HAL
open science

A Review on Damage Mechanisms, Models and Calibration Methods under Various Deformation Conditions

Jiali Lin, Yehan Liu, T. A. Dean

To cite this version:

Jiali Lin, Yehan Liu, T. A. Dean. A Review on Damage Mechanisms, Models and Calibration Methods under Various Deformation Conditions. International Journal of Damage Mechanics, 2005, 14 (4), pp.299-319. 10.1177/1056789505050357 . hal-00571146

HAL Id: hal-00571146

https://hal.science/hal-00571146

Submitted on 1 Mar 2011

HAL is a multi-disciplinary open access archive for the deposit and dissemination of scientific research documents, whether they are published or not. The documents may come from teaching and research institutions in France or abroad, or from public or private research centers.

L'archive ouverte pluridisciplinaire **HAL**, est destinée au dépôt et à la diffusion de documents scientifiques de niveau recherche, publiés ou non, émanant des établissements d'enseignement et de recherche français ou étrangers, des laboratoires publics ou privés.

A Review on Damage Mechanisms, Models and Calibration Methods under Various Deformation Conditions

J. LIN,* Y. LIU AND T. A. DEAN
Mechanical and Manufacturing Engineering
School of Engineering
The University of Birmingham
Birmingham, B15 2TT, UK

ABSTRACT: The development of microdamage under the deformation conditions of high temperature creep, cold metal forming, superplastic forming, and hot metal forming has been reviewed and discussed, and typical constitutive equations developed to model the individual damage mechanisms are summarized. Based on the microstructural analysis of the key damage features for metallic materials under a wide range of deformation conditions, a set of schematic diagrams is designed to illustrate the major types of damage mechanisms. This helps researchers and engineers to understand the major cause of failure of materials under different deformation conditions and to select simple and appropriate mechanism-based damage equations to predict the damage evolution. Further discussions are carried out on the dominant damage mechanisms in hot metal forming conditions and it is concluded that the dominant damage mechanism can be 'grain boundary (creep-type) damage' or 'plasticity-induced (ductile) damage' depending on the material microstructure and deformation rate. In the case of grain boundary damage in hot forming, the shape of microdefects is different from those in high temperature creep and superplastic forming although all of those result in intergranular failure of materials. Furthermore, damage calibration techniques for different conditions of plastic deformation are summarized and discussed.

KEY WORDS: damage mechanisms, damage calibrations, creep and viscoplasticity, cold and hot metal forming.

*Author to whom correspondence should be addressed.

1056-7895/05/04 0299–21 $10.00/0 DOI: 10.1177/1056789505050357
© 2005 Sage Publications

INTRODUCTION

THE PROGRESS OF failure in metals, under various loading conditions, is assumed to involve the degradation of structure due to nucleation and growth of defects, such as microvoids and microcracks, and their coalescence into macrocracks (Krajcinovic, 2000; Ibijola, 2002). This process, generically termed damage, was first used to predict the creep rupture of metals in-service at elevated temperatures (Kachanov, 1958). Although damage mechanics provides a measure of material degradation on a micromechanics scale, the damage variables are introduced to reflect average material degradation at a macromechanics scale and thus continuum damage mechanics (CDM) was developed. Extensive treatments of CDM have been presented in books by Rabotnov (1969), and Lemaitre and Chaboche (1990). In the theories, it is assumed that once the values of specified damage variables reach certain levels, the material cannot sustain the applied load any more and failure takes place. This has been used to study the failure of materials under high temperature creep (Hayhurst, 1972; Leckie and Hayhurst, 1977) and creep-cyclic plasticity (Chaboche, 1987; Lin et al., 1999). This concept has also been extended into metal forming to study the formability of materials, especially for cold forming (Gelin, 1998). It was discovered that void nucleation, growth, and coalescence take place under large plastic deformation and ductile void growth theories have been developed to predict failure of materials in metal forming (Bonora, 1997).

The macroscopic discontinuities in the workpiece are due to the development of microvoids or microcracks that grow under large straining conditions (Gelin, 1998; Bellenger and Bussy, 1998). From observations at a microscopic level, the main cause of the damage in metal forming is ductile fracture or intergranular fracture occurring in crystallographic materials (Bellenger and Bussy, 1998; Krajcinovic, 2000). Ductile fracture results from decohesion of the matrix material around inclusions or second-phase particles, then the microcavities or microvoids grow with increasing tensile strains and tend to form macroscopic cavities or cracks leading to macroscopic fracture (Lemaitre, 1984). Such a mechanism mainly occurs in cold forming conditions. In hot forming conditions, i.e., when the deformation temperature is greater than about half the melting temperature (T_m), the main cause of damage development is normally believed to be associated with intergranular damage (Gelin, 1998) that corresponds to the nucleation of cavities at grain boundaries (Semtiatin, et al., 1998). This phenomenon is controlled mainly by the viscoplastic strain rate and might be responsible for damage development in superplastic forming processes, where very low strain rates are applied (Du and Wu, 1995). However, at high temperature and high strain rate deformation, e.g., fatigue tests

on stainless steel at 625°C at 15–50 Hz (high strain rates), transgranular cracks were observed (Ahsam, 1997).

Based on the concept of Kachanov (1958), many constitutive equations have been developed to describe the phenomenological aspects of the creep damage process (Leckie and Hayhurst, 1977). In addition to rupture times, secondary and tertiary creep behavior of metallic materials can be well predicted using the phenomenological equations in which the material is treated as a continuum. Since the detailed process of degradation of the material is not examined, assumptions or postulates are made to describe the rate of damage evolution. The usual assumptions have a certain generality, which allows the resulting equations to be fitted to creep experimental data with a degree of success (Zhang and Lee, 1993); but they are not based on microstructural observations or physical reasoning. The material constants in these equations do not have clear physical meanings and the dominant damage mechanisms cannot be modeled using the equations. Material scientists studying creep damage are not content with this vague description of damage. The dissatisfaction is reinforced when attempts are made to model the growth of voids or cracks during creep, which can lead to equations that do not appear to resemble those of the continuum treatment. The weakness of the approach is further demonstrated by the obvious experimental fact that there are several mechanisms of creep damage, while the continuum equations appear to describe only one. Thus the research on creep damage has been extended into the area of categorizing damage mechanisms. Significant contributions have been made by Dyson (1990), who created diagrams to show identified creep damage mechanisms. Mathematical representations of the corresponding damage mechanisms, damage evolution, and their effects on creep deformation have been studied and developed. Based on the development of the understanding of the damage mechanisms, physically inspired, multivariable damage models have been proposed and used for the modeling of creep rupture of materials (Hayhurst et al., 1994).

Although the damage concept has been introduced to study the failure of materials in cold forming ($T < 0.3T_m$) since 1969 (Rice and Treacy, 1969) and microdamage has been observed in hot forming ($T < 0.6T_m$) by Cottingham (1966) and Dieter et al. (1966), research on the identification of dominant damage mechanisms, linking with texture evolution during deformation, and on the development of mechanism-based damage constitutive equations is behind that in creep damage. Current review papers only discuss the development of damage evolution and modeling methods for a particular deformation type, such as high temperature creep damage, ductile damage, or superplasticity damage. Thus one of the objectives of this paper is to design a set of diagrams to enable the distinct

damage mechanisms arising in various deformation conditions to be compared. The corresponding modeling methods, constitutive equations, and calibration techniques are also discussed and compared.

