Dwell Fatigue, I: Damage Mechanisms

Tarun Goswami

Cessna Aircraft Company, Materials and Processes #178 MS 1
2617 South, Hoover Road, Wichita, KS 67215-1200
Adjunct Associate Professor, Department of Mechanical Engineering,
Wichita State University, Wichita, KS 67260-0035, USA

ABSTRACT

The mechanisms controlling deformation and failure under high temperature creep-fatigue conditions of materials are examined in this paper. The materials studied are solder alloys, copper alloys, low alloy steels, stainless steels, titanium alloys, and Ni-based alloys. The deformation and failure mechanisms were different (fatigue, creep, oxidation and their interactions) depending upon test and material parameters employed. Deformation mechanisms, such as cavity formation, grain boundary damage, intergranular (IG) and transgranular (TG) damage, oxidation, internal damage, dislocation cell concentration, and oxide mechanisms are very important in order to gain more knowledge of fatigue behavior of materials. The observed mechanisms can be categorized as follows:

- Depending upon the test temperature, higher NCR resulted with higher strain range, dwell time and lower strain rates. The damage was due to creep-fatigue interaction by mixed (TG) and/or (IG) creep damage by cavity formation, and oxidation striated surface damage. Oxidation damage was found to depend upon a critical temperature and compression and tension dwell periods in a cycle.
- Dwell sensitivity was effective only below a certain strain range; once this threshold was exceeded NCR value was not affected by further increase in dwell time.
- Microstructure changed depending upon test temperature, dwell period, and strain range. Triple point cracking favored mechanisms such as cavitation. New metal precipitates formed depending upon the temperature, strain range and dwell time. Some precipitates were beneficial in blocking the grain boundary damage from creep, whereas other precipitates changed the dislocation sub- structure, promoting more damage. New cells formed during tests that coarsened with longer dwell times. In some cases dynamic strain aging occurred enhancing fatigue behavior of materials.
- Depleted regions developed due to high temperature exposure, which was a function of dwell time applied in a cycle, and material composition that aided in the formation and/or propagation of (IG) cracks.
- Dwell cycles evolved mean stresses in tension and compression directions. Mean stresses in tension were more deleterious and caused dwell sensitivity.
- Dwell sensitivity was also dependent on material conditions, and discontinuities present in a material. These parameters together with test parameters produce damage interactions in a particular fashion that evolve different micro-mechanisms.

The dwell sensitivity micro-mechanisms are summarized in this paper.

Keywords: dwell sensitivity, tensile dwell, compressive dwell, transgranular (TG), intergranular (IG), internal damage, creep damage, and oxidation.

INTRODUCTION

The operating conditions of many engineering components are known to be complex. In such applications the service loads change with respect to time at the startup and shutdown conditions. While a major portion of the operation requires the loads to be held at near steady conditions for a period of time or dwell time, frequent start-stop cycles cause thermal transients and failure in low cycle fatigue mode. The period of time during which the loads are held near steady conditions (or constant) is known as a dwell time. Creep may occur during the steady loading conditions while the startup and shutdown cycles produce the fatigue damage. Therefore, dwell sensitivity is a material behavior in which the fatigue resistance of materials reduces compared to non-dwell, continuous fatigue response. Damage mechanisms under dwell sensitivity conditions are discussed in this paper. Dwell sensitivity studies may play an important role in material selection and design of such critical equipment.

In an earlier paper, denoted by I here /1/, a list of the materials studied was provided together with high temperature low cycle fatigue (HTLCF) behavior, and dwell sensitivity maps. An attempt was made in this paper to review and summarize the damage mechanisms under various dwell and non- dwell conditions for all the materials studied in I. No such elaborate attempt has been made in the literature to summarize dwell sensitivity damage mechanisms for these materials. Part two of this paper, Paper II, which will present stress evolution as a result of various dwell cycles, is in preparation.

Dwell times are applied in tension, compression, or in both directions (see Fig. 1, I) and are called tensile dwell (t/0), compression dwell (0/t), unbalanced dwell (t_1/t_2) and balanced dwell (t/t) cycles. The degradation resulting from applied dwell cycles in a particular direction is not very well understood /2-6/. The dwell sensitivity in forged titanium alloys caused disk failures in commercial aircraft engines in the early 1970s. Therefore, titanium alloys in forged conditions were investigated more thoroughly in the literature than other materials, for example solder alloys. Deformation mechanisms are a result of creep-fatigue and environmental interactions and are discussed in this paper so that common features, if they exist, can be identified under similar dwell conditions for different materials.

DAMAGE MECHANISMS

The materials examined in I belonged to the following categories: solder alloys, copper alloys, low alloy steels, stainless steels, titanium alloys, and superalloys. The details of the tests, mechanical properties, HTLCF dwell data, and dwell sensitivity maps were summarized in Tables 1-4 in I. Associated mechanistic features under different test waveforms are discussed below.

Solder Alloys

1. 96.5% Pb-3.5%Sn

Solder joints are susceptible to start-up and shut-down cycles. Thermal transients (thermal fatigue situation) arise from frequent start-stop cycles. A dwell cycle occurs as the solder joint allows passage of electrical current between a start-stop cycle. Fatigue results from the mechanical support that the joint provides to a surface mounted component soldered on to the printed circuit board. Therefore, solder joints fail in low cycle fatigue (LCF) mode. Deformation mechanisms under these service conditions are very important in order to predict the life of a solder joint.

Limited LCF tests on the bulk solder materials were reported in the literature. The cyclic life of solder alloys, in general, depends on the test frequency, stress or strain range, hold time, and temperature /7/. Wild (1974) reported the fatigue behavior of solder alloys. A summary of that study is presented here. As the frequency increased from 1.1×10^{-3} to 8.3×10^{-2} Hz, the fatigue life increased for the Sn/Pb alloys, as opposed to the following alloy types, namely: 50Sn50In, 42Sn58Bi and 50Pb50In. The effect of grain size on the thermal-mechanical fatigue response of solder alloys was not determined. This, however, is an important parameter, as the grain size changes with aging. Aging occurs as a result of either dwell times or with an increase in non-dwell test time (time between stop to start cycle or associated with lag time between successive assembly stages) in the case of Sn/Pb alloys.

Once the test temperature exceeds over one-half the absolute melting temperature, a reduction in frequency has been found to decrease the number of cycles to failure in isothermal fatigue. This behavior is due to the creep deformation, cavity formation and growth, and environmental attack. A strong frequency influence was observed at intermediate levels /8,9/. Frequency, in general, means number of cycles per second (in Hz); for the solders it was assumed as the reciprocal of twice the ramp time, thus, fixing frequency as strain rate variable. The duration of a cycle is a function of two variables: ramp-time and dwell-time. These two strain rate or time variables produce profoundly different results on the fatigue life of solder alloys at a given strain rate. Also, the fatigue life of a joint tested in tension is nearly 100 times less than that tested in shear /7/. A tensile strain range of 0.577% therefore corresponds to the same damage produced by 1.0 percent shear strain range.

For the solder stated above /8/, as the frequency reduced, damage intensified in the grain boundaries since time per cycle during fatigue increased. A mixture of TG and IG cracking was observed, which in turn depends upon the strain rate at which tests were performed. Grain boundary cracking observed was a result of increased environmental effects. This may be the main damage mechanism for dwell cycles. At very low frequency, the damage mode changed from mixed TG – IG to predominantly IG damage. The TG mode of cracking was due to lower temperatures and higher frequencies and can be in either Stage I or II, depending upon the nature of deformation at the crack tip. A plastic blunting process /10/ or accumulation of damage at the crack tip by micro low cycle fatigue process /11/ were assumed to be possible controlling mechanisms.

Solder alloys, like many other alloys found in I (for example, copper based alloys; AMZRIC and NARloy-Z, low alloy steels: 1Cr-Mo-V, 1.25Cr-Mo and 9Cr-1Mo steels, austenitic stainless steels: SS304, SS304L, SS316 and SS316L and other derivatives, tantalum alloys: ASTAR-811 and T-111, and superalloys: MAR M002, Rene 80, Inconel 617, IN 100, PWA 1480 and MA 754) are tensile dwell sensitive. As the dwell time in the tension direction increases, the life decreases. However, compressive dwell sensitivity was documented for 2.25Cr-Mo steel, Ti-6Al-4V, and IMI 829, superalloys Waspaloy, Rene 95 and MAR M 002 below a certain temperature. The number of cycles to failure with increasing dwell time, in both directions, was observed to saturate (higher NCRs) for many materials as shown in I. Since this study covers dwell sensitivity under creep-fatigue conditions, it may be difficult to isolate dwell sensitivity and temperature sensitivity. With a small increase in temperature, the homologous temperature increases considerably for solder alloys, where conventional materials, e.g., steels and titanium alloys, do not operate. Other phenomena such as oxidation may play a role in the degradation of materials in such scenarios. Oxidation was also observed in solder alloys occurring as a result of tensile dwell cycles. Often, the reduction in fatigue life was attributed to the transition from the failure mode from TG to cavitated IG mode. Lower and higher magnified micrographs (see /12/, Fig. 8 (a, b)) show typical growth of ductile holes and IG facets respectively seen within the holes. This has been interpreted in terms of less secondary grain growth and precipitate coarsening. It may be noted that some solder alloys contain regions of rich Pb and/or Sn zones and that the microstructure varies for two bulk solder alloys, and for a solder joint (described in /13/, Fig. 1 (a, b, c)). These regions, also due to aging, produce features such as dimples and facets and respective IG damage. In application, failure features do not show the striations, but IG failures /14/. In a Cu-clad-invar system, multiple crack origins for 63Sn/37Pb solder joint (Fig. 12) were documented /14/. Cracks seen on several locations were the regions where increased spheroidization and particle growth occurred. The crack origins appeared to be associated with voids, and/or the interfaces between the large areas of Pb-rich and Sn-rich phases (Fig. 13 (a, b)).

