creep curves and the corresponding creep strain curves measured by extensometer and LVDT. There is no contribution to DCPD from other creep damage, since smooth bars do not accumulate a significant amount of intergranular cavitation. Because the potential probes are widely spaced, there is not enough sensitivity to measure crack growth, even for long cracks.

The DCPD curves in V-notch René 95 specimens show crack initiation and propagation. Because net section stresses are very small, there is no measurable deformation in the specimen to change the DCPD signal, and cavitation is so highly localized at the notch root that it does not contribute to the measured potential. It is possible to relate the increase in potential to an increment in length of an axially symmetric circumferential crack at the notch root [125, 131], but asymmetric cracking in René 95 prevents the quantitative determination of crack growth.

In U-notch specimens of HIP-processed René 95, potential drop curves closely follow the accumulation of deformation across the notch, as described in Section 7.3. In EIF-processed material, potential drop creep curves follow the strain accumulation curves for about half of the specimen lifetime. After this, the potential curves accelerate due to nonuniform creep strain (such as necking) and to the accumulation of a significant amount of creep cavitation. In EIF specimens, the large amount of grain boundary cavitation causes a large increase in the resistivity of the material in the notch. Conversely, cavitation in HIP material is more widely spaced (on prior particle boundaries rather than on grain boundaries); the presence of such a small amount of cavitation has only a minimal effect on the resistance of the notch.
Figure 7.6. Calculated and Experimental Potential Drop Creep Curves for Smooth Specimens of HIP René 95. (a) Test 22: 100 ksi, (b) Test 29: 112 ksi, (c) Test 25: 127 ksi.
Figure 7.7. Calculated and Experimental Potential Drop Creep Curves for Smooth Specimens of EIF René 95. (a) Test 23: 100 ksi, (b) Test 30: 112 ksi, (c) Test 26: 127 ksi.
Figure 7.8. Calculated and Experimental Potential Drop Creep Curves for U-Notch Specimens of HIP René 95. (a) Test 27: 100 ksi, (b) Test 13: 120 ksi, (c) Test 6: 130 ksi (cont.).
Figure 7.8. Calculated and Experimental Potential Drop Creep Curves for U-Notch Specimens of HIP René 95. (d) Test 18: 145 ksi, (e) Test 9: 160 ksi.
Figure 7.9. Calculated and Experimental Potential Drop Creep Curves for U-Notch Specimens of EIF René 95. (a) Test 28: 100 ksi, (b) Test 5: 130 ksi, (c) Test 3: 130 ksi (cont.).
Figure 7.9. Calculated and Experimental Potential Drop Creep Curves for U-Notch Specimens of EIF René 95. (d) Test 17: 145 ksi, (e) Test 12: 160 ksi, (f) Test 1: 160 ksi.
8. Summary and Conclusions.

Creep rupture tests were carried out on smooth and notched specimens of René 95, at 704°C (1300°F) in laboratory air. Two notch geometries were used, a blunt notch (round notch or U-notch, $k_t=1.6$), and a sharp notch (the British Standard Notch or V-notch, $k_t=4.2$). The René 95 had been produced by powder metallurgy techniques, consolidated into turbine disk form by two different methods: hot isostatic pressing (HIP) and extrusion and isothermal forging (EIF). Creep tests were monitored using the direct current (DC) potential drop (PD) technique to obtain a continuous record of creep deformation and damage accumulation. An extensometer and LVDT were used in conjunction with the DCPD measurements to record creep curves for the smooth bar specimens. This study was undertaken for two purposes. The first was to investigate the effects of specimen geometry (smooth bar, blunt notch, and sharp notch) and of processing history (i.e., HIP vs. EIF) on the creep rupture properties of René 95. The second purpose was to evaluate the usefulness of the DCPD technique to monitor the accumulation of creep damage and deformation in a creep brittle alloy.

Summary

The following observations and explanations were made of the René 95 creep rupture behavior:

- René 95 showed moderate notch strengthening (longer rupture lives than smooth bar specimens at a given stress) in blunt notch tests but severe notch weakening in sharp notch tests. Smooth specimens exhibited low ductility at fracture; rupture ductility decreased further as notch acuity increased.
- Each smooth bar specimen failed after the initiation of a single crack at the surface of the specimen, probably by oxidation and cracking of a carbide precipitate; no intergranular cavitation or microcracking was detected beneath the surface. Blunt notch specimens did not undergo surface crack initiation; damage accumulated within the notch in the form of cavities at prior particle boundaries (PPBs) in HIP material and at grain boundaries in EIF material. These cavities grew, coalesced into facet cracks, and merged into larger macroscopic cracks which grew to failure. Sharp notch specimens exhibited early crack initiation at the notch root surface.
- Notched specimen behavior is attributed to the time- and position-dependent stress distributions across the notch throat and to the interaction of the stress at the specimen surface with the air environment. Smooth bars are subjected to a non-varying stress; atmosphere-assisted crack initiation occurs at the surface of
the specimen. Sharp notches have severe stress concentrations at the notch root. Under the influence of the environment, cracks initiate readily at the notch root surface before these stresses can be redistributed across the notch. These cracks grow quickly to failure. The stress concentration in blunt notch specimens can relax at the root of the notch, eliminating the environmental initiation of surface cracks. Damage occurs deep within the notch, where the redistributed stresses are greatest. As the sharpness of the notch increases, the ability of the material within the notch to deform plastically decreases, and the resulting rupture strains are decreased.

- There is little or no effect of processing history on rupture lives. The crack propagation path depends on the powder consolidation method used: HIP specimens failed along PPBs, while EIF specimens failed along grain boundaries. The density of damage in HIP U-notch specimens was much less than that in EIF U-notch specimens, because the cavities are much more widely spaced (100 μm powder particle size in HIP material vs. 5 μm grain size in EIF material). Despite this difference in cavity densities, the initial cavity size and spacing is similar, so the time to coalesce cavities into macroscopic cracks is about the same for HIP and EIF specimens. Therefore, processing history has a negligible effect on U-notch specimen life. The times to initiate surface cracks in smooth and V-notch specimens are significantly greater than the times to propagate these cracks to a critical size. Although creep crack propagation is faster in HIP specimens than in EIF specimens, the rupture lives are essentially the same because the initiation times are nearly the same.

- EIF specimens show greater rupture ductilities than HIP specimens. Higher cooling rates during heat treatment result in smooth and precipitate-free grain boundaries in the EIF material. EIF specimens are therefore able to accumulate strain through grain boundary sliding as well as through power law creep.

- The potential drop technique was able to monitor creep strain during smooth bar tests. The PD probes were attached too far apart on the specimens to resolve crack initiation or propagation. In sharp notch specimens, the DCPD method could resolve crack initiation for very small crack lengths. Because of asymmetrical crack growth, only a qualitative measure of crack propagation is possible using DCPD. In U-notch specimens, creep deformation can be followed using the DCPD technique, and a simple procedure can be used to extrapolate a rupture life for the unfailed notch of a specimen by comparing the
potential curves from both notches. The amount of cavitation and microcracking in HIP U-notches was too small to be resolved through PD measurements; the EIF material accumulated a sufficient amount of creep damage to have a significant effect on the PD feedback.

