



Title	Review on High Temperature Creep Behavior of Metals
Author(s)	Enjyo, Toshio
Citation	Transactions of JWRI. 1974, 3(1), p. 111-120
Version Type	VoR
URL	https://doi.org/10.18910/11954
rights	
Note	

The University of Osaka Institutional Knowledge Archive : OUKA

<https://ir.library.osaka-u.ac.jp/>

The University of Osaka

Review on High Temperature Creep Behavior of Metals[†]

Toshio ENJYO*

Abstract

This review deals principally with the creep deformation behavior of metals at high temperatures in which the mechanism of self-diffusion is involved in the creep rate controlling process. Such temperatures are generally above one half the absolute melting temperature. The main explanation is about creep in pure metals but there are some problems on solid solution alloys.

Thermally activated process of the creep deformation based on the average behaviour of large number of dislocations is proposed to explain stress and temperature dependency of steady state creep rate, activation energy and activation area of creep deformation, correlation between internal stress and effective stress and creep deformation mechanism.

1. Introduction

The high temperature creep characteristics of metals are considerably different from their behavior at low temperature. The creep deformation is controlled primarily by thermally activated process at elevated temperature. So, the creep-rate at such high temperatures is strongly dependent on stress and temperature. The strain-time behaviour of metals during creep can take many forms, depending on the metals and conditions to which it is subjected. The high temperature creep of metals is generally characterized by four stage as shown schematically in Fig. 1. The stage I ($0 - \epsilon_1$) of creep is known as inverse transient creep or incubation period and is the period where the density of isolated dislocations increase rapidly with no cells being formed¹⁾ and then creep rate is

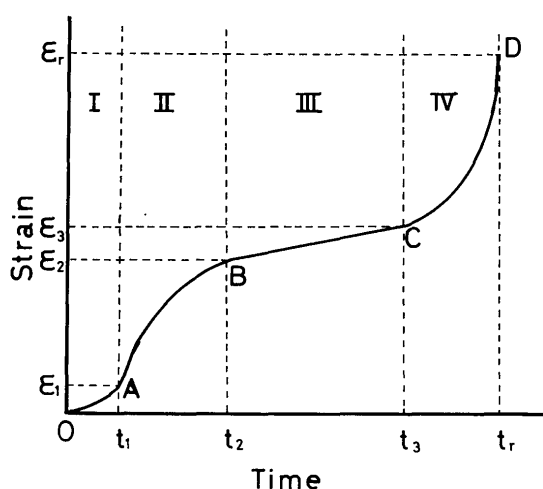


Fig. 1. General form of creep curve.
 Stage I : Inverse transient creep (Incubation period)
 Stage II : Transient creep (primary creep)
 Stage III : Steady state creep (secondary creep)
 Stage IV : Tertiary creep

rapidly increasing with time. The stage II ($\epsilon_1 - \epsilon_2$) of creep is known as transient creep or primary creep and is the period where the metal is work-hardened and then the creep rate is decreasing with time. The stage III ($\epsilon_2 - \epsilon_3$) of creep is known as steady state creep or secondary creep and is the period where work-hardening is completely balanced by recovery and the substructure remains in a stable state. These substructures are strongly dependent on the creep stress and only in a minor way on temperature. Therefore, the creep rate is constant in this stage. The stage IV ($\epsilon_3 - \epsilon_r$) of creep before fracture is known as tertiary creep and is frequently associated with localized necking and also with the formation of cavity.

The transient creep with both stage I and II is called sigmoidal transient-creep, and that with only stage II is called normal transient-creep. For example, on the Cu-Al system alloys, normal transient creep behaviour changes to sigmoidal transient creep behaviour with decreasing stress and temperature and with increasing aluminum content of the alloys as showing in Fig. 2.¹⁾

The present review is particularly concerned with the steady state creep at temperatures above one half the absolute melting temperature ($0.5 T_m$).

2. The Stress Dependence of High Temperature Creep Rate

On discussing the high-temperature creep of metals, of particular interest is the making clear the stress and temperature dependence of steady-state creep rate $\dot{\epsilon}_s$. A number of investigations have been made to reveal the stress dependence of $\dot{\epsilon}_s$ in various metals and under various conditions of stresses and

[†] Received on Dec. 25, 1973

* Professor

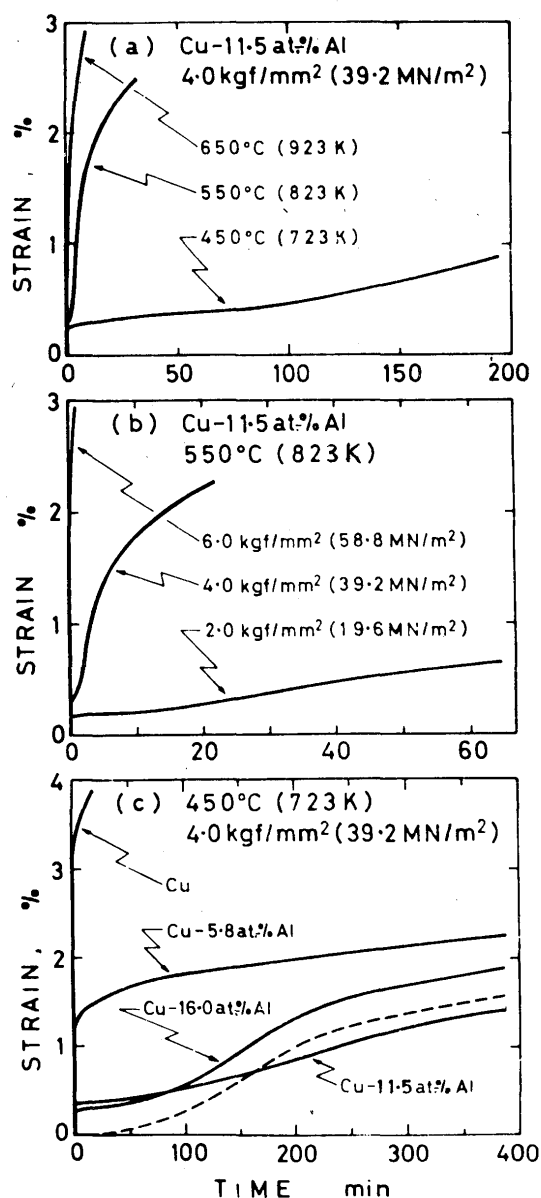


Fig. 2. Creep behavior in copper-aluminum alloys at an early stage of transient creep, as a function of: (a) temperature, (b) stress, and (c) aluminum content.¹¹

temperatures. It has been shown that $\dot{\epsilon}_s$ can be described by a power law equation of the following type at the constant temperature,

$$\text{power law; } \dot{\epsilon}_s = A_1 \sigma^n \quad (1)$$

In this equation, σ is the creep stress and A_1 and n are constants independent of stress and n is usually called a stress exponent. And n has been found about five for pure metals in spite of the crystal structure.^{2), 3), 4)}

It has been reported that in some solid solution alloys n decreases from about five toward about three with increasing solute atom content.^{5)~7)} Fig. 3⁶⁾ shows the steady-state creep rate versus stress for Au-Ni.

