A Physics-Based Understanding of Hold-Time Effects on Fatigue

X.J. Wu
National Research Council Canada, Ottawa, Ontario, Canada

Abstract: It has been long known that hold-time during fatigue can cause a significant life debit, but the problem has been mostly dealt with empirical approaches. In this paper, the basic deformation/damage mechanisms are reviewed, and a unified model is presented by considering fatigue crack propagation in coalescence with hold-time induced damage such as wedge cracks or creep cavities. Thus, a nonlinear damage accumulation rate equation for the process is derived. Experimental data from the literature are used to compare with the descriptions of the model.

1. Introduction

With demands for advanced gas turbine engines that maximize the firing temperature to increase fuel efficiency and lower greenhouse gas emission, components are required to operate at higher and higher temperatures such that creep damage is imparted to those components such as disks that traditionally experience low cycle fatigue (LCF). For life cycle management, it also needs to consider the effects of variable-amplitude thermomechanical loads with hold/dwell times in the operating history to optimize the maintenance schedule and to predict the remaining life, which has become the essence of a new paradigm called *Prognostic Health Management (PHM)*.

Historical lessons from the impeller failure of RB 211 (1970) to the more recent turbine disk failure of the engine that powers B767 at LAX (June 2008) have shown that complex loading interactions at either low or high temperatures could induce catastrophic failure. Therefore, it is imperative to understand the fundamental mechanisms of dwell/creep-fatigue interactions such that a more accurate and reliable method can be established to take that into account to prevent unexpected component failure.

The conventional dwell/creep-fatigue damage assessment methods can be classified into three major approaches: i) the linear damage rule (LDR) which combines the Miner [1] and Robinson's [2] rules in a linear summation form, ii) the frequency-modified equation [3], and iii) the strain range partitioning (SRP) method by NASA [4]. The linear fraction rule, however, disregards whether the hold time is in stress control or strain control, which can produce different effects, via creep or stress relaxation. Therefore, it can be grossly erroneous for some cases [5]. The other two approaches seem to be only applicable to strain controlled cyclic processes, which are somewhat based on the Coffin-Manson relationship [6, 7]. In the open literature to date, there are perhaps more than a hundred lifing equations outgrowing from the aforementioned approaches [8], but they are largely empirical, and therefore valid on a case-by-case basis. The problem still lies in the lack of an understanding of the intricate interactions between different damage modes in a complex loading cycle.

This paper reviews the fundamental mechanisms of fatigue and dwell/creep induced damage and then presents a unified model of dwell/creep-fatigue based on the predominant deformation and crack growth mechanisms.

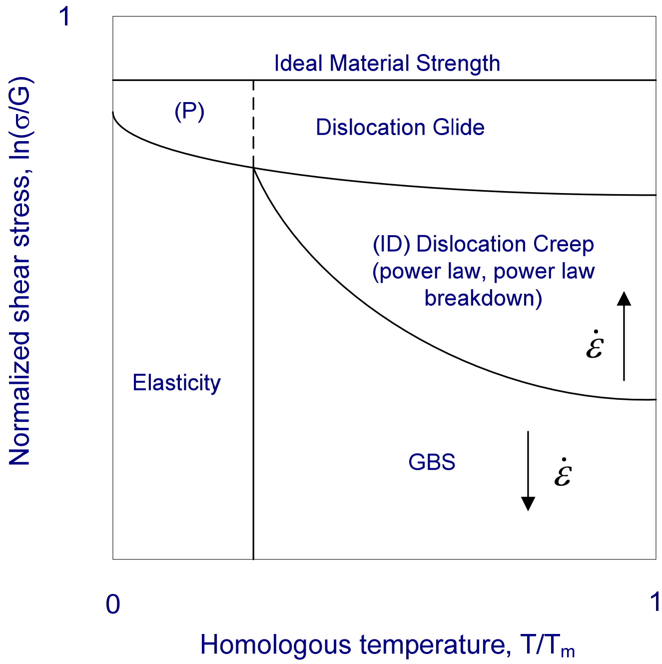


Figure 1 : Schematic deformation mechanism map.