For 37Pb-63Sn /15/ solder, surface cracks were documented for up to 3.5% total strain range under axial fatigue tests. The cracks were noticed to align with the slip bands at about 45° to the axis of the applied stress (Stage I cracking). The number of surface fatigue cracks, or secondary cracks, reduced as the strain range decreased. In those cases, slip bands were much finer than at higher strain ranges. Discontinuities in the material accelerated the early nucleation of fatigue cracks. These fatigue cracks propagated in the material and at some point a more dominating crack developed whose growth culminated in the final failure of the specimen. During this process, cracks coalesce with material discontinuities and other imperfections. Radial shear ridges observed on the specimens indicate several initial sites and final failure was a result of crack growth from these sites with no striations (Figs. 9-11, /16/). The fracture path was predominantly TG through the tin-rich matrix. Less lead-rich matrix was involved in the deformation process.

This solder was found to be a softening type with the application of fatigue cycles. Stress lowered as the cycles progressed. A decrease in load level may result from several factors such as a change in microstructure resulting from aging /7,17/. Such a change in microstructure does not alter the load bearing area of the specimen, whereas the cracks alter the load bearing area of the specimen associated with fatigue cracking. Since there were multiple crack nucleation sites, it was not possible to correlate number of secondary cracks with the load change. Figure 2 /6/ plots the data and shows that these mechanisms are responsible for reducing the NCR values for tensile dwell cycles. Compressive dwell cycles ideally reverse the slip systems, causing a beneficial effect in the fatigue life and NCR values improved at least four times at the same strain range and dwell times.

Near eutetic Sn-Pb solders, the kind discussed above, have distinct areas of separated Pb and Sn rich zones and cracking through the Sn-rich matrix was documented to be the mechanism for deformation and fracture /18,19,15/. Fine grained, near eutectic, solders under high strain rate fatigue undergo an initial stage of coarsening mechanism /15,18/ where the cracks formed propagated through the regions of coarsened microstructure. However, coarsening did not occur in the case of tests at lower strain range. Also, coarsening did not occur for the solders with coarse microstructure /19/.

For solders Sn-Pb and Sn-Pb near eutectic systems, tensile dwells were more damaging; their mechanisms were discussed earlier. In those cases, dwell times caused the most damage early on, followed by a phase in which increased dwell time did not affect the life. A number of materials (e.g., 9Cr-1Mo) behave the same way – discussed in I and elsewhere /6/. Stage I cracking, new precipitation, aging, coarsening of microstructure, dislocation systems by transmission electron microscopy (TEM) are not investigated for the leaded solders. More research needs to be undertaken to document these features for lead-free solders.

Copper Alloys

1. AMZIRC and NARloy-Z tested at 538°C

Copper alloys AMZIRC and NARloy-Z possess low strength and high ductility. It was pointed out (I, Fig. 2-3) that the plastic strain components were much higher than the elastic strain components for these two alloys. Therefore, plastic strain accumulation constitutes a major component of LCF life. The damage mechanisms for AMZIRC, a copper-base alloy /20/, were not investigated in detail, whereas Shih /21/ reported the microstructure (Figs. 29, 30 (a, b)), scanning and transmission electron microscopic results (Figs. 32-47) for NARloy-Z. Since both materials exhibit similar mechanical properties and have been tested with very high strain components, failure mechanisms in one are expected to be similar in the other. A summary of damage mechanisms is presented in this paper for NARloy-Z that may be extended to AMZIRC.

Several mechanisms were reported for the frequency effects and dwell times. The mechanisms controlling deformation and failure for NARloy-Z were due to creep, environmental interactions (oxidation), localization of damage at the triple points, and transition from TG to IG cracking. A combination of strain rates, frequency, strain range and dwell time causes surface damage by IG mode. This, however, within a distance of three to four grain size (ASTM No. 3 and 150 μ) transitions to TG crack growth and striations were seen on the fracture surface. However, other combinations of test parameters such as higher strain range (5%), tensile dwell, lower strain rates (see tests 20) promote IG damage, resulting from grain boundary sliding or creep. During these combinations of test parameters the interactions among creep, environment, internal cracking and cavitation produce damage at multiple sites resulting in their link-up. A lower life occurs in those cycles than compressive dwells. For compressive dwell cycles normal TG cracks were noticed. Compressive dwells at lower strain ranges with higher strain rates produced sub-grains. Sub-grain formation may have raised the fatigue resistance and dwell sensitivity deformation mechanisms were TG cracking dominated by striations. The life under compressive dwell and continuous fatigue cycles was similar at lower strain rates. The fracture surfaces for these two conditions were quite similar as well, resulting in higher normalized cycle ratios at higher strain ranges than at lower strains with the same dwell periods. Tensile dwells produced lower NCR values.

The three modes in which damage occurred were as follows:

1) Creep effects dominated in the case of slow-fast cycles in which strain rates are low in tension but faster in compression return half-cycle.
2) Cyclic strain damage became more important for the fast tensile strain rates (fast-slow tests). Slow-fast tests were more damaging than fast-slow tests.
3) Environmental interactions or oxidation damage may play a part depending upon the test temperature and dwell time applied. Internal damage may interact and link up with IG crack or with surface damage, documented by /21/.

The extent of damage produced during a cycle was also in the same order under the three modes. Creep damage caused a larger reduction in life than either cyclic strain or environmental damage.

A varied substructure was documented in different grains for the tests with low strain range, low rate of cycling, and tensile dwells (/21/, Fig. 47 (a-b)), which shows random distribution of dislocations. Dislocations were pinned as a result of silver precipitation that had coarsened during dwell time tests. Sub-grain formation and growth have also been documented for this material for some grains, which was due to different grain orientation and produced different shear stresses on active slip planes. Accordingly, less favorable or low shear stresses on the active slip planes did not enable the dislocation to overcome the precipitate barriers except by thermally activated cross slip and/or climb mechanisms. As a result dislocations were arranged into a low energy configuration like sub-grain boundaries. This arrangement depended upon the time and applied stress/strains. The former produces cross slip and may influence relative orientation whereas the latter may control the sub-grain size. Correlation between re-crystallization and different dwell conditions has not been investigated, which is important for studies on damage mechanisms.

Low Alloy Steels

1. 1Cr-Mo-V Steel

Dwell fatigue data of low alloy steels were compiled /22/ in a tabular format from where trends in the creep-fatigue behavior of low alloy steels were determined. The dwell cycles were tensile, compressive, balanced and unbalanced cycles. In one study /23/, different dwell cycles discussed above were applied. The damage mechanisms were summarized by Plumbridge *et al.* /24/ (Figs. 3-7, 10), /25/ (Figs. 1-6). This author had access to the data and metallographic samples and re-examined them. The damage mechanisms showed the distinction of creep and fatigue damages with dwell times at 565°C. Continuous fatigue cycles produced no evidence of grain boundary sliding, IG damage or cavitation at all strain ranges (0.55 go 2%). However, when a dwell time was applied in the tension direction the grain boundary damage was exhibited. Cavities formed depending on the test parameters such as strain range, strain rate, dwell period, and temperature. When the same dwell period was also applied in the compression direction (balanced dwell cycles), the grain boundary damage and creep-cavitation reduced considerably. This enhanced the NCR values for balanced dwell cycles (I, Fig. 4). In an unbalanced dwell cycle when a fraction of tensile dwell time was applied in the compression direction it reduced the damage, IG cracks and their numbers. For a tensile dwell cycle the NCR was much lower compared to an unbalanced dwell cycle. Oxidation damage was also observed to take place with creep and fatigue damage. It may not be possible to comment on the interactions of the three damage mechanisms, and their individual contributions.

