Lawrence Berkeley National Laboratory
Recent Work

Title
EXPERIMENTAL CORRELATIONS FOR HIGH-TEMPERATURE CREEP

Permalink
https://escholarship.org/uc/item/31p4z5v2

Authors
Mukherjee, Azaiya K.
Bird, James E.
Dorn, John E.

Publication Date
1968-10-01

University of California

Ernest O. Lawrence
Radiation Laboratory

EXPERIMENTAL CORRELATIONS FOR HIGH-TEMPERATURE CREEP

Amiya K. Mukherjee, James E. Bird, and John E. Dorn

October 1968

DISCLAIMER

UCRL-18526

UNIVERSITY OF CALIFORNIA

Lawrence Radiation Laboratory
Berkeley, California

AEC Contract No. W-7405-eng-48

EXPERIMENTAL CORRELATIONS FOR HIGH-TEMPERATURE CREEP

Amiya K. Mukherjee, James E. Bird, and John E. Dorn

October 1968

UCRL-18526

UNIVERSITY OF CALIFORNIA

Lawrence Radiation Laboratory

Berkeley, California

EXPERIMENTAL CORRELATIONS FOR HIGH-TEMPERATURE CREEP

by

Amiya K. Mukherjee[+], James E. Bird[*] & John E. Dorn[++]

+Associate Professor, Department of Mechanical Engineering, University of

  California, Davis, California.

 *Research Assistant, IMRD of Lawrence Radiation Laboratory, University of

  California, Berkeley, California.

++Professor Materials Science, Department of Mineral Technology and Senior

  Research Metallurgist, Inorganic Materials Research Division, Lawrence

  Radiation Laboratory, University of California, Berkeley, California

ABSTRACT

Substantial strides have been made over the recent past in uncovering the role of many of the important factors that determine the physical mechanisms of high-temperature creep, but many significant details yet require clarification. The apparent activation energies for high-temperature creep of metals and dilute solid-solution alloys are independent of creep stress and strain. In general they are in good agreement with the activation energy of self diffusion. It has been firmly established that the rate is controlled by the rate of atom mobility and is proportional to the diffusivity. The substructural changes in creep are insensitive to the temperature and depend on the time and stress. During steady-state creep the significant substructural details remain constant. The exponent of the stress dependence of the steady-state creep rate has usually a value of about five, although many exceptions are known.

A complete solution of the creep problem in all its generalities cannot yet be given. This is due to a lack of detailed knowledge of effects as such factors as modulus, grain size, stacking-fault energy, subgrain size and other substructural features, etc. When the temperature dependence of modulus is taken into account the true enthalpy of activation for creep in pure metals is in excellent agreement with the activation enthalpy for self-diffusion. The activation energy for creep does not depend on the jog energy or the stacking-fault energy. Although some experimental observation suggests that the creep rate might increase with an increase in stacking-fault energy, this question is not yet satisfactorily resolved. Whereas the subgrain size decreases with increasing

stress it is not known how it depends on modulus or on the stacking-fault energy or how it affects the creep rate. Recent evidence suggests that the effect of grain size on creep rate is relatively small.

No satisfactory theoretical model exists for primary creep where the structure continually changes with time. Two kinds of models have been suggested for steady-state creep: the motion of jogged screw dislocations and the climb of edge dislocations. The screw dislocation models suffer from various handicaps, and major modifications are needed to obtain the correct activation energies and stress laws. The climb models appear to account more readily for the observed activation energies and stress laws for creep. However, some of the assumptions in several climb models, e.g., dislocation pile-ups, effectiveness of Lomer-Cottrell barriers, etc., that have been used to develop the stress laws are questionable. Furthermore, assumptions regarding constancy of dislocation sources independent of the stress level are not very tenable.

The eventual solution of the problem of steady-state creep will depend on a more intimate understanding of the formation, nature and recovery of the substructure. More complete empirical data are needed on the creep rates. Also more detailed and systematic investigations are needed on the substructures that are developed with specific emphasis on their significance to the creep mechanism.

rapidly approach immeasurably small rates as the gliding dislocations become arrested at higher energy barriers that might result from long-range stress fields, attractive junctions, etc. If, however, the temperature is above about $T_m/2$, one-half of the melting temperature, creep nevertheless continues to take place despite the fact that glide dislocations now become arrested at quite high energy barriers. In contrast, however, it is just at these high temperatures where creep continues ad infinitum until it is modified by such auxiliary phenomena as grain boundary fissuring, necking and grain boundary cavitation which lead to rupture. The significant observation to be made here is that at $T_m/2$ diffusion first becomes reasonably rapid. Above this temperature dislocations are known to acquire a new degree of freedom in their motion; edge-components of dislocations are no longer confined to glide exclusively on their original slip planes and may climb to new planes. Although other suggestions have been made from time to time, it seems inevitable that climb itself must necessarily play a role in any model for high-temperature creep; it need not, however, be the rate-controlling mechanism. For example if climb were extremely rapid and some glide processes much slower, creep might then take place in an almost fully climb-recovered substructure as controlled by a slow-glide mechanism.

It is the objective of this paper to evaluate the present status of experimental knowledge on the high-temperature creep of some metals and alloys with particular reference to the significance of climb of dislocations. A number of reviews, symposia and monographs are now available on this as well as the much broader areas of interest in creep[1-13]. The viscous-creep

mechanism concerned with the stress-directed diffusion of vacancies in poly-crystalline aggregates, first described by Nabarro[14] and later further elaborated by Herring[15], will not be considered here. Rather emphasis will be given to examples of creep of coarse-grained metals under stresses where dislocation mechanisms contribute principally to the creep. Furthermore the creep of dispersed phases, as e.g. SAP, which often gives results that differ substantially from those usually obtained in single phase metals and therefore appears to be the result of special mechanisms, will also be omitted from consideration.

This report is presented in three major sections. The first will be concerned with the effects of the independent variables of stress and temperature on the creep rate and the possible influence of crystal structure, modulus of elasticity, stacking-fault energy and grain size on these data. In the second section emphasis will be given to a review of the major substructural changes that attend high-temperature creep. And thirdly a summary will be presented on the theoretical implications of the known experimental facts relative to the validity of various proposed high-temperature creep mechanisms.

II. MECHANICAL BEHAVIOR

The dependence of the creep rate, $\dot{\epsilon}$, on the stress, σ, and temperature, T, is basic to an understanding of creep. Any acceptable theory of creep must account well for all such reliable data. Several factors, however, serve to complicate the picture. Frequently test materials are not adequately characterized relative to purity, grain size and other pertinent quantities. Good data require excellent control of the temperature and stress. In order to cover a wide range of stresses at a given constant temperature, some tests must be conducted over very long periods of times. As will be described later, the subgrain size developed at the secondary stage of creep, although independent of the test temperature, depends on the stress. Consequently, the effect of stress may not easily be separated from the effects of coincident substructural changes. Furthermore the proportion of grain-boundary shearing to the total creep strain increases as the stress is lowered. Other factors such as crystal structure, moduli of elasticity, stacking-fault energy, diffusivity, etc. also require detailed consideration.

A. Stress Law

The effect of stress on the secondary creep rate will be emphasized here, first because there are more data in this area than others and secondly because most existing creep theories are concerned principally with the secondary stage of creep. Although other empirical functions have been suggested, the secondary creep rate can usually be represented as a power of the stress according to

$$\dot{\epsilon}_s = \sigma^n f\{T\} \tag{1}$$

In general the exponent n is substantially constant over a wide range range of stress and temperature. At high stresses the secondary creep rate appears to increase more rapidly than that given by a constant value of n. Occasionally, at quite low stress levels,[16,17] the secondary creep rate is somewhat higher than the value obtained by extrapolation of Eq. 1. In some metals this is believed to arise [17,18] from the greater percentage of grain boundary shearing accompanying grain straining at such low stresses.