2. The Basic Deformation and Damage Mechanisms

In general, for a polycrystalline engineering material, deformation regimes can be summarized into a deformation map, following Frost and Ashby [9], as shown schematically in Figure 1. Elastic (E) and rate-independent plasticity (P) usually happens at low temperatures (i.e. $T < 0.3\ T_m$, where T_m is the melting temperature). In this regime, it is well understood that the material failure mechanism mainly occurs by fatigue, except for ultimate tensile fracture and brittle fracture. As temperature increases, dislocations are freed by vacancy diffusion to get around the obstacles so that time-dependent deformation manifests. Time dependent deformation at elevated temperatures is basically assisted by two diffusion processes—grain boundary diffusion and lattice diffusion, the former assists dislocation climb and glide along grain boundaries, resulting in grain boundary sliding (GBS), and the latter assists dislocation climb and glide within the grain interior, resulting in intragranular deformation (ID) such as the power-law and power-law-breakdown.

Based on the above deformation mechanisms, the total inelastic strain in a polycrystalline material can be decomposed as follows:

$$\varepsilon_{in} = \varepsilon_g + \varepsilon_{gbs} \tag{1}$$

In the following sections, we will examine the roles of each deformation component played in the damage process with respect to either fatigue or creep.

2.1 Fatigue Damage

Fatigue damage is known to be associated with formation of persistent slip bands (PSB) or dislocation substructures that evolve during cyclic deformation [10]. Fatigue crack nucleation tends to occur at the free surface or at surface/subsurface inclusions/pores where intrusion/extrusion of persistent slip bands (PSB) accumulates because of the presence of material discontinuity.

Using the continuously distributed dislocation theory, Tanaka and Mura first developed a fatigue crack nucleation life equation for isotropic materials, as [11]

$$N_c = \frac{4(1-\nu)w_s}{\mu b}\frac{1}{\Delta\gamma^2} \tag{2a}$$

where w_s is the surface energy, b is the Burgers vector, μ is the shear modulus and ν is Poisson's ratio.

Following the same concept, Wu recently determined the fatigue crack nucleation life for anisotropic materials, as [12]

$$N_c = \frac{4w_s}{F_{ij}\,\Delta\gamma_i\,\Delta\gamma_j b} \tag{2b}$$

where F_{ij} is the material's characteristic elastic matrix defined by the eigenvalue problem.

Crystal anisotropy can have a strong effect on the formation of dislocation pile-ups, particular in materials that exhibit coupling ($F_{ij} \neq 0$, i $\neq$j) between mode-I, II and III, at the crack nucleation stage.

2.2 Cold Dwell Damage

Dwell-fatigue usually refers to fatigue with hold-times at ambient temperatures. In the low temperature regime of the deformation mechanism map, Figure 1, dwell damage is most likely to form as dislocation pile-ups in materials with microstructural inhomogeneity. It has been found that high strength titanium alloys such as IMI 685, IMI 829 and IMI 834, and Ti6242 have the most pronounced dwell sensitivity. Dwell fatigue of titanium alloys is often accompanied with faceted fracture along the basal planes of α phase, which is believed to be driven by dislocation pile-up [13].

Dislocation pile-ups can be effectively classified as Zener-Stroh-Koehler (ZSK) cracks with an energy release rate of [14]

$$G = \frac{b_T^{(i)} F_{ij} b_T^{(j)}}{8\pi a} \tag{3}$$

where b_T is the total Burgers vector and a is the half crack length.

It has been shown using dislocation kinetics that when dislocation pile-up is driven by deformation at a strain rate $\dot{\gamma}_p$, having an escape rate κ by climb, the ZSK crack size is given by [15]

$$l_z = 2a = \frac{\overline{F}_{22} b^2}{16\pi w_s} \left(\frac{\dot{\gamma}}{\kappa} \right)^2 \left[1 - \exp(-\kappa\tau) \right]^2 \tag{4}$$

2.3 Creep Damage

Under sustained loading at high temperatures, materials creep in three stages: i) primary, ii) secondary, and iii) tertiary. Wu and Koul developed an entire-creep-curve model based on the involved deformation mechanisms [16, 17]. During short-period hold times, GBS is mainly responsible for producing interganular damage, i.e., creep cavity/crack, during the transient (primary plus secondary) creep, since tertiary creep would never start upon short cycle repeats. Thus, the creep cavity/crack size can be expressed as

$$l_c = \varepsilon_{gbs} d \tag{5}$$

where d is the grain size, and ε_{gbs} is grain boundary sliding strain.