Conclusions

The following conclusions can be drawn about the creep rupture of René 95, and about the testing methods employed:

(1) René 95 is a strong but very brittle alloy. For blunt notches, some degree of notch strengthening is possible, but for sharper notches, this alloy is severely weakened. Once initiated, creep cracks propagate very rapidly, especially in the stress field at the root of a notch.

(2) The DCPD technique is useful for measuring creep deformation in smooth and U-notch specimens, and can detect initiation of a small crack at a sharp notch. Asymmetric crack growth prevents the accurate measurement of crack length in V-notch specimens. Potential drop can be used to monitor the extent of cavitation and microcracking in EIF U-notch specimens, in which a significant amount of damage occurs. Much less creep damage accumulates in HIP specimens; the DCPD method is not sensitive enough to measure this damage.

(3) The choice of powder consolidation method has little effect the rupture life of a laboratory specimen. Processing history does affect the location of creep cavitation and the path of a propagating crack as well as crack growth rates, and it affects the strain at failure in the specimens tested. The designer of a turbine engine may find extruded and isothermally forged (EIF) René 95 preferable to hot isostatically pressed (HIP) René 95, due to slower creep crack growth rates, a greater tolerance for creep damage, and slightly improved ductility in the EIF alloy. The EIF material also has a more homogeneous microstructure, resulting in less variability of tensile and creep rupture properties.
Appendix A. HP85A BASIC Program to Monitor DCPD Testing.

The HP85A BASIC program which controls acquisition of direct current (DC) potential drop (PD) test data is presented on the following pages. Operation of this program is not complicated; a brief description is given below. Although the computer does not control the temperature of the furnace, the program monitors the two thermocouples and displays the temperatures while the specimen heats up and during the test. This allows the operator to adjust the temperature controller precisely, to stabilize the temperature, and to eliminate any thermal gradients along the length of the specimen, before the load is applied.

A schematic of the DCPD data collection system is presented in Figure 5.6, and Figure A.1 shows a flow chart of the data collection routine. During each data collection cycle, the program reads each active channel on the multiprogrammer. These channels include two thermocouples, a reference probe, and up to four additional voltages, which may be DCPD probes, an LVDT to monitor specimen elongation, or any other measurable potential. The computer stores these values and elapsed time on a disk in the disk drive. When each data file is filled, the program formats and begins entering data in a new file on the disk. When all data files on the disk are full, the computer switches to the other disk in the drive and leaves a message on its internal printer telling the operator to replace the full disk with an empty one. Before storing data on the second disk, the program makes sure that the disk is empty. If the disk is not new, data storage proceeds on the tape drive of the HP85A, and the computer prints a message to alert the operator. If the tape fills up, or if there is no tape cartridge in the tape drive, the data is output on the printer.

During each cycle, data is numerically tabulated and graphically plotted on the CRT screen of the HP85A. The operator can follow the progress of the test through this “on-the-fly” display of data. After storing and presenting the data during each cycle, the computer program checks for fracture of the specimen. Failure is indicated by an “infinite” potential: about 1 V is recorded for an open circuit, much greater than the maximum of about 5 mV recorded for a notch potential. The program also searches for any error condition in the I/O devices, and continually monitors for errors in its routine operation. A message is printed to alert the operator to specimen failure or computer error; data collection continues in order to minimize loss of information. For example, a broken potential drop lead or a faulty spot weld could cause the computer to indicate a spurious failure of the specimen; a computer error may be unrelated to data collection from the continuing creep test, and the operator may be able to salvage the data if it has been affected.

In addition to performing these tasks, the program provides the operator with various cues before and during the test. These include reminders to activate ice point reference
junctions on the thermocouples, to check the current from the power supply before applying the load, and to insert new disks at the start of the test.

Figure A.1. Flow Chart for Computerized Data Acquisition.
The Following is a Recent Version of the HP-85A BASIC Computer Program to Monitor Creep Rupture Testing.