Thermal-mechanical fatigue tests /26/ showed that as the temperature, dwell time and the strain range increased, the dwell effects caused lower NCR. The tests started in compression were found to be less damaging than the tests in the tension direction /27,28/. Similar observations made for stainless steel and copper alloys are discussed in this paper.

2. 1.25Cr-Mo Steel

Micro-cracks were documented /29/ for continuous fatigue tests between 0.7 to 1% total strain range. However, no micro-cracks occurred at 0.5% total strain range or under for continuous fatigue and dwell fatigue cycles with 1 and 10 minutes tensile dwell times. The external surfaces of all the specimens were contaminated with oxidation damage, forming a thick layer of oxides. The thickness of this oxide layer increased as dwell periods increased, for example, 10 minutes. Circumferential cracks were found to generate on the oxides. However, the depth of these cracks was found to be very small, within the oxides and not into the matrix. Circumferential cracks forming from the oxide layers were also reported for 1Cr-Mo-V steel by Ellison *et al.* /23/, who found that as the time of dwell in tension increased, internal grain boundary cracks developed. The longer the dwell period was in tension, the higher the number of internal and external cracks generated. Longer dwell periods generated cavitation damage for this material, which was the case for other low alloy steels, titanium alloys and copper alloys. Other features were similar to those observed for 1Cr-Mo-V steel.

3. 2.25Cr-Mo Steel

Several conditions are used widely in the power plants, namely: annealed (A) condition, in which the microstructure is primarily of the ferritic type, normalized and temper condition (N&T), in which the microstructure is tempered bainitic, and quenched and tempered (Q&T) condition, which contains a martensitic type of microstructure. Details of the tests, material parameters, and HTLCF behaviors were presented elsewhere /22/. As the microstructure changed, the dwell sensitivity of this particular material also changed. Among other test and material parameters; microstructure, waveforms, strain- rates, environment (e.g., oxidation) and low temperature-low stress creep were found to be major contributors to the damage development. The damage produced by the same dwell time, temperature and strain rate was different for different conditions. A compressive dwell was found to cause more damage in A and N&T conditions, whereas tensile dwell was more damaging for Q&T condition. In the former conditions, oxidation interaction with fatigue and creep damage resulted in striated surface damage and lowered the life, whereas in the latter case cavitation damage was a candidate mechanism occurring within the temperature range of 540-565°C at low stress levels. Generally, with higher material strength, a transition from oxidation-fatigue-creep dominated damage to creep-cavitation damage was observed.

A threshold temperature was identified /22/, above which the oxidation damage was found to occur. It was observed to be in a range of 250-450°C. Brinkman /30/ applied a dwell in compression direction and found that oxidation damage was enhanced. This was also accompanied by increased mean stress and dimples were observed. The density of dimples depends upon the test parameters where slower strain rates and longer dwell periods promote dimples /31/. Numerous other studies /32-34/ showed oxidation damage.

A few TEM studies were made on the dislocation structures (/34/, Fig. 9) under different temperatures. For all the cases considerable disorientation of cells was noticed. Within a temperature range of 450°C average cell size was approximately 0.7 micrometer. However, as the temperature increased to 550°C considerable coarsening took place and average cell size doubled. The density of carbide particles increased at higher temperatures. Some of these studies are also linked with the chemical composition such as the present of P weight percent in the composition (Janovec *et al.* /35/). Additional dwell sensitivity damage mechanisms were reported for 9Cr-1Mo steel in hot rolled and forged conditions. A summary has been prepared describing dwell sensitivity damage mechanisms for 9Cr-1Mo steel that will be useful in understanding the damage mechanisms for other low alloy steels.

Hot rolled and hot forged 9Cr-1Mo steels were studied and damage mechanisms reported /36,37/. Hot rolled material exhibited a longer life than hot forged material. Tests in lab-air and vacuum indicated a clear pronounced environmental effect. The tensile properties (yield strengths) were higher for the hot rolled material than the hot forged material. As a result, the ratio of elastic and total strain range at the same strain range of 0.5% was higher, 0.64 for the hot forged material, than the hot rolled material where it was 0.5. The TEM studies /36/ on this material indicated that the microstructure consisted primarily of packets of parallel laths containing a high density of dislocations. The width of the laths varied almost 50% for the two conditions, which was wider in the case of hot forged material. A larger precipitate ($M_{23}C_6$) was found to be oriented along the prior austenite grain boundaries and lath boundaries. This material was found to be of a softening type, interpreted in terms of combination of dislocation annihilation and rearrangement, and perhaps also in part as a result of a reduction of strength as Mo-C-Mo clusters transfer to Mo-C pairs. At the initial stages of HTLCF tests, a gradual drop in strength may be due to the coarsening of precipitates.

In another study /37, examinations made by TEM found the following:

1) Under normalized and tempered condition, fatigue life was reduced drastically as a function of tensile dwell time for up to 1 hour. Little additional reduction in life occurred for longer dwells (1-3 hr). This was due to softening effects of cyclic plasticity, diminishing with longer dwells. Time of thermal aging reduced the HTLCF properties of this material.
2) Environmental interactions with creep and fatigue reduced the incubation period for crack nucleation; however, they accelerated the rate of crack propagation. At lower temperatures (538°C), no creep cavitation was observed. However, as the temperature increased (593°C), cavitation was documented.
3) Cyclic lives for a 30min tensile dwell cycle were the same in air and vacuum. This was due to creep dominated IG cracking occurring at 593°C.
4) Microstructure consisted of similar features noted in the former study.
5) In addition, for aged materials, recovered regions were identified suggesting localized microstructural instability. New recrystallized grains (dislocation free) formed adjacent to prior austenite grain boundaries.

In the case of Cr-Mo steels, Mo has been found to create Mo-C clusters /38/, which impart interaction solid solution hardening. Since these clusters are less mobile, during dynamic strain aging they enhance strengths and perhaps fatigue resistance. Such a study was reported for 1Cr-Mo-V steel /39/, 2.25 Cr-Mo steels /34,38/ and recently for 9Cr-1Mo steels. Since dynamic strain aging may be beneficial from the fatigue resistance standpoint, it restricts the spread of plasticity during Stage I fatigue crack growth. In the case of 2.25 Cr-Mo steels, the spread of strain – life curve appears to be highest at around 427°C; it was nearly 50°C higher where peak ultimate tensile strength was reported. Also, within this band of test temperatures, 250-450°C, oxidation was documented to occur in 2.25Cr- Mo steels depending upon the dwell periods. This phenomenon may indicate the beneficial effect dynamic strain aging has on the creep-fatigue response within 250-450°C temperature ranges. It was speculated that competition between the oxidation and dynamic strain aging results in the domination of one mechanism over another and oxidation has been found to accelerate the growth of damage and cause a reduction in life. These mechanisms take place during the dwell time tests and, as the tests progress, the time at a particular temperature and stress increases, which increases the thickness of the deposited oxide layers. The oxides have been seen to crack on the surface and create striated surface from where fatigue cracks develop. In the case of 1Cr-Mo-V steels tensile dwells develop mean tensile stresses which accrue more strains per cycle, and a reduction in life occurs. The opposite was found for 2.25Cr-Mo steels, where compressive dwell cycles developed tensile mean stresses. Such interactions, competition mechanisms, among various dwell conditions are not very well understood. The IG damage and cavitation with longer tensile dwells at or above 593°C were believed to be the reason for lower NCRs in those cases.

Stainless Steels

1. Stainless Steel SS 316

Triangular waveforms produced the longest lives at higher strain rates (6.7×10^{-3} /s) /40/. The life decreased as the strain rates decreased and dwell time in tension increased. Other waveforms were slow-fast, saw-tooth and fast-slow, where the former produced lower lives than the latter. A reduction in life was greater at 600°C than at 700°C. The slow-slow, triangular waveforms, tensile dwell, and slow-fast saw-tooth tests failed by IG fracture (see Fig. 7). Other wave-shapes (see /40/ for wave shapes) by TG mode showed well-defined striations. Extensive grain boundary micro-cracks by cavity and wedge type cracks were documented for tensile dwell and slow-fast waveforms. These features were not found for slow-slow triangular waveshapes (Fig. 8). The mechanisms provided for tensile dwell cycles and dimples for the solder alloys were found for this material. Under slow-slow waveforms, even though the cracks formed on the specimen surface, the grain boundary sliding provided a path for the crack propagation.

Dwell fatigue life was much higher in stainless steels. Cyclic lives were on the order of a magnitude lower for 1Cr-Mo-V and 2.25Cr-Mo steels. Figures 5 and 7 (I) show that the NCR values were from 0.004 to 0.008 at 0.55% total strain range for low alloy steels and 0.4 and above when extrapolated at the same strain for SS 316 at 600°C and 700°C.