Under stress control, GBS would accumulate with time as the ratchet strain. Under cyclic strain controlled conditions, l_c can be stabilized once the entire hysteresis behaviour is stabilized.

3 The Dwell/Creep-Fatigue Interaction Model

By dwell/creep-fatigue interaction, a fatigue crack may form at the surface either by intrusion/extrusion of persistent slip bands or by oxidation, which then propagates inside the material coalescing with dwell/creep damage of size *l* along the path. The step of coalescence repeats over an average distance of λ until final fracture. This damage process is schematically shown in Figure 2. Therefore, the total crack growth rate can be expressed as [18]:

$$\frac{da}{dN} = \left(1 + \frac{l_c + l_z}{\lambda}\right)\left(\frac{da}{dN}\right)_f \qquad (6)$$

where both l_z (Eq. 4), and l_c (Eq. 5) are added together to include the competition between the dwell and creep mechanisms in a broad temperature range.

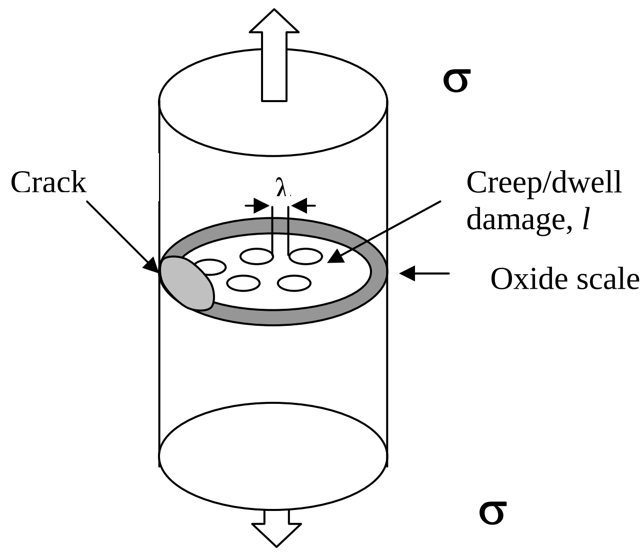


Figure 2. A schematic of damage development in a specimen cross-section.

In the presence of environmental effects, the total crack growth rate is

$$\frac{da}{dN} = \left(1 + \frac{l_c + l_z}{\lambda}\right)\left\{\left(\frac{da}{dN}\right)_f + \left(\frac{da}{dN}\right)_{env}\right\} \qquad (7)$$

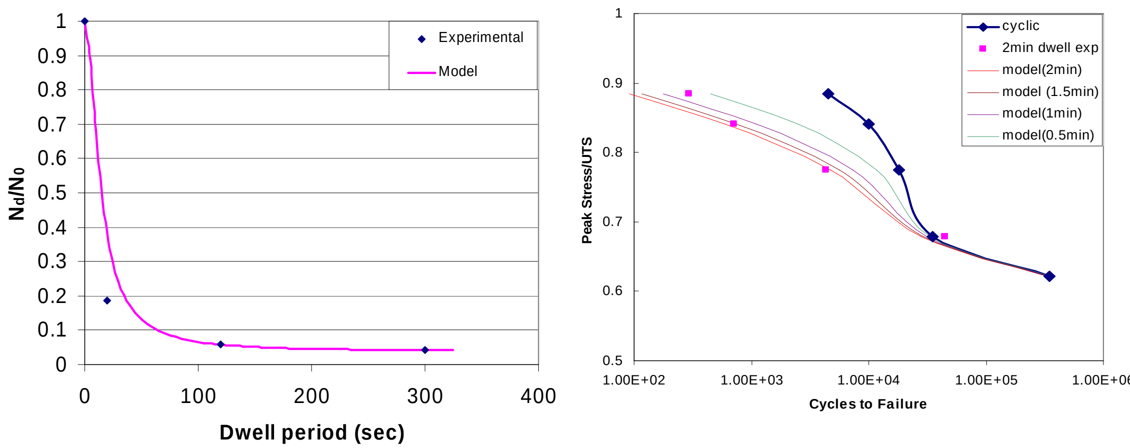
4 Discussion

4.1 *Dwell-Fatigue*

For constant-amplitude dwell fatigue at ambient temperature, substituting Eq. (4) into (6) and neglecting the creep damage (l_c), we have

$$N = \frac{N_f}{\left(1 + \frac{F_{22} b^2}{16\pi\lambda w_s}\left(\frac{\dot{\gamma}}{\kappa}\right)^2 [1 - \exp(-\kappa\tau)]^2\right)} \qquad (8)$$

