Plastic Flow Behaviour of Inconel 690 Super Alloy Under Compressive Impact Loading

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This paper studies the dynamic plastic flow behaviour, fracture characteristics, and microstructural evolution of Inconel 690 alloy under impact loading conditions. Compressive impact tests are performed using the compressive split-Hopkinson bar at strain rates ranging from 2.3×10^3 to $8.3 \times 10^3 \text{ s}^{-1}$ at room temperature. The effects of strain rate on the dynamic flow response, work hardening characteristics, strain rate sensitivity, and thermal activation volume are evaluated. A constitutive law based on the Zerilli-Armstrong model is proposed to describe the impact flow behaviour. The relationships between flow stress, dislocation, and twinning are analyzed and discussed in relation to the loading conditions. The evolutions of the dislocation and twinning substructures are investigated using transmission electron microscopy. Damage initiation and the fracture mechanisms are evaluated by scanning electron microscopy. The flow stress-strain response is shown to be significantly dependent on the strain rate, which not only causes obvious changes of the work hardening rate, strain rate sensitivity, and activation volume, but also influences the dislocation tangling and deformation twin substructures. The results indicate that the catastrophic failure at high strain rates results from the formation of localized shear bands. Finally, it is shown that the proposed constitutive law provides accurate predictions of the stress-strain relationship.

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1. Introduction

Inconel alloy 690, a high-chromium nickel alloy, has good formability, high strength and excellent corrosion resistance. As such, it finds extensive use in applications such as combustion systems, turbines, and nuclear reactors. In such applications, components are typically required to be tolerant of high stress impact loading, which may cause material degradation and lead to catastrophic failure during service operation. In order to design optimal Inconel 690 structural components, particularly when they are to be applied in critical systems, or are intended for use under arduous operating conditions, developing a thorough understanding of the material's impact properties and fracture characteristics as a function of strain rate is essential.

It has been reported that strain rate plays an important role in the impact flow behaviour of various materials. Further, it has been found that the strain rate sensitivity increases steeply above a critical strain rate of approximately 10^3 s^{-1} .¹⁻³⁾ In most cases, and particularly in FCC metals, the flow stress increases gradually with strain rate below the critical strain rate, but increases dramatically with the strain rate once the critical strain rate has been surpassed. Regazzoni *et al.*⁴⁾ proposed that this transition is prompted by the change in the dominant dislocation motion rate-controlling mechanism from the cutting of point obstacles (thermal activation) to viscous phonon drag. However, Zerilli and Armstrong⁵⁾ suggested that this behaviour can be more accurately interpreted as a result of enhanced rates of dislocation generation and twin formation.

The flow stress-strain response reflects the existence of different microstructural states during deformation. Two well-known deformation mechanisms, *i.e.* dislocation slip and twinning, dominate plastic flow when materials are subjected to impact loading. The onset of twinning, however,

requires the application of a greater stress than dislocation slip. At high strain rates, dislocation multiplication occurs more rapidly and provides a strengthening effect.^{6,7)} Dislocation multiplication and twinning both enhance the flow strength of metals.^{8,9)} Various investigators have attempted to characterize Inconel 690 in terms of its elevated temperature hardening mechanisms,¹⁰⁾ fatigue crack growth rate,¹¹⁾ machinability,¹²⁾ corrosion properties,¹³⁾ and tensile properties.¹⁴⁾ However, little research has been reported regarding this alloy's impact behaviour and fracture response. Consequently, the present study adopts the split-Hopkinson bar technique to evaluate the effects of strain rate on its impact flow, fracture response, and microstructure. Additionally, the interrelationships between the flow properties and microstructures of Inconel 690 are established, and the variations of fracture surface features induced by different impact loading rates are presented and analyzed.

2. Experimental Procedure

Inconel 690 with a nominal composition of 29.33% Cr, 10.01% Fe, 0.29% Ti, 0.19% Al, 0.10% Cu, 0.03% Co, 0.02% Mo, 0.02% W, 0.01% C (in mass %), and a balance of Ni was supplied by Inco Alloys, Inc. (USA) in the form of solution-annealed plate. Cylindrical specimens of height and diameter 5 mm were machined with an electrodischarge machine (EDM). All specimen surfaces were then mechanically polished with emery paper to #1500 and subsequently buff-polished with Al_2O_3 powder (0.3 μm particle size) to assure a close contact during impact deformation testing. Both ends of the specimen were lubricated with molybdenum disulfide grease to minimize friction effects.