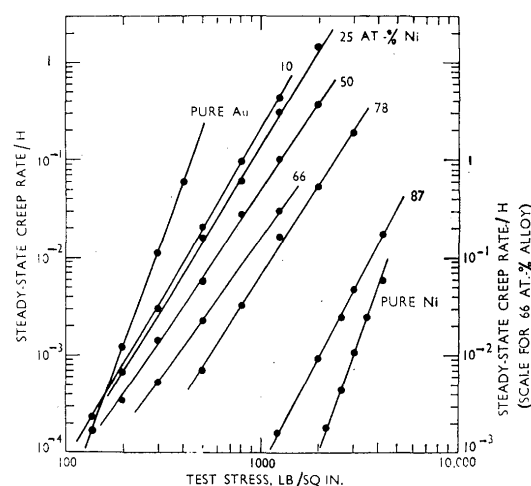


Fig. 3 (a). Relationship between creep rate and stress for gold-nickel alloy at 860°C.⁶⁾

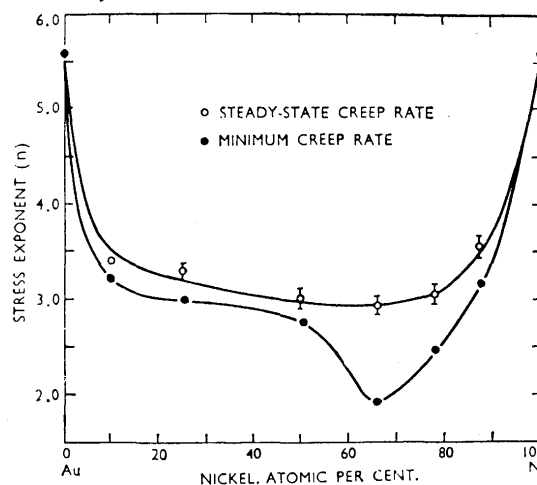


Fig. 3 (b). Relationship between stress exponent at 860°C and composition.⁶⁾

solid solution alloys at 960°C. As shown in Fig. 3, the creep data of pure Au and pure Ni are represented by straight line with a slope of approximately five and the all solid solution alloys of Au-Ni binary system fall along separate parallel lines with a approximately three. However, it is now known that the value of n do not follow the three power law by alloying in some solid solution alloys as shown in Table 1 where various solid solution alloys are separated into two categories.

Cannon and Sherby⁸⁾ proposed that the decreasing of n value by attending of solute atom is closely connected with the difference of atomic radius between the solvent atom and solute atom and with the elastic modulus of the material. That is, a considerable decreasing of n by attending solute atom is expected when the difference of atomic radius is large and the elastic modulus considerably decreases by alloying.

On the other hand, Jonas and Sellars⁹⁾ found out that the stress exponents of steady-state creep rate for

Table 1. Power law dependence for creep of solid solution alloys.

Class I				Class II			
Alloy	Solute atom at. %	Stress exponent n	Reference	Alloy	Solute atom at. %	Stress exponent n	Reference
Al-Mg	2.5	3.2~3.6	Aramaki ¹¹⁾	Al-Zn	≈ 4	4.4~5.6	Endo ⁶⁾
Au-Ni	10~90	3.0~3.5	Sellars ²⁾	Cu-Al	≈ 15	4.6~5.6	Kikuch ⁷⁾
In-CD	≈ 5.1	3.0~3.2	Weertman ³⁾	Ni-Cu	≈ 90	3.8~5.3	Monma ⁸⁾
In-Pb	12	3.0~3.5	Weertman ³⁾	Ni-Cr	≈ 30	4.6~5.2	Monma ⁹⁾
In-Sn	≈ 7.9	3.5	Weertman ³⁾	Ni-W	≈ 9.2	3.9~7.2	Monma ¹⁰⁾
In-Hg	6.0	3.6	Weertman ³⁾	Ni-Fe	≈ 80.3	4.6	Karashima ¹¹⁾
In-Tl	≈ 36	3.2~3.5	Weertman ³⁾	Pb-Bi	≈ 16	4.5	Weertman ³⁾
Pb-In	≈ 64	3.3~3.6	Weertman ³⁾	Fe-Cr	≈ 26.6	4.6~5.0	Imai ¹²⁾
Pb-Sn	≈ 19.4	3.1~3.5	Weertman ³⁾	Fe-Si	5.8	4.5	Barrett ¹³⁾
Pb-Cd	≈ 4.8	3.8	Weertman ³⁾	Cu-Si	≈ 6.9	6.8~7.0	Singh ¹⁴⁾
Cu-Au	27	3.4	Davies ⁴⁾	Ni-Co	≈ 69	~7	Davies ¹⁵⁾
Cu-Zn (Beta Brass)	52.5	2.9	Brown ⁵⁾	Cu-Zn (Alpha Brass)	≈ 30	5.0	Bonesteel ¹⁶⁾

- (1) T. Aramaki, T. Endo, T. Enjyo, and M. Adachi; Japan Inst. Metals vol. 33 (1969) p. 349
 (2) C. M. Sellars and A. G. Quarrell; J. Inst. Metals vol. 90 (1962) p. 329
 (3) J. Weertman; Trans. Met. Soc. AIME vol. 218 (1960) p. 207
 (4) R. G. Davies; Trans. Met. Soc. AIME vol. 221 (1961) p. 1280
 (5) N. Brown and D. R. Lenton; Acta Met. vol. 17 (1969) p. 669
 (6) T. Endo, T. Nomura, T. Enjyo and M. Adachi; J. Japan Inst. Metals vol. 35 (1971) p. 427
 (7) S. Kikuch, M. Kajitani, T. Enjyo and M. Adachi; J. Japan Inst. Metals vol. 37 (1973) p. 228
 (8) K. Monma, H. Suto and H. Oikawa; J. Japan Inst. Metals vol. 28 (1964) p. 258
 (9) K. Monma, H. Suto and H. Oikawa; J. Japan Inst. Metals vol. 28 (1964) p. 253
 (10) K. Monma, H. Suto and H. Oikawa; J. Japan Inst. Metals vol. 28 (1964) p. 304
 (11) S. Karashima, T. Motomiya and H. Oikawa; Tech. Rep. Tohoku Univ. vol. 33 (1968) p. 193
 (12) Y. Imai and T. Murata; J. Japan Inst. Metals vol. 30 (1966) p. 740
 (13) C. R. Barrett and W. D. Nix; Acta Met. vol. 13 (1965) p. 1247
 (14) N. N. Singh and C. R. Barrett; Trans. Met. Soc. AIME vol. 245 (1969) p. 2469
 (15) C. K. L. Davies, P. W. Davies and B. Wilshire; Phil. Mag. vol. 12 (1965) p. 827
 (16) R. M. Bonesteel and O. D. Sherby; Acta Met. vol. 14 (1966) p. 385

copper and copper 10 at. % Ni solid solution alloy are about five and that for copper 10 at. % gold solid solution alloy is about three. And then, they advanced the other opinion⁹⁾ in which this change in stress exponent has been taken to indicate that the effect of nickel is to reduce the rate of the controlling recovery process, whereas the addition of gold changes the controlling process to one of viscous glide of dislocation, as proposed by Weertman.^{5), 10)}

For such a change in rate-controlling process, a strong interaction between solute atoms and dislocations is required.