Creep-fatigue behavior of SS 316 and SS 304 steels was found to depend upon the strain rate. With faster strain rates (0.001 /s) other material parameters such as grain size did not influence life. However, as the strain rate decreased (below 10^{-5}), life decreased with an increase in grain size. Dwell fatigue life was a function of grain size. Fracture mode was TG for the faster strain rate tests where striations were documented. With a decrease in the strain rate, fracture mode changed from TG and striations to mixed mode, and as the grain size increased, IG damage occurred. For coarse grained materials, dwell cycles produced IG fracture. In these, and other austenitic stainless steels, dimples were observed to be similar to creep-rupture damage with cavity type cracks.

In summary, the tendency towards IG fracture increased with an increase in the grain size. For stainless steels, the same characteristic that their fatigue resistance increased with an increase in temperature within the range of 600-800°C was observed for this material. This behavior was due mainly to the cell type of dislocation substructure up to 600°C and was sub-boundary type at lower strain rates and higher temperatures (700°C). These observations point to recovery occurring with increased temperatures. On the other hand, for other stainless steels, SS 321 and 347, dislocation substructures were cell-type, which do not promote recovery. Cyclic hardening occurred for the two latter steels, which allowed grain boundary sliding, cavity formation, and IG fracture.

Microscopic observations of SS 316L specimens tested under isothermal (IF) and thermal-mechanical fatigue (TMF) showed multiple crack origins. Several cracks propagated to a large length and were a significant percentage of the main crack causing failure. Many shorter secondary cracks formed that were present over the entire gauge area. A rough estimate given by Shi and Pluvinage /41/ is that approximately 50% of LCF life contributes to the crack propagation phase as striations were reported (see Fig. 6) for a thermal-mechanical, out-of-phase cycle. Under IF, continuous fatigue cycle failure was found to occur by growth of a surface crack and fractures could be characterized by ductile fatigue striations.

In the SS 316L, when tested with the same strain rate with total strain range of 0.8% and 2.4%, isothermal fatigue tests were more damaging than the TMF tests. As a result in Paper I (Fig. 10) lower NCR values were shown for IF tests than TMF ones. The duration of a dwell time was found to be more damaging for this material under IF than under TMF since in the IF case the test temperature is kept constant during the entire test rather than only at peak strains as with TMF. As a result, NCR values were distributed in a large range of the NCR scale.

The low cycle fatigue resistance of several stainless steels such as SS 304, SS 316 and SS 348 were investigated very extensively by Conway *et al.* for the US Atomic Energy Commission /42/. They examined the effect of strain rate, dwell times and temperature on the cyclic life and fracture mode. A summary of their observations for SS 304 is outlined in this paper. The fracture mode by IG or TG crack propagation was found to depend upon the strain rate, strain range, dwell time and the temperature at which the tests were conducted. Within the range of strain rates 4.2×10^{-5} to 4.2×10^{-3} /s, as the strain rate decreased the fracture mode changed from TG to predominantly IG. At higher temperature range of 816°C, IG mode of cracking was observed at all strain rates and dwell times. The length of IG crack (0.8 – 1.2 mm) increased from the initiation point as the temperature increased to 816°C and the strain rate decreased. However, at lower temperatures, for example 650°C, the crack length was found to be much smaller (0.2 – 0.6 mm).

The normalized cycle ratio (NCR) and strain range distribution shown in I reveals that as the plastic strain range increased, NCR increased. There, the difference between the life of dwell cycle and continuous fatigue cycle was small. In some cases, for example the tension only dwells (6 minutes) at 2% strain range showed better life than the continuous fatigue /6/. At higher strain ranges and dwell periods, IG damage leads to the failure, whereas at lower strains and temperature, a mixed distribution of IG and TG damage occurs. When IG damage was found in both the continuous fatigue and dwell time sequences, the NCR was found to improve. It was observed that as the damage modes in continuous fatigue and dwell time sequences were different, the NCR was a result of different mechanisms and varied significantly. Also, at lower strain ranges the NCR varied significantly (see Fig. 11 /6/).

The cyclic behavior of this material was similar to other materials, where an initial hardening was followed by a significant (50% of fatigue life) stable phase and softening towards the end of the fatigue life. This was the case for Ni 201: as cracking progressed, the stresses dropped. High rate strain cycling, total strain range above 2%, resulted in premature buckling. Even though not all materials tested were studied for this particular mode, SS 304L behaved similarly to other materials as discussed for Inconel 617 /43/. As a result, at higher strain ranges with compressive dwell cycles, an improvement in the life was observed. The life was higher than the continuous fatigue life. The SS 304L was found to be tensile dwell sensitive material. However, an increase in temperature lowered the life. The NCR values decreased with compressive and tensile dwells. When inelastic strain and temperature – either in combination or independently – increased, the cyclic life decreased. Since an increase in temperature lowers the tensile strength, yield strength of the material decreases, therefore less elastic strain accumulated during every cycle, resulting in higher plastic strain accumulation per cycle. This led to a modest life reduction in the high strain, low-life region (where higher NCR values are observed) compared to low-strain, long-life region where NCRs are lower (Fig. 12).

Numerous other studies have been made to study features such as cavity formation /44-46/ and their growth for stainless steels. Therefore, in this paper only relevant features controlling deformation and failure under dwells are briefly discussed. Significant mechanisms such as cavity formation are dependent upon strain range (typically less than 2%, low strain rates) and the conditions which promote the mechanisms such as dislocation pile-up, grain boundary sliding, and vacancy cluster caused by tensile dwells. During a tensile dwell cycle, stress relaxation occurs which produces an additional amount of tensile strain and reduces life. Models have been developed utilizing the cavity formation in terms of nucleation and growth. These models were applied in the creep-fatigue, independent of creep, and creep conditions.

Oxidation in stainless steels plays a role in damage mechanisms and life /28/. As the temperature increases the thickness of the oxide layer increases. An oxide wedging mechanism was developed from these observations, where oxides formed during a tensile dwell were though to cause a wedging action between the crack faces and during the faster compression return cycle. Oxides break and produce a crack in the matrix.

Cyclic deformation, within the temperature range of 900-1700°F, causes carbon from the supersaturated solution to reject. Sensitization to IG corrosion and deterioration of mechanical properties take place as a result. Usually, as seen also for low alloy steels, some metal carbides of the form $M_{23}C_6$ form, predominantly chromium carbides. However, iron and molybdenum (Mo) may substitute for chromium. These carbides precipitate along the grain boundaries, slip bands and along non-metallic inclusions. Since carbon is smaller than chromium, it is more mobile and forms its precipitates, whereas Cr depleted regions develop closer to the grain boundaries. Those regions from which Cr is depleted are susceptible to corrosion and IG corrosion. During continuous fatigue, no or very few Cr-carbides form during the end of the test (since time at temperature and stress have increased) compared to dwell cycles. Tensile dwell cycles cause depleted regions from where IG damage takes place. IG corrosion, cavity formation and grain boundary sliding also take place from depleted regions. Depletion also occurs in superalloys with tensile dwell cycles and other cycles. The NCR values, plotted for austenitic stainless steels, show this trend, where tensile dwell cycles were more damaging than other cycles.

Titanium Alloys

1. Ti-6Al-4V

The damage mechanisms were not published in the literature. However, fatigue growth characteristics were examined very extensively in the literature. Song and Hoeppner /4/ provide a summary of the dwell fatigue crack growth mechanisms for various alpha-beta titanium alloys. The high temperature low cycle fatigue and general FCG mechanisms were common when hydrogen reaction and formation of extensive slip lines were considered. For another titanium alloy, IMI 685, Eylon and Hall /47/ found an improvement in LCF resistance as a result of finer grain size. However, fine grain enhances the fatigue crack growth rates for a series of aluminum, titanium and other materials. Surface damage produced the early cracks from the regions of local discontinuities and fracture surface contained sites of straight shear traces and cleavage like planes. A hydrogen diffusion mechanism or hydrogen assisted cracking proposed for steels /48/ was reported extensively for FCG tests in titanium alloys as well. It is interesting to note that some workers related dimples (at higher stress intensity factors), IG cracking at lower stress intensity factors, quasi-cleavage, and micro-void coalescence fracture modes depending upon the microstructure, the crack tip stress intensity factor range and concentration of hydrogen. However, these mechanisms, in high temperature isothermal fatigue situation, are not very well known and need to be examined.

2. IMI 829

Plumbridge /49/ reported dwell sensitivity damage mechanisms in IMI 829. This is a compressive dwell sensitive material, where compressive dwell cycles produced positive mean stress, and stress ratio imbalance. This produced internal damage as compressive dwell time increased together with strain range and temperature. The crack path resulting from the dwell containing cycles was found to be tortuous and TG, however, with the exception of tensile only dwell cycles, where IG propagation occurred later in life and cavitation damage was found to develop (Figs. 2-5, 8). All the cracks were found to be IG since cavities may nucleate on platelets aligned approximately on plates of maximum shear (Fig. 5). Damage mechanisms for Ti-6Al-4V /50-52/ and IMI 829 /49,53,54/ were reported in the literature.