DAMAGE MECHANISMS AND MODELS

Damage Due to Multiplication of Mobile Dislocations

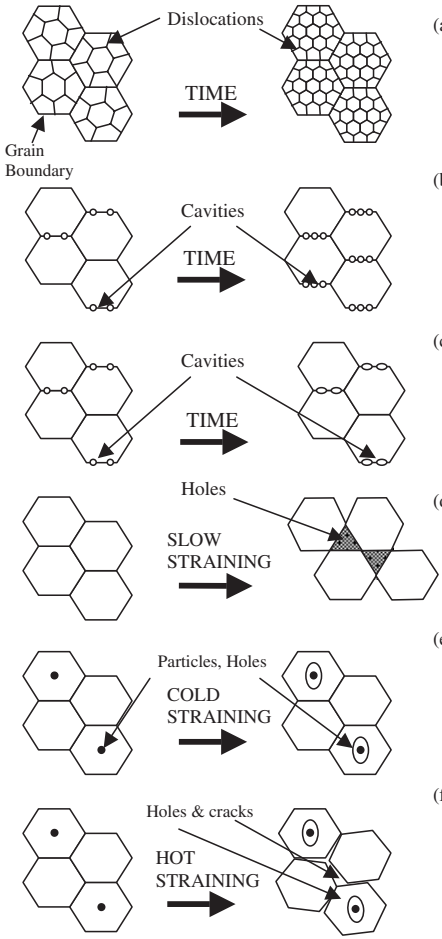
The microdamage mechanism responsible for this behavior is a progressive accumulation of mobile dislocations as metal creep proceeds at high temperatures (Dyson, 1988). It has been termed mobile dislocation strain-softening (Othman et al., 1993) and is shown schematically in Figure 1(a). Creep in these materials is therefore not controlled by dislocation recovery but is best thought of in terms of the kinetics of dislocation multiplication and subsequent motion (Ashby and Dyson, 1984a). By treating the velocity of dislocation around γ' particles as one of diffusive drift–whose spacing is very much less than that of the dislocations, secondary and tertiary creep in superalloys under various loading conditions can be modeled by the introduction of mobile dislocation damage. A typical damage model has been represented by Lin et al. (1993) as:

$$\dot{\omega}_1 = C(1 - \omega_1)^2 \dot{\varepsilon}_c \tag{1}$$

where, ω_1 $(= 1 - \rho_i/\rho)$ is the mobile dislocation damage and its evolution increases proportionally to creep rates $\dot{\varepsilon}_c$. ρ_i is the initial dislocation density, which is influenced by the materials processing route. ρ is the current dislocation density. The parameter C reflects the propensity of the material for entering tertiary creep and its magnitude is inversely proportional to ρ_i. This damage causes the softening of materials, i.e., increasing creep rates, but does not determine the failure of materials (Dyson, 1988).

Damage Due to Creep Constrained Cavity Nucleation and Growth

The damage mechanism, shown in Figure 1(b), is grain boundary cavity nucleation and growth. Its presence or absence is strongly sensitive to alloy composition and processing route; for example, it is clearly absent in single crystals. Its presence reduces the load bearing section and so accelerates creep and this, in turn, increases the rate at which the damage grows. At low stresses, the damage is void-like; at high stresses, the voids may link to form grain boundary cracks (Dyson and McLean, 1983). Many metals can fracture by this mechanism alone, though it is more usual for other



(a) **Mobile Dislocation**. Damage develops due to plastic deformation and recovers at high temperature. This occurs in creep and plastic deformation.

(b) **Cavity Nucleation**. This is creep-controlled and occurs at grain boundaries during low stress high temperature creep. Creep rate is very low, e.g. $\dot{\varepsilon} \propto 10^{-15}\ s^{-1}$.

(c) **Continuum Cavity Growth.** This is creep and grain boundary diffusion-controlled and occurs at grain boundaries at high stress high temperature creep. Creep rate is relatively high, e.g. $\dot{\varepsilon} \propto 10^{-10}\ s^{-1}$.

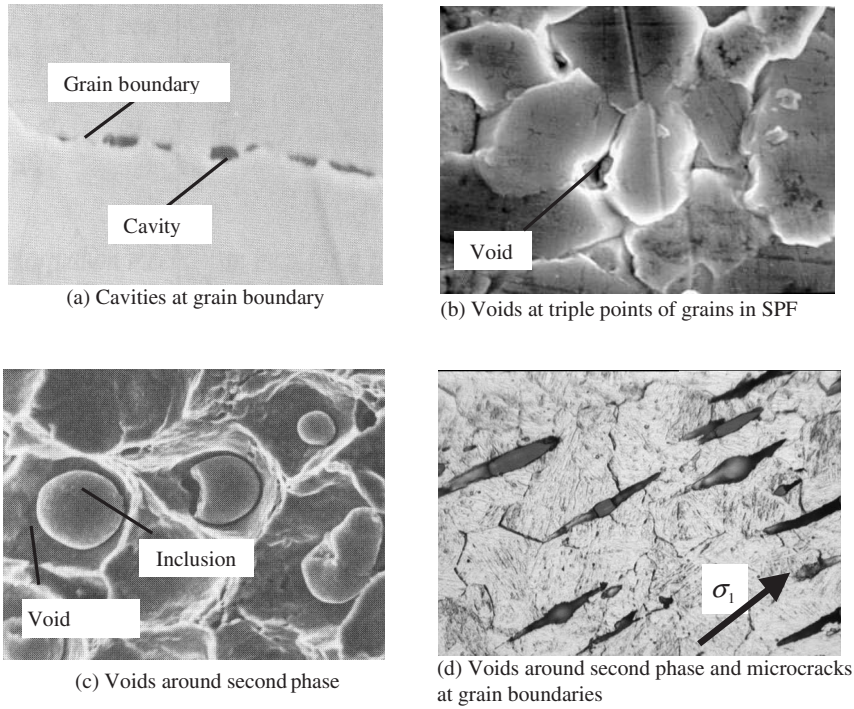
(d) **Superplastic Void Growth**. Damage created at triple points of grain boundaries, which occurs in superplastic deformation. Typically, $\dot{\varepsilon} \propto 10^{-4}\ s^{-1}$.

(e) **Ductile Void Growth**. This is plasticity-induced damage. Holes are nucleated and grown around second phase particles in cold metal forming.

(f) **Ductile Void Growth and Micro-Cracking**. This occurs at hot forming with high strain rates. Damage is generated at both grain boundaries and in second phases. Typically $\dot{\varepsilon} \propto 0.1\ s^{-1}$.