As shown in Fig. 1 the values of n vary from about 3 to about 7 for various solid solution alloys and pure metals. Most frequently pure metals give values of 4.2 < n < 6.9 which suggests [19] an average value of n = 5. Some alloys [20,21] (e.g. Al - Mg and Ni - Au) exhibit values of n $\simeq$ 3 whereas others [22,23] (e.g. Fe - Si and Fe - Al) have values as high as about 6 < n < 7.

The observed variations in n for metals appear to bear no systematic relationship with crystal structure, grain size, or stacking-fault energy or any other known factor. However the data for those alloys which give n $\simeq$ 3 are usually in fair agreement with Weertman's viscous glide theory.

B. Apparent Activation Energies

Since creep is a thermally-activated process, its temperature dependence must arise principally from an Arhenius type of expression. Early investigations[3] revealed that the temperature dependence was frequently given by the expression

$$\dot{\varepsilon}_s = f\{\sigma\} \, e^{-\frac{Q_c'}{RT}} \qquad (2)$$

with rather good accuracy. The apparent activation energy, Q_c', was shown to be insensitive to the stress; it was in very close agreement with the activation energy for self diffusion, H_D. More recent investigations[8-13] have invariably served to confirm the validity of Eq. 2, and to reaffirm the agreement of Q_c' with H_D.

C. Modulus of Elasticity

When Eqs. 1 and 2 are combined the empirical relationship

$$\dot{\varepsilon}_s \, e^{\frac{Q_c'}{RT}} = A' \, \sigma^n \tag{3}$$

results. A', however, differs greatly from metal to metal. McLean and Hale[24] made the important observation that variation in A' amongst the various metals and alloys seems to depend systematically on their moduli of elasticity. They have shown that the correlation of secondary creep rates among the various metals and alloys is considerably improved when Eq. 3 is rewritten as

$$\dot{\varepsilon}_s \, e^{\frac{Q_c}{RT}} = A'' \left(\frac{\sigma}{G}\right)^n \tag{4}$$

where G is the shear modulus of elasticity. Similarly Sherby[13,19] suggested the empirical relationship

$$\frac{\dot{\varepsilon}_s}{D} = A''' \left(\frac{\sigma}{E}\right)^n \tag{5}$$

where D is the self diffusivity and E is Young's modulus of elasticity, as
a basis for correlating the secondary creep rates of metals. When the temperature
variation of E was included, slightly improved correlations were obtained.

Despite the virtues of Eq. 5, two major objections can be raised against
it. First it is not dimensionally correct. In order to permit unbiased comparisons,
the term preceeding the equality sign should be multiplied by the square of some
significant length, which might have different values for the various metals.
Secondly Eq. 5 is not wholly consistent with the requirements for thermally-
activated or diffusion-controlled mechanisms of creep. The driving force of any
thermally-activated creep mechanism depends on a free-energy or chemical-potential
gradient resulting from the applied stress or a local stress concentration. The
insensitivity of the apparent activation energy to the stress merely reflects the
fact that the usual exponential terms in the gradient expression reduce to the
linear case. This suggests a relationship such as

$$\dot{\varepsilon}_s = A'''' \left(\frac{\sigma}{G}\right)^{n-1} \left(\frac{\sigma \Omega}{kT}\right) e^{-\frac{H_c}{RT}} \qquad (6)$$

for simple cases where the entropy is not a fuction of the temperature and where
Ω is an activation volume and H_c is the activation enthalpy for creep; the shear
stress, G, enters as a result of dislocation interactions. As will be shown, when
the temperature variation of G is taken into consideration, H_c, using Eq. 6, is in
excellent agreement with H_D for most of the available data on pure metals. This
emphasizes that high temperature creep is diffusion-controlled.

The value of H_c is directly determinable from Eq. 6 or from the temperature sensitivity of the creep rate:

$$H_c = \frac{\Delta \ln \dot{\varepsilon}}{\Delta\left(-\frac{1}{RT}\right)} + R \ln \frac{T_2}{T_1}\left(\frac{G_2}{G_1}\right)^{n-1} \Bigg/ \left(\frac{1}{T_1} - \frac{1}{T_2}\right) \tag{7}$$

Since high-temperature creep is diffusion-controlled, comparisons of the secondary creep rates might be based on

$$\frac{\dot{\varepsilon}_s kT}{DGb} = A\left(\frac{\sigma}{G}\right)^n \tag{8}$$

where the Burgers vector, b, is tentatively taken to be the significant dimension in either diffusion or deformation to provide the required dimensionless expression.

The transition from Eq. 5 to Eq. 8 does not alter the significance of n, and changes only slightly the temperature effects. Nevertheless the change is not trivial when comparisons are made between various metals in an attempt to uncover what other factors might be involved in creep.

Eq. 8 is yet largely empirical and attempts only to systematize the effects of D, T, G and b on the secondary creep rate. If other factors are also pertinent, Eq. 8 will have to be reformulated so as to take these factors into appropriate consideration.

D.   Comparison with Experimental Results

A detailed evaluation of the nominal correlation of experimental data with

the trends suggested by Eq. 8 suffers a number of severe handicaps:

1.   Few constant stress creep data are available because most tests have

      been conducted under constant load.

2.   Often creep data are obtained at temperatures below or above the

      temperature range where diffusion-controlled creep occurs.

3.   Data on the unrelaxed shear modulus of elasticity at high temperatures are

      often not available.

4.   Data on self diffusivities are not always accurate.

5.   In some tests grain growth and even recrystallization took place at the

      higher test temperatures. These changes are known to produce abnormally

      high creep rates in instances where they have been observed. Occasionally

      investigators neglected to comment or even check on these possibilities

      when reporting their data.

6.   Precautions to limit possible errors due to high temperature oxidation were

      exercised in only a few cases.

In general the test data used here (vide Tables I and II) were obtained from

constant-stress tests. The Mo, Ta, and some W Data, however, were obtained

under constant-load conditions and then reported in terms of the true stress

under steady-state conditions. The other W data were obtained from constant-

load tests. The Ni data from reference M were obtained from constant-load tests

whereas those from reference K refer to constant-stress data. Similarly the Ag

data from reference Q and S refer to constant-load tests whereas those given in

reference R were obtained under constant stress.

Table I. Summary of Mechanical Behavior

Metal	References for Creep Data	References for Modulus Data	References for Diffusion Data	n	H_c, ΔT tests* H_c	Data Source	H_c, Graphical Method** H_c	Data Source	H_D	$\dfrac{\dot{\varepsilon}_s kT}{DGb}$ when $\sigma/G = 2\times10^{-4}$
Al	A,B,C,D	E	F	4.4	34.0	C	$\begin{cases} 34.0^{-0.6}_{+1.6} \\ 34.0^{-0.9}_{+0.4} \end{cases}$	A B	34.0	1.8×10^{-10}
Cu	G,H,D	I	J	4.8	48.4	G	50 ± 2	G	47.1	1.3×10^{-12}
Au	K	I	L	5.5	48 ± 5	K	___		41.7	2.6×10^{-12}
Ni	K,M,N,D	O	P	4.6 5.2	66.5 64 ± 6	N K	66 ± 4	M	66.8	3.7×10^{-12}
Ag	Q,R,S,D	I	T	~5.3	43	Q	40-60	Q	44.1	5.7×10^{-14}
Pb	U,V	W	X	~4.2	24.2 ± 2.5	V	___$^+$	U	24.2	1.1×10^{-11}
Pt	Y	W	z††	7.6	___		66 ± 4	Y	66.6	1.2×10^{-11}
Alpha Fe	a,b	c†	d	6.9	63.0	a	___		60-64	‡2.2×10^{-11}
Fe-4%Si	e	c†	f	5.75	71.4	e	___		54.4	___
Ta	g	g	h	4.2	114 ± 4	g	110 ± 12	g	110.0	2.2×10^{-10}
Mo	i,j	k	l	4.3	~93	j	109***	i	92.2	5.0×10^{-12}
W	m,n,o	k	p††	5.4-6.2	___				140.3	~7×10^{-10}
β-Tℓ	q	W	r	5.8	~21	q	___$^+$	q	20.0	5.1×10^{-11}

Table I. Continued

Metal	References for			n	H_c, ΔT tests*		H_c, Graphical Method**		H_D	$\dfrac{\dot\epsilon_s kT}{DGb}$ when σ/G=2X10^{-4}
	Creep Data	Modulus Data	Diffusion Data		H_c	Data Source	H_c	Data Source		
α-Tℓ	q	W	r	5.3	21.0	q	21±2	q	22.7	9.1 X 10^{-12}
Cd	s	t	u	4.3	19±2	t	20.3 ± 0.2	s	19.1	‡2.1 X 10^{-13}
Zn	v,w	W	x	6.1	21.6±1	w	24.5±1.5	v+w	24.3	4.0 X 10^{-13}

* H_c recalculated from ΔT tests using Eq. 7b.