At the present time, the alloying effects on the stress dependence of high temperature steady-state creep rate are so complicated that they have not clearly solved.

Moreover, the value of n approaches one under conditions which are very low stresses and temperatures near the melting point. In such cases, the steady state creep deformation mechanism is not due to dislocation motion and climb but rather is due to a stress-

directed atom migration process mechanism which is known as the Nabarro-Herring^{11), 12)} creep.

On the high stress creep above about $\dot{\epsilon}/D = 10^9$ (D is the diffusion coefficient), vacancy concentration is greater than the thermal equilibrium number because one might generate many excess vacancies by dislocation intersection process, and so steady-state creep rate greater than those predicted from power law equation at such high stress range.

Beside a power law, the stress and steady-state creep rate are related by an exponential law^{14), 15)} or hyperbolic sine law^{2), 16)} at constant temperature.

$$\text{exponential low; } \dot{\epsilon}_s = A_2 \exp(\beta \sigma) \quad (2)$$

$$\text{hyperbolic sine low; } \dot{\epsilon}_s = A_3 (\sinh \alpha \sigma)^n \quad (3)$$

where A_2 , A_3 , α , β and n are constants independent of stress. If high temperature creep rate is controlled by the single deformation mechanism, there should be only one state equation. The reason why these three equation usually have been reported is incomprehensible. However, the equations of power law and

exponential law are satisfied by the an equation of hyperbolic sine law.¹⁷⁾ For values of $\alpha\sigma < 0.8$ (low stress level), Eq. (3) approximately reduces to Eq. (1) and $A_3\alpha^n = A_1$.¹⁷⁾ For values of $\alpha\sigma > 1.2$, Eq. (3) approximately reduce Eq. (2), therefore $A_3/2^n = A_2$ and $n\alpha = \beta$.¹⁷⁾

3. The Apparent activation Energy for High-Temperature Creep Deformation

Various experimental evidence showing that creep deformation is a thermally activated process has been available. The high temperature steady-state creep rate can be described by Arrhenius equation and namely the deformation process show the temperature dependence of creep rate through the factor $\exp(-\Delta H_c/RT)$ at constant stress, where ΔH_c is the apparent activation energy for the process which is controlling creep rate, R is the gas constant and T is the absolute temperature. Consequently, a power law (Eq. (1)), an exponential law (Eq. (2)) and a hyperbolic sine law (Eq. (3)) which are presented in above section are improved to Eq. (4)~(6) as a function of stress and temperature

$$\text{power law ; } \dot{\epsilon}_s = B_1 \sigma^n \exp(-\Delta H_c/RT) \quad (4)$$

$$\text{exponential law ; } \dot{\epsilon}_s = B_2 \exp(\beta\sigma) \exp(-\Delta H_c/RT) \quad (5)$$

$$\text{hyperbolic sine law ;}$$

$$\dot{\epsilon}_s = B_3 (\sinh \alpha\sigma)^n \exp(-\Delta H_c/RT) \quad (6)$$

In these equation, B_1 , B_2 , B_3 , α , β and n are constants independent of stress and temperature. Therefore, the deformation mechanism controlling the creep rate involve micromechanisms on the atomic scale. The apparent activation energy ΔH_c is determined on the assumption that structure of crept metal remain nearly constant within small temperature intervals at constant stress and that the apparent activation energy is independent of stress and strain. In generally, temperature-differential creep test method on based on these assumption have been used for determining apparent activation energy. Namely, a

sample is subjected to creep deformation at temperature T_1 at constant stress until a some strain is reached and at the point the temperature is changed abruptly to T_2 which may be slightly above or below T_1 . And then the apparent activation energy of steady state creep can be calculated from the ratio of the steady state creep rate immediately before and after the small change in temperature. Thus

$$\Delta H_c = \frac{\partial \ln \dot{\epsilon}}{\partial (-\frac{1}{RT})} \approx \frac{\ln \frac{\dot{\epsilon}_2}{\dot{\epsilon}_1}}{\frac{1}{R} (\frac{1}{T_1} - \frac{1}{T_2})} \quad (7)$$

where $\dot{\epsilon}$ is the steady state creep rate, T is the absolute temperature and R is the gas constant.

Fig. 4¹⁸⁾ shows the variation of the apparent activation energy for creep in pure polycrystalline aluminum with temperature between 78 and 850°K. As shown in this figure, three distinct ranges can be discerned. The apparent activation energy of creep ΔH_c increases sharply from about 5,000 to 27,500 cal/mol at low temperature below about 0.27 T_m , and remained relatively constant of about 28,000 cal/mol between about 0.27 and 0.43 T_m . It appears insensitive to temperature between about 0.5 and 1.0 T_m and then its constant value is about 35,500 cal/mol which is nearly equal to value of activation energy for aluminum self-diffusion. However value of apparent activation energy for creep increase only slightly with increasing temperature in this region because the elastic modulus decreases with increasing temperature.¹⁹⁾ The apparent activation energz for creep at temperatures above 0.5 T_m is found to be nearly equal to the activation energy for its self-diffusion by experimental results on various pure metals. Therefore at this temperature range, the nonconservative motion of jogs in screw dislocations,^{16), 20)~24)} climbing of edge dislocations,^{25)~27)} and the other diffusion controlled mechanism are considered as steady-state creep rate controlling mechanism. (The creep rate controlling mechanism will be described later) Fig. 5²⁸⁾ shows the relation between activation energy for self-diffusion

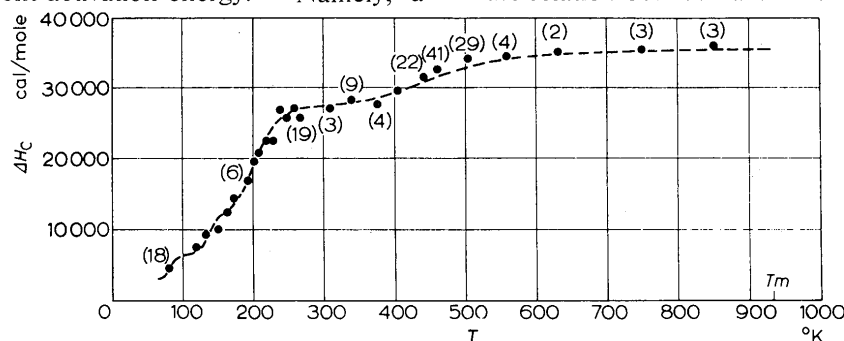


Fig. 4. Apparent activation energies for creep of pure aluminum as a function of the absolute temperature.
● Experimental data (Numbers in parenthesis refer to number of determinations made; only one determination made on all other points)¹⁸⁾