10 ! "RUPTURE" ! 9 DEC 86
20 ! Set Up I/O
30 OPTION BASE 10 DIM Z$[32]
40 Z$="********************
  "****" @ DIM A(7),G(5)
50 B,E0,G,G1,S1=0 @ S0=16 @ D$=
60 "*
60 RESET 7 @ CLEAR 7 @ CLEAR
70 WAIT 3000
80 LOCAL LOCKOUT 7
90 OUTPUT 723 ;"F1T4R-2RAZ1N5T3
100 OUTPUT 709 ;"CMON1"
110 CLEAR @ DISP Z$
120 DISP "Where do you want to start?" @ GOSUB 130 @ GOTO 23 @
130 DISP " 0: Up to Temperature"
140 DISP " 1: Potential Measurement"
150 DISP Z$ @ BEEP
160 INPUT Y @ RETURN
170 CLEAR @ DISP Z$ @ BEEP
180 DISP "Turn on Cold Junction"
190 DISP " Compensators,"
200 DISP "Then Press CONTINUE"
210 DISP Z$ @ PAUSE
220 RETURN
230 IF Y=1 THEN 520
240 ! Up to Temp
250 CLEAR @ DISP Z$
260 DISP "Enter Desired Temperature"
270 DISP "(degrees C and millivolts)
11s)"
280 DISP Z$ @ BEEP @ INPUT T,V
290 GOSUB 170
300 OUTPUT 709 ;"SLIST 101,102"
310 ON KEY# 1 GOTO 480
320 FOR I=2 TO 3
330 TRIGGER 709 @ TRIGGER 723
340 ENTER 723 ; A
350 A(I)=A(I)+1000 @ NEXT I
360 A2$,A3$=" 
370 A2=704+(A(2)-29.246)/.043
380 A3=704+(A(3)-29.296)/.043
390 IF ABS(A(2)-V)<.1 THEN A2$="*"
400 IF ABS(A(3)-V)<.1 THEN A3$="*
410 DISP Z$ @ DISP USING 440 ; "Set Value",V,T
420 DISP USING 440 ; "Temp 1",A(2),A2,A2$
430 DISP USING 440 ; "Temp 2",A(3),A3,A3$
440 IMAGE 10A,3DZ.3D,4DZ.2D,XX,A
450 WAIT 4000
460 IF A2$="*" OR A3$="*" THEN 3
470 DISP "At Temperature; Key 1 to go on" @ GOTO 320
480 OFF KEY# 1
490 CLEAR @ DISP Z$
500 DISP "Now what do you want to do?" @ GOSUB 130
510 IF Y#1 THEN 310
520 ! Graphic Set-up
530 CLEAR @ DISP Z$
540 DISP "Enter Test Designation 
  @ P=1 @ DISP Z$
550 INPUT T$
560 DISF "Enter X-axis 
maximum" @ Disp Z$
570 DISF "(24 hours suggested)" 
580 DISF @ BEEP @ INPUT T0 
590 GOSUB 660 @ WAIT 3000 
600 CLEAR @ DISP Z$
610 DISF "Is plot correct?" 
620 DISF "0=no" 
630 DISF "1=yes" @ BEEP 
640 DISF @ INPUT Y 
650 IF Y=1 THEN 890 ELSE 520 
660 ! Frame Drawing Subr. 
670 DEG @ GCLEAR 
680 SCALE -37,220,-53,140 
690 MOVE 0,-28 @ DRAW 0.140 @ DR 
   AW 220,140 @ DRAW 220,-28 @ 
   DRAW 0,-28 
700 FOR I=0 TO 8 STEP 2 
710 MOVE I*22,-22 @ IDRAW 0,-6 
720 LDIR 0 @ IMOVE -10,-12 
730 IF P>10 THEN IMOVE -8,0 
740 LABEL VAL$(P-1)$&"."&VAL$(I) 
750 MOVE (I+1)*22,-28 @ IDRAW 0, 
   3 @ NEXT I 
760 MOVE 216,-40 @ IF P>9 THEN I 
   MOVE -8,0 
770 LABEL VAL$(P)$ 
780 MOVE 65,-53 @ LABEL "Time/"& 
    VAL$(T0)$&" hrs" 
790 MOVE -30,-50 @ LDIR 90 
800 LABEL "Change in Potential 
(mV)" 
810 LDIR 0 
820 FOR I=0 TO 4 
830 MOVE 6.28*I @ IDRAW -6.0 
840 IF="0."&VAL$(I) 
850 IMOVE -25,-4 
860 LABEL I$ @ NEXT I 
870 MOVE 10,125 @ LABEL T$ 
880 RETURN 
890 ! Set Up Data Storage 
900 GOSUB 170 @ CLEAR @ DISP Z$
910 DISF "Is data to be stored o 
   n disc?" 
920 DISF "0:No, bypass disc" 
930 DISF "1:Yes, store data" 
940 DISF @ BEEP @ INPUT D1 
950 IF D1=0 THEN 1020 
960 DISF @ DISP "Put discs in 
   and press CONT" @ BEEP 
970 DISF @ PAUSE 
980 CLEAR @ DISP Z$ @ BEEP 
990 DISF "How many disc drive s 
   (2/4)?" 
1000 DISF @ INPUT D 
1010 IF D#2 AND D#4 THEN 980 
1020 DISF @ DISP "How many da 
   ta Per cycle (5-8)?" 
1030 DISF @ BEEP 
1040 INPUT MR N=M-1 
1050 IF M<5 OR M>8 THEN 1020 
1060 ON ERROR GOSUB 3350 
1070 IF D1=0 THEN D=1 
1080 V=3 @ W=M*8 @ G$="abc" 
1090 OFF KEY# 1 
1100 FOR Y=1 TO 5 
1110 O(Y)=0 @ NEXT Y 
1120 ON N-3 GOTO 1130,1140,1150, 
1160 
1170 ! Main Body of Program 
1180 FOR D0=1 TO D 
1190 D$="0:"&VAL$(INT((D0-1)/2)) 
   &VAL$(2*FP((D0-1)/2)) 
1200 FOR V1=1 TO V 
1210 F$=T$&G$[V1,V13]&D$ 
1220 IF D1=0 THEN 1300
1230 PRINT "Creating '"F$:"'
1240 ON ERROR GOTO 2710
1250 IF G=1 THEN ASSIGN# 1 TO *
1260 CREATE F$,*U,W
1270 ASSIGN# 1 TO F$
1280 ON ERROR 'OSUB
1290 Q(00)=0
1300 IF G=0 THEN GOSUB 1370
1310 FOR U1=1 TO U-1
1320 GOSUB 1670
1330 NEXT U1
1340 PRINT F$&"full"
1350 PRINT USING "3DZ.3D,3A" ; T
1360 NEXT V1
1370 NEXT DO @
1380 ! Check PD Before Startin3
1390 OUTPUT 709 ; "CRESET 1; SLIS T 100-106"
1400 WAIT 50 @ FOR I=1 TO 7
1410 IF I>5 THEN A(I)=0
1420 IF I>5 THEN 1450
1430 TRIGGER 709 @ TRIGGER 723
1440 A(I)=A(I)*1000
1450 NEXT I @
1460 DISP Z@ @ GOSUB 2480
1470 DISP Z@ @ DISP "Are Potenti als Correct?"
1480 DISP " 0=No"
1490 DISP " 1=Yes"
1500 DISP Z@ @ BEEP @ INPUT Y
1510 IF Y=1 THEN 1570
1520 CLEAR @ DISP Z$
1530 DISP "Adjust Current and"
1540 DISP "Press CONTINUE"
1550 DISP Z$
1560 BEEP @ PAUSE @ GOTO 1370
1570 CLEAR @ DISP Z$
1580 DISP "New test (start timer ) : '1'"
1590 DISP "Continued test: '0'"
1600 DISP Z@ @ BEEP @ INPUT Y
1610 IF Y=0 THEN RETURN
1620 CLEAR @ DISP Z$
1630 DISP "Apply Load and"
1640 DISP "Press CONTINUE"
1650 DISP Z@ @ BEEP @ PAUSE
1660 SETTIME 0.1 @
1670 ! Data Collection Subr.
1680 IF U1<20 OR D$=":T" THEN 17
1690 WAIT 6000 @ ALPHA @ WAIT 60 00
1700 IF D0=6 THEN WAIT 20000
1710 DISP & DISP USING 1720 ; TIME/3600+(DATE-1)*24
1720 IMAGE "Taking Data, " ,DDZ.3 D," hrs..."
1730 OUTPUT 709 ; "CRESET 1 ; SLI ST 100-106"
1740 WAIT 50
1750 ! Take Data
1760 FOR I=1 TO N
1770 TRIGGER 709 @ TRIGGER 723
1780 ENTER 723 ; A(I)
1790 A(I)=A(I)*1000
1800 NEXT I
1810 T=TIME/3600+(DATE-1)*24
1820 IF G#0 THEN 1940
1830 ! Initial Values for Plot
1840 ON N-3 GOTO 1880,1870,1860, 1850
1850 A7=A(7)
1860 A6=A(6)
1870 A5=A(5)
1880 A4=A(4)
1890 A1=A(1) @ G=1 @ PRINT
1900 PRINT "Initial Values ";
1910 PRINT USING "3DZ.3D,3A" ; T ," hr"
1920 PRINT ALL @ GOSUB 2480
1930 NORMAL @ PRINT @ GOTO 1950
1940 GOSUB 2480
1950 ! Plot Data