** H_c estimated from $\dot\epsilon_s TG^{n-1}/\sigma^n$ vs. 1/T plot.

*** There was much scatter in $\dot\epsilon_s TG^{n-1}/\sigma^n$ at four test temperatures; this is the only value of H_c that could be assigned to fit within all four bands of scatter.

† G calculated from random Young's modulus values.

†† D used in calculations is an average of experimentally determined values, as recommended by this reference.

+ Data is too scattered to assign a value of H_c.

‡ Extrapolated from higher stresses

Note: For the fcc metals Al, Cu, Au, Ni, and Ag values of elastic compliances vs. temperature are available. For these metals the unrelaxed shear modulus G was calculated from the relation $G = \sqrt{\tfrac{1}{2} C_{44}(C_{11}-C_{12})}$. The resulting value represents the isotropic shear modulus in the (111) plane[25] and provides a very good approximation to the effective G for dislocations on the (111) plane according to anisotropic dislocation theory (see ref. y). For all five metals G varied linearly with temperature over the temperature range indicated in Table II. For the other metals analysed G was estimated from data on Young's modulus E as a function of temperature and from the value of Poisson's ratio μ at 20°C (obtained from ref. z), according to the relation G = E/2 (1+μ). The Young's modulus values were linearly extrapolated from lower temperatures when it was apparent that relaxations had unduly lowered their value.

Table II. Pre-History of Creep Specimens

Metal and Ref.	Reported Purity in % and Impurities in ppm	Thermomechanical History Prior to Testing	Grain Size (mm)	Temperature Range for which $H_c \simeq H_D$ According to ΔT Tests	Temperature Range of Creep Tests Used in Fig. 5-7	Testing Method	Comments
Al(a) (A)	99.995% 20 Fe 17 Cu 10 Si,2 Mg 6 Na + Ca	Specimens annealed at 650°F for ½ hr.	0.05 or 1 - 3	$T > 410°F$ Ref (C)	500-1100°F. (260-593°C) 400°F tests not used	Const. stress tensile tests in air	Marked grain growth was observed in aluminum tested at 900 & 1100°F.
Al(b) (B)	99.996%	After machining of tensile specimens, 4 hour anneal at 500°C	4	$T > 210°C$ Ref(C)	250-550°C. Tests at 150°C and 200°C were not used	Tensile testing in air	Tests were run at a strain rate which increased systematically with specimen extension. Strain rate varied ~25% during course of tests; steady state creep stress increased about 5%. To compensate for this variation, the steady-state stress was recorded at a constant value of strain, $\varepsilon = 0.4$.
Cu(a) (G)	99.99% 40 Fe 8 S, 6 Ag 5 Ni 4 Pb	Specimens initially cold-drawn rods; annealed in vacuum ½ hr. at 650°C	0.03	$T=700°C$ Ref(G)	700°C. Tests at $T \leq 630°C$ not used	Const. stress tensile tests in air	Tests made at $T \leq 630°C$ not included because ΔT tests showed $H_c < H_D$. Grain growth probably occurred during testing.
Cu(b) (H)	99.995% 20 O 8 Ag 2 H 2 Ca	Material hot-rolled, recrystallized, cold-rolled; annealed in air or H_2 at 1000°C for 1 hr.	1.0	$T=700°C$ Ref(G)	700-950°C	Const. stress tensile tests in H_2	Tests made at temperatures between 660-700°C were not included because recrystallization occurred during testing.

Table II. Continued

Metal and Ref.	Reported Purity in % and Impurities in ppm	Thermomechanical History Prior to Testing	Grain Size (mm)	Temperature Range for which $H_c \simeq H_D$ According to ΔT Tests	Temperature Range of Creep Tests Used in Fig. 5-7	Testing Method	Comments
Au (K)	99.99% <10 non-metallic <80 metallic	Final heat-treatment $\frac{1}{2}$ hr. at 980°C	0.47	T=860°C Ref. (K)	860°C	Const. stress compression tests in air	
Ni(a) (M)	99.75% Principal impurity is Co; traces of Fe,Mn, Mg,Al	As received material machined into tensile specimens; annealed 1 hr. at 1150°C under helium	0.33	T > 750°C Ref.(N)	800-1100°C	Const. load compression tests	Tests were made in air and helium. All 800°C, 900°C tests were used; only tests performed in He at 1100°C were used because of extensive oxidation during testing in air.
Ni(b) (K)	99.96% 50 O 10 S <100 C 240 Co+ Cu+Fe	Final Heat-treatment: 10 hrs. at 1010°C	0.22 after recrystallization	T=820-940°C Ref.(K)	860°C	Const. stress compression tests in air	Nickel recrystallized several times before minimum steady-state creep rate was reached; values reported are for minimum $\dot{\varepsilon}_s$ after these recrystallization periods.
Ag(a) (Q)	99.95%	Silver wire was annealed as received for $\frac{1}{2}$ hr. at 900°C	≥0.5	Little information available. ΔT test at 700°C in this investigation. Shows $H_c \simeq H_D$	620-736°C	Const. load tensile tests in air	Grain size in silver unreported. Alloys containing 0.2 and 0.3 atomic percent Mg annealed under identical conditions resulted in grain size of 0.5mm.

Table II. continued

Metal and Ref.	Reported Purity in % and Impurities in ppm	Thermomechanical History Prior to Testing	Grain Size (mm)	Temperature Range for which $H_c \cong H_D$ According to ΔT Tests	Temperature Range of Creep Tests Used in Fig. 5-7	Testing Method	Comments
Ag(b) (R)	99.97% <300 metallic impurities	Wire, cold-drawn from 0.06" to 0.02" dia., annealed in H_2 at 700°C under 70 psi load for 1/2 hr.	0.017	same as above	600°C	Const. stress tensile tests in air	Data reported as strain rate vs. time. Only tests in which a steady state creep rate was reached were considered. Tests 75, 76, 78, 79, 81, 86 used.
Ag(d) (S)	99.999%	Wire specimens annealed in vacuum at T> testing temperature. Specimens tested with either a clean or slightly oxidized surface.	∿0.03	same as above	545-590°C	Tensile tests in vacuum under a single const. load.	The mean of the $\dot{\epsilon}_s$ values observed at each test temperature was considered: $\dot{\epsilon}_s$=0.00057 min^{-1} @ 590°C, 0.00043 @ 578°C, and 0.00016 @ 545°C.
Al(c) (D)	99.996% Cu, Fe, Mg	Anneal 1 hr. at 550°C in vacuum	1.0	Same as Previously Listed	358&418°C	Const. stress tensile tests in vacuum	
Cu(c) (D)	99.96% 10 each Al, Ni, Mg, O	Anneal 1 hr. at 975°C in vacuum	0.6		674&775°C		Recrystallization occurred during 775°C test; minimum creep rate, not $\dot{\epsilon}_s$, reported.
Ni(c) (D)	99.99%	Anneal 1 hr. at 1100°C in vacuum	0.4		978°C		
Ag(c) (D)	99.992% 80 Cu	Anneal 1 hr. at 850°C in vacuum	0.7		591&680°C		