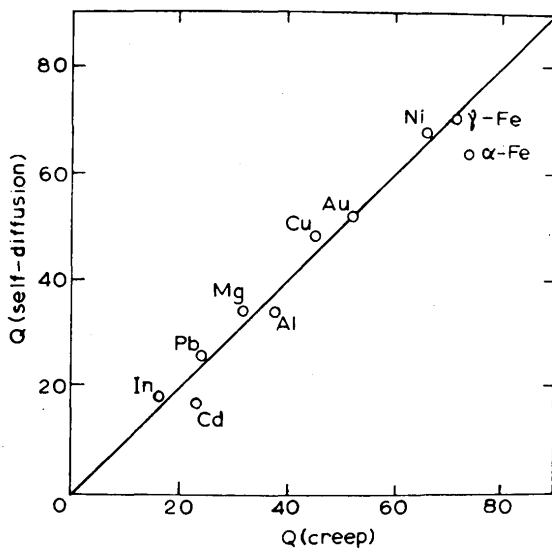


Fig. 5. Comparison of the activation energies of creep and of self-diffusion for a number of metals. The units for both scales are Kcal/mole.²⁸⁾

and apparent activation energy for high temperature creep in a number of pure metals.

On comparatively dilute solid solution alloys, there is little effect of composition on apparent activation energy of steady state creep at high temperature.^{7), 29) ~ 31)} However, as shown in Fig. 6,⁶⁾ an

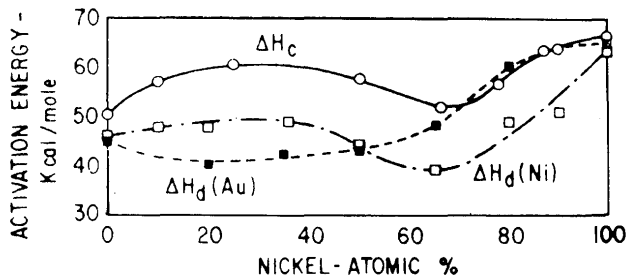


Fig. 6. Variation of the apparent activation energy for creep and self-diffusion in gold-nickel alloys.⁶⁾

apparent activation energy for creep changes with composition on alloys of Ni-Au complete solid solution binary system. The changes in apparent activation energy of creep with composition parallels the change in that of diffusion for nickel in the alloys, however apparent activation energy of creep is greater than that of diffusion of nickel by about 10,000 cal/mol. The apparent activation energy of creep agrees well with that for gold, at the range above 70 at%Ni. But these results is not clearly solved.

Fig. 7³²⁾ shows an apparent activation energy of creep for alpha iron determined by temperature-differential creep test. As shown in Fig. 7,³²⁾ the same apparent activation energy of creep was obtained throughout the primary and secondary stage of creep. This fact suggests that creep rate of both transient state range and steady state range are controlled by same deformation mechanism, notwithstanding that the structure is changing in transient creep range and is constant in steady state creep range.

4. The Internal Stress and Effective Stress on Creep Stress

The internal stress and the effective stress is called respectively athermal component or long range stress field and thermal component or short range stress field in other words. Seeger³³⁾ the first made a propose on investigation of yield stress of single crystal of metal in 1954 that the flow stress of metals can be considered to consist of two component; the one is internal stress τ_μ which depends slightly on temperature through the temperature dependence of the shear stress μ and the other one is effective stress τ^* which is more sensitively depend on temperature and strain rate $\dot{\gamma}$. Thus the flow stress τ_f is expressed as

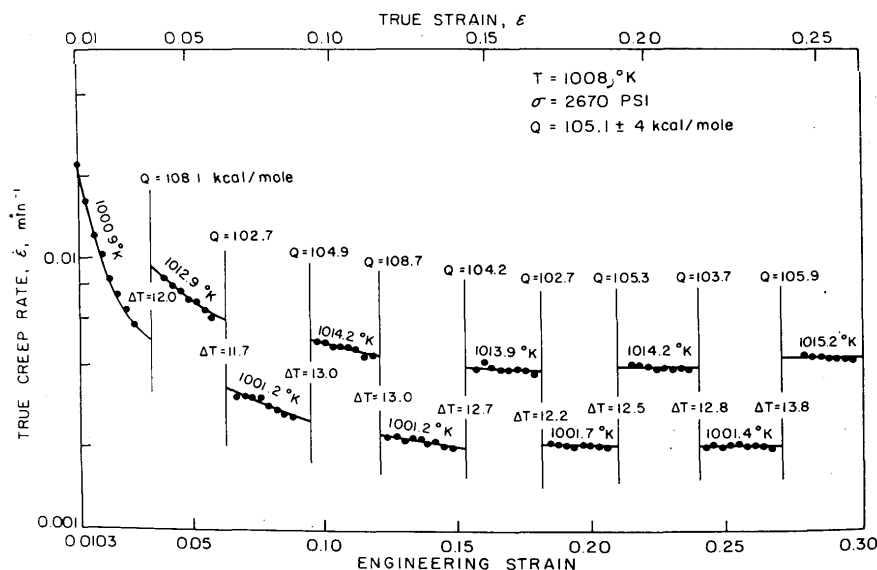


Fig. 7. Typical evaluation of the apparent activation energies.³²⁾

$$\tau_f = \tau_\mu + \tau^*(T, \dot{\gamma}) \quad (8)$$

This propose has been supported by cottrell,³⁴⁾ Haasen³⁵⁾ and other various investigators.^{36)~38)} Then, Corrad³⁹⁾ developed a detailed thermodynamic theory of thermally activated deformation on the foundation of the opinion that the flow stress consists of internal stress and effective stress. **Fig. 8**^{37), 39)} illustrates the relation between internal stress and effective stress against a dislocation moving through the crystal lattice. Namely, a moving dislocation will encounter the short-range obstacles (effective stress) superimposed on the long-range stress field (internal stress) which is considered to be a fluctuating field with positive and negative values of wavelength λ . Here, a possitive stress apposes the moving of disloection and a negative stress assists it. In generaly, it is believed that stress field of internal stress extend over a relatively large distance namely greater than about 10 atomic diameter while stress fields of effective stress is associated with short-range obstacle namely less than about 10 atomic distance. Therefor, the short-range stress fields can be overcome by tthe combined action of effective stress and thermal fluctuation, however the energy associated with overcoming long range stress field is so large that thermal fluctuations to make no significant contribution.