1960 X=T/T0*220 MOD 220
1990 Y7=(A(7)-A7+.2) MOD .6*280-.28
2000 Y5=(A(5)-A5+.15) MOD .6*280-.28
2010 Y4=(A(4)-A4+.125) MOD .6*280-.28
2020 Y1=(A(1)-A1+.1) MOD .6*280-.28
2030 P1=INT(T/T0+.1)
2040 IF P=P1 THEN 2060
2050 P=P1 @ GOSUB 660
2060 ON N-3 GOTO 2140,2110,2090, 2070
2070 MOVE X-1,Y7-1 @ IDRAW 2,2
2080 MOVE X-1,Y6 @ IDRAW 2,0
2090 MOVE X-1,Y5 @ IDRAW 1,1
2100 IDRAW 1,1 @ IDRAW -1,1
2110 MOVE X-1,Y4-1 @ IDRAW 2,0
2120 IDRAW 0,2 @ IDRAW -2,0
2130 IDRAW 0,-2
2140 MOVE X,Y1 @ DRAW X,Y1
2150 ! Store Data
2160 IF D1#1 THEN 2340
2170 ON ERROR GOTO 2770
2180 IF U1=1 THEN 2230
2190 READ# 1,U1-1 : Y
2200 ON N-3 GOTO 2240,2260,2280, 2300
2210 PRINT# 1,U1 : T,A2,A3,A(1), A(4)
2220 GOTO 2310
2230 PRINT# 1,U1 : T,A2,A3,A(1), A(4),A(5)
2240 GOTO 2310
2250 GOTO 2310
2260 PRINT# 1,U1 : T,A2,A3,A(1), A(4),A(5)
2270 GOTO 2310
2280 PRINT# 1,U1 : T,A2,A3,A(1), A(4),A(5),A(6)
2290 GOTO 2310
2300 PRINT# 1,U1 : T,A2,A3,A(1), A(4),A(5),A(6),A(7)
2310 IF D$"=T" THEN GSTORE "PLOT T"&VAL$(P)&D$
2320 Q(D0).G1=0
2330 ON ERROR GOSUB 3350
2340 GOSUB 3410
2350 ! Check for Broken Spec.
2360 IF ABS(A(1))>30 OR ABS(A(4))>30 OR ABS(A(5))>30 THEN 2380
2370 IF ABS(A(6))>130 OR ABS(A(7))>30 THEN 2380 ELSE RETURN
2380 IF B=1 THEN RETURN
2390 BEEP @ PRINT ALL @ DISP Z$
2400 DISP "TEST OVER" @ BEEP
2410 DISP U1:" Data Cycles"
2420 DISP USING "3DZ.3DsK" T," hr"
2430 DISP Z@ NORMAL
2440 IF D1#0 AND D$"=T" THEN 2460
2450 MOVE 0,0 @ COPY @ GOTO 2470
2460 GSTORE "PLOTf"&D$
2470 B=1 @ RETURN
2480 ! Display Data on CRT
2490 ON N-3 GOTO 2600,2570,2540, 2500
2500 DISP USING 2620 ; A(1)
2510 DISp USING 2630 ; "V1",A(4)
2520 DISp USING 2630 ; "W3",A(6)
2530 GOTO 2640
2540 DISp USING 2630 ; "W4",A(7)
2550 DISp USING 2630 ; "W3",A(6)
2560 GOTO 2640
2570 DISp USING 2630 ; "W3",A(7)
2580 DISP USING 2630 ; "Ref",A(1)
2590 DISP USING 2630 ; "V1",A(4)
2600 DISP USING 2630 ; "V2",A(5)
2610 DISP USING 2630 ; "W4",A(7)
2620 DISP USING 2630 ; "W3",A(6)
2630 DISP USING 2630 ; "W3",A(7)
2640 DISP USING 2630 ; "W3",A(6)
2650 DISp USING 2630 ; "W3",A(7)
2660 DISp USING 2630 ; "W3",A(6)
2670 DISp USING 2630 ; "W3",A(7)
2680 DISp USING 2630 ; "W3",A(6)
2690 DISp USING 2630 ; "W3",A(7)
2700 DISp USING 2630 ; "W3",A(6)
2710 DISp USING 2630 ; "W3",A(7)
2720 DISp USING 2630 ; "W3",A(6)
2730 DISp USING 2630 ; "W3",A(7)
2740 DISp USING 2630 ; "W3",A(6)
2750 DISp USING 2630 ; "W3",A(7)
2760 DISp USING 2630 ; "W3",A(6)
2770 DISp USING 2630 ; "W3",A(7)
2780 DISp USING 2630 ; "W3",A(6)
2790 DISp USING 2630 ; "W3",A(7)
2800 DISp USING 2630 ; "W3",A(6)
2810 DISp USING 2630 ; "W3",A(7)
2820 DISp USING 2630 ; "W3",A(6)
2570 DISP USING 2620 ; A(1)  
2580 DISP USING 2630 ; "V1", A(4)  
2590 GOTO 2640  
2600 DISP USING 2620 ; A(1)  
2610 GOTO 2640  
2620 IMAGE "Reference", 3DZ.4D  
2630 IMAGE "T", A(2)  
2640 A2=704+(A(2)-29.296)/.043  
2650 A3=704+(A(3)-29.296)/.043  
2660 DISP USING 2680 ; A(2), A2  
2670 DISP USING 2680 ; 2..A(), A3  
2680 IMAGE "Temp", D.40.." mV"  
2690 RETURN  
2700 ! Disc Errors  
2710 OFF ERROR & ON ERROR GOSUB 3350  
2720 IF ERRN#63 THEN 2840  
2730 IF Q(DO)<3 THEN PRINT F""  
2740 Q(DO)=Q(DO)+1/2  
2750 IF D0=5 AND Q(5)<3 THEN 135  
2760 GOTO 2900  
2770 OFF ERROR & ON ERROR GOSUB 3350  
2780 IF ERRN#66 THEN 2810  
2790 IF Q(DO)<3 THEN PRINT F""  
2800 GOTO 2900  
2810 IF ERRN#71 THEN 2840  
2820 IF Q(DO)<3 THEN PRINT "End of "&F""  
2830 GOTO 2970  
2840 IF ERRN#72 THEN 2960  
2850 IF Q(DO)<3 THEN PRINT "Record: "&F""  
2860 GOTO 2870  
2870 Q(DO)=Q(DO)+1/2  
2880 IF G1=0 THEN GOSUB 2930  
2890 G1=1  
2900 IF Q(1)<3 OR Q(2)<3 THEN 13  
2910 IF D=2 THEN 3180  
2920 IF Q(3)<3 OR Q(3)<3 THEN 13  
2930 PRINT USING "3DZ.3D.3A" ; T  
2940 PRINT ALL @ GOSUB 2480  
2950 NORMAL @ RETURN  
2960 IF ERRN#60 THEN 2990  
2970 IF Q(DO)<3 THEN PRINT D""  
2980 GOTO 3120  
2990 IF ERRN#62 AND ERRN#130 THEN 3020  
3000 IF Q(DO)<3 THEN PRINT D""  
3010 GOTO 3120  
3020 IF ERRN#126 THEN 3050  
3030 IF Q(DO)<3 THEN PRINT D""  
3040 GOTO 3120  
3050 IF ERRN#65 AND ERRN#128 THEN 3080  
3060 IF Q(DO)<3 THEN PRINT D""  
3070 GOTO 3120  
3080 IF ERRN#129 THEN 3110  
3090 IF Q(DO)<3 THEN PRINT D""  
3100 GOTO 3120  
3110 PRINT "Disc Error"; ERRN; "on Line"; ERRL  
3120 Q(DO)=Q(DO)+1  
3130 IF G1=0 THEN GOSUB 2930  
3140 G1=1  
3150 IF Q(1)<3 OR Q(2)<3 THEN 13  
3160 IF D=2 THEN 3180  
3170 IF Q(3)<3 OR Q(4)<3 THEN 13  
3180 IF Q(5)<2 THEN 3260
3190 D$=":T" @ D0=5
3200 V=INT(217596/W) @ V=1
3210 PRINT Z$
3220 PRINT "Replace Discs and"
3230 PRINT "Press KEY 1"
3240 ON KEY# 1 GOTO 1080
3250 PRINT Z$ @ GOSUB 1200
3260 PRINT ALL @ DISP Z$
3270 DISP "Replace Discs and"
3280 DISP "Press KEY 1"
3290 DISP Z$ @ OFF KEY# 1
3300 ON KEY# 1 GOTO 3340
3310 NORMAL @ D1=0
3320 D0=6 @ CRT IS 2
3330 GOSUB 1200 @ GOTO 3330
3340 D1=1 @ CRT IS 1 @ GOTO 1080
3350 ! Non-Disc Errors
3360 OFF ERROR @ ON ERROR GOSUB 3350
3370 IF ERRN=EO THEN RETURN
3380 BEEP @ PRINT Z$
3390 PRINT "Error":ERRN:"on Line ";ERNL
3400 EO=ERRN @ RETURN
3410 ! SPOLL to Check I/O
3420 S=SPOLL(709)
3430 IF S=17 OR S=16 OR S=S0 THEN N 3470
3440 PRINT Z$ @ PRINT "Scanner P
3450 problem (SPOLL#16 or 17)"
3460 PRINT "SPOLL=";S @ BEEP
3470 S0=S @ S=SPOLL(723)
3480 IF S=0 OR S=S1 THEN 3530
3490 PRINT Z$
3500 PRINT "DVM Problem (SPOLL#0
3510 )"
3520 PRINT "SPOLL=";S @ BEEP
3530 S1=S @ RETURN
3540 END
Appendix B.  Correction Procedures for DCPD Measurements