Figure 1. *Schematics showing dominant damage mechanisms for particular deformation conditions.*

mechanisms to contribute also. Grain boundary cavity formation is a kinetic phenomenon and its influence on deformation resistance and on fracture mode under arbitrary stress state depends critically on cavity nucleation rate and growth rate. When both rates are high, there is a potential for a strong coupling between cavitation and creep rate through the mechanism of creep constrained cavity growth, leading to rapid tertiary creep (Dyson and Loveday, 1981). This damage normally occurs during long-term high temperature creep at low stress levels. If the failure of a material is due to this mechanism, its ductility is very low. Figure 2(a) shows cavities observed



(a) Cavities at grain boundary

(b) Voids at triple points of grains in SPF

(c) Voids around second phase

(d) Voids around second phase and microcracks at grain boundaries

Figure 2. *Damage in (a) high temperature creep; (b) superplastic forming; (c) cold forming; and (d) hot forming.*

at grain boundaries. A typical equation for modeling damage evolution is given by Dyson (1990) and Lin et al. (1993):

$$\dot{\omega}_2 = D\dot{\varepsilon}_c \tag{2}$$

where, $D = \varepsilon_f/3$ and ε_f, the strain at failure under uniaxial tension, is a material constant. The damage variable $\omega_1 (= \pi d^2 N/4)$ describes grain boundary creep constrained cavitation. N is the number of constrained grain boundary facets per unit area and d the cavity diameter. Creep cavitation can either be nucleation or growth controlled, which is linear to strain rate $\dot{\varepsilon}_c$, and the failure criterion is $\omega_2 = 1/3$ (Dyson, 1990).

Damage Due to Continuum Cavity Growth

This type of damage is schematically shown in Figure 1(c). It is similar to cavity nucleation and growth, shown in Figure 1(b), and occurs during high

temperature creep, but at high stress levels. The difference in fracture mode, resulting from the two damage mechanisms, shown in Figure 1(b) and (c), was observed by Dyson et al. (1981). Again, creep damage occurs as voids or cracks often on grain boundaries as shown in Figure 2(a). A void can grow during creep by diffusion of atoms away from it, or by the plastic flow of the material, which surrounds it, or by the combination of both (Ashby and Dyson, 1984b). If the void growth is controlled by boundary diffusion alone, matter diffuses out of the growing void and plates onto the grain boundary. If surface diffusion is rapid, matter is distributed quickly within the void allowing its shape to remain near-spherical.

A void can grow by power-law creep of the surrounding matrix. Toward the end of life, when the damage is large, this mechanism always, ultimately, takes over. A model has been developed by Cocks and Ashby (1982) to calculate void growth rate based on grain boundary diffusion mechanisms. In simple tension, the zone within the damaged area, as shown in Figure 4, extends a little faster than the rest of the material, by a factor of $1/(1-\omega_3)$, where $\omega_3 = r_h^2/l^2$ is the damage due to continuum cavity growth. Here, l is the spacing of the growing voids and r_h the void radius. The matter is also constrained by its surroundings so that it dilates, causing the void to grow in volume, thereby increasing the damage. The rate of damage growth due to continuum cavity growth has been given by Cocks and Ashby (1982), as:

$$\dot{\omega}_3 = [1/(1-\omega_3)^n - (1-\omega_3)]\dot{\varepsilon}_m \tag{3}$$

where, $\dot{\varepsilon}_m$ is the minimum creep rate and n a constant.

Damage Due to Superplastic Void Growth

Superplasticity can be observed for some metals with fine grain size (<10 μm) deforming at an appropriate strain rate and temperature (Pervezentsev et al., 1992; Lin and Dunne, 2001). The superplastic behavior of metals has been used to form complex shaped lightweight engineering components, especially for aerospace applications (Mabuchi and Higashi, 1999). The dominant deformation mechanism for this type of plastic deformation is grain rotation and grain boundary sliding (Kim et al., 1999). Such relative displacement of grains is accommodated by the distribution of matter within the mantle adjacent to grain boundaries (Ridley, 1989; Pilling, 1985). When the accommodating process fails to meet the requirements imposed by deformation rate, stresses at grain boundaries are not relaxed sufficiently and consequently cavities nucleate (Vetrano et al., 1999).

When a cavity is present at a grain boundary, either nucleated during superplastic deformation or preexisting this deformation, it grows by one of the following mechanisms:

(i) Stress-directed vacancy diffusion (Raj and Ashby, 1975), which is also termed conventional diffusion. Under an optimum deformation condition (a suitable temperature and strain rate), the overall void growth in most superplastic materials is dominated initially by this mechanism. The void volume increases as vacancies are directed into the void along a contiguous grain boundary. This mechanism is dependent upon the cavity radius: the growth rate decreasing with increasing cavity radius.
(ii) Superplastic diffusion (Ridley, 1989). This is significant only in fine-grained materials, with a void diameter greater than the average grain size but less than $\approx 10\,\mu m$, which are deforming at a strain rate of $1 \times 10^4\,s^{-1}$ or less. This mechanism occurs as vacancies are directed into a cavity along more than one grain boundary. However, if the strain rate is greater than $1 \times 10^4\,s^{-1}$, void growth is superseded by the following damage mechanism.
(iii) Plastic deformation (Chokshi and Langdon, 1987). The occurrence of plastic deformation is due to the translation of neighboring grains which tends to enlarge grain boundary cavities. In general, this plastic-strain-controlled damage mechanism is the most important to overall void growth in the majority of superplastic materials.

The total volume of voids in a superplastic material can be obtained using precision density measurements or quantitative metallography. When the result is determined and presented in terms of the void volume fraction, its relationship with strain allows the dominant void growth mechanism to be identified. For the conventional diffusional growth, the volume fraction increases linearly with strain. For superplastic diffusional growth, the cube root of void volume fraction increases linearly with strain. Finally, for plastic-strain-controlled growth, the void volume fraction increases exponentially with strain (Raj and Ashby, 1975; Ridley, 1989).

Based on the dominant damage mechanisms of superplastic materials discussed above, two types of damage evolution equations have been developed. One assumes that the mechanism of void growth is due to stress-directed vacancy diffusion. The deformation of the cavitated material is described as a cavitated cylinder, the radius of which is equal to the cavity spacing l. For the behavior of the slab containing a spherical cavity of radius r_h, the proposed mechanism needs to account for the extent of the diffusion zone indicated by r_d. Assuming that the load shed by the diffusion zone is negligible, the slab can be seen as containing a hole of an 'effective' size r_d.

Thus the effective damage can be defined as $w_4 = r_d^2/l^2$ and the damage law proposed by Lin et al. (2002) is

$$\dot{\omega}_4 = D_1\omega_4^{n_1}\left(\dot{\varepsilon}_p\right)^{n_2} + D_2\left(\dot{\varepsilon}_p\right)^{-n_3}\cosh\left(D_3\varepsilon_p\right) \tag{4}$$

where, D_1, D_2, D_3, n_1, n_2, and n_3 are material constants. ε_p is the plastic strain.