As shown in Fig. 8,^{37), 39)} the creep rate will be controlled by the thermally activated overcoming process of the strongest short-range obstacle on the top of the opposing long-range stress fields. On the creep at elevated temperature, the creep stress will be assisted by thermal fluctuations and deformation can occur at a stress τ which is less than τ_0 in Fig. 8.^{37), 39)} For the polycrystalline metals, τ can be approximated by $\sigma/2$ where σ is the creep stress and the shear strain γ is approximated by $\epsilon/1.4$ where ϵ is the tensile strain.

The various experimental techniques have been proposed for measuring internal stresses in a creep deformation of metals.^{33), 40)~55)} For the determination

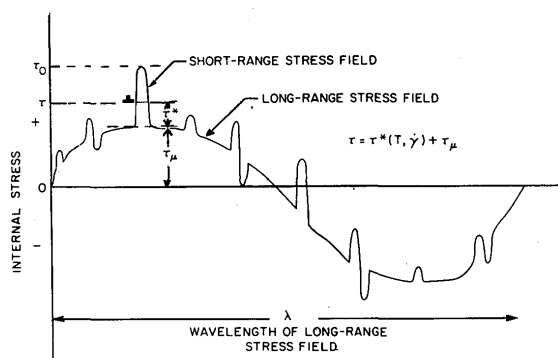
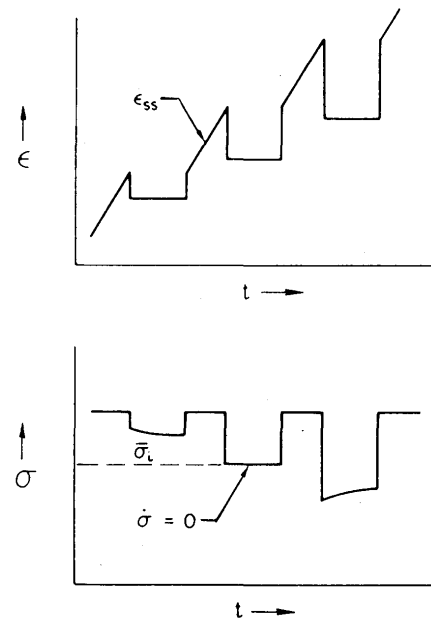
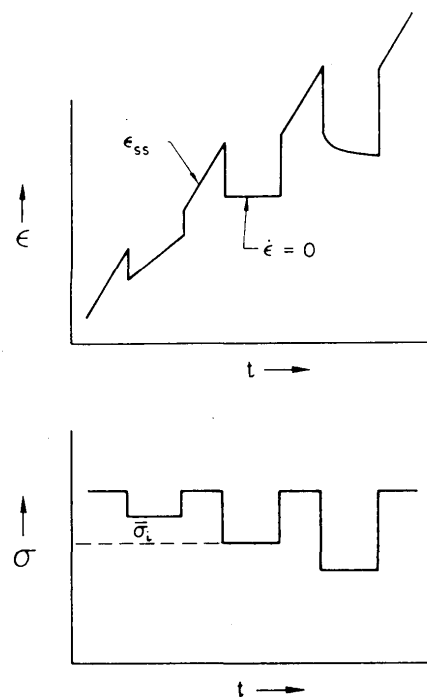


Fig. 8. Internal stress fields encountered by a dislocation moving through the crystal lattice.^{37), 39)}

of internal stress in high temperature creep stress, one must be especially careful that the structure must be maintained constantly during measurement. Therefor, the dip test is often used for measurement of them. **Fig. 9**⁵⁶⁾ (a) and (b) schematically illustrate the stress transient dip test and the strain transient dip test respectively. In the stress transient dip test, the sample is crept under a constant stress to the point at which



a) STRESS TRANSIENT DIP TEST



b) STRAIN TRANSIENT DIP TEST

Fig. 9. "Dip test" for defining the average internal stress.⁵⁶⁾

a measurement is to be made. Then, the creep stress is reduced suddenly and sample gage length hold constant. When the reduced stress equals the internal stress $(d\sigma/dt)_\epsilon = 0$, the other hand $(d\sigma/dt)_\epsilon < 0$ for a smaller stress reduction and $(d\sigma/dt)_\epsilon > 0$ for a larger stress reduction as shown in Fig. 9(a). In the strain transient dip test, the sample is crept under a constant stress to the point at which a measurement is to be made. Then the creep stress is reduced suddenly and held at new reduced stress. When the reduced stress equals the internal stress, creep rate is equal to zero, the other hand, the instantaneous strain rate after reducing stress is positive for a smaller stress reduction and that is negative for a larger stress reduction as shown in Fig. 9 (b).

The origins of internal stress has been discussed by various investigators for a long time. Seeger, Diehl, Mader and Rebstock⁵⁷⁾ proposed that internal stress is ascribed to back stress due to pile up dislocations on the slip plane and Solomon and Nix⁵⁸⁾ proposed that it occurs by the interaction between mobile dislocations and elastic stress field of other dislocations both mobile and immobile. Jonas⁵⁹⁾ proposed that the internal stress is concerned in either with isolated dislocations within cells or with cell boundaries, and Lloyd and Embury⁶⁰⁾ proposed too that it originates from the cell structure which occurs by plastic deformation. Recently, Hasegawa Ikeuchi and Karashima¹⁾ proposed that internal stress arises mainly long-range stress fields of cell boundaries consisting of dislocation tangles according to their experimental results for copper-16 at% Aluminum solid solution alloy as shown in Fig. 10.¹⁾ Fig. 11 schematically illustrates the relative internal stress level (σ_i/σ) in steady state of high temperature creep as a function of temperature, stress, solute atom concentration or stacking fault energy. Where σ is a creep stress and σ_i is a internal stress in σ . The value of (σ_i/σ) decreases with an increasing in temperature, creep stress, solute atom concentration and stacking fault energy. For example, Fig. 12⁶¹⁾ shows these relations on the high temper-

ature steady state creep of copper and copper-Aluminum solid solution alloys.