Low strain rate compression tests were carried out using a tensile testing machine at strain rates of 10^{-3} , 10^{-2} and 10^{-1} s^{-1} . Compression impact tests were performed on a compressive split-Hopkinson pressure bar apparatus¹⁵⁾ at strain rates ranging from 2.3×10^2 to $8.3 \times 10^3 \text{ s}^{-1}$. During

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compressive impact testing, the cylindrical specimens were located between two high-strength tool steel bars, *i.e.* the incident bar and the transmitter bar, through which a compression wave was passed. Due to variations of mechanical impedance as the wave passes through the specimen, the incident wave (ε_i) is partially reflected (ε_r) and partially transmitted (ε_t). From one-dimensional elastic-wave propagation theory, it can be shown that the compression strain (ε), strain rate ($\dot{\varepsilon}$), and stress (σ) in the specimen are given by:

$$\varepsilon = \frac{-2C_0}{L} \int_0^t \varepsilon_r dt \quad (1)$$

$$\dot{\varepsilon} = \frac{-2C_0}{L} \varepsilon_r \quad (2)$$

$$\sigma = \frac{EA}{A_s} \varepsilon_t, \quad (3)$$

where C_0 is the longitudinal-wave velocity in the split-Hopkinson bar, L is the effective gage length of the specimen, E is the Young's modulus of the split bar, and A and A_s are the cross-sectional areas of the split bar and specimen, respectively.

Following testing, transmission electron microscope (TEM) specimens were sliced perpendicularly to the compression axis and ground to a thickness of 0.2 mm using emery paper with grade from #800 to #1500. Subsequently, discs of 3 mm diameter were extracted from these ground slices using an EDM process. The discs were electrolytically polished at 20 V D.C. in a twin-jet polishing machine using a solution of 10% perchloric acid and 90% methanol at -20°C . The thin foils were then examined under a scanning transmission electron microscope (STEM) operated at 300 kV. In accordance with the method outlined by Ham,¹⁶⁾ the dislocation density was determined as $\rho = 2n/Lt$, where n is the number of intersections between a dislocation and a random set of lines of length L , and t is the foil thickness. Meanwhile, the twin density was defined as the average number of twins per unit area within the visible region of the thin foil specimen. Finally, the fracture appearances were observed under a scanning electron microscope (SEM) operated at 15 kV.

3. Results and Discussion

3.1 Flow stress-strain response

Figure 1(a) plots the true stress-strain response of the compressive flow for Inconel 690 at room temperature as a function of strain rate. Although the strain rate sensitivity at low strain rates (between 10^{-3} and 10^{-1} s^{-1}) is quite small, in the impact region, the flow stress is seen to be more sensitive to the strain rate. Specifically, the flow stress increases by 50% as the strain rate increases from 2.3×10^3 to $8.3 \times 10^3 \text{ s}^{-1}$. This enhancement of flow stress with strain rate indicates that Inconel 690 is suitable for high velocity impact applications. Different strain rates result not only in differences in the flow stress, but also in the fracture strain. As shown in Fig. 1(a), fracturing does not occur at strain rates below $3.8 \times 10^3 \text{ s}^{-1}$. However, at strain rates exceeding $5.1 \times 10^3 \text{ s}^{-1}$, the alloy fails in a catastrophic manner and the fracture strain is seen to decrease gradually with increasing

(a)

(b)

Fig. 1 (a) Typical true stress-strain curves of Inconel alloy 690 at different strain rates, (b) Variation in work hardening rate of Inconel 690 as a function of strain and strain rate.

strain rate. The observed stress-strain response in Fig. 1(a) can be described by a power law of the form: $\sigma = A + B\varepsilon^n$ where A represents a constant (commonly equal to the yield strength), B is the material constant, and n is the work hardening coefficient. Table 1 presents the variations of A , B and n with strain rate, and also indicates the corresponding fracture strain and fracture stress data. Although the yield stress and the fracture stress both increase with strain rate, the material constant, B , and the work hardening coefficient, n , show no significant variation.

As implied by the tendencies of the stress-strain curves in Fig. 1(a), the work hardening rate is significantly affected by the strain rate in the range 2.3×10^3 to $8.3 \times 10^3 \text{ s}^{-1}$. Fig. 1(b) presents the variation of the work hardening rate,

Table 1 Mechanical properties of Inconel 690 impacted at different strain rates.

