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Model for healing of creep cavities in nickel-based superalloys under hot isostatic pressing

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ABSTRACT

A healing model for creep cavities under hot isostatic pressing (HIP) is proposed, according to the healing characteristics and thermodynamic theory. The stress gradient around the cavity is considered as a driving force for atomic diffusion in this shrinkage model. The effects of HIP temperature and hydrostatic pressure on the healing kinetics of creep cavities are studied by numerically integrating the shrinkage rate equation. A useful healing map is constructed, showing the healing time as a function of temperature and pressure. This model provides a more realistic basis for the selection of appropriate HIP parameters to rejuvenate the creep damage present in high-temperature components.

1. Introduction

Nickel-based superalloys are widely utilized in manufacturing gas turbine components such as disks and blades, which required excellent microstructural stability and mechanical properties at elevated temperatures [1,2]. Creep-induced cavities are inevitably formed along grain boundaries due to the high temperatures and complex stresses, resulting in reduced strength. Hot isostatic pressing (HIP) technology has been successfully applied in the rejuvenation of turbine blades through sintering creep cavities [3–5]. There has been much interest in developing models to describe the healing kinetics of creep cavities and hence to provide a basis for optimizing HIP parameters (e.g., temperature, pressure and soaking time) of the rejuvenation schedule. Some researchers [6,7] have derived the cavity sintering models from the powder sintering theory and others have obtained some rigorous analytical solutions for the cavity shrinkage based on a robust continuum theory [8,9]. In addition, several models [10–12] have been formulated based on the reverse effect of the diffusional growth of cavities under externally applied stress. However, the theoretical predictions do not agree well with the experimental results, since the formulation of the shrinkage models cannot reflect the healing behavior under HIP.

The purpose of the present work is therefore to propose a cavity healing model based on the observed healing characteristics during HIPing. The effects of HIP temperature and pressure on the cavity shrinkage are also investigated by numerically integrating the shrinkage rate equation.

2. Experimental

The investigated samples were removed from the mid-height section of an aero-engine turbine blade which had been operated for 1000 h. They were made from a cast polycrystalline superalloy K465, which has the chemical composition (wt.%): 0.15C, 8.72Cr, 9.87Co, 10.31W, 1.57Mo, 1.03Nb, 5.29Al, 2.65Ti and balance Ni. It is well known that creep cavities can be effectively healed under appropriate HIP pressure (~150 MPa) and HIP temperature (higher than γ'-solvus temperature) [5]. The HIP process in the present study was conducted at 1453 K/200 MPa/4 h in a laboratory HIP apparatus (ABB, Quintus). The selected HIP temperature, well below the γ'-solvus temperature of K465 alloy (~1503 K), was designed to reveal the healing characteristics during HIPing.

3. Cavity shrinkage model

Fig. 1 shows the typical healing characteristics in the vicinity of a creep cavity after HIP treatment at 1453 K/200 MPa for 4 h. The

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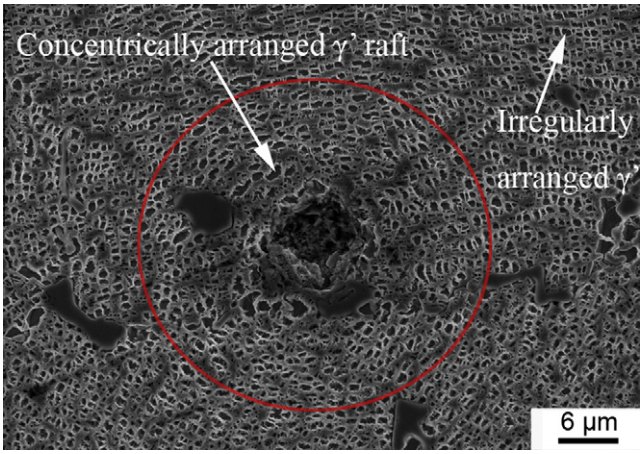

Fig. 1. Concentrically arranged γ' raft structure in the healing zone of the creep cavity under HIP condition of 1453 K/200 MPa/4 h.