Table II. continued

Metal and Ref.	Reported Purity in % and Impurities in ppm	Thermomechanical History Prior to Testing	Grain Size (mm)	Temperature Range for which $H_c \simeq H_D$ According to ΔT Tests	Temperature Range of Creep Tests Used in Fig. 5-7	Testing Method	Comments
Pb (U)	99.999%	Tensile specimens made from 0.1 in. wire	Single crystal	T> ~ 320°K Ref.(V)	320-593°K	Const. stress tensile tests in helium	Two single crystals were tested to avoid recrystallization during testing. Two tests at high stress gave creep rates too high to be shown on Fig. 5
Pt (Y)	> 99.98% <100 Fe <100 Pd	Cold-drawn wire, 0.38 mm dia., annealed in air at 1400°C for 10 min.	0.1 to 0.2	No ΔT test results available	1100-1550°K	Const. stress tensile tests in air	Most test results were from primary stage of creep. $\dot{\epsilon}_s$ could be found only for tests conducted at stresses $\leq$ 2450 psi when 1550°K $\leq$ T $\leq$ 1250°K
Fe (a)	99.98% 120 O 40 S 30 P 10 C 10 N	Cold worked specimens recrystallized in argon at 840°C	0.1	T > 227°C Ref. (a,b)	605-753°C	Const. stress tensile tests in argon	The activation energy of creep at temperatures near the magnetic transition increases sharply, but creep is still diffusion-controlled.
Fe-4% Si (e)	Fe 96.02% Si 3.92% 700 O 140 C 10 N 20 Si, P	Cold worked specimens recrystallized in argon at 975°C	0.1	$H_c \neq H_D$	377-627°C	Const. stress tensile tests in argon	The activation energy of creep at temperatures near the magnetic transition increases sharply, but creep is still diffusion-controlled.

-18-

Table II. continued

Metal and Ref.	Reported Purity in % and Impurities in ppm	Thermomechanical History Prior to Testing	Grain Size (mm)	Temperature Range for which $H_c \simeq H_D$ According to ΔT Tests	Temperature Range of Creep Tests Used in Fig. 5-7	Testing Method	Comments
Ta (g)	Partial Chemical Analysis: 60-170 C 60-80 O 3-4 H 15 N	Electron-beam melted, cold-forged, rolled, swaged. Annealed in vacuum for 8 hr. at 2050°C to make interstitial content uniform among specimens tested	3	$T \geq 1690°C$ Ref. (g)	1690-2500°C	Const. load tensile tests in vacuum & He atmospheres	Engineering creep rates were converted to true strain rates, and the instantaneous average stresses corresponding to these rates were calculated with the assumptions of constant specimen volume and uniform extension.
Mo (i)	99.9% <100 each Si, Al Cr, Fe Zr <10 each Mn, Mg Co, Ni 130-200 C 110-180 N 30-35 O	Specimens manufactured from powder by pressing, sintering, and swaging; annealed in He at testing temperature for 1/2 hr. prior to loading	∿15	$T \geq 1180°C$ Ref. (j)	1650-2500°C	Const. load tensile test in He atmosphere	Comment for Ta is applicable. O_2 content diminshed from 35 to 25 ppm during testing at 1650°C. O_2 content equilibriated at 5 ppm in those specimens annealed at $T \geq 2250°C$ prior to testing. Some test specimens developed abnormally large grains and low creep rates; these results were not included. Possible contamination from W during testing (See ref. aa).
W(a) (m)	99.95% <100 Si, Fe, Ti <10 Ca, Mn 30 C 70-110 N	Same as for Mo	∿0.04		2250-2800°C	Const. load tensile tests in He atmosphere	No analysis made for oxygen content. Comment for Ta is applicable. Threaded ends of specimens were coated with ThO_2.

Table II. continued

Metal and Ref.	Reported Purity in % and Impurities in ppm	Thermomechanical History Prior to Testing	Grain Size (mm)	Temperature Range for which $H_c \simeq H_D$ According to ΔT Tests	Temperature Range of Creep Tests Used in Fig. 5-7	Testing Method	Comments
W(b) (n)	99.973% 100 O 40 Al 30 C, Si 10 N,F,Cl Mg, P, Ca Fe, Mo 5 Ni, Cr 3 Ta, Cs	Tungsten produced from powder, sintered in H_2, swaged at 1500°C, and recrystallized in vacuum at 2400°C for 1/2 hr.	0.036		2250-2400°C	Tensile tests in vacuum	The flow stress during these tests reached a nearly steady-state value. The ultimate flow stress was used as an approximation to the steady-state creep stress.
W(c) (o)	99.995% 20 Ta 6 C, N 3 O <2 Al, Fe Mo, Ni, Si	Electron beam melted & purified, extruded and swaged.	2-4		3500°F	Const. load tensile tests in vacuum	Recrystallization occured during testing.
W(d) (o)	99.99% 25 Fe 15 Mo 10 N, Si 9 Al 5 C,O,Ni	Arc-melted, machined, extruded, and swaged.	0.05 to 2.0		3500°F & 4000°F	Const. load tensile tests in vacuum	Recrystallization occured during testing. $\dot{\varepsilon}_s \sim L^{0.43}$ where L is recrystallized grain size. $\dot{\varepsilon}_s$ values have been normalized to a grain size of 0.4 mm.
α-Tl β-Tl (q)	99.98% 80 Pb 30 Fe 30 Cu	Cast rods were extruded at 25°C, annealed at 215°C for 1 hr.	α-Th 1.0 β-Th ?	T>80°C	62-280°C	Const. stress compression in silicone oil bath	Grain size of β-Tl assumed to be about the same as for α-Tl.

-20-

Table II. continued

Metal and Ref.	Reported Purity in % and Impurities in ppm	Thermomechanical History Prior to Testing	Grain Size (mm)	Temperature Range for which $H_c \simeq H_D$ According to ΔT Tests	Temperature Range of Creep Tests Used in Fig. 5-7	Testing Method	Comments
Cd (s)	99.995%	Annealed 3/4 hr. at 300°C	0.25	60°C <T<∼ 150°C when H_c recalculated from Eqn.(7). Ref. (t)	60-150°C	Const. stress compression in air	$H_c \simeq H_D$ over this temperature range according to ref. (s).
Zn(a) (v)	99.97% 110 Pb 60 Cu 50 Cd 20 Fe, Si 13 Mg	Annealed at 400°C for 1/2 hr.	0.8	130°C<T<254°C Ref. (v)	140≗254°C	Const. stress compression in air	
Zn(b) (w)	>99.99% 30 Fe 20 Cd 5 Pb	Rolled at 150°C, annealed in argon at 395°C for 1/2 hr.	1.5	76°C<T<∼200°C Ref. (w)	130-190°C	Const. stress compression in air	