Fracture characteristics of titanium alloys are very complex. Microstructure sensitivity in these alloys has been extensively examined in the literature. The crack path, which is observed to be tortuous, was modeled in terms of structure sensitivity, where a reversed plastic zone has been modeled with respect to the Widmanstätten packet size (or grain size). Transmission electron microscopic studies /54,55/ revealed that cyclic straining at room temperature produced an increased dislocation density and significant inhomogeneity of deformations. Some alpha colonies appeared to be unaffected; however, long planar slip bands formed which extended across several laths and sometimes an entire α-colony. With the inclusion of dwell times in compression (0/15 min), two types of precipitates were noticed, one of smaller size (15-20 nm) within the lath matrix and the other bigger (50 nm) along the slip bands and lath boundaries. Some of these precipitates were identified as silicides. Other workers /54/ interpreted both precipitate sizes as silicides. In addition, hydrides were also documented in the literature. It is not known very well what mechanisms controlled development of silicide precipitates that may either be beneficial or detrimental to fatigue behavior. These results also imply that if the test temperature increased by 50°C, the precipitation process observed during dwell cycles may be a cavity nucleation and growth process. Cavity formation and growth have been documented for IMI 829 under creep-fatigue tests, Ti-6Al-4V under dwell conditions and this author's recent work has indicated cavity formation under elevated temperature fatigue crack growth process within the temperature range of 285 and 345°C.

Superalloys

1. MAR M 002

Dwell fatigue damage mechanisms were reported in /56-58/. This particular study was focused to investigate position of cracks with respect to grain boundaries, primary MC carbides, and secondary carbides $M_{23}C_6$ that nucleate on the grain boundaries. As a result, crack formation, numbers, size and distribution were studied using hollow specimens. The fracture surface was generally perpendicular to the stress axis, although near the outer wall surface it was at an angle of 45° for the samples tested at 750°C. Multiple TG cracks were observed on the gauge area forming from the external surface. The external surface of the specimens was coated. However, IG cracks were seen from the internal surface of the uncoated hollow specimens. The number of cracks, mixed TG and IG fracture, depended on test temperature, strain range and dwell times. As the temperature and tensile dwell time increased to five minutes at nearly 0.9% total strain range, primary MC carbides were forming at the IG sites and secondary $M_{23}C_6$ carbides were noticed in the grain boundaries. At 750°C there was also evidence of some porosity developing. Cracks were documented on the coatings; however, these cracks in most cases stopped at the coating-matrix interface. Larger secondary cracks and primary cracks leading to failure crossed the coating-matrix intersection. These features were also found at 850°C and 1000°C, where IG damage increased with temperature. As the test temperature increased, oxidation damage was taking place. Features such as depleted areas were observed based on qualitative optical observations. Increased amounts of IG damage were produced in these depleted areas.

Compressive dwell cycles (0/5 min) at nearly 1% total strain range had no or limited influence on the damage production at 750 and 850°C. The NCR values were closer to 1.0 for these cycles. However, at 1000°C, oxides grew rapidly and wider cracks were documented on both external and internal surfaces. Another feature relates to the aluminide coating, which starts to diffuse with a rise in temperature. At 1000°C this was clearly seen on the photomicrographs (/56/, Figs. 4.25-4.63) and produces roughness on the surface. The carbides that were present and documented in the grain boundaries were visible under continuous fatigue conditions. Application of dwell times resulted in their diffusion/depletion. These deformation mechanisms cause beneficial effects on the damage progression and an enhanced NCR value was observed.

Tensile dwells were found to influence the damage production significantly for MAR M 002 at all temperatures. The cracking increased significantly with the tensile dwell cycles, as did internal damage and IG damage. Wedge cracks were found to nucleate at a 45° angle or higher to the stress axis. Growth of such cracks continued by the TG mode when the grain boundaries were approximately parallel to the stress axis. The internal surface wall cracks can nucleate either by TG or IG mode. Tensile dwell cycles promoted cracking of the surface coating. At 750 and 850°C, the fracture mode was mainly fatigue dominated. However, at 1000°C, unlike at 750 and 850°C, secondary carbides precipitated on the grain boundaries. These precipitates were very limited. Internal damage in the form of wedge cracks was documented in the grain boundaries. Depletion of carbides was documented qualitatively only at 1000°C (/58/, Figs. 9-10). The NCR values as a result of tensile dwell cycles at 1000°C were found to be in the lowest range on the dwell sensitivity map (I, Fig. 10). A particular damage mode, for example, TG and IG, associated the following:

1) TG damage, limited surface damage, oxidation, internal damage, coating cracks, wedge cracks and precipitation of secondary carbides promote higher NCRs and fatigue resistance, and
2) IG damage, higher surface damage, oxidation damage, internal damage, coating cracks, wedge cracks, depletion of carbides and less precipitation of secondary carbides produce low NCRs and lower fatigue resistance.

Deformation mechanisms within the temperature range of 750-850°C were similar in a similar study on MAR M 002 /59/, in that TG damage occurred. However, as the temperature increased to a range of 950-1040°C, mainly IG damage was documented which significantly altered the fatigue resistance of MAR M 002. Goswami /58/ reported depletion, and oxides, which form IG cracks together with lack of precipitates as discussed earlier, aid in the deterioration of creep-fatigue resistance.

Quantitatively, relating these mechanisms with NCR and HTLCF life will be very difficult, since very little is known about the interaction mechanisms under the conditions of creep-fatigue-oxidation and other processes occurring internally. However, a few modes are worthy of note:

1) Beyond a transition temperature, IG damage together with carbide precipitation, depletion of microstructure, and oxidation, cause deterioration in creep-fatigue behavior of MAR M 002.
2) When test temperatures were lower than this transition temperature, NCR values were found to saturate and therefore were higher where other mechanisms were active.

2. Waspaloy

The damage mechanisms were not published by /60/. However, in a separate study other damage mechanisms were reported for this material at 650°C. Those features are summarized here. Clavel and Pineau /61/ conducted tests at room temperature and 650°C, where cyclic lives were 5-10 times higher at room temperature than at 650°C. Data were presented in the form of a Coffin-Manson plot for room temperature tests; however, a break in the slope of plastic strain range and life was observed at plastic strain amplitude larger than 1%. The LCF tests conducted at room temperature, 650°C and 700°C showed initial presence of crystallographic features, facets, followed by striations. Striations were present in the fatigue crack growth samples in the region where a steady crack growth rate prevails (CGR were within the range of 10^{-6}-10^{7} m/cycle). Waspaloy was found to be of a hardening type, whereas many other materials, for example Inconel 718, are of the softening type. Towards the higher range of low cycle fatigue lives (within the range of 10^4 cycles) equivalent deformation in the case of fatigue crack growth corresponds to near threshold of the long crack growth phase. Characteristics of inhomogeneous deformations resulting from micro-twins were documented /62/. However, as the LCF lives reduce (10^2-10^4) particle shearing was documented. With an increase in plastic strain range (0.42%), where cyclic life of 2000 cycles was observed, micro-twins nucleated at mid-life span. At very low cycle regimes (less than 500 cycles) deformation was by propagation of intense twinning for Waspaloy; however, for Inconel 718 it was by intense planar slip bands. The intensity of twin-forming increased as the plastic strain range increased (> 2%) and twins were recorded as early in life as after the second cycle. In this band, where cyclic life is very low, the slope in the Coffin-Manson line (which is plastic strain and life plot) was very low. Maximum damage is localized within this regime and by the mode in which deformation accrued.

Figure 11 in I shows that the compressive dwells were most damaging and the observations made from the above can be extended for cyclic lives within the range of several hundred cycles. As a result the NCR values recorded were found to be the lowest compared to other dwell cycles in which tensile dwells were applied. A few unbalanced cycles with dwell periods of 100 seconds in tension and 10 seconds in compression produced intermediate NCRs as well as lives. It is imperative that as dwell time is applied plastic strain range per cycle increases, resulting in production of twins and damage localization.

3. Rene 95

This is a high-strength, nickel based superalloy used primarily for applications in gas turbine engine disks in the turbine section. In addition to a conventional strengthening mechanism by high density gamma prime precipitation (0.5 μm) and a lesser degree of solid solution hardening, Rene 95 also derives its strength from the residual dislocation substructure introduced during thermo-mechanical processing (TMP). This TMP produces a complex microstructure called necklace structure, in which large warm duplex grains (75 μm) are surrounded by a necklace of fine recrystallized grains (4 μm) (Figures 4-7 /21/). Such TMP enhances the tensile and creep rupture properties of this material. The microstructures, TMP and mechanical properties of Rene 95 were published extensively /63-66/.