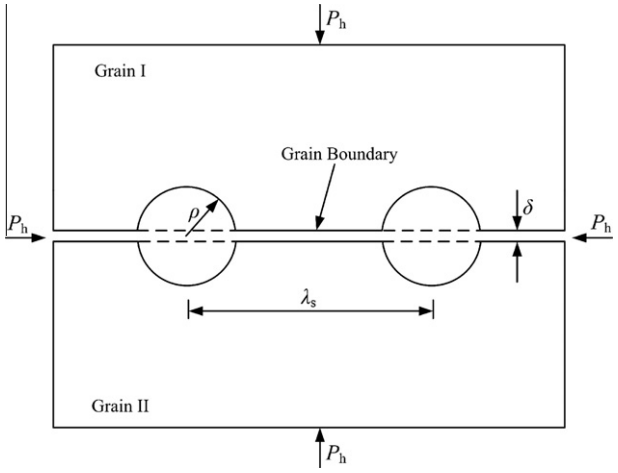

Fig. 2. Schematic representation of the shrinkage model for creep cavities along the grain boundary under the hydrostatic pressure.

creep cavity is nucleated at the grain boundary during high-temperature service. The γ' precipitates in the healing zone of the cavity (marked by the circle) are concentrically arranged and elongated along the circumferential direction, whereas those remote from the cavity are irregularly precipitated in the matrix. From a mechanical perspective, the local complex stress state induced by the presence of the cavity can be represented as the equivalent uniaxial tension in the radial direction. The service-exposed alloy before the HIP treatment has a negative value of lattice misfit [13]. Hence, the product of the equivalent stress and lattice misfit is negative, indicating that the concentrically oriented γ' raft is N-type rafting [14]. As pointed out in our previous work [4], the observed rafting behavior is controlled by atomic diffusion, which results in the cavity healing during HIP. Moreover, the size of the rafted zone reflecting the effective diffusion field of diffusional creep is determined by the resultant stress gradient [4].

Therefore, two driving forces are assumed to be operative for the sintering of creep cavities. These are surface tension due to the curvature of cavity surface and the stress gradient around the cavity. The most efficient sink for vacancies is assumed to be grain boundary, rather than other possible sinks (e.g., dislocation cell walls and the boundaries around inclusions or between dendrites). Vacancies are transported from the cavity surface to the grain boundary through grain-boundary diffusion, and simultaneously atoms diffuse in the direction opposite to that of vacancy diffusion. It should be pointed out that the effect of lattice diffusion is neglected, because it does not play significant role in comparison to grain-boundary diffusion.

We simplify the analysis in a two-dimensional model as illustrated in Fig. 2. Consider an array of spherical cavities of radius ρ and spacing λ_s along a grain boundary of width of δ. Each cavity is assumed to be surrounded by a vacancy sink midway between the adjacent cavities. The model is subjected to a remotely applied hydrostatic pressure P_h. The difference in chemical potential between the grain boundary and the cavity surface, caused by the surface tension, drives the atoms depositing onto the cavity surface. It can be determined from the Gibbs–Thomson relationship as follows:

$$\Delta\mu = \frac{2\gamma_s\Omega_A}{\rho}, \tag{1}$$

where γ_s is the surface energy of the cavity, Ω_A is the atomic volume. The associated atomic flux is then given by Nernst–Einstein equation, i.e.,

$$J_s = -\frac{D_{gb}}{kT\Omega_A}\nabla(\Delta\mu) = \frac{4D_{gb}\gamma_s}{kT\rho\lambda_s}, \tag{2}$$

where D_{gb} is the grain-boundary diffusion coefficient, k is the Boltzmann's constant, T is the absolute temperature. The term D_{gb} is described by the Arrhenius-type equation as $D_{0b}\exp(-Q_b/RT)$, where D_{0b} and Q_b are the frequency factor and the activation energy for grain-boundary diffusion, respectively.

The other driving force, i.e. the stress gradient around the cavity, can be obtained from the Lame's solution of a thick spherical shell model. Due to the spherical symmetry of the mechanical model, the magnitude of the effective stress is expressed in a spherical coordinate system as [15]:

$$\sigma_{eff}(r) = \frac{P_h}{1-(\rho/a)^3}\frac{3\rho^3}{2r^3}, \tag{3}$$

where r is the radial distance, a is the grain size of polycrystalline alloys. Hence, the effective stress gradient at the cavity surface (i.e., $r = \rho$) is then written as:

$$\frac{d\sigma}{dr} = -\frac{9P_h}{2\rho[1-(\rho/a)^3]}. \tag{4}$$

Likewise, the atomic flux induced by the applied hydrostatic pressure can be given by the Nernst–Einstein equation, i.e.,

$$J_\sigma = -\frac{D_{gb}}{kT\Omega_A}\frac{d\mu_h}{dr}, \tag{5}$$

where $d\mu_h/d\rho$ is the chemical potential gradient, which is given by:

$$\frac{d\mu_h}{dr} = \Omega_A\frac{d\sigma}{dr}. \tag{6}$$

Substituting Eqs. (4) and (6) into Eq. (5), we have

$$J_\sigma = \frac{D_{gb}}{kT}\frac{9P_h}{2\rho[1-(\rho/a)^3]}. \tag{7}$$

Therefore, the total atomic flux along the grain boundary can be estimated by the linear superposition of these two flux components determined by Eqs. (2) and (7), i.e.,

$$J = \frac{D_{gb}}{kT}\left\{\frac{9P_h}{2\rho[1-(\rho/a)^3]} + \frac{4\gamma_s}{\rho\lambda_s}\right\}. \tag{8}$$