References for Table I and Table II

A. I. S. Servi and N. J. Grant, Trans. AIME, 191 (1951) 909.

B. J. Weertman, J. Mech. and Phy. Solids, 4 (1956) 230.

C. O. D. Sherby, J. L. Lytton, and J. E. Dorn, Acta Met., 5 (1957) 219.

D. C. R. Barrett and O. D. Sherby, Trans. AIME, 233 (1965) 1116.

E. P. M. Sutton, Phy. Rev., 91 (1953) 816.

F. T. S. Lundy and J. F. Murdock, J. Appld. Phy., 33 (1962) 1671.

G. P. Feltham and J. D. Meakin, Acta Met., 7 (1959) 614.

H. C. R. Barrett and O. D. Sherby, Trans. AIME, 230 (1964) 1322.

I. Y. A. Chang and L. Himmel, J. Appld. Phys., 37 (1956) 3567.

J. A. Kuper, H. Letaw, L. Slifkin and C. Tomizuka, Phy. Rev., 98 (1955) 1870.

K. C. M. Sellars and A. G. Quarrell, J. Inst. Met., 90 (1961-62) 329.

L. S. Makin, A. Rowe and A. LeClaire, Proc. Phy. Soc., 70B (1957) 545.

M. J. Weertman and P. Shahinian, Trans. AIME, 206 (1956) 1223.

N. J. E. Cannaday, R. J. Austin, R. K. Saxer, Trans. AIME, 236 (1966) 595.

O. G. A. Alers, J. R. Neighbours, and H. Sato, J. Phys. Chem. Solids, 13 (1960) 40.

P. R. Hoffman, F. Pickus and R. Ward, Trans. AIME, 206 (1956) 483.

Q. G. R. Leverant, F. V. Lenel and G. S. Ansell, Trans. ASM, 59 (1966) 890.

R. R. P. Carreker and R. W. Guard, "Creep of Silver Wires," General Electric Co. Report No. 55-RL-1414, (Nov. 1955).

S. K. Schröder, A. J. Giannuzzi and G. Gorsha, Acta. Met. 16 (1968) 469.

T. C. Tomizuka and E. Sonder, Phy. Rev., 103 (1956) 1182.

U. J. Weertman, Trans, AIME, 218 (1960) 207.

References for Table I and Table II

V. C. D. Wiseman, O. D. Sherby, and J. E. Dorn, Trans. AIME, 209 (1957) 57.

W. W. Köster, Z. Metallkunde, 39 (1948) 1.

X. N. H. Nachtrieb and G. S. Handler, J. Chem. Phys, 23 (1955) 1569.

Y. R. P. Carreker, Jr., J. Appld. Phy., 21 (1950) 1289.

Z. F. Cattaneo, E. Germagnoli, and F. Grasso, Phil. Mag. 7 (1962) 1373.

a. Y. Ishida, C. Y. Cheng and J. E. Dorn, Trans. AIME, 236 (1966) 964.

b. C. Y. Cheng, Report, Metallurgy Division, Argonne National Laboratory, Argonne, Illinois (April 1968).

d. Data of Figure 3.

e. C. Y. Cheng. A. Karim, T. G. Langdon and J. E. Dorn, Trans. AIME, 242 (1968) 890.

f. R. Borg, Univ. of California, Lawrence Radiation Laboratory, private communication.

g. W. V. Green, Trans. AIME, 233 (1965) 1818.

h. R. L. Eager and D. B. Langmuir, Phy. Rev., 89 (1953) 911.

i. W. V. Green, M. C. Smith and D. M. Olson, Trans. AIME, 215 (1959)1061.

j. A. H. Clauer, B. A. Wilcox, and J. P. Hirth, "Correlation of Dislocation Substructure with Creep Properties in Refractory Metals," Report BMI-X-517, Battelle Memorial Institute, Columbus, Ohio (1968)

k. P. E. Armstrong and H. L. Brown, Trans. AIME, 230 (1964) 962.

l. J. Askill, "Tracer-Diffusion Studies in Molybdenum," International Conference on Diffusion in BCC Metals, Metals Park: Am. Soc. Met. (1965) 247.

m. W. V. Green, Trans. AIME, 215 (1959) 1057.

-24-

References for Table I and Table II

n. G. W. King and H. G. Sell, Trans. AIME, 233 (1965) 1104.

o. W. D. Klopp, W. R. Witzke, and P. L. Raffo, Trans. AIME, 233 (1965) 1860.

p. K. G. Kreider and G. Bruggeman, Trans. AIME, 239 (1967) 1222.

q. O. D. Sherby, Trans. AIME, 212 (1958) 708.

r. G. A. Shirn, Acta Met. 3 (1955) 87.

s. J. E. Flinn and F. A. Duran, Trans. AIME 236 (1966) 1056.

t. C. R. Barrett, A. J. Ardell and O. D. Sherby, Trans. AIME, 230 (1964) 200.

u. F. Wajda, G. Shirn and H. Huntington, Acta. Met., 3 (1955) 39.

v. J. E. Flinn and D. E. Munson, Phil. Mag., 10 (1964) 861.

w. W. J. McG. Tegart and O. D. Sherby, Phil. Mag., 3 (1958) 1287.

x. G. Shirn, E. Wajda and H. Huntington, Acta. Met., 1 (1953) 513.

y. L. J. Teutonico, Phil. Mag., 15 (1967) 959.

z. W. Köster and H. Franz, Met. Rev., 6 (1961) 1.

aa. P. N. Flagella, J. AIAA, 5 (1967) 281.

Comparison of creep test data using Eq. 8 is justified only when creep is diffusion-controlled. Yet it is well established that creep at lower temperatures can take place where H_c is less than the activation energy for self diffusion H_D. Several of the investigations used in this comparison reported creep data at temperatures below which $H_c \simeq H_D$, according to ΔT tests. For this reason these data were not included in the comparison. Investigations of hcp metals have shown that at very high temperatures $H_c > H_D$; likewise data obtained at these temperatures were excluded. The temperature range over which $H_c \simeq H_D$ according to ΔT tests is reported in Table II as well as the temperature range of the data that were used in the comparison.

Data on the unrelaxed shear modulus at elevated temperatures are not generally available so that in most cases it was estimated from Young's modulus of elasticity and Poisson's ratio. The Young's modulus values were linearly extrapolated from lower temperatures when it was apparent that relaxations had unduly lowered their value. For the few fcc metals for which single crystal elastic compliances are known as a function of temperature, the shear modulus corresponding to the (111) plane[25] was calculated and used in the analysis.

Two methods were used to obtain values for the activation enthalpy of creep. Where data on ΔT tests were available, H_c was calculated from Eq. 7. In addition, for each of the metals examined, data were plotted according to Eq. 6 and H_c obtained graphically. In cases where creep tests were reported at many different temperatures, values of H_c could be assigned with confidence,

but for several metals only approximate values could be assigned. As shown in Table I, H_c is in better agreement with the activation energies H_D of self diffusion of metals than was $Q_c'^{(11)}$. The Fe-4% Si alloy data must be discounted in this context because of the presence of complexities arising from short-range ordering which results in a higher activation energy for creep than that for the diffusivity of Fe.

The significance of diffusion to high-temperature creep is further emphasized by two striking examples: As illustrated in Fig. 2, Sherby[19] has shown that the secondary creep rates of austenite under a constant stress and temperature increase with carbon content in a manner that closely parallels that for the effect of C on the self-diffusivity of Fe in austenite. Such an effect of alloy softening cannot be easily explained on any other basis. As shown in Figs. 3 and 4, the secondary creep rate for α - Fe at a constant stress shows almost the same variation through the magnetic-transformation range of temperatures as does the self-diffusivity of iron in α - Fe. Two factors contribute to this effect, namely the temperature variation in the modulus of elasticity and in the free energy, H - TS, for self diffusion. Over the magnetic transformation range, both the enthalpy H, and the entropy S become functions of temperature due to changes in magnetic state. It has been shown[26-29] that changes in H - TS play the major role in affecting the apparent activation energies for creep over the magnetic transformation range; the changes in modulus of elasticity provide a small additional effect.

Current knowledge on the effect of stress on the secondary creep rate is much less substantial than for temperature. The results for the series of cases that were examined are plotted in accord with Eq. 8 in Figs. 5 to 7. The validity of a power law over a wide range of stresses is sustained. The values of n, however appear to vary somewhat from metal to metal.

It is known that in some cases n decreases with alloying. But in other instances alloys have values of n approaching 7. These major differences cannot be ascribed to scatter.

The variations of n that have been observed for rather pure metals, however, usually appear to have a smaller range than those found for alloys. They fail to exhibit any systematic variation with any known factor. It appears that one major current experimental problem on high-temperature creep concerns a clarification of the stress laws for pure metals.

Two possible alternate attitudes might be taken regarding the observed variations for the currently accepted values of n. The first suggests that n is influenced by a number of uncontrolled factors such as grain size, subgrain size, stacking-fault energies, impurities, oxidation, etc., the combined effects of which result in what falsely appears to be a random scatter in its value. The second alternative suggests that n is the same physical constant for all pure metals, namely about 5, which is dictated by a single creep mechanism and that the observed scatter in its value is due to experimental inaccuracies. Although latter attitude appears somewhat attractive, at the present level of knowledge it cannot be wholly endorsed without further experimental confirmation.