The voltage $V$ measured between two points along a conductive object is related to the current $I$ passing between the two points and the resistance $R$ of the object by Ohm's Law:

$$V = IR = \int_{L}^{\lambda} \frac{I\rho}{A(\lambda)} d\lambda \quad (B.1)$$

where $\rho$ is the resistivity of the material, $\lambda$ is the distance along the object, and $A(\lambda)$ is the cross-sectional area at the location $\lambda$. At time 0 and at time $t$, the measured notch potential $V_n$ and the reference potential $V_{ref}$ can be defined by:

$$V_n(0) = I(0) \rho(0) G(0) \quad (B.2a)$$
$$V_n(t) = I(t) \rho(t) G(t) \quad (B.2b)$$
$$V_{ref}(0) = I(0) \rho(0) G_{ref}(0) \quad (B.3a)$$
$$V_{ref}(t) = I(t) \rho(t) G_{ref}(t) \quad (B.3b)$$

where $G$ is a geometric factor with dimensions of $1/\text{length}$, defined by:

$$G = \int_{L}^{\lambda} \frac{1}{A(\lambda)} d\lambda \quad (B.4)$$

For example, a smooth cylindrical object of length $L$ and diameter $D$ has a geometric factor defined by:

$$G = \frac{L}{A} = \frac{4L}{\pi D^2} \quad (B.5)$$

The geometric factor is affected by deformation, which elongates the object and reduces its cross section, and by cracking and internal cavitation, which further reduce its cross sectional (current carrying) area.

Potential drop probes are spot welded onto the specimen as shown in Figure B.1. The DCPD probes are located a distance $L_p$ apart, so the potential is somewhat greater than that which would be measured if the probes could be affixed at the exact edges of the notch, separated by a distance $L_n$. This measured potential $V_m$ can be separated into two components:

$$V_m = V_{\text{notch}} + V_{\text{straight section}} \quad \text{or}$$
The reference probe can be defined as:

\[ V_{\text{ref}} = I \rho \frac{L_{\text{ref}}}{A} \]  

(B.7)

Therefore, the notch potential is given by:

\[ V_n = V_m - \frac{L_p}{L_{\text{ref}}} V_{\text{ref}} \]  

(B.8)

This notch potential is further corrected for current and temperature fluctuations as described below.

The reference probe is located along a region of the specimen which presumably does not experience creep strain or damage, so

\[ G_{\text{ref}}(0) = G_{\text{ref}}(t) = G_{\text{ref}} \]  

(B.9)

The corrected notch potential \( V_c \), to be used in data analysis, is given by:

\[ V_c(t) = I(0) \rho(0) G(t) \]  

(B.10)
in order that the initial and current potentials can be compared on the basis of a true constant current and resistivity. Combining Equations B.10 and B.2b:

\[ V_c(t) = \frac{I(0)}{I(t)} \frac{\rho(0)}{\rho(t)} V_m(t) \]  

(B.11)

By substituting Equations B.3a and B.3b into B.6, one obtains:

\[ \frac{I(0)}{I(t)} \frac{\rho(0)}{\rho(t)} = \frac{V_{ref}(0)}{V_{ref}(t)} \]  

(B.12)

These last two equations are combined to obtain the correction factor:

\[ V_c(t) = \frac{V_{ref}(0)}{V_{ref}(t)} V_m(t) \]  

(B.13)

This relationship corrects for fluctuations in applied current and in resistivity (e.g., due to changes in temperature).

Figure B.2 shows a plot of the reference potential and the measured and corrected notch potentials for a typical test (Test 6, HIP U-Notch). The measured potentials (open symbols) remain close to their initial values during the entire test. The reference potential steadily decreases, indicating a decreasing applied current. When the notch potentials are corrected according to the technique described above, they are transformed into steadily increasing, nearly linear curves, which are typical of HIP specimens. Corrected potential data are not totally free from scatter, but in general the scatter is greatly reduced.
Figure B.2. Measured Notch and Reference Potentials and Corrected Notch Potentials Illustrating Effectiveness of Correction Procedure.
Appendix C. Statistical Analysis of Creep Rupture Data.

Fitting of experimental data to statistical models, and analysis of the fitting parameters, has been performed using standard, well-established procedures [140-142]. The mathematical derivations are long and beyond the scope of this volume; a brief explanation is provided below, followed by computer-calculated statistical models that fit the experimental data.