Strain rate $\dot{\epsilon}$ (s^{-1})	Yield Stress (A) (MPa)	Materials constant (B) (MPa)	Work hardening coefficient (n)	Fracture strain	Fracture stress (MPa)
10^{-3}	280	626	0.46	—	—
10^{-2}	284	640	0.46	—	—
10^{-1}	289	666	0.47	—	—
2.3×10^3	519	779	0.58	—	—
3.8×10^3	596	773	0.57	—	—
5.1×10^3	669	768	0.56	0.84	1343
6.6×10^3	769	750	0.55	0.78	1408
8.3×10^3	861	762	0.55	0.74	1479

$\theta = (d\sigma/d\epsilon)$, as a function of true strain for each strain rate. It is clear that at each strain rate, the work hardening decreases with increasing strain. An important feature of Fig. 1(b) is the significantly enhanced work hardening of the material under impact conditions. Comparing the effects of strain rate on the work hardening rate for a fixed strain in the impact regime, it is found that at small strains (0 to 0.1), the work hardening rate increases only slightly with increasing strain rate. However, at strains exceeding 0.1, dynamic softening has an increasing effect on the work hardening rate. Lower work hardening rates correspond to higher strain rate loading conditions, indicating that high-speed deformation generates thermal softening. This factor plays a key role in determining the deformation behaviour and work hardening rate.

3.2 Strain rate dependence and activation volume

The dependence of flow stress on strain rate can be illustrated by plotting the flow stress values, at specific strain, as a function of the logarithmic strain rate. Figure 2(a) shows the variation of true stress with strain rate for fixed strains of 0.1, 0.3, 0.5, and 0.7. Note that for comparison purposes, the low strain rate loading data (10^{-3} to $10^{-1} s^{-1}$) are also plotted. As expected, two distinct regions of strain rate sensitivity can be identified. The transition between these regions takes place at a strain rate of $10^3 s^{-1}$. It is noted that the flow stress increases gradually at strain rates below $10^3 s^{-1}$, but that at higher strain rates, it increases dramatically. Furthermore, comparing the slope of the flow stress at different strains, it is found that the slope varies slightly with strain rate. The rapid increase of strain rate sensitivity at strain rates above $10^3 s^{-1}$ suggests that the domination of dislocation generation and twin formation in the deformation process brings about a change in the dislocation motion rate-controlling mechanism.⁵⁾ This supposition is confirmed by the morphological observations of the dislocation and twin structures discussed below.

The strain rate sensitivity, as defined by the slope of the flow stress versus the strain rate, is calculated from the expression $\lambda = (\partial\sigma/\partial \ln \dot{\epsilon})_{\epsilon,T}$ using the data of Fig. 2(a). The strain rate sensitivities for the two strain rate ranges, *i.e.* 10^{-3} to $10^3 s^{-1}$ (λ_s) and 10^3 to $8.3 \times 10^3 s^{-1}$ (λ_d), are listed in Table 2. It is observed that λ_d is higher than λ_s for a given true strain level. As the true strain increases, both λ_d and λ_s are

(a)

(b)

Fig. 2 (a) Influence of strain rate on flow stress under different plastic strains, (b) Strain rate sensitivity and activation volume as functions of work hardening stress.

Table 2 Comparison of the dynamic (λ_d) and static (λ_s) strain rate sensitivity obtained at different true strain levels.

Strain rate sensitivity (MPa)	True strain			
	0.1	0.3	0.5	0.7
λ_s	7.4	7.0	6.5	6
λ_d	244	234	226	223

seen to decrease. In order to correlate the dynamic strain rate sensitivity and the activation volume with different work hardening characteristics and to track the corresponding microstructural evolution during deformation, the strain rate sensitivity and activation volume are typically plotted against the work hardening stress ($\sigma - \sigma_y$). On the assumption that the strain rate, $\dot{\epsilon}$, is controlled by a thermal activation process

with a free energy of activation, G^* , an Arrhenius-type law in the form of $\dot{\varepsilon} = \dot{\varepsilon}_0 \exp[-\Delta G^*/(K_b T)]$ can be used to describe the deformation behaviour. The impact strain rate sensitivity, λ_d , and activation volume, v^* , can then be related by:

$$(\delta \ln \dot{\varepsilon} / \delta \sigma^*)_T = -(\delta G^* / \delta \sigma^*) / K_b T = v^* / K_b T \quad (4)$$

where σ^* is the increment of the flow stress, *i.e.* $(\sigma_2 - \sigma_1)$, in which the flow stresses σ_2 and σ_1 are obtained from tests conducted at constant strain rates of $\dot{\varepsilon}_2$ and $\dot{\varepsilon}_1$, respectively, and are calculated at the same value of strain and at the same temperature. K_b is the Boltzmann constant.