E.   Grain Boundary Sliding

During the creep of polycrystalline metals at elevated temperatures,
deformation may occur by the relative translation [30-32] of one grain with
respect to another as well as by deformation within individual grains. Usually
associated with the phenomenon of grain-boundary sliding is the process of grain-
boundary migration[31-33]. Initial sliding occurs in the immediate vicinity of
the boundary. With increases in time, the sliding gradually takes place over
a zone of finite thickness. The material within the zone generally shows
evidence of progressive grain-boundary migration and grain distortion. Usually
finer subgrains form in the vicinity of the boundary than elsewhere in the grain.
This may arise from local stress concentrations. The shear displacement in grain-
boundry sliding is non-uniform along the length of the boundary [31,32]. This
non-uniformity arises from restraints imposed at the grain corners, and from
blockage due to ledges along the boundary [32-34].

McLean and Farmer[35] have shown that the average grain-boundary dis-
placement vs. time curves in Al are similar in shape to the total creep curves.
The ratio $\frac{\varepsilon_{gb}}{\varepsilon_t}$, the strain due to grain boundary sliding to the total creep
strain, was found to increase with increasing temperatures[35] or decreasing
stresses[30]. Harper, et al.[18] observed the same trends but noted that below
a very low critical stress the ratio decreased with further decrease in stress.
Fazan, et al. [31], and Martin, et al.[36], found that $\varepsilon_{gb}/\varepsilon_t$ was independent
of temperature for a constant stress, i.e., the effect of temperature merely
reflects the fact that lower stresses were employed at higher temperatures.
McLean[30] and Barrett[37] observed that $\varepsilon_{gb}/\varepsilon_t$ increases as the grain size

decreases. This increase is greater than would be expected purely from the
increase in grain-boundary area associated with the finer grain size. Except
at very low stresses the ratio $\varepsilon_{gb}/\varepsilon_t$ is usually small.

It is interesting to compare the values of the activation energy obtained
from grain-boundary sliding tests with those obtained for the total creep strain.
Such a comparison is shown in Table III for Al, Cu and Sn where Q_s is the acti-
vation energy for grain-boundary sliding obtained from surface measurements, Q_k
the value obtained from Kê-type[38] measurements of the anelastic grain-boundary
effects, and Q_c' is the activation energy for creep.

Table III

A Comparison of Activation Energy for Grain Boundary Sliding, Q_s, for Internal
Friction, Q_k, and for Total Creep, Q_c'.
(after Conrad, Ref. 5)

Metal	Q_s Kcal/mole	Q_k Kcal/mole	Q_c' Kcal/mole
Al	33-37	32-38	34-36
Cu	33	32-40	28-37
Sn	19.2	19.0	21-26

These data reveal that small-angle anelastic sliding, large scale grain-boundary
sliding and deformation of grains by creep exhibit about the same activation
energies. The constant value of $\varepsilon_{gb}/\varepsilon_t$ and the similarity between activation
energies for grain-boundary sliding and crystal deformation suggest that grain-
boundary sliding is related to crystal deformation.

F. Grain Size Effects.

Experimental observations concerning the effect of grain size on steady-state creep rates are both varied and sometimes conflicting. Numerous early investigators found that a decrease in the grain size resulted in a subsequent decrease in the high temperature creep strength[40-42]. Others[11,19,39,43,44] found that $\dot{\varepsilon}_s$ at first decreased with increasing grain size and then increased with a further increase in grain size beyond some optimum value.

It is not always evident as to whether grain size effects arise from secondary factors that might have been modified coincident with the treatments that were employed in developing different grain sizes. Changes in grain size might be accompained by the following auxiliary modifications:

1. Type and degree of preferred grain orientation.

2. Distribution of impurities.

3. Geometry and regularity of the grain boundaries.

4. Density and distribution of dislocations.

Modifications 3 and 4 are unimportant since they are adjusted over the primary stage of creep as a result of grain-boundary migration and shearing or as result of subgrain formation to almost fixed conditions during steady-state creep.

Much of the data on grain size effects has been reviewed by Conrad (Ref. 5) and by Garofalo[11]. More recently Barrett, Sherby and Lytton[45] investigated the creep of both randomly oriented, and strongly cube-textured Cu over comparable ranges of grain size. Their data on the randomly oriented specimens revealed that the secondary creep rate was independent of the method of preparation and

and also grain size for all grain diameters greater than 0.1 mm. As the grain diameter was decreased below about 0.1 mm a slight increase was obtained in the secondary creep rate. Since all grain sizes exhibited the same subgrain size at the secondary stage in these constant stress constant temperature tests, the authors concluded that crystallographic creep was unaffected by grain size and that the increase in the secondary creep rate when the grain diameter was decreased below 0.1 mm arises from the effects of grain-boundary sliding. It must then also follow that under the employed test conditions the contribution of grain-boundary sliding was negligible for grain sizes greater than 0.1 mm in diameter. The cube-textured Cu, however, exhibited a lower secondary creep rate under identical test conditions than the randomly oriented material. This might be attributed to both orientation effects and higher resistances to grain-boundary sliding in the lower angle grain-boundaries of the cube-textured Cu.

Current evidence [13] suggests that the effect of grain-size per se on the secondary creep rate is small except for very low stresses and very fine grain sizes. Increased creep rates with decreasing grain size appear primarily attributable to grain-boundary sliding. Auxiliary effects arising from texture and impurities might account for some of the trends and inconsistencies previously reported on grain-size effects.

G. Stacking-Fault Energies

The universal agreement between H_c and H_D clearly reveals that the stacking-fault energy in no way affects the activation energy for high-temperature creep. Recently, however, Sherby et al.[13,46-48] have suggested that the stacking-fault energy nevertheless affects the secondary creep rate, not in terms of the activation energy but in a much more subtle way. Several issues concerning

the experimental verification of this suggestion warrant detailed consideration:

1. It was assumed that the secondary creep rates of various metals could be correlated on the basis of Eq. 5 despite the fact that the terms are not dimensionless and the expression is not wholly in agreement with theoretical expectations.

2. It was, in effect, tacitly assumed that the reported variations in n are due to random scatter and that, n has the same constant value for all metals.

3. Values of $\dot{\epsilon}_s/D$ for a series of metals at a constant value of σ/E varied by a factor of about 10^4, increasing with increasing stacking-fault energy.

4. The stacking-fault energy was assumed to be independent of the temperature.

5. The results suggested that the stacking-fault energy. γ, appears to have affected the secondary creep rate according to the expression

$$\frac{\dot{\epsilon}_s}{D} = A^v \gamma^{3.5} \left(\frac{\sigma}{E}\right)^\eta \qquad (9)$$

Several objections to the form of Eq. 9 are immediately apparent even when n is assumed to be the same for all metals. They include items covered by one and two cited above. Furthemore A^v, which was treated as a constant in the comparison, has the dimension of the reciprocal of a length squared and multiplied by an energy per cm^2 to the power 3.5. Finally the formulation of the equation in terms of a power of the stacking-fault energy contains the physically unacceptable implication that creep rates would be infinite for metals having infinite stacking-fault energies. Obviously, if the stacking-

fault energy does affect the secondary creep rate, its effect cannot be as represented by Eq. 9.

A less objectional approach to possible effects of stacking-fault energies might be based on the more acceptable relationship given by Eq. 8. The following assumptions need yet be made:

1. The grain size effect is small.

2. Although n appears to vary from 4.2 to 5.5, the same value of $n \simeq 5$ is assumed to apply to fcc metals.

3. The most accurate values of $\frac{\dot{\varepsilon}_s kT}{DGb}$ for comparison occur near the center of the σ/G stress range, say at $\sigma/G = 2 \times 10^{-4}$

With these assumptions Eq. 8 might be qualifiedly adjusted to read

$$\frac{\dot{\varepsilon}_s kT}{DGb} = A^{vi} \phi\left\{\frac{\gamma}{Gb}\right\}\left(\frac{\sigma}{G}\right)^n \tag{10}$$

With these assumptions the correlation shown in Fig. 8 was obtained in terms of a dimensionless stacking-fault energy parameter γ/Gb.