The linear regression model assumes that the dependent variable $Y$ is linearly related to the independent variable $x$ by:

$$ Y = \alpha + \beta x $$  \hspace{1cm} (C.1)

The measured values of $Y$ can be related to $x$ according to linear regression by:

$$ Y_i = a + b x_i + \varepsilon_i $$  \hspace{1cm} (C.2)

where $a$ and $b$ are estimates of the fitting parameters $\alpha$ and $\beta$ (i.e., y-intercept and slope). $\varepsilon_i$ is the error in the particular measurement, assumed to be normally distributed with a mean of 0 and a variance of $\sigma^2$. The best estimates for $\beta$, $\alpha$, and $\sigma^2$ are given by:

$$ b = \frac{n \sum (x_i Y_i) - (\sum x_i)(\sum Y_i)}{n \sum x_i^2 - (\sum x_i)^2} $$  \hspace{1cm} (C.3)

$$ a = \bar{Y} - b \bar{x} $$  \hspace{1cm} (C.4)

$$ s^2 = \frac{n-1}{n-2} (s_Y^2 - b^2 s_x^2) $$  \hspace{1cm} (C.5)

where the variances of the $x$ and $Y$ data arrays are given by:

$$ s_x^2 = \frac{1}{n-1} \left( \sum x_i^2 - n \bar{x}^2 \right) $$  \hspace{1cm} (C.6)

$$ s_Y^2 = \frac{1}{n-1} \left( \sum Y_i^2 - n \bar{Y}^2 \right) $$  \hspace{1cm} (C.7)

The correlation coefficient, $\rho$, an indicator of the "goodness-of-fit" of the statistical model, is estimated by $r$ as follows:

$$ r = \frac{s_x}{s_Y} b $$  \hspace{1cm} (C.8)

The value $r^2$ represents the fractional part of the variation in $Y$ which is accounted for by the model: for perfect correlation, $r^2=1$ or $r=\pm1$; the sign of $r$ is the same as the sign of the slope. If $r=0$, either there is no correlation, or the fitted line is horizontal.
Experimental data in this study were fitted as follows. The independent variable is applied stress, which in turn determined the dependent variables rupture life, strain at failure, and minimum creep rate. Straight lines can be fitted to the natural logarithms of these variables according to the regression formula:

$$\log (Z) = a + b \log (\sigma)$$  \hspace{1cm} (C.9)

which transforms to:

$$Z = A \sigma^b$$  \hspace{1cm} (C.10)

where $A = \exp (a)$. This is the form of most equations that relate rupture life and minimum (or steady state) creep rate to applied stress.

Separate sets of data can be compared as described below [140, 141]. To decide whether the slopes and intercepts of the lines for two populations are statistically different, one examines the following variables:

$$T_b = T_{b_1-b_2} = \frac{b_1 - b_2}{s_{b_1-b_2}}$$ \hspace{1cm} (C.11)

$$T_a = T_{a_1-a_2} = \frac{a_1 - a_2}{s_{a_1-a_2}}$$ \hspace{1cm} (C.12)

where

$$s_{b_1-b_2}^2 = s_p^2 \left[ \frac{1}{n_1 - 1} s_{x_1}^2 + \frac{1}{n_2 - 1} s_{x_2}^2 \right]$$ \hspace{1cm} (C.13)

$$s_{a_1-a_2}^2 = s_p^2 \left[ \frac{1}{n_1} + \frac{1}{n_2} + \frac{x_1^2}{n_1 - 1} s_{x_1}^2 + \frac{x_2^2}{n_2 - 1} s_{x_2}^2 \right]$$ \hspace{1cm} (C.14)

The test variables $T_b$ and $T_a$ are described by a Student's t-distribution with $n_1+n_2-4$ degrees of freedom. If the value of $T_b$ ($T_a$) does not exceed the value of the t-distribution for the desired confidence level (i.e., the confidence in a decision whether to reject a hypothesis which is actually true, often 0.95 or 0.99), then the hypothesis that the slopes $b_1$ and $b_2$ (intercepts $a_1$ and $a_2$) are equal cannot be rejected; therefore the fitting parameters are not significantly different between the two straight-line models.

The above t-tests determine whether the slopes and intercepts, independently of each other, of two models are the same. There is a more powerful method [141] which takes the
fitting parameters together to determine whether the two lines are coincident. Before describing this method, a few terms must be defined:

\[ \bar{Y} = \frac{1}{n} \sum_{i} Y_i \]  
\[ \text{Average value of } Y \]  
(C.15)

\[ Y_{Ri} = a + b x_i \]  
\[ Y \text{ predicted by regression model} \]  
(C.16)

\[ (Y_i - \bar{Y}) \]  
\[ \text{Total variation in } Y \]  

\[ (Y_{Ri} - \bar{Y}) \]  
\[ \text{Variation in } Y \text{ explained by regression} \]  

\[ (Y_i - Y_{Ri}) \]  
\[ \text{Variation in } Y \text{ due to error} \]  

In addition, the total sum of squares (SSY), the regression sum of squares (SSR), and the error (or residual) sum of squares (SSE) are given by:

\[ \text{SSY} = \sum_{i} (Y_i - \bar{Y})^2 = \sum_{i} (Y_i^2) - n\bar{Y}^2 \]  
(C.17)

\[ \text{SSR} = \sum_{i} (Y_{Ri} - \bar{Y})^2 = \sum_{i} (Y_{Ri}^2) - n\bar{Y}^2 \]  
\[ \sum_{i} (a + bx_i)^2 - n\bar{Y}^2 \]  
(C.18)

\[ \text{SSE} = \sum_{i} (Y_i - Y_{Ri})^2 = \sum_{i} (Y_i^2) - \sum_{i} Y_{Ri} Y_i \]  
\[ \sum_{i} (Y_i^2) - \sum_{i} (a + bx_i) Y_i \]  
(C.19)

\[ \text{SSY} = \text{SSR} + \text{SSE} \]  
(C.20)

The correlation coefficient \( r \) and sample variance \( s^2 \), described earlier, can alternatively be described as:

\[ r = \sqrt{\frac{\text{SSR}}{\text{SSY}}} \]  
(C.21)

\[ s^2 = \frac{\text{SSE}}{(n - 2)} \]  
(C.22)

The pooled estimates for \( \beta, \alpha, \) and \( \sigma^2 \) are:

\[ b_p = \frac{(n_1 - 1) s_{x_1}^2 b_1 + (n_2 - 1) s_{x_2}^2 b_2}{(n_1 - 1) s_{x_1}^2 + (n_2 - 1) s_{x_2}^2} \]  
(C.23)
Here, the subscripts 1 and 2 refer to the two different populations being pooled (for example, the HIP and EIF specimens). These pooled values are not the same as those calculated by taking all data points as a single population.

The following statistic determines whether the two lines should be considered coincident:

\[
F = \frac{\frac{SSR_p - SSR_T}{2}}{\frac{SSE_p}{DF_p} + \frac{(n_1 - 2)s_1^2 + (n_2 - 2)s_2^2}{n_1 + n_2 - 4}}
\]

where the subscript \( P \) refers to the pooled populations and the subscript \( T \) refers to taking the two populations together as a single population; \( DF \) means degrees of freedom. The parameter \( F \) varies as an F-distribution with \( 2 \) and \( n_1 + n_2 - 4 \) degrees of freedom. If \( F \) does not exceed the tabulated value of the standard F-distribution for the desired confidence level, the hypothesis that the two lines are coincident is accepted.