Khaleel et al. (2001) developed a damage law for superplastic materials based on the void nucleation and growth around particles, either at grain boundaries or within grains, which is similar to the ductile damage equations by Gurson (1977). The damage parameter, ω_5, represents the volume fraction of voids and the evolution equation is given as:

$$\dot{\omega}_5 = \eta(1-\omega_5)\dot{\varepsilon}_p + \frac{F\left(\varepsilon_p\right)}{(1-\omega_5)}\sigma\dot{\varepsilon}_p \tag{5}$$

The parameter η is usually taken as a constant and F is a monotonic function of plastic strain ε_p, indicating that the rate of nucleation increases with increased strain. The first term of the equation contributes more significantly to the damage accumulation at the initial stage of deformation (low damage) and the second at the late stage (high damage).

Void growth occurring at triple points of grains is often found in superplastic forming, especially for aluminum alloys, as shown in Figure 2(b) and schematically represented in Figure 1(d). Superplastic deformation rates are much higher than creep rates. The fine-grained microstructure at high temperatures encourages grain boundary sliding and grain rotation. This dominant deformation mechanism causes the voids to be created at triple points of grains. Although the two types of damage equation described above can be used to model the softening of superplastic materials with good accuracy, there are no physically based equations available to model the void nucleation and growth at triple points due to grain rotation and grain boundary sliding.

Damage Due to Ductile Void Growth

In cold metal forming processes, multiplication of dislocations is the dominant deformation mechanism and no grain boundary sliding takes place (Li et al., 1994; Brust and Leis, 1992). This causes voids to be nucleated around second phases (Zheng et al., 1996), normally within grains, as shown in Figures 1(e), 2(c) and (d). Once a microvoid has been nucleated in a plastically deforming matrix, by either the debonding or

cracking of a second-phase particle or inclusion, the resulting stress-free surface of the void causes a localized stress and strain concentration in the adjacent plastic field (Tvergaard, 1990). With continuing plastic flow of the matrix, the void will therefore undergo volumetric growth and shape change, which amplifies the distortion imposed by the remote uniform strain-rate field (Staub and Boyer, 1996). The early work on the development of damage models was based on the modeling of the rate of change in the radius of the void, $\dot{R}_k$ ($k = 1, 2, 3$), in the principal directions of strain rates for the remote strain rate field (Rice and Tracey, 1969). Based on an average concept, the average rate of void growth in a unit cell is given (Li et al., 1994) by:

$$\dot{R} = R \cdot N \exp\left(\frac{3}{2}\frac{\sigma_m}{\sigma_Y}\right)\dot{\varepsilon}^p_{eq} \tag{6}$$

where, R is an initially spherical isolated void of radius, N constant. In this case, voids are assumed to be spherical.

Once the void is no longer spherical, the rates of change of three radii in the three directions of the principal plastic strain rates are different as the ellipsoidal void has different stiffness in these directions, the Rice and Tracy's formula with a power law of the radii ratio is induced by Boyer et al. (2002), where the spacing of the voids is sufficiently great. The growth of the void can have the form of (Boyer et al., 2002):

$$\dot{R}_k = \left[(1 + E)\dot{\varepsilon}_k + D\left(\frac{2}{3}\dot{\varepsilon}_L\dot{\varepsilon}_L\right)^{1/2}\right]R \tag{7}$$

where $(k, L) = 1, 2, 3$. The values of $(1 + E)$ and D are related to strain hardening of materials and are given by Thomason (1990). The void growth model is implemented into a yield function to model the shrinkage of yield surface of materials in plastic deformation.

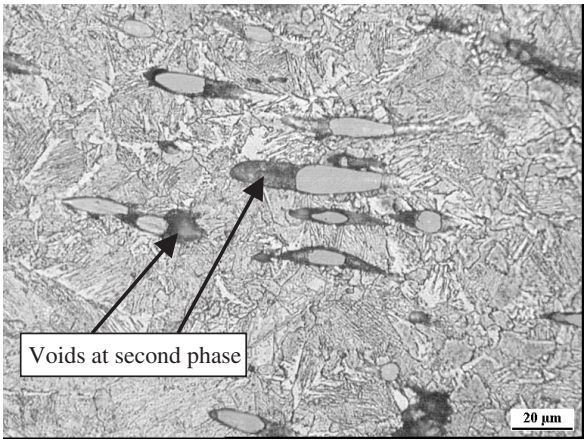
DAMAGE EVOLUTION IN HOT, HIGH STRAIN RATE METAL FORMING

According to the above analysis, it can be seen that the dominant damage mechanisms vary significantly between different deformation conditions (Bogatov, 2003). Creep-type damage is modeled through the analysis of diffusion during high temperature creep and the constitutive equations are developed based on the continuum damage mechanics theory. However, the constitutive equations predicting damage evolution in cold forming are

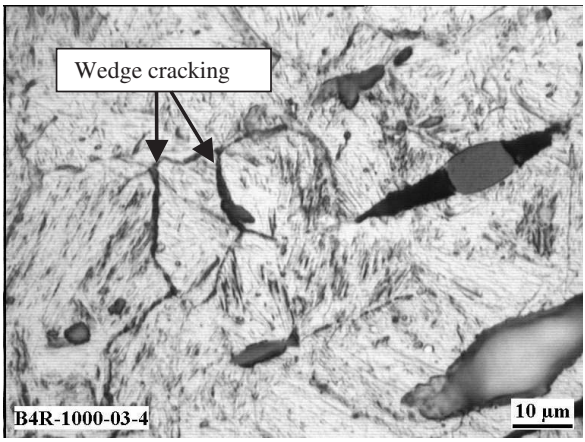
based on the modeling of void shape changes under plastic deformation. Although the two damage analyses use different models of microstructural change, both are able to model material softening due to microdamage.

Hot metal forming temperatures normally are above $0.6T_m$ and strain rates are often higher, e.g., $10\,s^{-1}$. The high temperature causes grain boundary diffusion, which facilitates grain rotation and grain boundary sliding. This deformation mechanism becomes more dominant if the grain size is small, temperature is high, and deformation rate is low, such as in superplastic materials and deformation conditions (Lin and Dunne, 2001). Damage occurs mainly around grain boundaries, as with 'creep-type' damage. However, in most hot metal forming processes, the initial material grain size is large and deformation rate is very high, e.g., the value is $\approx 10^3$ more than that of superplastic deformation rates and $\approx 10^{12}$ more than that of creep rates, which reduces grain boundary diffusion and sliding, and dislocation-based plasticity occurs. This results in plasticity-induced damage, which takes place around inclusions or hard particles. Due to the lack of appropriate evidence to demonstrate the damage features under hot metal forming conditions, a series of hot tensile tests have been carried out on a free-cutting steel at different strain rates. The microstructure of the tested specimens was examined and two typical micrographs are shown in Figure 3.