Unlike Waspaloy, which was found to be of the hardening type following exposure to fatigue cycling, Inconel 718 and René 95 are found to be of the softening type, in that the tensile stresses gradually reduce in their magnitude with the application of fatigue cycles. Since these materials soften generally, therefore, significant increases in inelastic strains occur as the test progresses. The creep or inelastic strain was a measure of the decrease in tensile stress during a test divided by its modulus at the corresponding test temperature. With a tensile or compressive dwell, the magnitude of the stress relaxation from softening was found to be significant. More relaxation (on an order of magnitude or more) with the cycles containing compressive dwells was reported in I. It was also noted that as the strain range decreases, the amount of softening, which accumulates creep/inelastic strain, increases. This may point towards a reason why dwell sensitivity was observed at lower strain ranges than at higher strain ranges in which NCR values were observed to saturate. It may be noted from Fig.12 that lower NCRs were associated with lower strain ranges with 20 cycles a minute tests, and tensile, compressive and balanced dwell tests. Dwell periods when applied in tension were shifting the mean stresses in the compression direction and vice-versa.

Shih /21/ in 1982 investigated further the mechanisms controlling failure of René 95 samples tested by Hyzak and Bernstein /63/. Continuous fatigue cycles involved crack formation by TG mode, and shifted to mixed mode in the crack propagation phase (Fig. 10 (b)), and striations were recorded. Carbides of the MC type were found to nucleate the cracks. Facets were reported by /64-66/ on the tensile and fatigue fracture surface of necklace René 95, speculated to be by micro-twinning in the warm worked grains. Others /65,66/ documented facets and provided different interpretations for different materials. Since René 95 was also strengthened by dislocation substructure, stated earlier, claims made in /21/ may be applicable in René 95, that facets were seen as a separation along dislocation channels which in fact are slip bands formed by an extensive planar slip of dislocations. In the warm worked regions, even though a higher density of dislocation and slip bands was present, they were very effective in dispersing the slip. Therefore, early formation of slip bands with HTLCF cycling forced defamation to take place more homogeneously, as compared to coarser planar slips that occur in conventional superalloys (e.g., Inconel 718). As the cycles progress, the intensity of slip systems increases and cracking (facets) along the slip systems were seen in the crack propagation phase. In the case of necklace René 95, a decrease in life was not as a result of IG cracking, which may have been due mainly to the tortuous nature of the crack path, which slows the propagation of cracks following this path.

Failure mechanisms exhibited were IG mode at the beginning of the LCF phase followed by mixed IG + TG deformation. Some of these features were dependent on the dwell times and their direction and strain range at which dwell was applied. At lower strain ranges, MC carbides played a role in crack formation at the early stages of LCF cycling under tensile dwells. As a result NCR increased for most dwell conditions.

The actual mechanisms which control the deformation and failure under different dwell conditions: for example, NCR values of t/0 cycles were similar to or higher than continuous fatigue cycles (0/0) and NCR values of compressive and balanced dwell cycles were low meaning lower life, are not fully understood. One may argue positive mean stresses in compresses dwell shift stresses in tension and cause dwell sensitivity. However, mechanisms that cause this behavior must be investigated. Therefore, interaction among fatigue, creep, and oxidation only was taken into account and other potential mechanisms and their role in accelerating damage were ignored.

4. René 80

René 80 is a high strength, nickel based superalloy used extensively in gas turbines as turbine blades for operation in temperatures up to 1010°C. The dwell sensitivity damage mechanisms were not published /67/. However, in a separate study, metallography and LCF damage mechanisms were discussed for 871°C, which is one of the test temperatures at which data is presented in I. A brief overview of the damage mechanisms observed there is presented here.

René 80 was tested with several combinations of heat treatments that allow surface coating and enhance other tensile and creep rupture properties. The exposure at 982°C caused coarsening of γ' precipitates. Under stress hold, 97 MPa, significant coarsening occurred, in that cubes present in the microstructure transformed into plates lying normal to the tensile axis (/68/, Figs. 1-6 and 9-13). There was also thickening of γ' precipitates on the grain boundaries. Since any forced load fluctuation process, like fatigue, creates dislocations, the arrangement of dislocations is subject to achieving a minimum strain energy condition, even though minimum energy configuration is influenced by the degree of misfit between the precipitate and matrix. Antolovitch *et al.* documented dislocations in hexagonal arrays /68,69/. Higher dislocation densities occurred at HTLCF (Figs. 11,12). A significant different in HTLCF life was noted with dislocation sub-structures with conventional heat treatment.

Thermally exposed specimens contained IG crack path morphologies, whereas stress exposed specimens exhibited a mixed TG + IG mode. In the latter case the size of cracks was larger and boundaries covering the cracks were lightly oxidized. This may be due to the higher γ'/carbide structure development with stress exposure which caused reduced fatigue performance.

Since there were numerous test and material variables involved in both the studies, for example, stress exposure, aging, thermal exposure, machining following the exposure, the HTLCF performance of René 80 was found to be different in each case, even though similar dislocation sub-structures were created by different procedures. A similar effect was observed for MAR M 002: when an aging heat treatment was applied, the HTLCF life was considerably reduced. The exposure in this case altered the γ' configuration, where γ' coarsens in the matrix which also alters the carbides and their precipitation rates. As a result grain boundaries change, in which oxides may penetrate and interact with the constituents that either cause a degradation or otherwise.

Frequency was found to determine the mode of deformation by TG and/or IG modes. For the continuous cycle types TG damage was found predominantly at both temperatures. However, as the test temperature increased to 1000°C IG damage and creep damage by grain boundary sliding was observed. As the frequency decreased, creep damage or creep-fatigue interactions reduced the life. For tensile dwells as well as compressive dwells the life was reduced considerably. However, NCR of tensile dwell cycles was found to be minimum in I. A large number of surface cracks were reported in the coated conditions, which in the case of continuous fatigue did not affect the life as much as in the case of dwell.

5. Inconel 617

The loading conditions, such as continuous fatigue cycling, did not produce any change in the physical shape of the specimens /43/. However, when tensile dwell cycles were applied, they produced bulging. Compression dwell cycles caused necking. These observations are true for many materials, for example, SS 304 and tantalum based materials and Inconel 800H /70/. In the case of Ni-base alloys such effects as bulging and necking were minimal, due mainly to their higher strength and lower ductility. As the strain rates decreased the damage modes changed from TG to IG mode; the same was true for dwell times in tension from 1 min to greater than 10 min. Tensile dwell conditions (larger than 1 min) produced IG cracks in the interior of the specimens. However, compression dwells caused dimples and fracture followed by necking. The development of secondary carbides was a function of loading condition employed.

Introduction of tensile dwell (1 min) increased the volume fraction and density of $M_{23}C_6$ carbides together with some M_6C and other secondary carbide precipitation. An increase in dwell time (10 min) caused rapid coarsening of grain boundary precipitates and increased inter-particle spacing. Longer dwell cycles, for example 120 min, did not produce any further damage and mechanisms for failure were similar to other cycles. Regardless of loading conditions, the alloy exhibited sub-grain formation in the majority of the grains. Multiple slip in some of the grains was also observed at 950°C. Tensile dwell exhibited some tendency to form dislocation tangles in the matrix and dislocation pile-ups at the grain boundaries. Despite dislocation pile-ups no IG cracking could be observed.

6. IN 100

The dwell sensitivity damage mechanisms were not published in the literature /67/. From a separate study /71/, a features are summarized which may help in the understanding of dwell sensitivity behavior. Since IN 100 has many product forms ranging from castings to powder metallurgically produced, the observations made in one study and condition cannot be applied to another form and condition. For a cast product, a decrease in total strain range caused a smaller inelastic strain build up, which produced a tensile mean stress. The absolute value of yield stress was smaller at 900°C, where the mechanical strain was minimum. The peak yield stress was observed at 700°C, where the mechanical strain achieved its highest value. This implies that dwell cycles are more damaging in tension direction than in compression. Deformation modes were dependent on the temperature, strain range, and dwell time. Interdendritic areas were oxidized and IG damage propagated. In the case of powder IN 100, material discontinuities may dictate the damage formation and growth kinetics. A similar trend was noted for 2.25Cr-Mo steel, where tensile properties were optimum at 450°C, and was due to smaller cell structure that increased with temperature, causing the material to accumulate more damage.

7. PWA 1480 Single Crystal Tested at 1015 and 1050°C

The dwell sensitivity damage mechanisms were dependent on the pores present for this material /72/. Cracks originated from the pores in most cases and linked with other pores, whereas in some cases surface crack formation and growth was a possible mechanism. The crack faces were oxidized and the pores aided in the growth and formation of surface cracks.

TMF OP cycles (when normalized by IF) show the greatest dwell sensitivity. It is not possible to conclude whether PWA under <001> orientation is more dwell sensitive, as the data contained numerous variables.