Equation (8) shows that the fatigue life is reduced when a dwell period is imposed on fatigue loading. It can be inferred from Eq. (8) that the dwell sensitivity is mostly influenced by the ratio of dislocation gliding strain rate to the climb rate, where the stress and temperature dependence arise, and it is an exponential function of dwell time. This means that the dwell-effect will be more detrimental when the ratio of dislocation glide to climb is large, particularly in materials with fewer active slip systems at low temperatures. Figure 3 (a) and (b) show the comparison of the model (parameter values are given in ref. [15]) with the experimental data reported by Bache et. al [13] on IMI 834. As temperature increases though, the climb will overwhelm glide such that dislocation pile-up can rarely form, and hence the dwell damage becomes minimal, but cavities may start to grow. Basically, this is the essence of “cold dwell” vs. “hot creep”.



(a) (b)

Figure 3. Comparison of Eq. (8) with the experimental data on IMI 834 [14]: a) Normalized dwell fatigue life as a function of dwell time, b) S-N curves with different dwell times.

4.2 Creep-Fatigue

The creep-fatigue interaction refers to the effect of cyclic-hold interactions at high temperatures where creep damage can be significant. NASA has developed a number of test methods (see Appendix A for details) and proposed the SRP approach to deal with the life prediction for a deformation process that may involve a combination of different strain components manifested from those tests [4]. In this approach, four strain components: plastic strain reversed by plasticity (*pp*), creep strain reversed by creep (*cc*), plastic strain reversed by creep (*cp*), and creep strain reversed by plasticity (*pc*), have to be identified from a complex hysteresis cycle, which, in practice, is often not easy. For simplification of the matter, it is attempted to use the physics base deformation decomposition rule, Eq. (1), to correlate with those strain components in their respective tests, based on the following rationale:

- The grain deformation, $\Delta\varepsilon_g$, when proceeds in a cyclic manner, will lead to transgranular damage accumulation, such as persistent slip bands and fatigue cracking, and therefore, it is equivalent to the *pp* strain plus any *cp* or *pc* strains in a asymmetric cycle.

- For short-period holds, *cc*, *pc* and *cp* types of inelastic strains are contributed mainly from GBS during the transient creep, which contribute to intergranular fracture, as discussion in section 2.3.

Therefore, basically, the SRP approach can be consolidated by Eq. (1), expressed in terms of $\Delta\varepsilon_g$ and $\Delta\varepsilon_{gbs}$, as shown in Table 1.

Under constant amplitude cycling conditions, Eq. (6) can be integrated into the form (in this case, neglecting dislocation pile-ups, i.e., let $l_z = 0$):

$$N = \frac{N_f}{1 + \frac{\varepsilon_{gbs} d}{\lambda}} \tag{9}$$

and the pure LCF life, N_f, is correlated to $\Delta\varepsilon_g$ through Eq. (2).

Table 1. Consolidation of SRP

Test Type*	$\Delta\varepsilon_g$	$\Delta\varepsilon_{gbs}$
HSRC	pp	0
CCCR	pp+pc	pc
TCCR	cp+cp	cp
BCCR	pp	cc
THSC	pp+cp+cc	$\Delta\sigma/E$**
CHSC	pp+pc+cc	$\Delta\sigma/E$**

* See Appendix A for the terminology.

** $\Delta\sigma$ is the stress relaxation range in this test.

Comparisons of Eq. (9) (parameter values are given in ref. [18]) with the experimental data on Rene 80 (in high vacuum) and IN100 (coated) as reported by NASA [19] are shown in Figure 4 and 5, respectively. For the bulk failure of these two materials under the test conditions, environmental effects can be neglected. Therefore, Eq. (9) is suitable to describe the creep-fatigue behaviour very well. The present model unifies the SRP concept with the physics-based deformation decomposition rule, Eq. (1), and the associated deformation mechanisms.