This F-test for coincidence is a more powerful test than the t-tests for slope and intercept, because, while the t-tests treat each parameter individually and independently, the F-test is a single simultaneous test. Conceivably there could be instances in which the slopes and intercepts of two fitted lines were judged identical by the t-tests, while the result of the F-test showed that the lines were nonetheless not the same. Such a discrepancy only arises in borderline cases; for most pairs of similar lines the two methods will agree.

These statistical techniques provide the best estimations for fitting parameters when the number of data points exceeds about 25. The judgments about the coincidence of two fitted lines also are more powerful (more reliable) when larger sample populations are involved. The calculations based on the experimental data in this thesis are based on relatively small populations, so any conclusions regarding the differences in HIP and EIF material behavior must be considered with this in mind.

Table C.1 gives calculated equations for creep behavior for HIP and EIF U-notch specimens. According to the t and F tests described above, slight differences in the behavior of HIP and EIF specimens are not statistically significant; there are no significant differences
between the equations calculated using only the failed notch data and those calculated using both failed and unfailed notch data. Table C.2 shows equations calculated for HIP and EIF smooth bars; obviously the HIP and EIF treatments behave differently, so no statistical tests were carried out.

Table C.1. Calculated Creep Parameters: U-Notch Specimens.

<table>
<thead>
<tr>
<th></th>
<th>Failed Notch Only</th>
<th>Failed and Unfailed Notches</th>
</tr>
</thead>
<tbody>
<tr>
<td>HIP</td>
<td>( t_R = 4.43 \times 10^{15} \sigma^{-6.59} )</td>
<td>( t_R = 5.26 \times 10^{15} \sigma^{-6.61} )</td>
</tr>
<tr>
<td></td>
<td>( R^2 = 0.985 )</td>
<td>( R^2 = 0.974 )</td>
</tr>
<tr>
<td>EIF</td>
<td>( t_R = 4.94 \times 10^{13} \sigma^{-5.62} )</td>
<td>( t_R = 4.79 \times 10^{13} \sigma^{-5.60} )</td>
</tr>
<tr>
<td></td>
<td>( R^2 = 0.963 )</td>
<td>( R^2 = 0.964 )</td>
</tr>
<tr>
<td>All</td>
<td>( t_R = 1.93 \times 10^{14} \sigma^{-5.92} )</td>
<td>( t_R = 2.73 \times 10^{14} \sigma^{-5.98} )</td>
</tr>
<tr>
<td></td>
<td>( R^2 = 0.951 )</td>
<td>( R^2 = 0.951 )</td>
</tr>
<tr>
<td>HIP</td>
<td>( m_c = 1.17 \times 10^{-20} \sigma^{-6.05} )</td>
<td>( m_c = 9.83 \times 10^{-21} \sigma^{-6.07} )</td>
</tr>
<tr>
<td></td>
<td>( R^2 = 0.988 )</td>
<td>( R^2 = 0.974 )</td>
</tr>
<tr>
<td>EIF</td>
<td>( m_c = 7.11 \times 10^{-21} \sigma^{-6.19} )</td>
<td>( m_c = 9.50 \times 10^{-21} \sigma^{-6.12} )</td>
</tr>
<tr>
<td></td>
<td>( R^2 = 0.916 )</td>
<td>( R^2 = 0.913 )</td>
</tr>
<tr>
<td>All</td>
<td>( m_c = 5.80 \times 10^{-21} \sigma^{-6.21} )</td>
<td>( m_c = 6.01 \times 10^{-21} \sigma^{-6.19} )</td>
</tr>
<tr>
<td></td>
<td>( R^2 = 0.941 )</td>
<td>( R^2 = 0.936 )</td>
</tr>
<tr>
<td>HIP</td>
<td>( t_R(m_c)^{1.088} = 8.712 \times 10^{-7} )</td>
<td>( t_R(m_c)^{1.080} = 9.930 \times 10^{-7} )</td>
</tr>
<tr>
<td></td>
<td>( R^2 = 0.996 )</td>
<td>( R^2 = 0.990 )</td>
</tr>
<tr>
<td>EIF</td>
<td>( t_R(m_c)^{0.878} = 4.136 \times 10^{-5} )</td>
<td>( t_R(m_c)^{0.887} = 3.506 \times 10^{-5} )</td>
</tr>
<tr>
<td></td>
<td>( R^2 = 0.983 )</td>
<td>( R^2 = 0.987 )</td>
</tr>
<tr>
<td>All</td>
<td>( t_R(m_c)^{0.835} = 6.939 \times 10^{-5} )</td>
<td>( t_R(m_c)^{0.927} = 1.535 \times 10^{-5} )</td>
</tr>
<tr>
<td></td>
<td>( R^2 = 0.777 )</td>
<td>( R^2 = 0.936 )</td>
</tr>
</tbody>
</table>

Table C.2. Calculated Creep Parameters: Smooth Bar Specimens.

<p>| | | |</p>
<table>
<thead>
<tr>
<th></th>
<th></th>
<th></th>
</tr>
</thead>
<tbody>
<tr>
<td>HIP</td>
<td>( m_c = 2.995 \times 10^{-32} \sigma^{12.0} )</td>
<td>( R^2 = 0.9999 )</td>
</tr>
<tr>
<td>EIF</td>
<td>( m_c = 7.047 \times 10^{-20} \sigma^{20.0} )</td>
<td>( R^2 = 0.821 )</td>
</tr>
<tr>
<td>HIP</td>
<td>( t_R(m_c)^{1.095} = 1.134 \times 10^{-6} )</td>
<td>( R^2 = 0.893 )</td>
</tr>
<tr>
<td>EIF</td>
<td>( t_R(m_c)^{0.599} = 2.474 \times 10^{-3} )</td>
<td>( R^2 = 0.984 )</td>
</tr>
</tbody>
</table>
Appendix D. Computer Modelling of Notched Bar Behavior.

A first order approximation of the creep behavior of a notched specimen can be made using simple computer calculations, based on the work of Cailletaud, Policella, and Baudin [123]. This rough model is based on simple assumptions, but it gives adequate predictions of potential drop behavior based on uniform creep deformation. The difference between experimental data and these predicted curves will indicate approximately the contribution of non-strain damage (i.e., cavitation and cracking) to potential drop measurements.