Using this relation, the values of strain rate sensitivity and activation volume can be calculated and presented as a function of the work hardening stress, as shown in Fig. 2(b). A linear relationship is observed between the dynamic strain rate sensitivity, λ_d , and the work hardening stress. The enhancement of dynamic strain rate sensitivity with increasing work hardening stress can be attributed to the increased rate of dislocation multiplication and twin formation with increasing strain and strain rate during the deformation process. Meanwhile, the activation volume is seen to decrease rapidly as the work hardening stress increases. This implies that the ability of thermal activation to assist the movement of dislocations is restricted under conditions of increasing strain and strain rate. It should be noted that although the results of Fig. 2(b) and Table 2 are calculated from the formulation $\lambda = (\partial \sigma / \partial \ln \dot{\varepsilon})_{\varepsilon, T}$, Fig. 2(b) is based on the data of Fig. 1(a) for strain rates ranging from 2.3×10^3 to $8.3 \times 10^3 \text{ s}^{-1}$ and true strains ranging from 0.05 to 0.7 continuously, whereas Table 2 is based on Fig. 2(a) for strain rates ranging from 10^3 to $8 \times 10^3 \text{ s}^{-1}$ and true strains of 0.1, 0.3, 0.5 and 0.7. In addition, according to Fig. 1(a), the strain rate sensitivity, λ_d , is calculated by first selecting a true strain in the range of 0.05 to 0.7 under strain rates ranging from 2.3×10^3 to $8.3 \times 10^3 \text{ s}^{-1}$ through the formulation of $\lambda_d = (\partial \sigma / \partial \ln \dot{\varepsilon})_{\varepsilon, T}$, then identifying the corresponding work hardening stress under the same strain and strain rate range. This process is repeated iteratively with different values of true strain from 0.05 to 0.7. By plotting the calculated strain rate sensitivity, λ_d , with the corresponding work hardening stress for each true strain, the relationship between the strain rate sensitivity and the work hardening stress is obtained.

3.3 Deformation constitutive equation

Since the flow stress of most materials is strongly affected by the deformation parameters of strain, strain rate and temperature, numerous approaches have been proposed for developing constitutive relationships to describe stress-strain behaviours under high strain rate loading. The present study adopts the Zerilli-Armstrong model¹⁷⁾ to evaluate the impact response of Inconel 690. This model is a physically-based relationship derived from dislocation mechanisms and is expressed in the form:

$$\sigma = \sigma_a + B e^{-\beta T} + B_0 \varepsilon^{1/2} e^{-\alpha T} \quad (5)$$

where:

$$\beta = \beta_0 - \beta_1 \ln \dot{\varepsilon} \quad (6)$$

$$\alpha = \alpha_0 - \alpha_1 \ln \dot{\varepsilon} \quad (7)$$

$$\sigma_a = \sigma_G + k \ell^{-1/2} \quad (8)$$

Fig. 3 Comparison of measured and predicted stress-strain curves over different strain rate ranges.

Furthermore, σ is the von Mises equivalent stress, ε is the von Mises equivalent strain, $\dot{\varepsilon}$ is the strain rate, T is the absolute temperature, ℓ is the average grain diameter, and σ_G reflects the effect of the solutes and the initial dislocation density. The quantities B , β_0 , β_1 , B_0 , α_0 , α_1 , k and σ_G are considered constant. In accordance with the parameter-determination procedure proposed by Zerilli *et al.*,¹⁷⁾ and using the data of Fig. 1(a), the coefficients and exponents of the constitutive equation at room temperature for Inconel 690 are found to be: $\sigma_a = 368 \text{ MPa}$, $B = 400 \text{ MPa}$, $k = 0.92 \text{ MN/m}^{3/2}$, $\ell = 36 \text{ }\mu\text{m}$, $\beta_0 = 26.7 \times 10^{-3} \text{ (K}^{-1}\text{)}$, $\beta_1 = 2.98 \times 10^{-3} \text{ (K}^{-1}\text{)}$, $B_0 = 454.8 \text{ MPa}$, $\alpha_0 = 1.08 \times 10^{-4} \text{ (K}^{-1}\text{)}$, and $\alpha_1 = 2.17 \times 10^{-4} \text{ (K}^{-1}\text{)}$. Figure 3 compares the true stress-strain curves predicted by the Zerilli-Armstrong model using the above constants with those obtained experimentally. It can be seen that good agreement exists between the two sets of results.