5 Conclusions

A unified model is proposed to deal with dwell/creep-fatigue interactions over a broad temperature range, based on the physics based deformation decomposition rule and the associated mechanisms. The model considers fatigue crack nucleation and propagation in coalescence with creep/dwell damages (cavities or wedge cracks) along its path until final fracture.

The model describes the dwell sensitivity as a function of dwell time, stress, temperature and microstructure. Also, for creep-fatigue interaction, the model has been shown to reconcile the SRP concept.

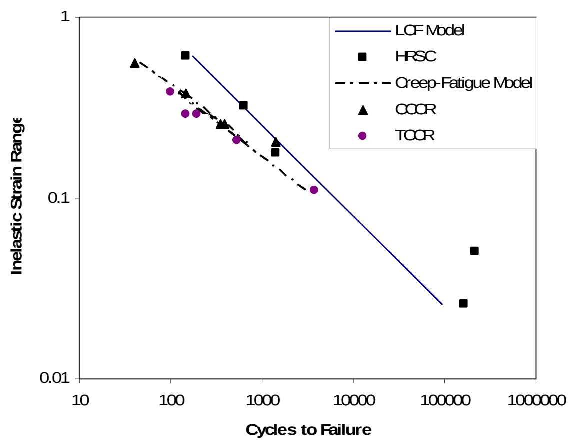


Figure 4. Comparisons of Eq. (9) with experimental data for Rene 80 at 871^{o}C.

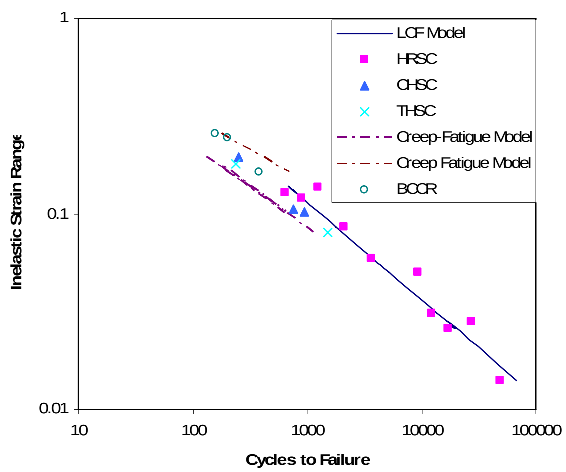


Figure 5. Comparisons of Eq. (9) with experimental data for IN 100 (coated) at 1000^{o}C.

Acknowledgments

The author would like to acknowledge the Defence Research and Development Canada of the Department of National Defence, Canada, for the financial support.

References

[1] M.A. Miner, Accumulative damage in fatigue, *Transactions*, J Appl Mech 67 (1945) A159-A167.

[2] E.L. Robinson, Effect of temperature variation on the long-time rupture strength of steels, Transactions ASME 74 (1952) 777-780.

[3] W.J. Ostergren, A damage function and associated failure equations for predicting hold time and frequency effects in high temperature low cycle fatigue, Journal of Test and Evaluation 4 (1976) 327-339.

[4] G.R. Halford, J.F. Saltsman, M.H. Hirschberg, Ductility-Normalized Strain Range Partitioning Life Relations for Creep-Fatigue Life Predictions, NASA TM-73737, 1977.

[5] R. Lagneborg, R. Attermo, The effect of combined low cycle fatigue and creep on the life of austenitic stainless steels, Metall Trans 2A (1971) 1821-1827.

[6] L.F. Coffin, A study of the effects of cyclic thermal stresses on a ductile metal, Transactions ASME 76 (1954) 931-950.

[7] S.S. Manson, Behavior of Materials under Conditions of Thermal Stress, NACA TN-2933, 1954.

[8] G.R. Halford, Evolution of creep-fatigue life prediction models, The Winter Annual Meeting of The American Society of Mechanical Engineers, ASME AD- 21 (1991) 43-57.

[9] H.J. Frost, M.F. Ashby, Deformation Mechanisms Maps, Pergamon Press, Oxford, 1982.

[10] N. Thompson, J.J. Wadsworth, Advances in Physics 7 (1958) 72-169.

[11] K. Tanaka, T. Mura, A dislocation model for fatigue crack initiation, J Appl Mech 48 (1981) 97-103.