Care was exercised to include all of the published measurements of γ/Gb for pure Ag, Au, Cu, Al, and Ni which have been obtained through direct measurement of stacking-fault energy. The value of $\frac{\gamma}{Gb}$ for Pb has been estimated only and may be in error. Data on Pt was not included in Fig. 8 since its value of n is so much greater than that assumed in (2) above. The incorporation of the shear modulus into the stacking-fault energy parameter γ/Gb appears necessary not only to provide a dimonsionless parameter but also to give consistent trends in the correlation. For example, the much lower creep rates of Ni as compared to Al cannot be rationalized on the basis of differences in stacking-fault energy alone since $\gamma_{Ni} > \gamma_{Al}$. With the use of the reduced stacking-fault energy γ/Gb, however, they can. Similar comments

can be made for Pb, which is believed to have a stacking-fault energy nearly
as low as that of Ag. The correlation of Fig. 8 is physically tenable in
that it seems to predict a finite creep rate for infinite values of stack-
ing fault energy.

The values of $\frac{\dot{\epsilon}_s kT}{DGb}$ for the bcc metals at the same value of σ/G range
from about 10^{-9} for W to about 5×10^{-12} for Mo. Since the stacking
fault energies of bcc metals should be quite large, one might expect, on
the basis of the correlation found in fcc metals, and neglecting other
possible factors, that $\frac{\dot{\epsilon}_s kT}{DGb}$ would be approximately 10^{-10} or greater in
all of these metals. The experimental creep data for bcc metals, however,
is not as well known as that for fcc metals. There exist large unexplained
differences between different investigations on the same metal, as the
four investigations on W summarized in Fig. 6 show. Thus it is not yet
clear whether the differences in $\frac{\dot{\epsilon}_s kT}{DGb}$ for bcc metals arise from experi-
mental inaccuracies, or from effects other than stacking-fault energy. If
other effects do exist, they might equally affect the correlation given
in Fig. 8 for the bcc metals.

III. CREEP INDUCED SUBSTRUCTURES

The decreasing creep rate over the primary stage for constant conditions
of stress and temperature must necessarily be a reflection of the substruc-
tural changes that take place; the constant creep rate over the secondary-stage
of creep demands, equally, that all significant substructural features must
have reached a steady state. It is not yet clear, however, just what sub-
structural details might be significant in this respect. Immediately upon loading
dislocation entanglements and cells are produced which bear some resemblance
to those observed in strain-hardened metals at low temperatures. During the
primary stage of creep these cells and entanglements disperse either

through recrystallization or what is more pertinent here through dynamic recovery processes such as cross slip and climb. The dispersal of the original entanglements and cells, although probably facilitated by cross slip, is largely due to polygonization resulting from the climb of edge dislocations. Cross-slip is known to be dependent on the stacking-fault energy which might, for this reason as well as others, play some role in creep.

Contradictory evidence exists regarding the effect of stacking-fault energy on the rate of polygonization. Using x-ray analysis Franks and McLean[49] have demonstrated that Al polygonizes more rapidly than the lower stacking-fault energy metal Cu during the process of creep. But the temperature at which they tested their Cu specimen was in the range for which $H_c < H_D$. Furthermore they did not investigate the effect of strain-rate on polygonization. Barrett[50] has shown that Cu does form well defined subgrains when the rate of straining is slow enough. Hardwick and Tegart[51] have shown that in general metals with high stacking-fault energies polygonize rapidly during creep whereas metals having low stacking-fault energies tend instead to lower their energy by recrystallization. It appears that recrystallization occurs only when the driving force for recrystallization cannot be diminished rapidly enough through recovery processes.

Three substructural features are accessible for study by electron-transmission microscopy, etch-pitting techniques, optical microscopy, and x-ray analyses. They might be classified as follows:

1. The disposition and density of dislocation in the interior of subgrains.

2. The nature of the sub-boundary walls and the subgrain disorientations

    across them.

3. The subgrain size.

These issues will be discussed with major emphasis on the steady-state

conditions.

A. Dislocations in Subgrains

The dislocations that are present in the interior of the subgrains

during high-temperature steady-state creep[24,47,52,67] are usually more or less

randomly positioned. Dipoles and loops are noted occasionally. The absence

of piled-up dislocation arrays and entanglements is highly significant. In

the few cases that have been examined the density of dislocations in the

subgrain interiors increases roughly with the square of the stress as shown

in Fig. 9.   Within the scatter of these data, the well-known relationship for

the equilibrium distribution of dislocations under and applied stress, namely

$$\frac{\sigma}{2} = \alpha\, G b \sqrt{\rho} \qquad\qquad (11)$$

seems to be valid. The value of α however appears to be somewhat greater

than that of 0.20 to 0.25 usually obtained in stressed metals at low tempera-

tures.

B.   Sub-boundaries

Subgrains in polycrystalline aggregates are demarked from each other

by low-angle tilt and twist boundaries of a complex nature. In a recent

investigation on the creep of single crystals of Mo, however, Clauer, Wilcox, and Hirth[53] have demonstrated that tilt boundaries predominate. In general the formation of subgrains arises principally as a result of polygonization due to the climb of edge dislocations under conditions where cross-slip is facile. Although early investigations suggested that the angle between the individual subgrains continued to increase over the secondary stage of creep [11,30], more recent studies[11,51,54-56], covering examples that involve extensive deformation, suggest that the subgrain misorientation eventually reaches some constant limiting value. This observation is consistent with the concept that the significant substructural details must remain substantially constant over the secondary stage of creep.

The need for a more detailed study on subgrain misorientation under secondary-stage creep (including the possible effects of stresses, stacking-fault energies and purity on the degree of misorientation) should be emphasized here. If, for example, the misorientation of the subgrains increases over the secondary stage of creep, it becomes apparent that such misorientation is not a significant substructural variable since it does not affect the creep rate. But if the misorientation remains constant over much of the secondary stage it may be a significant variable and then should be taken into consideration.

C.   Subgrain Size

Early investigations suggested that the subgrain size increased with increasing values of test temperatures and with decreasing strain rate[10,11]. Shortly thereafter, however,[10,11], it was shown that the subgrain size could preferably be related more simply to the stress alone, regardless of test temperature of the secondary creep rate. Higher stresses result in finer subgrain sizes. The limited data currently available[13], shown in terms

of dimensionless parameters in Fig. 10, suggest that the subgrain size is related

almost linearly to the reciprocal of the stress. The results also indicate that

stacking-fault energy might be a significant variable in determining the subgrain

size, the subgrain diameter being larger for a higher γ/Gb metal at a com-

parable value of σ/G.

It is significant that the same subgrain size is developed in a given metal

undergoing steady-state creep, regardless of its previous work-hardened or precrept

condition[10]. When, for example, as shown in Fig. 11, a metal is first crept

under a high stress so as to develop a fine subgrain size and the stress is then

decreased to a lower value, the creep rate decreases abruptly and then increases

with time until the secondary creep rate is a simultaneous migration of sub-

boundaries and readjustment of the substructure to give the unique and now larger

steady-state subgrain size appropriate to the lower stress[57]. It also follows

that the density of dislocations within the subgrains decreases to the value

appropriate to the lower stress and it is probable that the subgrains also

acquire a consistent disorientation.

Although the effect of stress alone was emphasized in the above discussion,

it appears likely that crystal structure, which dictates what types of dis-

locations are present, the shear modulus of elasticity, which in part determines

dislocation interactions, and the stacking-fault energy, which appears to be

important in polygonization, might also be significant in establishing the

substructural features during steady-state creep. The emphasis given here

to the stress merely reflects the fact that some rather tangible, although

incomplete, evidence exists on the effects of this variable.

It is interesting now to speculate on which of the three above-mentioned

substructural features might be most important in steady-state creep. When,

after establishing steady-state at a high stress, the stress is abruptly
decreased, the substructure at first remains close to that developed by the
higher initial stress. Under these conditions the initial creep rate should
nearly correspond to that for a structure which has a high density of dislocations
in the subgrain and a fine subgrain size. During the period over which the
dislocation density decreases and the subgrain size increases to that appropriate
to the lower stress, the creep rate increases to its new steady value. If creep
were exclusively due to the dislocations in the body of the subgrain, the creep
rate should have decreased and not increased during the transition period. On
this basis it appears that the creep-rate may not be so much dependent on the
dislocations in the body of the subgrains. The observed trend suggests that
creep rates might increase as a result of increases in the subgrain size.