Figure 3 shows the microstructure after deformation at (a) a strain rate of $10\,s^{-1}$ and (b) a strain rate of $0.01\,s^{-1}$ at the same temperature of 1000°C. Microdamage can be observed at both grain boundaries and around second phases. However, it can be seen clearly that more damage exists at grain boundaries, for the low strain rate tests and almost all the voids are around second-phase regions at the higher strain rate test. The plasticity-induced damage, Figure 3(b), is similar to that observed in cold metal forming. However, the grain boundary damage is different from 'creep-type' and 'superplastic-type' damages. In creep damage, a multiplicity of very small voids arises at a single grain boundary (see Figure 2(a)) due to long-term diffusion under low stress, and, under superplastic conditions, voids are mainly at triple points (see Figure 2(c)) due to grain rotation and grain boundary sliding under medium stress and strain rates. In hot metal forming, both strain rate and stress are high. The high strain rate (short time) deformation results in little grain boundary diffusion taking place and reduces the chance for grain rotation and grain boundary sliding. Thus multiple voids are not found at grain boundaries nor are voids observed at grain triple points. Instead, microwedge cracking can be observed at grain boundaries for the hot forming conditions. It is believed that the grain boundary diffusion makes the grain boundary weaker and microcracks form under high flow stress.



(a) $\dot{\varepsilon} = 10.0\ s^{-1}$



(b) $\dot{\varepsilon} = 0.01\ s^{-1}$

Figure 3. *Comparison of damage evolution in hot forming ($T = 1000°C$) at (a) high and (b) low strain rates.*

The two types of damage arising in hot forming are schematically illustrated in Figure 1(f) and experimental observations are shown in Figures 2(d), 3(a) and (b). The proportion of the two types of damage varies with strain rate, temperature, and grain size. If microvoids and cracks are linked together, macrocracking occurs. No physically based constitutive equations are available to model these two hot forming damage mechanisms which occur simultaneously.

DAMAGE CALIBRATION TECHNIQUES

Significant efforts have been made on the identification of damage mechanisms for different materials under different plastic deformation conditions. Microvoids have been observed within solids and different damage levels at various deformation stages identified. However, the quantification of damage accumulation during plastic deformation is a very difficult task. Research has been carried out to identify microvoid accumulation in superplastically deformed aluminum alloys through microstructural examinations (Ridley, 1989). Figure 4 shows the relation of volume fraction of voids with plastic deformation of AA7475 for different strain rates. In general, void volume fraction increases with plastic strain. For high strain rate deformation, voids are relatively big and the volume fraction reaches about 2% close to failure for $\dot{\varepsilon} = 10^{-3}\,\mathrm{s}^{-1}$. But at the low strain rate, the void volume fraction close to failure is significantly lower at about only 0.25% for $\dot{\varepsilon} = 10^{-4}\,\mathrm{s}^{-1}$. In creep, the strain rate, e.g., $\dot{\varepsilon}_c = 10^{-15}\,\mathrm{s}^{-1}$, is much lower than that in superplastic deformation and the void volume fraction close to failure should be much lower than the void volume fractions observed at superplastic deformation conditions. This indicates that the microstructural examination of void growth is very difficult to relate the assumption of damage variables,

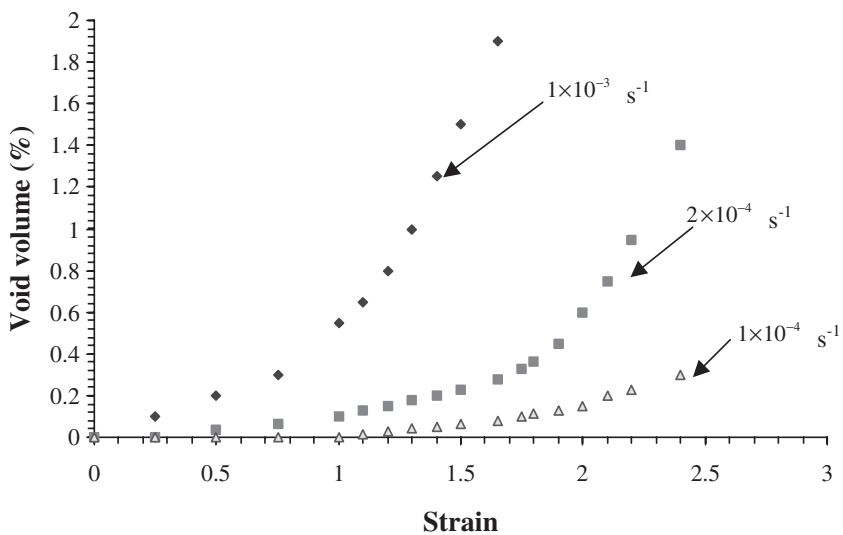


Figure 4. *Relations of cavity volume fraction to strain for different strain rates in superplastically deforming AA7475 alloy (Ridley, 1989).*

such as defined by Rabotnov (1969). Thus it is difficult to determine the damage evolution equations through the microstructural examination results. Another method used for the quantification of damage is to measure the change of value of Young's modulus in a loading, unloading, and reloading sequence (Chaboche, 1987). At present, the commonly used methods to calibrate the damage evolution are through the fitting of mechanical test results (Li et al., 2002), such as creep curves, stress–strain curves, etc., for different loading conditions, which is discussed in this paper for individual cases.

Calibration of Damage Evolution from Creep Curves

To determine the creep rupture behavior of a material at an elevated temperature, normally creep tests are carried out for a number of constant load/stress levels. For example, the symbols in Figure 5 show creep test results for an aluminum alloy at 150°C for three effective stress levels of 280, 300, and 320 MPa (Lin et al., 2003) and three stress states, they are (1) pure tension $(\sigma_{12}/\sigma_{11}) = 0$, (2) combined tension and torsion $(\sigma_{12}/\sigma_{11}) = \sqrt{3}/3$, and (3) pure torsion $(\sigma_{12}/\sigma_{11}) = \infty$, where σ_{11} is the axial stress due to tension and σ_{12} the shear stress due to torsion. It is believed that the acceleration of creep strain rates at the tertiary stage is due to the material damage. Thus the level of damage during the creep process for the material can be calibrated through the determination of creep damage constitutive equations by fitting the creep curves using optimization techniques. A detailed description of the techniques and a complete set of creep damage constitutive equations are given by Lin et al. (2003) and the best fitted results are shown by the solid curves in Figure 5.

Calibration of Damage Evolution from Cyclic Loading Test Results

Cyclic plasticity/viscoplasticity tests are normally carried out for the calibration of fatigue damage and for the determination of fatigue damage constitutive equations. These cyclic tests can be either strain range controlled or stress range controlled. Figure 6(a) shows experimental results for pure copper, (Dunne and Hayhurst, 1992), under strain range controlled cyclic viscoplastic loading conditions. It can be seen clearly that, at the initial stage of the cyclic tests, the stress range increases. This is due to isotropic hardening, which takes place only over the first few cycles. The drop of the stress range at the late stage of the cyclic tests is believed to be due to the accumulation of fatigue damage and also creep damage if the values of strain and temperature are sufficiently high. These experimental results have been used to calibrate damage development in cyclic