[12] X.J. Wu, Evolution Life in Metallic Materials, AN-SMPL-2007-0070, Institute for Aerospace Research, National Research Council Canada, Ottawa, Canada, 2007.

[13] M.R. Bache, M. Cope, H.M. Davies, W.J. Evans and G. Harrison, Dwell sensitive fatigue in a near alpha titanium alloy at ambient temperature, Int J Fatigue 19 (1997) S83-S88.

[14] X.J. Wu, A continuously distributed dislocation model of Zener-Stroh-Koehler cracks in anisotropic materials, Int J Solids & Struct, 42 (2005) 1909-1921.

[15] X.J. Wu, P. Au, Deformation kinetics during dwell fatigue, Mater Sci Technol 23 (2007) 1446-1449.

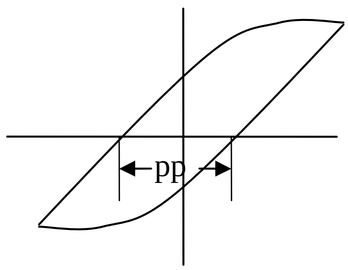
[16] X.J. Wu, A.K. Koul, Grain boundary sliding in the presence of grain boundary precipitates during transient creep, Metall Trans 26A (1995) 905-913.

[17] X.J. Wu, A.K. Koul, Modeling creep in complex engineering alloys, in: H. Merchant (Ed.), Creep and Stress Relaxation in Miniature Structures and Components, TMS, Warrendale, PA, 1996, pp. 3-19.

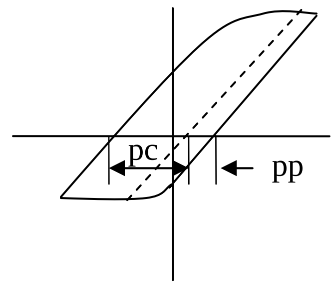
[18] X.J. Wu, A nonlinear model of dwell/creep-fatigue interaction, J Gas Turbine Powers (2008), in press.

[19] G. R. Romanoski, Jr., Mechanisms of Deformation and Fracture in High Temperature Low-Cycle Fatigue of Rene 80 and IN 100, NASA Contractor Report 165498, 1982.

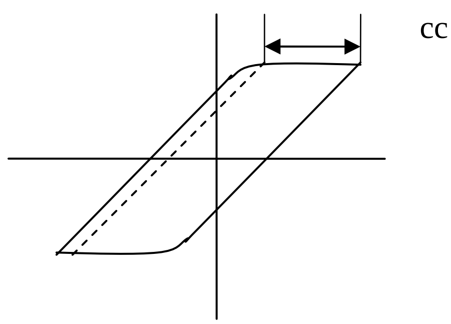
Appendix A The Cyclic Tests of SRP



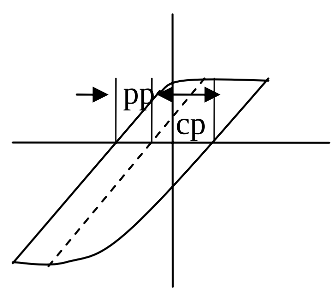
(1) High Rate Strain Cycle (HRSC)



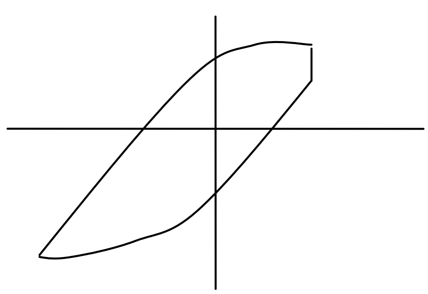
(2) Compressive Cyclic Creep Rupture (CCCR)



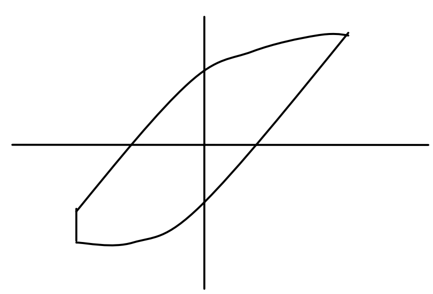
(3) Balanced Cyclic Creep Rupture (BCCR)



(4) Tensile Cyclic Creep Rupture (TCCR)



(5) Tensile Hold Strain Cycle (THSC)



(6) Compressive Hold Strain Cycle (CHSC)