The IF damage for PWA 1480 under the orientations specified produced cyclic softening, whereas TMF under both IP and OP produced hardening. Cracking occurred from the outside surface in TMF IP and OP tests. With the increase in number of cycles greater γ' agglomerations were observed /73/. Higher exposure times resulted in higher oxidation damage. At the same strain range TMF IP cycle produced higher life than TMF OP cycle with higher γ' agglomerations and oxidation. Negative mean stresses developed for TMF IP cycles, as opposed to positive mean stresses for TMF OP cycles. Positive mean stress was detrimental to life under TMF OP conditions. Stress relaxation occurred, amounting to nearly 30% of initial stress range for TMF IP cycles, whereas a similar extent of tensile stress developed for the TMF OP cycles. The exact interactive mechanisms that occur under IF and TMF are not very well characterized to address dwell sensitivity for PWA 1480. More research is needed to construct various boundaries of mechanistic aspects in terms of test parameters and material conditions.

8. MA 754

The fatigue life in the longitudinal orientation was higher than in long-transverse orientation. This behavior is similar to that found in René 150 /74/. Like other materials, for example solder alloys /7-9/, failures were as a result of surface deterioration, which progressed internal to the specimen gauge area. Under high cycle fatigue similar patterns were seen /75/. The LT specimens deformed by TG mode at the early and final stages of fatigue life, whereas for TL a mixed mode damage was observed during the crack propagation phase. The difference in the fatigue life and failure mode was interpreted in terms of decrease in tensile properties (modulus). For compression dwells, this material showed enhanced creep-fatigue resistance and also for the dwells in the balanced conditions. A similar trend was exhibited by stainless steels. TG failure mode dictated the failure. For both the conditions tensile dwells were most damaging due mainly to a shift in the crack propagation mode by IG mode. In general, IG damage in all the stages of damage growth is much faster than the TG and mixed conditions.

CONCLUDING REMARKS

The dwell sensitivity damage mechanisms examined in this paper considered materials from many alloy groups. In the case of planar slip materials, as pointed out in individual sections, slip systems generated during a fatigue process. These dislocations piled up against barriers like grain boundaries, precipitates and other secondary phases. Strain localization along the slip system occurred and caused the slip planes to crack. Low stacking fault energy, low temperature, low strain and presence of coherent precipitates favored stage I cracking. Stage I cracking occurs along planes aligned at 45° to the stress axis. A slight change in the direction of crystallographic fracture is a typical feature for this group of alloys. These were documented for solder alloys, titanium alloys, low alloy steels, and some nickel base alloys (MAR M 002 below 750°C).

For wavy slip materials, dispersal of slip to adjacent slip planes occurs by means of dislocation cross slip and climb mechanisms. These lead to a more homogeneous deformation pattern. The TG cracking perpendicular to stress axis occurs in stage II. Opposite conditions such as higher temperature, higher strain, higher stacking fault energy, promote this type of deformation. Since dislocation cross slip and climb are thermally activated phenomena, temperature plays a major role in the deformation process. Low temperature, low mean stresses, and high frequency promote this type of deformation where striations form. TG deformation and striations reported for low alloy steels, titanium alloys and copper alloys belong to this class of materials.

Higher temperatures and lower strain rates, in general, produce damage in IG mode. Grain boundary sliding and cavity formation were inter-related with creep mechanisms. When the grain boundaries align themselves at 45° to the stress axis through migration, grain boundary sliding is promoted. This process creates cavities, by vacancies, which in turn is a result of plastic deformation. Only after reaching a critical size do the cavities become stable thermodynamically. The IG fracture mechanism is extensively investigated within the creep framework and can be summarized in terms of grain boundary sliding and stress redistribution as a result of inhomogeneous cavitation. Stress becomes non-uniform and multi-axial, even when the remotely applied stress is purely uniaxial. Several mechanisms proposed for the grain boundary sliding and IG cavitation are discussed within the creep frameworks.

The dwell sensitivity damage mechanisms of materials examined in this study are summarized briefly as follows:

1) The dwell sensitivity was dependent on the materials, for example, microstructure, condition, and grain size. Only after a particular temperature was exceeded did the deformation in a material accrue by means of metal (M) and carbide ($M_{23}C_6$) precipitation, along the grain boundaries. These were of Cr, Mo, C and other alloying elements. As these precipitate formed, slip systems became more mobile and concentrated along the carbides, or the grain boundaries that promoted grain boundary sliding, cavity formation, and IG damage.
2) The damage was also dependent on the test temperature, strain rate, dwell time and direction in which it was applied. With the application of tensile dwell, stress relaxation occurred which lowered the material strength in that loading direction. Softening occurred with a combination of lower strain rates, longer dwell periods, lower stress range, and large relaxation strains (which is a result of strain transformation from elastic to plastic or inelastic) that promoted IG damage. Other features such as cavity formation, grain boundary sliding, etc., compounded the damage, and failure in this mode was faster than in TG mode.
3) Within a range of temperatures, the mechanical properties of some materials improved, for example, 2.25 Cr-Mo, Ni201, and SS316. Within this temperature range clusters of carbides formed, for example Mo-C, as a result of dynamic strain aging. This action enhanced the fatigue as well as tensile properties of materials within that range of temperatures. However, as this temperature was exceeded, more damage occurred as other processes such as environmental interactions and oxidation accelerated the growth of damage.
4) Oxidation formed on the surface due to exposure at high temperatures. Surface striated damage formed and crack intrusion by either IG/TG or mixed conditions was observed. As the strain range increased, the number of such oxide intruded cracks increased. Microstructure plays an important part, in that straight crack and bifurcated tortuous crack paths occur, indicating sensitivity effects. Longer dwell periods caused several cracks to grow inside the gauge area.
5) Dwell cycles produced a change in the mechanical properties of materials, where a dwell cycle either lowered the modulus or enhanced it. In most cases, as the test temperature increased the tensile strength decreased. With the application of a dwell cycle, stress range in both directions increased or decreased. An initial 5-15% of the test life was spent in achieving the stable hysteresis loops. Deformation in the material was in three modes, where TG, mixed mode, and predominantly IG fracture occur. These

regions depend upon the stress range applied in a cycle and material softening and hardening phenomena. Only after exceeding a critical stress can every cycle produce the damage in one of the following three modes: TG, mixed, and IG. These stresses are not identified for each material studied in this paper. It is imperative that lower stresses and strain rates produce IG damage where grain boundary sliding and cavity formation occur whereas, in the other ranges, mixed mode and TG fracture occur with defined striation patterns.

6) For some materials, with the application of tensile dwell, mean stress developed in the tension direction. Positive mean stresses produced cavitation and lower life. Positive mean stress was also produced during compression dwell. As exposure time at high temperature increased, oxide scales formed and cracking of the oxide surface was a common feature observed during dwell cycles. Longer dwell cycles caused more surface roughness.

