Dislocation Creep

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خزش نابجاپی

ایجاد تغییرفرم خزشی به وسیله نابجاپی در اثر نفوذ جای خالی اتمی و یا برخ، خزش نابجاپی نامیده می شود.

مکانیزم های خزش نابجاپی:

1- همکاری مهای عبور از گذر (by pass)

2- صعود نابجاپی (Dislocation Climb)

(Growan Loop)

3- کم حالت کردن (Bowing) و تشکیل حلقه نابجاپی

4- همکاری کردن (Shear)

(Staking Fault)

5- به وسیله تشکیل نقش چیدن (Anti-Phase Boundary)

6- به وسیله تشکیل مرز ضد فازی
Power-law Creep

For pure metals:

- This regime, characterized by a stress exponent $n=4-5$, $p=0$ and a value of $Q=Q_0$ ($Q_0$ is the activation energy for lattice self-diffusion).

- In this regime creep is controlled by recovery mechanisms, such as climb and annihilation of dislocations.

The edge dislocations can climb due to the extensive diffusional flow of vacancies; then the high temperature enhances the mobility of dislocation, and diffusion controls the creep strain rate.

In fact, the deformation results from a sequence of glide and thermally-assisted climb the slowest mechanism is rate controlling, and as long as glide is very easy (as in pure metals), creep is controlled by climb.
Dislocation climb

The result of the diffusion of vacancies is climb of dislocation from plane 4 to plane 3.

Temperature is sufficiently high that vacancies diffuse to the lowest row of atoms in the half-plane of dislocation. In the figure a vacancy in plane 6 diffuses to plane 4.
Dislocation Creep: Power-Law Creep

At high temperatures, dislocations acquire a new degree of freedom: they can climb as well as glide.

If a gliding dislocation is held up by discrete obstacles, a little climb may release it, allowing it to glide to the next set of obstacles where the process is repeated.

The glide step is responsible for almost all of the strain, although its average velocity is determined by the climb step.

Mechanisms which are based on this climb-plus-glide sequence we refer to as climb-controlled creep.

The important feature which distinguishes these mechanisms from other mechanisms is:

The rate-controlling process is the diffusive motion of single ions or vacancies to or from the climbing dislocation, rather than the activated glide of the dislocation itself.
Intermediate-stress regime (Power-Law Creep)

For solid solution alloys:
I. A first region of the intermediate-stress regime, where creep is controlled by climb \((n=4-5, \quad Q=Q_0)\)

II. A second region of the intermediate-stress regime, where creep is controlled by glide of dislocation in a “cloud” of solute atoms (“viscous drag”); in this regime climb is faster than viscous glide, and \(n=3, \quad p=0\) and \(Q=Q_0\)

III. A third region of climb-controlled creep \((n=4-5, \quad Q=Q_0)\)
Power-Law Creep

Three-Power-Law Viscous Glide Creep

This is due to the fact that the dislocations interact in several possible ways with the solute atoms, and their movement is impeded.

There are two competing mechanisms over this stress range:

1. **dislocation climb**
2. **dislocation glide**

*glide is slower and thus rate controlling*
More precisely, following the original model of Weertman, Viscous Glide Creep is described by the equation:

\[ \dot{\varepsilon}_{ss} \approx \frac{0.35}{A} G \left( \frac{\sigma}{G} \right)^3 \]

There are several possible viscous drag (by solute) processes in region II, or Three-Power-Law regime:

- Segregation of solute atoms in an atmosphere around the moving dislocations (Cottrel’s atmosphere) that must diffuse to follow the dislocation.
- Segregation of solute atoms to stacking faults (Suzuki effect).
- In solid solution alloys with short-range order, dislocation motion destroys the order creating an interface (Fisher's mechanism).
- Local ordering of solute atoms in the stress field of the dislocation (Snoek mechanism).
Dislocation Creep: Power-Law Creep

Creep deformation by dislocation due to

Vacancy diffusion or cutting

⇒ Dislocation creep

Stacking Fault Formation Mechanism

A perfect dislocation

Partial dislocation

Partial dislocation

Repulsive Force

Attractive Force

Surface tension energy

Stacking Fault Energy
At high stresses, when $n=11$, dislocations cut through $\gamma'$ particles by stacking fault (SF) formation and anti-phase boundary (APB) coupled dislocation pairs. The creep induced dense dislocation networks are mainly constrained at the $\gamma$–$\gamma'$ interface.

Since the stacking faults are lying on a common (111) glide plane, the matrix dislocations must, as totals, dissociate into partials at the $\gamma$–$\gamma'$ interface, producing a single superpartial gliding in the $\gamma'$ phase and a partial left at the interface which relaxes the high coherency stresses at $\gamma$–$\gamma'$ interfaces. Therefore, the partial reduces the overall energy of the dislocation interface configuration so that a stacking fault formation can be generated at the coherent interface. After passing the precipitate, the partial dislocations rearrange themselves to totals for further motion in the matrix.
Groups of partial dislocations may create a variety of fault configurations in \( \gamma' \) precipitates. Various types of fault configurations were observed in specimen creep tested under Shear condition.
Coherent interfaces are requirements for stacking fault formation. Otherwise, the resulting pileups of matrix dislocations at the interfaces generate local stresses and cause a transition from stacking fault to APB mechanism.

Observation of the microstructure of GTD-111 at the end of stage II indicates high densities of stacking faults in γ′ precipitates and increasing the probability of stacking fault interactions. An example of the microstructure is shown in the bellow figure where reactions of stacking faults from different {111} planes have created locks.
Also at this high creep strain and high stress in addition to stacking faults, a few pairs of \( a/2\langle 110 \rangle \) screw dislocations are observed on \{111\} glide planes inside \( \gamma' \) cubes. The paired dislocations, well separated in the matrix, are constricted upon entering the \( \gamma' \) particle due to the creation of the APB between them. Sometimes this constriction is so high that the dislocation pairs, which are resolvable in the matrix, are not observed in the \( \gamma' \) precipitates.