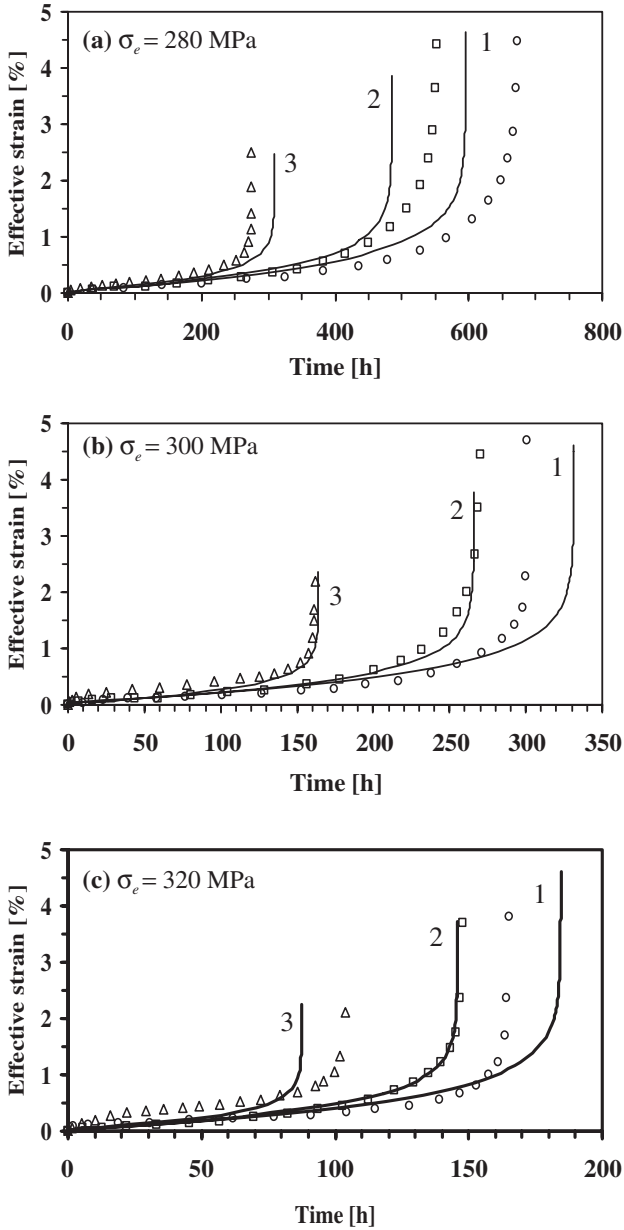
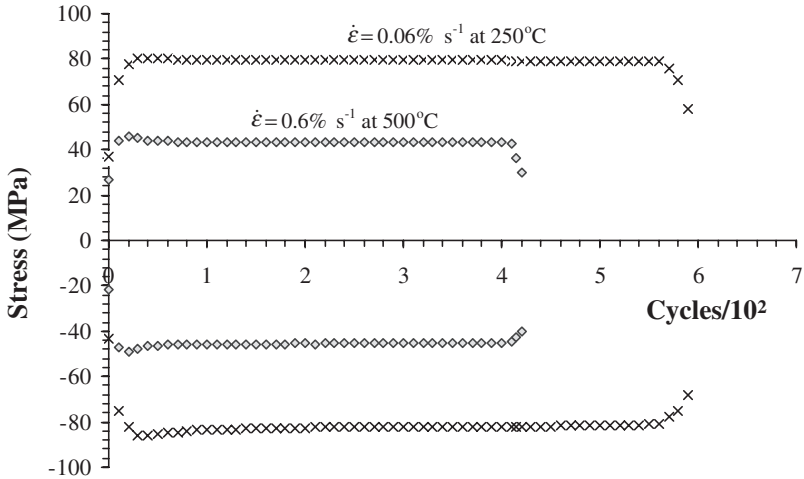
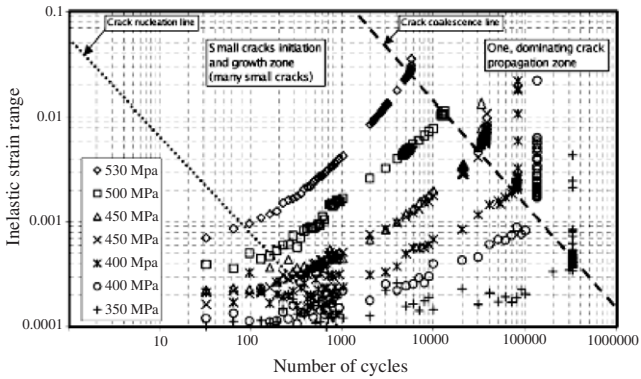


Figure 5. *Comparison of experimental (symbols) and computed (solid curves) effective strain creep curves for an aluminum alloy at 150°C for three effective stress levels and three stress-states. The number indicates types of stress-states: 1 – tension; 2 – tension and torsion; 3 – torsion (Lin et al., 2003).*



(a) Cyclic test results for copper over a strain range of 0.6% [Dunne and Hayhurst, 1992]



(b) Cyclic test results for steel at different stress amplitudes [Socha, 2003]

Figure 6. *Variation of (a) stress range for strain-controlled tests, and (b) strain range for stress-controlled tests against number of cycles in cyclic plasticity tests.*

viscoplasticity through the determination of the material constants in a set of cyclic viscoplasticity damage constitutive equations (Dunne and Hayhurst, 1992).

If the cyclic plasticity tests are stress range controlled, ratcheting takes place. This results in unequal compressive and tensile strains and an increased strain range. This is shown in Figure 6(b) for a steel tested at different stress ranges (Socha, 2003). It is believed that the increased strain range under the same cyclic stress range is due to material softening under

cyclic loading, resulting from the accumulation of microdamage. Based on this assumption, a new representation of damage variables is proposed as the measure of damage progression, as follows:

$$\omega = \sum \left(\frac{\Delta\varepsilon^i - \Delta\varepsilon_0^i}{\Delta\varepsilon_f^i - \Delta\varepsilon_0^i} \right)^{-1/C} \tag{8}$$

where, C is a constant, and $\Delta\varepsilon_0^i$, $\Delta\varepsilon^i$, and $\Delta\varepsilon_f^i$ are initial, current, and final values of the inelastic strain range. The damage variable varies from 0, the initial state, to 1, at which failure takes place. Compared to the damage definition given in Equation (1), the definition in Equation (10) provides a more convenient way to measure and calibrate the damage evolution for cyclic loading conditions, since the strain range variation over the cyclic loading is easier to measure. The small changes of effective stress (the effective cross-sectional area reduces due to materials damage) would result in significant changes in strain magnitudes. This enables the damage to be calibrated more accurately.

Calibration of Damage from Plastic and Viscoplastic Stress Strain Curves

In the early stage of elastic-plastic/viscoplastic deformation, the overall material hardening is mainly a result of the increase of dislocation density and grain growth (Lin and Dunne, 2001). At the later stage of deformation, softening due to damage dominates and flow stress decreases. The results of these phenomena can be seen in the experimental results shown by symbols in Figure 7 for Al–Zn–Mg deforming at a superplastic forming temperature of 515°C at strain rates of 2×10^{-3}, 5×10^{-4} and $2 \times 10^{-4}\,\mathrm{s}^{-1}$. The main purpose of implementing damage evolution equations into elastic-viscoplastic constitutive equations is to predict the ductility and reduction of flow stress in the late stage of deformation. Thus the procedure for calibrating damage evolution is to obtain a fit of calculated results with experimental data by determining the constants within a set of constitutive equations, using optimization techniques. The solid curves in Figure 7 show the best fit results, based on damage models and optimization procedures presented by Lin et al. (2002).