The area of the atomic diffusion perpendicular to the diffusion direction is $2\pi\rho\delta$, so that the rate of shrinkage in cavity volume can be calculated from the equation:

$$-\frac{dV}{dt} = J(2\pi\rho\delta)\Omega_A = \frac{2\pi\delta D_{gb}\Omega_A}{kT}\left\{\frac{9P_h}{2[1-(\rho/a)^3]} + \frac{4\gamma_s}{\lambda_s}\right\}. \tag{9}$$

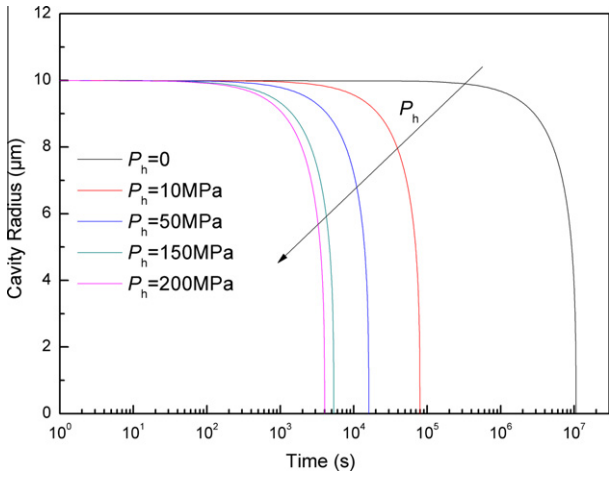

Fig. 3. Predicted healing kinetics of creep cavities at 1503 K under different hydrostatic pressures.

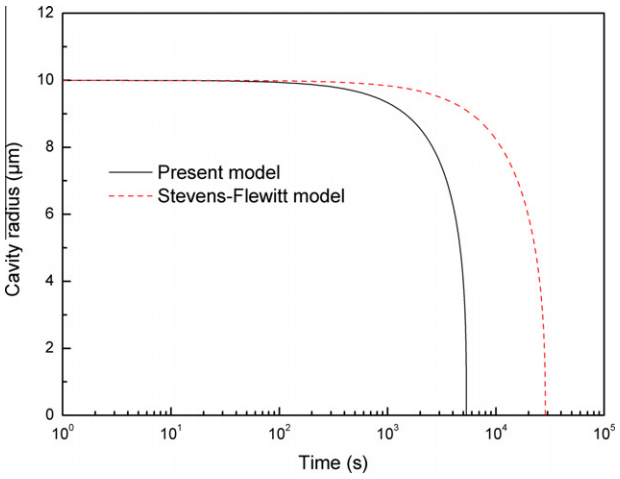

Fig. 5. Comparison of healing kinetics predicted by the present model and Stevens–Flewitt model under HIP (1503 K/150 MPa) for an initial cavity radius of 10 μm and spacing of 100 μm.

Since surface diffusion is faster than grain-boundary diffusion, it is reasonable to assume that the cavity remains rounded during healing. Due to $V = 4\pi\rho^3/3$, the healing rate of the creep cavity, characterized by the linear shrinkage rate of cavity, can be then expressed by

$$\frac{d\rho}{dt} = -\frac{D_{gb}\delta\Omega_A}{kT\rho^2}\left\{\frac{9P_h}{4[1-(\rho/a)^3]} + \frac{2\gamma_s}{\lambda_s}\right\}. \quad (10)$$

4. Results and discussion

Numerical integration of Eq. (10) by using Simpson's rule better reveals the kinetics of the evolving grain-boundary cavity during HIPing and enables the sensitivity of the healing time to various HIP parameters to be analyzed. A simplification for the diffusion parameters has to be made, due to a lack of sufficient data for the composite diffusion coefficients in the complex superalloy system. Since the cavity shrinkage is diffusion-controlled and nickel is the most abundant element which forms nickel solid solution, it is assumed that the diffusion of other elements in nickel can be neglected, and the diffusion data for pure nickel is considered to be representative of the nickel-based alloy. The relevant thermodynamic data for nickel [11] are $\gamma_s = 1.7\ J/m^2$, $D_{0b} = 7 \times 10^{-6}\ m^2/s$, $Q_b = 115\ kJ/mol$, $\delta = 5 \times 10^{-10}\ m$, $\Omega_A = 1.09 \times 10^{-29}\ m^3$.