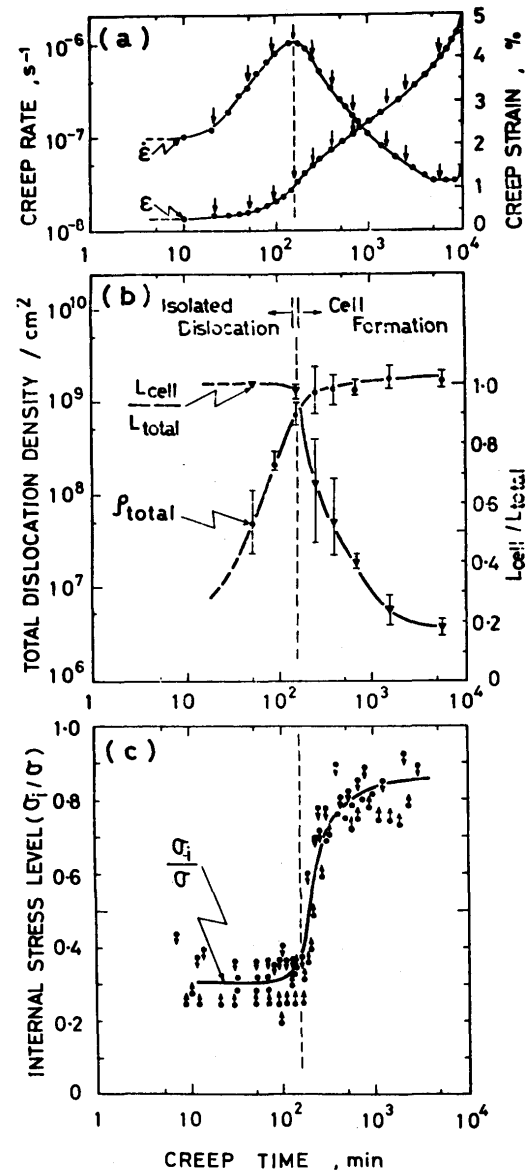


Fig. 10. Various parameters as a function of creep time. Cu-16 at% Al, 450°C, 39.2 MN/m² (4.0 kgf/mm²).
 (a) Creep strain (ϵ) and creep rate ($\dot{\epsilon}$).
 (b) Total dislocation density (ρ_{total}) and ratio of length of dislocations, (L_{cell}/L_{total}).
 (c) Average internal stress level (σ_i/σ).

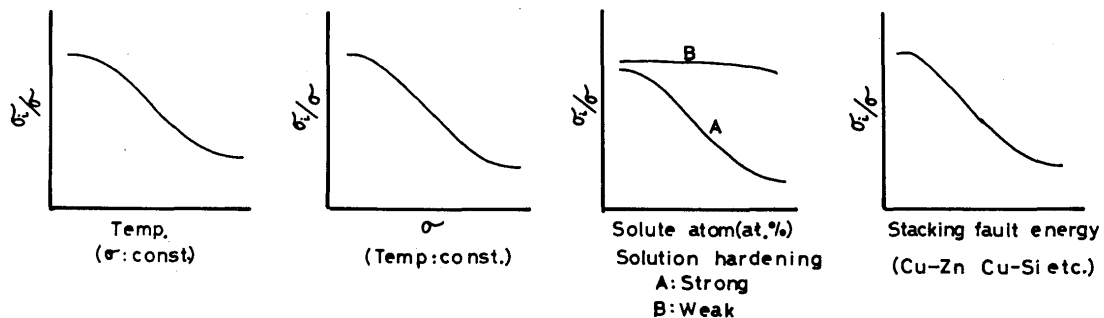


Fig. 11. The relative internal stress level as a function of temperature, applied stress, content of solute atom and stacking fault energy.

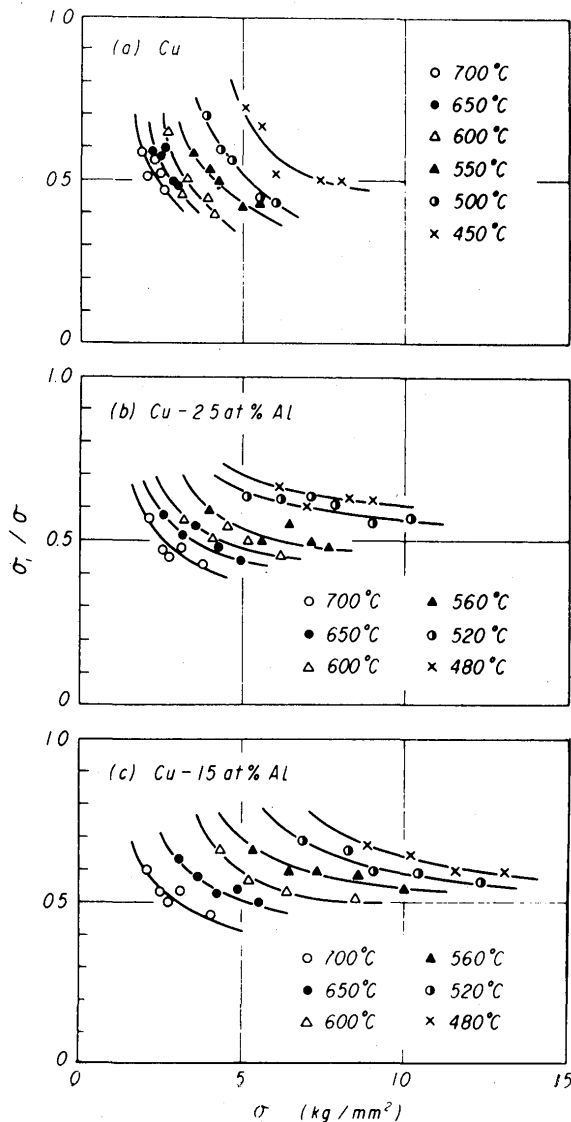


Fig. 12. The relative internal stress level as a function of applied stress during steady-state creep of Cu and Cu-Al alloys at 650°C.⁽⁶¹⁾

5. The Activation Area for High Temperature Creep Deformation

The high temperature creep rate is controlled only by a single thermally activated deformation mechanism and the energy which must be supplied by thermal fluctuations is given by the shaded area of the short range obstacle as shown in Fig. 8.^{(37), (39)} With increasing of temperature, τ can decrease at constant creep rate, until it become equal to τ_μ in Fig. 8.^{(37), (39)} Fig. 13^{(37), (62), (63)} illustrates a more detailed representation of a common type of thermal activation barrier. This energy barrier is not a function of τ^* and when a stress τ which is greater than τ_μ is applied, the dislocation moves up the force barrier to the position F given by $F = \tau^* b l^*$. Where b is the Burgers vector, and l^* is the length of the

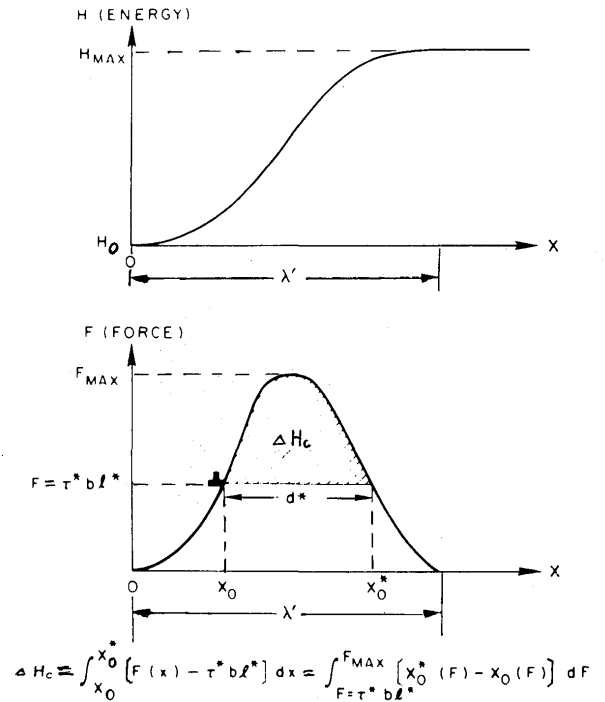


Fig. 13. A common type of thermal activation barrier.^{(37), (62), (63)}

segment between barriers.