3.4 Fracture behaviour and fractographic observation

A SEM technique was utilized to examine the fracture process and mode as a function of the strain rate. In the tested strain rate range, Inconel 690 fracture occurs only at strain rates exceeding $5 \times 10^3 \text{ s}^{-1}$ and is dominated by localized shearing along a plane inclined at an angle of approximately 45° with respect to the impact axis. It is known that during compression, planes of maximum shear stress are orientated at 45° to the axis of compression and that plastic deformation occurs along these planes. Under impact loading, shear instabilities develop and form a localized shear band during the final stage of deformation.

Examination of the fracture surfaces reveals that strain rate has a significant influence on the fracture characteristics. The fracture surfaces are characterized by dimple-like structures, which are indicative of a ductile fracture mode. Figures 4(a) and (b) present typical SEM micrographs of the specimens deformed at strain rates of $5 \times 10^3 \text{ s}^{-1}$ and $8.3 \times 10^3 \text{ s}^{-1}$, respectively. In Fig. 4(a), the fracture surface is characterized by a large number of shear dimples with diameters ranging

(a)

(b)

Fig. 4 Fracture surface of specimens deformed at strain rates of: (a) $2.3 \times 10^3 \text{ s}^{-1}$ and (b) $8.3 \times 10^3 \text{ s}^{-1}$.

from $0.5 \sim 4 \mu\text{m}$. These dimples are elongated along one direction, which suggests the presence of shear-dominated stress prior to failure. In the increased strain rate case of $8.3 \times 10^3 \text{ s}^{-1}$, Fig. 4(b) shows that the fracture surface exhibits a more brittle behaviour. Both the depth and the density of the dimples decrease with increasing strain rate, indicating a reduction in ductility and fracture resistance. This result correlates well with the flow curves of Fig. 1(a).

3.5 Dislocation and twin structure observations

The thin foils removed from the deformed specimens were observed by TEM to study the microstructural features associated with the various loading conditions. Figures 5(a) and (b) show the evolution of the dislocation substructure at strain rates of $2.3 \times 10^3 \text{ s}^{-1}$ and $8.3 \times 10^3 \text{ s}^{-1}$, respectively. It is clear that the strain rate has a significant multiplication effect on the final dislocation configuration and that the dislocation density increases with increasing strain rate. Figure 5(a) shows the dislocation substructure at a strain rate of $2.3 \times 10^3 \text{ s}^{-1}$. This substructure consists of planar dislocation arrays and loosely tangled dislocation cells. The cell walls are very thin and very few dislocation loops are

(a)

(b)

Fig. 5 Dislocation substructure of specimens deformed to a true strain of 0.7 at strain rates of: (a) $2.3 \times 10^3 \text{ s}^{-1}$ and (b) $8.3 \times 10^3 \text{ s}^{-1}$.

evident. At the increased strain rate of $8.3 \times 10^3 \text{ s}^{-1}$, Fig. 5(b) indicates a rapid multiplication of dislocations in the matrix. The dislocation density exceeds $4.23 \times 10^{15} \text{ m}^{-2}$. Furthermore, the cell walls are thicker and more well-defined since more dislocations exist in the walls. The large number of tangled dislocations increases the resistance to dislocation motion, and hence causes the flow stress to increase, as revealed by the true stress-strain curves of Fig. 1(a).