Fig. 3 shows the radius variation with sintering time at 1503 K under various HIP pressures ranging from 0 to 200 MPa for an initial cavity radius of 10 μm and spacing of 20 μm. The influence of the applied hydrostatic pressure on the healing time of cavity can be also seen. When no pressure is applied, i.e. $P_h = 0$ MPa, the surface tension is the only driving force for healing, resulting in a quite long time (about several 1000 h) required for complete shrinkage. That is to say, it is unrealistic for common heat treatments to heal creep cavities, which is time-consuming and costly. A reduction of three orders of magnitude in healing time is observed to occur

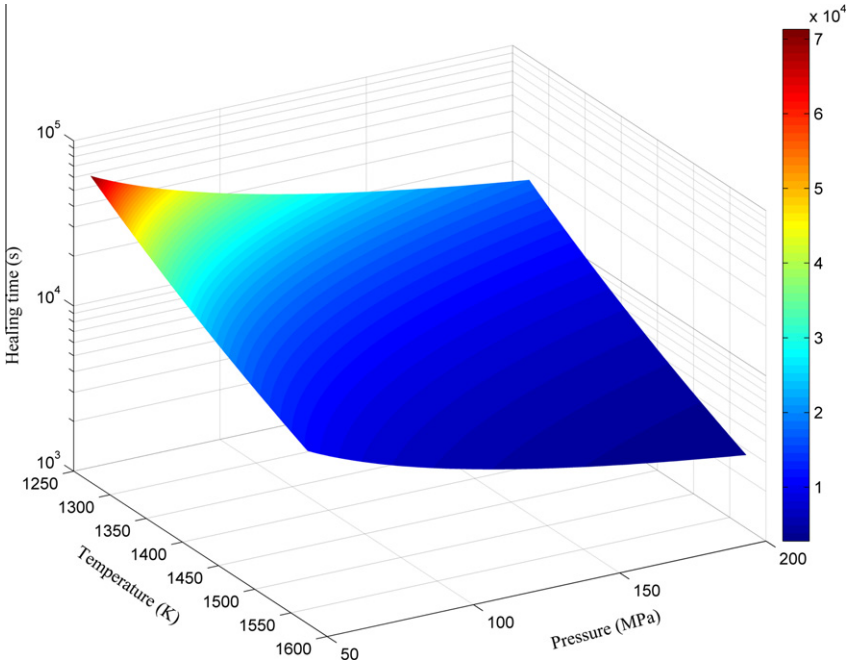

Fig. 4. Healing map showing the time required to sinter the cavity as a function of HIP temperature and pressure.

when increasing P_h from 0 to 150 MPa. It is also clear that there are few extra benefits obtained through increasing the applied pressure above 150 MPa.

A healing map shown in Fig. 4 is constructed, representing the time required to sinter the cavity as a function of HIP temperature and pressure. Healing time is significantly decreased with increasing the applied temperature and hydrostatic pressure. As can be seen from the map, it is more advisable to raise temperature rather than pressure to shorten the soaking time of HIPing, when the pressure is above about 100 MPa. It is worth noting that the selection of temperature needs to consider some metallurgical factors, e.g. the γ' and MC solvus temperatures. In addition, the predicted healing kinetics of the creep cavity in the present study is shown in Fig. 5, and is compared with the result given by another diffusion-controlled model proposed by Stevens and Flewitt [10,12]. The time required for complete healing under a typical HIP condition (1503 K/150 MPa) in the present model is much shorter than that of the latter model, though the trends of cavity shrinkage are basically similar. The predicted sintering time under the HIP condition in the Steven–Flewitt model is 28,500 s (~7.92 h), whereas the healing time in the present model can be reduced to 5330 s (~1.48 h). Meanwhile, the creep cavities in the nickel-based superalloys can be effectively healed in 2–4 h under the same HIP condition, as demonstrated by the results of the experiments in Refs. [4,5,16]. Therefore, the healing kinetics predicted by the present model is more consistent with the experimental results, and it can better describe the healing behavior of creep cavities during HIPing.

5. Conclusion

In this paper, a cavity healing model based on the thermodynamic theory is proposed, and the effects of HIP temperature and pressure on the healing kinetics are studied. The stress gradient around the cavity, acting as one of the driving forces, is considered in the sintering model. The predicted results are in general agreement with the HIPing experimental results for nickel-based superalloys. This model provides a more realistic basis for the determination of optimal HIP schedules to rejuvenate the creep damage in high-temperature components.

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