CONCLUSIONS

Damage constitutive equations for creep have been developed to model individual damage mechanisms during metal diffusion and creeping processes, which are known as mechanism-based creep damage constitutive

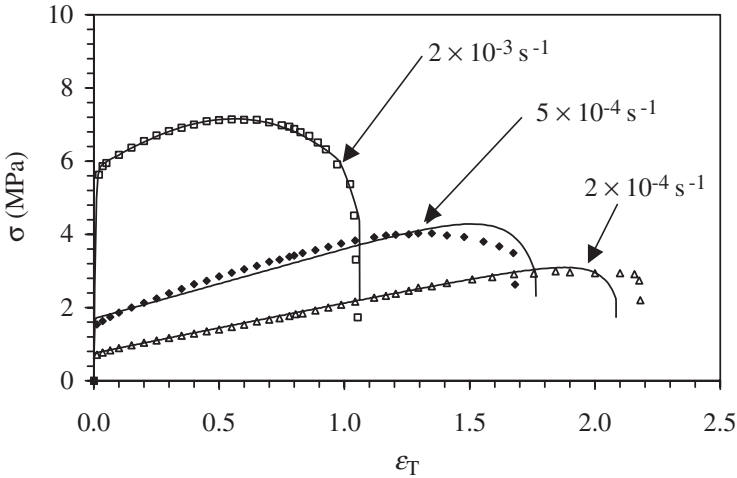


Figure 7. *Comparison of computed (solid curves) and experimental (symbols) stress–strain relationships for Al–Zn–Mg tested at a temperature of 515°C and different strain rates (Lin et al., 2002).*

equations. The 'plasticity-induced' damage models are developed based on the modeling of the void geometry evolution in cold metal forming conditions. Although efforts have been made to link the damage and texture evolution in cold metal forming processes, the 'metallurgy content' in ductile damage models is less than that in creep damage models.

Although there are some damage models available for superplastic deformation conditions, those models are not capable to predict damage evolution due to the dominant deformation mechanisms, i.e., grain boundary diffusion, grain boundary sliding, and grain rotation.

In hot metal forming conditions, damage appears at both grain boundaries and around second phases. The dominant damage mechanism, which depends on strain rate, temperature, microstructure, and material compositions, dynamically changes during forming processes due to dynamic recrystallization. The grain boundary damage in hot forming is not void-based, but microwedge cracking. Appropriate equations need to be developed to model the damage mechanisms and their interactions.

It also can be concluded that, at high temperature deformation conditions, the shape of the voids at grain boundaries varies from strain rates. Tiny voids are observed at grain boundaries for high temperature creep (e.g., $\dot{\varepsilon}_c = 10^{-15}\,\mathrm{s}^{-1}$), big voids at triple points for superplastic deformation (e.g., $\dot{\varepsilon} = 10^{-4}\,\mathrm{s}^{-1}$), and wedge cracks at grain boundaries for hot metal forming (e.g., $\dot{\varepsilon} = 10.0\,\mathrm{s}^{-1}$).

Significant research needs to be carried out on the development of physically based damage calibration methods, so that damage levels of a material can be measured easily from simple mechanical/materials testing results.

REFERENCES

Ahsam, Q. (1997). PhD Thesis, University of Birmingham, UK.

Ashby, M.F. and Dyson, B.F. (1984a). Creep Damage Mechanics and Micro-mechanisms, National Physical Laboratory, UK.

Ashby, M.F. and Dyson, B.F. (1984b). Creep Damage Mechanics and Micro-mechanisms, In: *ICF Advance in Fracture Research*, pp. 3–30, Pergamon Press.

Bellenger, E. and Bussy, P. (1998). Plastic and Viscoplastic Damage Models with Numerical Treatment for Metal Forming Processes, *J. of Materials Processing Technology*, **80**: 591–596.

Bogatov, A.A. (2003). The Mechanics of the Ductile Damage Under the Metal Forming, In: *The 6th Int. ESAFORM Conference on Metal Forming*, Salerno, Italy.

Bonora, N. (1997). A Non-linear CMD Model for Ductile Failure, *Eng. Fract. Mech.*, **58**(1/2): 11–28.

Boyer, J.C., Vidalsalle, E. and Staub, C. (2002). A Shear Stress Dependent Ductile Damage Model, *J. of Materials Processing Technology*, **121**: 87–93.

Brust, F.W. and Leis, B.N. (1992). A New Model for Characterizing Primary Creep Damage, *Int. J. Fract.*, **54**: 45–63.

Chaboche, J.L. (1987). Continuum Damage Mechanics: Present State and Future Trends, *Nuclear Engineering and Design*, **105**: 19–33.

Cocks, A.C.F. and Ashby, M.F. (1982). On Creep Fracture by Void Growth, *Progress in Material Science*, **27**: 189–244.

Chokshi, A.H. and Langdon, T.G. (1987). A Model for Diffusional Cavity Growth in Superplasticity, *Acta Metallurgica*, **35**: 1089–1101.

Cottingham, D.M. (1966). The Hot Workability of Low-carbon Steels, *Proceedings of The Conference on Deformation Under Hot working Conditions*, pp. 146–156.

Dieter, G.E., Mullin, J.V. and Shapiro, E. (1966). Fracture of Inconel under conditions of hot working, *Proceedings of The Conference on Deformation Under Hot Working Conditions*, pp. 7–12.

Du, Z. and Wu, S. (1995). A Kinetic Equation for Damage During Superplastic Deformation, *J. of Material Processing Technology*, **52**: 270–279.

Dunne, F.P.E. and Hayhurst, D.R. (1992). Continuum Damage Based Constitutive Equations for Copper Under High Temperature Creep and Cyclic Plasticity, *Proc. R. Soc. Lond.*, A, **437**: 545–566.

Dyson, B.F. (1988). Creep and Fracture of Metals: Mechanisms and Mechanics, *Rev. Phys. Appl.*, **23**: 605–613.

Dyson, B.F. (1990). Physically-based Models of Metal Creep for use in Engineering Design, In: Embury, J.D. and Thompson, A.W. (eds) *Modelling of Materials Behaviour and Design*, pp. 59–75, The Materials, Metals and Materials Society, London, UK.

Dyson, B.F. and Loveday, M.S. (1981). Creep Fracture in Nomonic 80A Under Triaxial Tensile Stressing, In: Ponter, A.R.S. and Hayhurst, D.R. (eds) *Creep in Structure*, pp. 406–421, Springer-Verlag, Berlin.

Dyson, B.F. and McLean, M. (1983). Particle-coarsening, σ_0 and Tertiary Creep, *Acta. Met.*, **30**: 17–27.

Dyson, B.F., Verma, A.K. and Szkopiak, Z.C. (1981). The Influence of Stress State on Creep Resistance: Experimentation and Modeling, *Acta Met.*, **29**: 1573–1580.