From an energetic point of view, the shear mechanism by APB coupled dislocation pairs occurs when matrix dislocations pileup at the interfaces in order to push the leading dislocation into the \( \gamma' \) cubes. Therefore, this mechanism, predominant at high stresses and high strains. So, it is expected that in the upper range of the shear region, APB-coupled dislocation pair is the dominant mechanism and in the lower range the SF formation is predominant.
At low stresses, when \( n=4 \), the dislocations are unable to cut through the particles. Deformation is based now on the movement of dislocations in the matrix (by pass). In this region it has been suggested either the dislocations are forced to overcome the particles by: 1- climb, or by: 2- looping. Also, deformation may occur predominantly in the grain boundary zones, i.e. by grain boundary sliding and deformation in regions of the grains adjacent to the boundaries.

The dislocation configurations in the grains, which are the representative of dislocation climb process, are illustrated in the Figure for GTD-111 specimens.
In the low stress range, in the region where the stress values are less than those required for dislocation bowing or looping around particles, dislocations climb and dislocation tangles at $\gamma - \gamma'$ interface are formed.

The figure shows dislocation configurations of a specimen creep tested in the steady-state region for 20 h at $982^\circ C/152$ MPa. Dislocations migrate mainly within $\gamma$ matrix in creep processes and many dislocation networks, after climbing over $\gamma'$ particles, exist at $\gamma - \gamma'$ interfaces. No evidence has been observed showing dislocations cut through $\gamma'$ particles.

Homogeneous distribution of low-density dislocations within matrix phase, network formation at the $\gamma - \gamma'$ interfaces with little or no dislocations are observed.
Orowan dislocation loops around $\gamma'$ precipitates are frequently observed in the specimens creep tested at low stresses only after prolonged creep life.

The looping stress depends on the size and spacing of precipitates that change due to aging during the creep tests.

Particle spacing increases as testing time or strain increase. The stress required for loop formation, or Orowan stress, decreases with increasing particle spacing. Hence, Orowan looping is a possible dislocation by-pass mechanism at the end of the longer term low stress creep tests. Such dislocation loops are believed to have been formed in the process of dislocation by-pass, this means that dislocations cannot shear $\gamma'$ particles in this condition.

The microstructure is characterized by the formation of the dislocation loops and tangles and deformation is confined within the $\gamma$ matrix.
With GTD-111 at high temperature and higher stress, in the bypass region, the dislocations generated at a source may bow between secondary $\gamma'$ precipitates and a few closed loops were seen around large secondary $\gamma'$ particles, as shown in the figure. The figure shows loops around $\gamma'$ particles deformed under creep condition 870°C/300 MPa.
Recovery: As it was mentioned in high temperature creep of two-phase alloys, the rate controlling process would be climb of dislocations over particles. At high stresses in the by-passing region, dislocations move past particles by looping. Loops build up and exert back stresses to prevent more looping until recovery occurs by dislocation climb of the loop near to the $\gamma'$. However, at low stresses, dislocations climb over $\gamma'$ particles without pile-up or bowing.

TEM microstructure of GTD-111, creep-deformed in steady state region at 982°C/152 MPa, showing dislocation climb mechanism, t=20 h.
With increasing exposure time at high temperature and under low stress creep conditions, e.g. 982°C/152 MPa, there is enough time for climbing and gliding and then arranging of dislocations at $\gamma - \gamma'$ interfaces and dislocation network formation.

TEM microstructure of GTD-111, creep-deformed in steady state region at 982°C/152 MPa, showing dislocation climb mechanism, $t=50$ h.
As shown in the Figure, it is clear that many dense networks of screw dislocations (three-dimensional dislocation networks) and dislocation configurations exist at \( \gamma - \gamma' \) interfaces in the specimen creep deformed at 982°C/152 MPa for 50 h. Also, at \( \gamma - \gamma' \) interfaces regular two-dimensional dislocation networks after climbing over \( \gamma' \) particles are seen clearly. Two dislocation groups of the type a/2 <110> may react with each other at \( \gamma - \gamma' \) interfaces and dislocation climb occurs more easily. The networks are formed by dislocations moving by climb around the \( \gamma' \) particles.
With further exposure time at the same creep condition, high-temperature and low-stress, dislocations inside $\gamma'$-particles form networks. The substructure consisting of subgrain in the matrix and primary $\gamma'$-particles could be seen. Such a substructure can be better observed in the specimens deformed for a longer time.

TEM microstructure of GTD-111, creep-deformed in steady state region at 982°C/152 MPa, showing dislocation climb mechanism, $t=70$ h.
A high-stress regime characterized by power law breakdown, where the conventional power law should be replaced with an exponential dependence of the strain rate on applied stress.

This regime is not particularly important in creep due the very high strain rate involved, but is rather a primary concern in hot-forming operations.
Power-law breakdown

At high stresses (above about $10^{-3} \mu$), the simple power-law breaks down:
The measured strain-rates are greater than previous eqn. predicts.

The process is evidently a transition from climb-controlled to glide-controlled flow.

There have been a number of attempts to describe it in an empirical way. Most lead to a rate-equation of the form:

$$\dot{\gamma} \propto (\sinh \beta' \sigma_s)^n \exp \left( \frac{-Q_{cr}}{RT} \right)$$

which at low stresses ($\beta' \sigma_s < 0.8$) reduces to a simple power-law, while at high ($\beta' \sigma_s > 1.2$) it becomes an exponential.