![Figure D.1. Notch Geometry (left) and Sectioning into Individual Disks (right).](image)

The measurements of the notch, illustrated in Figure D.1, are entered into the computer. The computer program slices the notch into a predetermined number of thin disks, where the $i^{th}$ disk has thickness $L_i$ and diameter $D_i$. Only stresses in the axial direction are considered; the triaxiality of the notch is ignored. There is no attempt to account for compatibility of strain between adjacent slices. The net section stress on the $i^{th}$ disk is defined as:

$$
\sigma_i = \frac{P A_i}{A_i A_i} = A_i \frac{A_i}{A_i} = \frac{D_i^2}{D_i^2} = \frac{\sigma_i}{D_i^2}
$$

where $P$ is the axial load on the specimen, $\sigma_i$ is the net section stress across the notch throat, $A_i$ and $A_i$ are the cross sectional areas of the notch throat and the $i^{th}$ disk, and $D_i = 2R_i$ is the diameter of the notch throat. The strain in each slice is assumed to be only a function of the net section stress in that slice; the multiaxial state of stress arising from the notch triaxiality is neglected. According to Norton’s Law, the strain rate in the $i^{th}$ disk is:

$$
\frac{\text{de}_i}{\text{dt}} = B \sigma_i^n
$$
where \( \varepsilon_i \) is the longitudinal strain in each disk, defined in terms of \( A_o, A_i, D_o, \) and \( D_i \) (the initial and present cross sectional areas and diameters of the \( \text{ith} \) disk) as:

\[
\varepsilon_i = \log \frac{L_i}{L_o} = \log \frac{A_i}{A_o} = \log \frac{D_i^2}{D_o^2} \quad \text{or} \quad \varepsilon_i = -\varepsilon_A = -2\varepsilon_D
\]  

(D.3a)

where \( \varepsilon_A \) is the areal strain (reduction in area) and \( \varepsilon_D \) is the diametral strain. Integrating Norton’s Law, the strain accumulated during time \( t \) in the notch throat is given by:

\[
\varepsilon_i = -\frac{1}{n} \log [1 - nB_o^i t]
\]  

(D.4a)

and the strain in the \( \text{ith} \) disk is:

\[
\varepsilon_i = -\frac{1}{n} \log [1 - nB_o^i t]
\]  

(D.4b)

For small values of strain, \( \varepsilon_i \) is related to \( \varepsilon_t \) by:

\[
\varepsilon_i = \varepsilon_i \frac{\sigma_i^n}{\sigma_t^n} = \varepsilon_i = \varepsilon_i = \varepsilon_i \frac{D_i^{2n}}{D_i^{2n}}
\]  

(D.5)

The total length of the notch after time \( t \) equals the sum of the thicknesses of the strained disks:

\[
L = \sum_i L_i \exp (\varepsilon_i)
\]  

(D.6)

For the \( \text{ith} \) disk, which has cross-sectional area \( A_i \) and length \( L_i \), the increment in potential drop signal due to an applied current \( I \) and material resistivity \( \rho \) is given as:

\[
dV_i^o = Ip \frac{L_i^o}{A_i} \quad \text{before strain accumulation}
\]

\[
dV_i = Ip \frac{L_i \exp (\varepsilon_i)}{A_i} = Ip \frac{L_i \exp (\varepsilon_i)}{A_i \exp (-\varepsilon_i)} = dV_i^o \exp (2\varepsilon_i)
\]  

when the strain has reached \( \varepsilon_i \). The superscript “\( o \)” refers to the initial state. The calculated initial and present values of potential drop are:
\[ V_0 = \sum_i dV_i^0 = I \rho \sum_i \frac{L_i^0}{A_i^0} \]  
\[ V = \sum_i dV_i = I \rho \sum_i \frac{L_i}{A_i} = I \rho \sum_i \frac{L_i^0}{A_i^0} \exp(2\varepsilon_i) \]

For the smooth bar geometry, Equation D.9b simplifies into \( V = V_0 \exp(2\varepsilon) \).

The computer predicts potential drop for two cases. First, for a given strain at the notch throat, the corresponding potential drop in the notch is calculated. Increments of strain and potential drop for each disk can be tabulated for a given value of strain, as in Table D.1, or the notch potential can be calculated for a series of strains, as in Table D.2. Second, for a given time, the strain and potential drop can be determined from Norton’s Law. An entire test can be reconstructed by taking increments of time from \( t=0 \) to \( t=t_R \), as shown in Table D.3.

For small values of strain, potential drop is directly related to notch throat strain by:
\[ \varepsilon_i = \kappa \log \frac{V}{V_0} \]  
For a smooth bar specimen, \( \kappa=0.5 \); for a notched bar, \( \kappa \) is a function of notch geometry and of stress exponent \( n \). Table D.4 and Figure D.2 present \( \kappa \) and \( 1/\kappa \) for a wide range of \( n \). For the values of \( n \) found for René 95 U-notch specimens (\( n=6 \)), \( \kappa \) is about 1.075. This value is used to convert minimum potential drop increase rates into minimum creep strain rates in U-notch specimens.

Plots of predicted potential drop vs. time/time to rupture are given in Figure D.3 for U-notch tests and in Figure D.4 for smooth bar tests; power law creep parameters used for the calculations are taken from Chapter 6.
Figure D.2. $\kappa$ and $\frac{1}{\kappa}$ for $n=0$ to 20, where $\varepsilon = \kappa \log \frac{V}{V_0}$

Figure D.3. Predicted Potential Drop Signal vs. Time/Time to Rupture for HIP and EIF U-Notch Specimens.
Figure D.4. Calculated Potential Drop Signal vs. Time/Time to Rupture for Smooth Bar Specimens: (a) HIP, (b) EIF, (c) HIP and EIF.
Table D.1. Notched Specimen: Calculated Incremental Strain and Potential Drop for Each Slice of Notch

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<tr>
<th>slice</th>
<th>X</th>
<th>R(X)</th>
<th>dV/dVo</th>
<th>dL/dX</th>
<th>Ec(X)</th>
</tr>
</thead>
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<td>0.11251</td>
<td>1.00200</td>
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<td>0.00100</td>
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</table>

Rn= 0.0760  Do= 0.2250  D/Do= 0.99950
Rt= 0.1125  Ao= 0.03975  A/Ao= 0.99900
Rs= 0.1875  Lo= 0.1744  L/Lo= 1.00038
H=25  I=10  Vo=10.3158  V/Vo= 1.00033
n= 6.00  Ec= 0.0010  ln(V/Vo)= 0.30093
Table D.2. Notched Specimen: Calculated Notch Potential as a Function of Strain

<table>
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<tr>
<th>Ec</th>
<th>D/Do</th>
<th>A/Ao</th>
<th>L/Lo</th>
<th>V/Vo</th>
<th>In(V/Vo)</th>
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Table D.3. Computer Simulated Notched Creep Rupture Test

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<th>Rt</th>
<th>R5</th>
<th>Ho</th>
<th>Lo</th>
<th>Vo</th>
<th>file: not stored</th>
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</table>

\[ \text{t-rupt} = 70.000 \text{ hr} \]
\[ \text{delta-t} = 0.700 \text{ hr} \]
\[ \text{Stress} = 130.00 \text{ ksi} \]

\[ \frac{dE}{dt} = 1.0000 \times 10^{-20} (\text{stress})^6.00 \]

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<th>t (hr)</th>
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<th>V/Vo</th>
<th>ln(V/Vo)</th>
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Table D.4. U-Notch: Strain-Potential Drop Calibration

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\[ Ec = K \ln(V/V₀) \]
References


127 T. Piness, private communication (December 1985).