The formation and arrangement of twins are also affected by the strain rate. It is thought that mechanical twins form when a highly tangled dislocation density restricts dislocation movement and causes various local areas to attain the level of critical stress required for twinning to take place. Figures 6(a) and (b) show the mechanical twins under two different strain rates. At the lower strain rate of $2.3 \times 10^3 \text{ s}^{-1}$, Fig. 6(a) shows the presence of very few mechanical twins and reveals that the twin distribution is nonhomogeneous. The importance of continued twin nucleation and growth can be seen in

(a)

(b)

Fig. 6 Twinning structure of specimens deformed to a true strain of 0.7 at strain rates of: (a) $2.3 \times 10^3 \text{ s}^{-1}$ and (b) $8.3 \times 10^3 \text{ s}^{-1}$.

the specimen deformed at $8.3 \times 10^3 \text{ s}^{-1}$ (Fig. 6(b)). Under this heavy deformation condition, the twin structure is more homogeneous and the twin density becomes more evident. These microstructural observations suggest that the applied strain rate plays a key role in enhancing dislocation multiplication and twinning. In order to understand the correlation between microstructural and macroscopic flow behaviours at different strain rates, Fig. 7 presents the dislocation density and twin density as a function of flow stress and strain rate. It is observed that both the dislocation density and the twin density increase as approximately linear functions of increasing strain rate and flow stress. A comparison of the slopes in Fig. 7 reveals that dislocation multiplication and twin formation have a strengthening effect on the flow resistance of Inconel 690.

Fig. 7 Flow stress and strain rate versus dislocation and twin density for Inconel 690.

4. Conclusions

This study has evaluated the response of Inconel 690 to compressive impact loading using the compression split-Hopkinson bar technique. The high strain rate data obtained in this study, together with the quasi-static measurement data, reveal that the flow stress of Inconel 690 increases by approximately 75% over strain rates ranging from 10^3 to $8.3 \times 10^3 \text{ s}^{-1}$. A rapid increase of strain rate sensitivity has been identified at strain rates exceeding 10^3 s^{-1} . The work hardening behaviour is also affected by the strain rate. Under large deformations, the work hardening rate decreases with strain rate in the impact region, indicating that a thermal softening process is induced. The enhancement of work hardening stress leads to an increase of strain rate sensitivity and to a decrease of the activation volume. The dislocation and twin substructures have been observed at all strain rates. The results have shown that the dislocation and twin densities both increase at higher strain rates. The variations of dislocation and twin structures tend to support the conclusion that Inconel 690 is highly strain rate sensitive. The fracture process is dominated by localized shearing. The appearance of the dimple structures observed on the fracture surfaces is significantly influenced by impact loading. A more brittle-like behaviour is observed with a higher impact rate. Finally, good agreement has been noted between the flow behaviour predictions of the constitutive model and the experimental results.

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REFERENCES

- 1) M. G. da Silva and K. T. Ramesh: *Mater. Sci. Eng.* **A232** (1997) 11–22.
- 2) A. Gilat and C. S. Cheng: *Exp. Mech.* **40** (2000) 54–59.
- 3) W. S. Lee and C. F. Lin: *Mater. Sci. Eng.* **A308** (2001) 124–135.
- 4) G. Regazzoni, U. F. Kocks and P. S. Follansbee: *Acta Metall.* **35** (1987) 2865–2875.
- 5) F. J. Zerilli and R. W. Armstrong: *Acta Metall.* **40** (1992) 1803–1808.
- 6) L. E. Murr: *Scr. Metall.* **12** (1978) 201–206.
- 7) W. S. Lee, J. C. Shyu and S. T. Chiou: *Scr. Mater.* **42** (2000) 51–56.
- 8) M. Blicharski and S. Gorczyca: *Met. Sci.* **12** (1978) 81–93.
- 9) F. Greulich and L. E. Murr: *Mater. Sci. Eng.* **A39** (1978) 81–93.
- 10) V. Venkatesh and H. J. Rack: *Mech. Mater.* **30** (1998) 69–81.
- 11) H. P. Park, Y. H. Kim, B. W. Lee and K. S. Rheem: *J. Nucl. Mater.* **231** (1996) 204–212.
- 12) I. A. Choudhury and M. A. El-Baradie: *J. Mater. Process. Technol.* **77** (1998) 278–284.
- 13) K. Stiller, J. Nilsson and K. Norring: *Metall. Mater. Trans. A* **27A** (1996) 327–341.
- 14) G. E. Fuchs and S. Z. Hayden: *Scr. Metall.* **25** (1991) 1483–1488.
- 15) U. S. Lindholm: *J. Mech. Phys. Solids*. **12** (1964) 317–335.
- 16) R. K. Ham: *Philos. Mag.* **6** (1961) 1183–1184.
- 17) F. J. Zerilli and R. W. Armstrong: *Proc. Metal and Ceramic Matrix Composites and Other Advanced Materials*, ed. by Y. D. S. Rajapakse and J. R. Vinson, (ASME, 1995) pp. 121–126.