REFERENCES

1. T. Goswami, *Mechanics of Materials*, **22**, 105-130 (1995).
2. J.E. Hack and G.R. Leverant, *Met. Trans. A*, **13A**, 1729 (1982).
3. G.F. Harrison, P.H. Tranter, M.R. Winstone and W.J. Evans, in: *Proceedings of Conference on Designing with Titanium*, Briston, Institute of Metals, 1986; p. 198.
4. Z. Song and D.W. Hoeppner, *Int. J. Fatigue*, **11** (1), 85 (1989).
5. W.J. Evans and M.R. Bache, *Int. J. Fatigue*, **16** (10), 443 (1994).
6. T. Goswami, *Int. J. Fatigue*, **21** (1), 55-76 (1999).
7. R.N. Wild, Some fatigue properties of solders and solder joints, in: *National Electronic Packaging Conference*, 1974; pp. 105-117.
8. S. Vaynman, E.F. Morris and D.A. Jeannotte, Chapter 4 in: *Solder Mechanics – A State of the Art Assessment*, D.R. Frear, W.B. Jones and K.R. Kinsman (Eds.), TMS, 1991; pp. 155-189.
9. H.D. Solomon, in: *Electronic Packaging – Materials and Processes*, J.A. Sortell (Ed.), ASM International, 1985; pp. 29-49.
10. L. Campbell, Fatigue Crack Propagation, ASTM STP 415, 1967; 131.
11. A. Saxena and S.D. Antalovich, *Metall. Trans.*, **6**, 1809 (1975).
12. H.D. Solomon, *Proc. 2nd ASM Electronic Packaging Conf.*, J.A. Sortell (Ed.), ASM International, 1986; pp. 29-47.
13. K.P. Jen and J.N. Majerus, *Trans. ASME J. Eng. Mater. Technol.*, **113**, 475-483 (1991).
14. J.M. Smeby, *IEEE Trans. Component, Hybrids and Manufacturing Technology*, 1985.
15. W.J. Plumbridge, Private communications and life assessment conducted for Faculty of Technology, Open University, Milton Keynes, England, 1997.
16. H. Jiang, R. Hermann and W.J. Plumbridge, *J. Material Sci.*, **31**, 6455-6461 (1996).
17. S. Vaynman, M.E. Fine and D.A. Jeannotte, *TMS AIME Annual Meeting*, Denver, P.K. Liaw and T. Nicholas (Eds.), 1987; p. 127.
18. D. Frear, D. Grivas, M. McCormack, D. Tribula and J.W. Morris Jr., *Proc. IIIrd Annual Conference on Electronic Packaging and Corrosion in Microelectronic*, M.E. Nicholson (Ed.), ASM International, 1987; pp. 269-274.
19. C.E. Cutiongco, S. Vaynman, M.E. Fine and D.A. Jeannotte, *J. Electronic Packaging*, **112**, 110-114 (1990).
20. R.H. Stentz, J.T. Berling and J.B. Conway, *AGARD CP 243*, 1978; Paper No. 12.
21. C.I. Shih, High temperature low cycle fatigue mechanisms for a nickel-base and a copper base alloy, NASA Contract Report 3543, 1982.
22. T. Goswami, *High Temperature Materials and Processes*, **14** (1), 1 (1995).
23. E.G. Ellison and A.J.F. Patterson, *Proc. I. Mech. Engs.*, **190**, 321 (1976).
24. W.J. Plumbridge, R.H. Priest and E.G. Ellison, Damage formation during fatigue-creep interactions, in: *Proc. IIIrd International Conference on Mechanical Behavior of Materials*, Cambridge, Smith *et al.* (Eds.), **2**, 129 (1979).
25. W.J. Plumbridge and D.A. Miller, *Res. Mechanica*, **3**, 1-8 (1981).
26. G. Degallaix, C. Korn and G. Pluvinage, *Fatigue Fract. Engng. Mater. Struct.*, **13** (5), 473-485

(1990).

27. S. Majumdar, *ASME Conference on Thermal Stress, Material Deformation and Thermo-Mechanical Fatigue*, H. Sehitoglu and S.Y. Zamrik (Eds.), 1987; p. 37.
28. S. Kalluri and S.S. Manson, Time dependency of strain-range partitioning life relationships, NASA Contract Report 174946, 1984.
29. S.D. Mann, Unpublished work, Electricity Commission of Victoria, Melbourne, Australia, 1991.
30. C.R. Brinkman, J.P. Strizak, M.K. Brooker and C.E. Jaske, *J. Nuclear Mater.*, **62** (2/3), 181 (1976).
31. T. Asayama, S. Cheng, Y. Tachibana and Y. Asadss, *Bulletin of JSME*, 1987.
32. A. Narumoto, *Japan-US Joint Seminar on Advanced Materials for Severe Service Applications*, Tokyo, Japan, 1986, K. Iida and A.J. McEvily (Eds.), 1987; p. 129.
33. H. Teranishi and A.J. McEvily, *Met. Trans.*, **10**, 1806-1807 (1979).
34. J. Polak, J. Helesic and M. Klesnil, *ASTM STP* 942, 1988; pp. 43-57.
35. J. Janovec, V. Magula, A. Holy and A. Vyrostkova, *Scripta Metallurgica et Materialia*, **26**, 1303-1308 (1992).
36. G. Ebi and A.J. McEvily, *Fatigue Engng. Mater. Struct.*, 7 (4), 299-314 (1984).
37. B.G. Gieseke, C.R. Brinkman and P.J. Masisz, The influence of thermal aging on the microstructure and fatigue properties of modified 9Cr-1Mo steel, in: *Proceedings of First International Symposium on Microstructures and Mechanical Properties of Aging Materials*, Chicago, 1992.
38. K.D. Challenger, A.K. Miller and C.R. Brinkman, *Trans. ASME, J. Eng. Mater. Tech.*, **103**, 7 (1981).
39. J.D. Baird and A. Jamieson, *J. Iron and Steel Institute*, **210**, 841-856 (1972).
40. K. Yamaguchi and K. Kanazawa, *Met. Trans.*, **11A**, 2019 (1980).
41. H.J. Shi and G.I. Pluvinage, *J. Fatigue*, **16** (9), 549 (1994).
42. J.B. Conway, R.H. Stentz and J.T. Berling, *Fatigue, Tensile, and Relaxation Behavior of Stainless Steels*, Technical Information Center, United States Atomic Energy Commission, 1975; pp. 33-81.
43. K.B. Rao, H.P. Meurer and H. Schuster, *Mater. Sci. Engng.*, **104**, 37 (1988).
44. W.D. Nix, *Mater. Sci. Engng.*, **A103**, 103-110 (1988).
45. D.A. Miller, C.D. Hamm and J.L. Phillip, *Mater. Sci. Engng.*, **53**, 233-244 (1982).
46. K.T. Rie, R.M. Schmidt, B. Slchner and S.W. Nam, *ASTM STP* 942, 1988; pp. 313-328.
47. D. Eylon and J.A. Hall, *Met. Trans.*, **8A**, 981-990 (1977).
48. C.D. Beachem, *Met. Trans.*, **3**, 437-451 (1972).
49. W.J. Plumbridge, *Fatigue Fract. Eng. Mater. Struct.*, **10** (5), 385-398 (1987).
50. G.R. Yoder, L.A. Cooley and T.W. Crooker, *Engng. Fracture Mech.*, **11**, 805-816 (1970).
51. E. Hornbogen and K.H. Zum Gahr, *Acta Metall.*, **24**, 581-592 (1976).
52. J.C. Chesnutt and J.C. Williams, *Met. Trans.*, **8A**, 514-515 (1977).
53. T. Goswami and W.J. Plumbridge, Unpublished work, Materials Engineering Department, University of Wollongong, Australia and Materials Discipline, Faculty of Technology, Open University, Milton Keynes, England (1989-1992).
54. A.P. Woodfield, P.J. Postans, M.H. Loretto and R.E. Smallman, *Acta Metall.*, **36**, 501-516 (1988).
55. W.J. Plumbridge, R. Smith and H.G. Dam, *Fatigue*, **90** (3), 1935-1940 (1990).
56. E.G. Ellison, W.J. Plumbridge and M.S. Dean, University of Bristol, Department of Mechanical Engineering Research Report 327, 1984.
57. E.G. Ellison and W.J. Plumbridge, *Fatigue Fract. Eng. Mater. Struct.*, **14** (7), 721-739 (1991).
58. T. Goswami, *High Temperature Materials and Processes*, **14** (2), 47 (1995).
59. V.T.A. Antunes and P. Hancock, *AGARD CP* 243, Paper No. 5, 1978.
60. G. Asquith and S.H. Springthall, *AGARD CP* 243, Paper No. 7, 1978.
61. M. Clavel and A. Pineau, *Mater. Sci. Engng.*, **55**, 157-171 (1982).
62. M. Clavel and A. Pineau, *Mater. Sci. Engng.*, **55**, 173-180 (1982).
63. J.M. Hyzak and H.L. Bernstein, *AGARD CP* 243,

Paper No. 11, 1978.
64. M.N. Menon, *J. Mater. Sci.*, **11**, 941 (1976).
65. M.N. Menon and W.H. Reimann, *J. Mater. Sci.*, **10**, 1571 (1975).
66. M.N.Menon and W.H. Reimann, *Met. Trans.*, **6A**, 1075 (1975).
67. G.R. Halford and A.J. Nachtigall, *AGARD CP* 243, Paper No. 2, 1978.
68. S.D. Antolovitch, P. Domas and J.L. Strudel, *Met. Trans.*, **10A**, 1859-1868 (1979).
69. S.D. Antolovitch, S. Liu and R. Baur, *Met. Trans.*, **12A**, 473-481 (1981).
70. H.P. Meurer, G.K.H. Gnirss, W. Mergler, G. Raule, H. Schuster and G. Ullrich, *Nucl. Technol.*, **66**, 315-323 (1984).
71. J.L. Malpertu and L. Remy, *ASTM STP* 942, 1988; pp. 657-671.
72. R.V. Miner, J. Gayda and M.G. Hebsur, *ASTM STP* 942, 1988; p. 371.
73. G.A. Swanson, I. Linask, D.M. Nissley, P.P. Norris, T.G. Meyer and K.P. Walker, Life prediction and constitutive models for engine hot section anisotropic materials program, NASA-CR-179594, 1987.
74. P.K. Wright and A.F. Anderson, in: *Superalloys*, J.K. Tien, S.T. Wlodek, H. Morrow III, M. Gell and G.E. Maurer (Eds.), ASM, 1980; pp. 689-698.
75. M.Y. Nazmy, ASTM *STP* 942, 1988; p. 385.