Gelin, J.C. (1998). Modelling of Damage in Metal Forming Processes, *J. of Materials Proc. Tech.*, **80–81**: 24–32.

Gurson, A.L. (1977). Continuum Theory of Ductile Rupture by Void Nucleation and Growth: Part-I yield Criteria and Flow Rule for Porous Ductile Metals, *J. Eng. Mat. & Tech.*, **99**: 2–15.

Hayhurst, D.R. (1972). Creep Rupture under Multi-axial States of Stress, *J. Mech. Phys. Solids*, **20**: 381–390.

Hayhurst, D.R., Dyson, B.R. and Lin, J. (1994). Breakdown of the Skeletal Stress Technique for Lifetime Prediction of Notched Tension Bars due to Creep Crack Growth, *Engineering Fracture Mechanics*, **49**(5): 711–726.

Ibijola, E.A. (2002). On Some Fundamental Concepts of Continuum Damage Mechanics, *Computer Methods in Applied Mechanics and Engineering*, **191**: 1505–1520.

Kachanov, L.M. (1958). *The Theory of Creep*, Kennedy A.J. (ed., English Translation), National Lending Library, Boston Spa.

Khaleel, M.A., Zbib, H.M. and Nyberg, E.A. (2001). Constitutive Modelling of Deformation and Damage in Superplastic Materials, *Int. J. of Plasticity*, **17**: 277–296.

Kim, J.S., Kim, J.H., Lee, Y.T. Park, C.G. and Lee, C.S. (1999). Microstructural Analysis on Boundary Sliding and its Accommodation Mode during Superplastic Deformation Ti-6AL-4V Alloy, *Materials Science and Engineering*, A**263**: 272–280.

Krajcinovic, D. (2000). Damage Mechanics: Accomplishments, Trends and Needs, *J. of Solids and Structures*, **37**: 267–277.

Leckie, F.A. and Hayhurst, D.R. (1977). Constitutive Equations for Creep Rupture, *Acta Metallurgica*, **25**: 1059–1070.

Lemaitre, J. (1984). How to use Damage Mechanics, *Nuclear Engineering and Design*, **80**: 233–235.

Lemaitre, J. and Chaboche, J.L. (1990). Solid Mechanics, Press Syndicate of the University of Cambridge, Cambridge, UK.

Li, Z.H., Bilby, B.A. and Howard, I.C. (1994). A Study of the Internal Parameters of Ductile Damage Theory, *Fatigue Fract. Engng. Mater. Struct.*, **17**(9): 1075–1087.

Li, B., Lin, J. and Yao, X. (2002). A Novel Evolutionary Algorithm for Determining Unified Creep Damage Constitutive Equations, *Int. J. of Mech. Sci.*, **44**: 987–1002.

Lin, J., Cheong, B.H. and Yao, X. (2002). Universal Multi-objective Function for Optimizing Superplastic Damage Constitutive Equations, *J. of Materials Proc. Tech.*, **125–126**: 199–205.

Lin, J. and Dunne, F.P.E. (2001). Modelling Grain Size Evolution and Necking in Superplastic Blow-forming, *Int. J. of Mech. Sci.*, **43**: 595–609.

Lin, J., Dunne, F.P.E. and Hayhurst, D.R. (1999). Aspects of Testpiece Design Responsible for Errors in Cyclic Plasticity Experiments, *Int. J. of Damage Mechanics*, **8**(2): 109–137.

Lin, J., Hayhurst, D.R. and Dyson, B.F. (1993). The Standard Ridges Uniaxial Testpiece: Computed Accuracy of Creep Strain, *J. of Strain Analysis*, **28**(2): 101–115.

Lin, J., Kowalewski, Z.L. and Cao, J. (2003). Modelling of Effects of Stress-states on Creep Damage for Copper and Aluminium Alloy, Dislocations, Plasticity and Metal Forming (Editor A.S. Khan), Neat Press, Maryland, USA. pp. 142–144.

Mabuchi, M. and Higashi, K. (1999). On accommodation of helper mechanism for superplasticity in Metal Matrix Composites, *Acta Mater.*, **47**(6): 1915–1922.

Othman, A.M., Hayhurst, D.R. and Dyson, B.F. (1993). Skeletal, Tertiary Creep in Circumferentially Notched Tension Bars undergoing Tertiary Creep Modelled with Physically-based Constitutive Equations, *Pro. R. Soc. Lond.*, **441**: 345–358.

Pervezentsev, V.N., Rybin, V.V. and Chuvil'deev, V.N. (1992). Overview No. 97: The Theory of Structural Superplasticity – I: the Physical Nature of the Superplasticity Phenomenon, *Acta Metallurgica Materiallia*, **40**(5): 887–894.

Pilling, J. (1985). Effect of Coalescence on Cavity Growth During Superplastic Deformation, *Materials Science and Technology*, **1**: 461–465.

Rabotnov, Y.N. (1969). In: *Creep Problems in Structural Members*, North Holland Publishing Company, Amsterdam.

Raj, R. and Ashby, M.F. (1975). Intergranular Fracture at Elevated Temperatures, *Acta Metallurgica*, **23**: 653–666.

Rice, R.J. and Tracey, D.M. (1969). On the Ductile Enlargement of Voids in Triaxial Stress Fields, *J. Mech. Phys. Solids*, **17**: 201–217.

Ridley, N. (1989). Cavitations and Superplasticity, In *Superplasticity*, *AGARD Lecture Series No. 168*, Essex: Specialised Printing Services Limited, pp. 4.1–4.14.

Semiatin, S.L., Seetharaman, V., Ghosh, A.K., Shell, E.B., Simon d, M.P. and Fagin, P.N. (1998). Cavitation during Hot Tension Testing of Ti–6Al–4V, *Materials Science and Engineering*, A**256**: 92–110.

Socha, G. (2003). Experimental Investigations of Fatigue Cracks Nucleation, Growth and Coalescence in Structural Steel, *Int. J. of Fatigue*, **25**: 139–147.

Staub, C.J. and C. Boyer, (1996). An Orthotropic Damage Model for Visco-plastic Materials, *J. of Materials Processing Technology*, **60**: 297–304.

Thomason, P.F. (1990). In: *Ductile Fracture of Metals*, Pergamon Press, UK.

Tvergaard, V. (1990). Material Failure by Void Growth to Coalescence, *Adv. App. Mech.*, **27**: 83–147.

Vetrano, J.S., Simonen, E.P. and Bruemmer, S.M. (1999). Evidence for Excess Vacancies at Sliding Grain Boundaries During Superplastic Deformation, *Acta Metallia*, **47**(15): 4125–4149.

Zhang, P. and Lee, H. (1993). Creep Damage and Fracture at High Temperature, *Eng. Fract. Mech.*, **44**(2): 283–288.

Zheng, M., Hu, C., Luo, Z.J. and Zheng, X. (1996). A Ductile Damage Model Corresponding to the Dissipation of Ductility of Metal, *Engineering Fracture Mechanics*, **53**(4): 653–659.