The apparent activation energy which must be supplied by thermal fluctuation to overcome the obstacle is given by ΔH_c which is shaded area under the curve at the force-distance diagram in Fig. 13.^{(37), (62), (63)} The work done by the creep stress during overcoming of obstacle is

$$W = \tau^* b l^* d^* \quad (9)$$

where d^* is the activation distance namely $d^* = x_0^* - x_0$ in Fig. 13.^{(37), (62), (63)} And then activation area is given by $l^* d^*$ and is namely the area swept by the portion of the dislocation which is directly involved in the overcoming of the barrier. ΔH_0 means the total area under the force-distance curve which correspond activation energy for a dislocation segment to overcome in the absence of effective stress τ^* . And then, ΔH_c is given by the difference between ΔH_0 and the total work done by the effective stress τ^* during overcoming the obstacle

$$\Delta H_c = \Delta H_0 - \left[\left(\int_{x_0}^{x_0^*} F(x) dx \right) + (\tau^* b l^* d^*) + \left(\int_{x_0}^{x_0^*} F(x) dx \right) \right] \quad (10)$$

or,

$$\Delta H_c = \Delta H_0 - v \tau^* \quad (11)$$

where v is simply the sum of the energy terms within the brackets divided by τ^* . And then, v has the dimensions of a volume and is called by total activation volume.

Consequently, a power law, an exponential law

and hyperbolic sine law which indicate the relation between steady state creep rate and creep stress are rewritten to following equations respectively

$$\text{power law } \dot{\epsilon}_s = C_1 \sigma^n \exp \left(-\frac{\Delta H_0 - \nu \tau^*}{RT} \right) \quad (12)$$

$$\text{exponential law } \dot{\epsilon}_s = C_2 \exp(\beta \sigma) \exp \left(-\frac{\Delta H_0 - \nu \tau^*}{RT} \right) \quad (13)$$

hyperbolic sine law

$$\dot{\epsilon}_s = C_3 (\sinh \alpha \sigma)^n \exp \left(-\frac{\Delta H_0 - \nu \tau^*}{RT} \right) \quad (14)$$

where $\nu \tau^*$ is the total mechanical work done to overcome a short range barrier, C_1 , C_2 , C_3 , α , β and n are constants independent of stress and temperature.

In general, the relaxation method^{(38), (64)~(68)} and the strain-rate suddenly change method⁽³⁰⁾ are used to determine an activation area of high temperature creep deformation, however it is difficult to measure them exactly.

6. Deformation Mechanism of High Temperature Creep

The correlation which exists between the activation energies of creep and of self-diffusion in metals as shown in Fig. 5⁽²⁸⁾ suggests that the mechanism of self-diffusion is involved in the creep rate controlling process. In consequence, the non-conservative motion of jogs in screw dislocation,^{(16), (20)~(24)} climbing of edge dislocations^{(25)~(27)} nonconservative motion of super-jog,⁽⁶⁹⁾ climbing of dipoles or dislocation roops⁽⁷⁰⁾ climbing of sub-boundaries⁽⁷¹⁾ and etc. has been presented as a possible rate-controlling process in steady state creep rate of metals at high temperature.

Among so many proposed deformation mechanism, the nonconservative motion of jogs in screw dislocation and climbing of edge dislocations are described briefly in this review.

Nonconservative motion of jogs in screw dislocation;

The screw dislocations contain thermal jogs and also acquire jogs by interaction with forest dislocations⁽⁷²⁾ during steady state creep deformation. The jogs in screw dislocation, of course, have edge dislocation segments and so those jogs which do not lie in the slip plane of the moving screw dislocation emit a stream of vacancies or interstitial atoms in their wakes to move with the dislocations. These jogs act as mobile pins which move with the dislocation. A vacancy emitting jog produces a vacancy on an interstitial-atom emitting jog absorb a vacancy when

the screw dislocation moves forward a distance b , the Burger's vector, and consequently the motion of screw dislocation with jog depends on the flow of vacancies to and from jogs. This fact predicts that an activation energy of steady state creep deformation is nearly equal to that of diffusion if steady state creep rate is controlled by nonconservative motion of jogs in screw dislocation.

Climbing of edge dislocation;

Weertman^{(25), (26)} suggests that the rate-controlling process in the high temperature creep of metals is the climbing of edge dislocations. The destruction of edge dislocations occurs when those of opposite sign on different slip planes climb towards each other and suffer annihilation upon meeting.^{(26), (73)} Some edge dislocations climb by creating vacancies, which diffuse to other edge dislocations which climb by absorbing vacancies. On the average, the number of vacancy-creating edge dislocation is equal to the number of vacancy-absorbing edge dislocation, and consequently the creep rate is controlled by the diffusion of vacancies between dislocations. Therefore, an activation energy of steady state creep deformation is nearly equal to that of diffusion too, if steady state creep rate is controlled by climbing of edge dislocation.

self-diffusion is

References

- 1) T. Hasegawa, Y. Ikeuchi and S. Karashima; Metal Science Journal vol. 6 (1972) p. 78.
- 2) F. Garofalo; Trans. Met. Soc. AIME vol. 227 (1963) p. 351.
- 3) E. M. Howard, W. C. Barmore, J. D. Mote and J. E. Dorn; Trans. Met. Soc. AIME vol. 227 (1963) p. 1061.
- 4) O. D. Sherby; Acta Met. vol. 10 (1962) p. 135.
- 5) J. Weertman; Trans. Met. Soc. AIME vol. 218 (1960) p. 207.
- 6) C. M. Sellars and A. G. Quarrell; J. Inst. Metals vol. 90 (1961-62) p. 329.
- 7) T. Aramaki, T. Endo, T. Enjyo and M. Adachi; J. Japan Inst. Metals vol. 33 (1970) p. 1030.
- 8) W. R. Cannon and O. D. Sherby; Met. Trans. vol. 1 (1970) p. 1030.
- 9) B. L. Jonas and C. M. Sellars; Metal Science Journal vol. 4 (1970) p. 96.
- 10) J. Weertman; J. Appl. Physics vol. 28 (1957) p. 1185.
- 11) F. R. N. Nabarro; In Report of a Conference on the strength of solids The physical Society, London (1948) p. 75.
- 12) C. Herring; J. Appl. Phys. vol. 21 (1950) p. 21.
- 13) O. D. Sherby and P. M. Burke; Mechanical Behavior of Crystalline Solids at Elevated Temperature. Progress in Material Science vol. 13 (1966) p. 369.
- 14) A. J. Kenedy; Processes of creep and fatigue in metals (Oliver and Boyd, London) (1962) p. 147.
- 15) D. Mclean; Mechanical properties of metals (Wiley, New York) (1962) p. 285.
- 16) C. R. Barrett and W. D. Nix; Acta Met. vol. 13 (1965) p. 1247.
- 17) F. Garofalo; Fundamentals of Creep and Creep-Rupture in Metals (The Macmillan Company, New York) (1965) p. 51.
- 18) O. D. Sherby, J. L. Lytton and J. E. Dorn; Acta Met. vol. 5 (1957) p. 219.

- 19) C. R. Barrett, A. J. Ardell and O. D. Sherby; Trans. Met. Soc. AIME vol. 230 (1964) 200.
- 20) J. Friedel; Phil. Mag. vol. 46 (1956) p. 1169.
- 21) P. B. Hirsch and D. H. Warrington; Phil. Mag. vol. 6 (1961) p. 735.
- 22) J. J. Holwes; Acta Met. vol. 15 (1967) p. 570.
- 23) W. D. Nix; Acta Met. vol. 15 (1967) p. 1079.
- 24) W. J. Duffin and F. A. Nichols; Acta Met. vol. 18 (1970) p. 1121.
- 25) J. Weertman; J. Appl. Phys. vol. 26 (1955) p. 1213.
- 26) J. Weertman; J. Appl. Phys. vol. 28 (1957) p. 362.
- 27) F. R. N. Nabaro; Phil. Mag. vol. 16 (1967) p. 231.
- 28) A. J. Kennedy; Processes of Creep and Fatigue in Metals (John Wiley and Sons, New York) (1963).
- 29) T. Endo, T. Nomura, T. Enjo and M. Adachi; J. Japan Inst. Metals vol. 35 (1971) p. 427.
- 30) T. Endo, T. Aramaki, T. Enjo and M. Adachi; J. Japan Inst. Metals vol. 32 (1968) p. 1276.
- 31) T. Endo, S. Watanabe, T. Enjo and M. Adachi; J. Japan institute of Light Metals. vol. 19 (1969) p. 310.
- 32) Y. Ishida, C. Y. Cheng and J. E. Dorn; Trans. Met. Soc. AIME vol. 236 (1966) p. 964.
- 33) A. Seeger; Z. Naturforsch 9a (1954) 758, 819, 856.
- 34) A. H. Cottrell and R. J. Atokes; Proc. Roy. Soc. vol. 233 (1955) p. 17.
- 35) P. Hassen; Phil. Mag. vol. 3 (1958) p. 384.
- 36) D. F. Stein and J. R. Low; J. Appl. Phys. vol. 31 (1960) p. 362.
- 37) H. Conrad; J. Metals vol. 16 (1964) p. 582.
- 38) J. W. Christian; Acta Met. vol. 12 (1964) p. 99.
- 39) H. Conrad; Mater. Sci. Eng. vol. 6 (1970) p. 265.
- 40) U. Essmann; Phys. Status Solidi vol. 6 (1965) p. 723.
- 41) F. Ramsteiner; Mater. Sci. Eng. vol. 1 (1966) p. 206.
- 42) G. B. Gibbs; Phil. Mag. vol. 13 (1966) p. 317.
- 43) F. W. Young Jr and F. A. Sherrill; Can. J. appl. phys. vol. 45 (1967) p. 757.
- 44) H. Kronmüller; Can. J. appl. phys. vol. 45 (1967) p. 631.
- 45) K. Zeyfang; phys. status Solidi vol. 24 (1967) p. 221.
- 46) J. C. M. Li; Can. J. Appl. phys. vol. 45 (1967) p. 493.
- 47) D. T. Peterson and R. L. Skaggs; Trans. Met. Soc. AIME vol. 242 (1968) p. 922.
- 48) H. Alexander; phys. Status Solidi vol. 27 (1968) p. 391.
- 49) G. Sargent and H. Conrad; Scripta Met. vol. 3 (1969) p. 43.
- 50) R. L. Jonas and H. Conrad; Trans. Met. Soc. AIME vol. 245 (1969) p. 779.
- 51) P. M. Kelley and J. M. Round; Scripta Met. vol. 3 (1969) p. 85.
- 52) C. N. Ahlquist and W. D. Nix; Scripta Met. vol. 3 (1969) p. 679.
- 53) A. A. Solomon; Rev. scient Instrum vol. 40 (1969) p. 1025.
- 54) A. A. Solomon, C. N. Ahlquist and W. D. Nix; Scripta Met. vol. 4 (1970) p. 231.
- 55) H. Conrad and K. Okazaki; Scripta Met. vol. 4 (1970) p. 259.
- 56) C. N. Ahlquist and W. D. Nix; Acta Met. vol. 19 (1971) p. 373.
- 57) A. Seeger, J. Diehl, S. Mader and H. Rebstock; phil Mag. vol. 2 (1957) p. 323.
- 58) A. A. Solomon and W. D. Nix; Acta Met. vol. 19 (1971) p. 373.
- 59) J. J. Jonas; Acta Met. vol. 17 (1969) p. 397.
- 60) D. J. Lloyd and J. D. Embury; Metal Sci. J. vol. 4 (1970) p. 6.
- 61) S. Kikuchi, M. Kajitani, T. Enjo and M. Adachi; J. Japan Inst. Metals vol. 37 (1973) p. 228.
- 62) A. Seeger; Phill. Mag. vol. 1 (1956) p. 651.
- 63) A. Seeger and P. Schiller; Acta Met. vol. 10 (1962) p. 348.
- 64) P. Feltham; Phil. Mag. vol. 6 (1961) p. 259.
- 65) F. W. Noble and D. Holl; Acta Met. vol. 12 (1964) p. 1089.
- 66) P. Rodriguez; J. Mater. Sci. vol. 3 (1968) p. 98.
- 67) F. Guiv; Scripta Met. vol. 3 (1969) p. 753.
- 68) C. N. Ahlquist; Scripta Met. vol. 5 (1971) p. 185.
- 69) T. Watanabe and S. Karashima; Trans. Japan Inst. Metals vol. 11 (1970) p. 159.
- 70) R. Chang; Physics and chemistry of ceramics (Gordon and Breach, New York) (1963) p. 275.
- 71) J. C. M. Li; Trans. Met. Soc. AIME vol. 227 (1963) p. 1474.
- 72) W. T. Read Jr; Dislocation in Crystals (Mc Graw Hill) (1953).
- 73) G. Shuech; Creep and Recovery (ASM, 1957) p. 199.