According to the data presented in Fig. 10 , however, the subgrain diameter
in fcc metals does not appear to increase with an increase in γ/Gb to the same
degree as does the steady-state creep rate. Other factors such as unique dis-
location reactions, purity, etc. might also influence the subgrain size, and
in this way account for the observed spread in $\dfrac{\dot{\varepsilon}_s kT}{DGb}$ at constant values of
σ/G for a series of metals. The above remarks merely emphasize the need for
definitive research in this area. Not only subgrain size but equally subgrain
misorientation and dislocation densities in the subgrains themselves need to be
investigated more thoroughly.

IV   THEORETICAL IMPLICATIONS

A.   General Discussion

Any realistic theory for high-temperature creep of metals must satisfac-
torily account for all of the experimental observations that have been documen-
ted in Sections II and III of this report. For the restricted ranges of
temperature and stress that were reviewed, these data are sufficiently coherent
to suggest that all might well be rationalized in terms of a single unified
theoretical approach. The variations of the power n of the stress law and the
effect of stress on the steady-state substructure, however, emphasize the
complexity of the theoretical problem and suggest that it may be unrealistic to
believe that high-temperature creep might be completely described in terms of
a single dislocation mechanism.

A complete theory for high-temperature creep must account for the generation
of dislocations, their motion through the lattice, formation of subgrains (including
the details of their composition, size, and misorientation), the annihilation of
dislocations, and the steady-state density and type of dislocations within the
subgrains. It must also rationalize the effects of stress, temperature, modulus
of elasticity, stacking-fault energies, impurities, etc. on these substructural
features and on the secondary creep rates. In this sense no complete and fully
adequate theory for high-temperature creep has yet been developed. All current
theories for creep have had the much less pretentious objective of accounting
semi-quantitatively for only the mechanical behavior of creep in terms of a
single rate-controlling mechanism. The substructural details essential to each
model were either assumed without experimental conformation or they were
expropriated from experimental data. None of the existing theories attempt to

predict the substructure. Since the creep rate usually changes many orders of magnitude as a result of the substructural changes that attend primary creep, it is obvious that the role of the substructure is far from a trivial one and thus demands much more consideration than it has thus far been accorded. On the other hand a semi-theory for high-temperature creep which is erected on experimentally confirmed substructural details which are judiciously incorporated into the theoretical analysis still has substantial virtue insofar as such semi-theories might account for the role of various auxiliary factors on the creep behavior.

Despite the current emphasis on dislocation models, there still remains extensive interest in the more phenomenological aspects of creep. Such interest stems from the hope of erecting a suitable phenomenological structure that will permit reasonable deductions from the factors that affect the creep rate without necessitating consideration of detailed, complex, and often unknown dislocation processes. In this sense a valid phenomenological approach might play the same role in creep as does classical thermodynamics in chemistry and physics insofar as it might permit correlations of phenomenon without demanding a detailed picture of mechanisms. The original concept of Bailey[58], subsequently extended by Orowan[59], and discussed by other investigators[60], that the secondary creep rate results from a balance between strain-hardening and recovery, yet appears attractive.

The statement that the secondary creep rate is determined by a balance between the rate of recovery and the rate of strengthening is tautoglogical and therefore not debatable. The real issue concerns comparisons of what recovery rates with what strengthening rates. In an interesting series of investigations Mitra and McLean[61,62] have shown that the high-temperature

secondary creep rate can be deduced from a stress-sensitive recovery rate at reduced stresses and a strengthening rate determined by the room temperature stress-strain curve. It has been shown, however, that recovery under reduced stresses leads, in its initial stages, to migration and coalesence of sub-grain boundaries[57], in a distinctly different manner from that which prevails during secondary creep. Furthermore there is a difference between the room-temperature strain-hardened substructure and the substructure developed at the secondary state of high-temperature creep. Therefore the observed correlation appears to have been somewhat fortuitous; the fact that it was obtained, however, deserves consideration: The "strength" during high-temperature creep might be accounted for by Eq. 11 which differs from the relationship that applies to room temperature only insofar as α seems to be slightly greater for creep. If further, the recovery rates under a reduced stress are similar to those that apply during secondary creep despite the difference in details, the observed correlation might yet be justified.

B. High Temperature Creep Theories

In order to provide some background for the discussion in the following subsection on the theoretical implications of the previously documented high-temperature creep data, a brief outline will be presented here on the various creep theories that have been proposed. The theories fall into two major categories, those that refer to glide controlled mechanism such as the motion of jogged screw dislocations and the nodal theory, and those that refer to dynamic recovery by climb.

1. Motion of Jogged Screw Dislocations

As screw dislocations glide, their jogs can only advance by the nonconservative emission and/or absorption of vacancies and/or interstitials. In view of

the high energy of formation of interstitials it has generally been conceded that the vacancy mechanism is much more probable. Vacancy generation and absorption models fall into two major categories, those that are based on thermal activation, e.g. Motts[63] original model, and those that are diffusion-controlled, as suggested by Friedel[64], Hirsch and Warrington[65] and Barrett and Nix[52]. Although the various models differ somewhat in detail from one another and are frequently given special interpretations, the factors of major interest here are at least sufficiently similar in all cases to permit a description of this process in terms of the recent dislocation by Gibbs[66] who has shown that to a good approximation,

$$\dot{\gamma}_e = \rho_s \, 6D \left(e^{\frac{\tau \ell_j b^2}{kT}} - 1 \right) \tag{12a}$$

$$\dot{\gamma}_a = \rho_s \, 6D \left(1 - e^{-\frac{\tau \ell_j b^2}{kT}} \right) \tag{12b}$$

$$\dot{\gamma}_c = \rho_s \, \frac{\pi}{2} \frac{D_c b}{\ell_j} \left(\sinh \frac{\tau \ell_j b^2}{kT} \right) \tag{12c}$$

where $\dot\gamma$ is the secondary shear strain rate and the subscripts e, a, and c refer to cases of exclusively vacancy-emitting jogs, exclusively vacancy-absorbing jogs, and equal numbers of both types of jogs respectively on the screw dislocations. The term ρ_s, τ, λ_j and D_c refer respectively to the density of the "mobile" screw dislocations, the applied shear stress, the mean distance between jogs, and the core diffusivity.

For these equations to correctly describe high-temperature creep they must predict a temperature dependence for $\dot\gamma$ which is approximately equal to that for bulk diffusion. Neglecting the temperature dependence of ρ_s, which is probably that of the elastic constants, it is clear that the temperature dependence of $\dot\gamma$ arises primarily from D or D_c and from the jog separation λ_j. In the case that all jogs on screw dislocations are produced by mechanical inter-section, λ_j should be independent of temperature, and the activation energy for creep should approximate that for D or D_c. Under conditions of very slow creep, however, it seems likely that some of the jog population will be produced by other processes, including perhaps thermal activation. In the case when all jogs are thermally produced, λ_j is proportional to $e^{\frac{U_j}{kT}}$, where U_j is the energy to form a jog of one atomic height. Then the effect of temperature upon λ_j is opposite to its effect on diffusivity, so that the apparent activation energy for $\dot\gamma$ is reduced below that of the diffusivity. Since the activation energy for creep is close to that of bulk diffusion, one can conclude that (1) Eq. 12c cannot be a rate-controlling mechanism for creep, and (2) Eq. 12a and 12b can be rate-controlling only when most jogs are produced by athermal processes. Weertman [90] has shown that the most likely jogged screw mechanism is that of Eqn. 12b when jogs are produced mechanically, since vacancy-

absorbing jogs are produced in greater numbers by intersection processes.

One of the primary difficulties of the jogged-screw dislocation models is prediction of the stress dependence of creep. At lower values of stress, when the value of $\frac{\tau \ell_j b^2}{kT}$ is small, the bracketed expression in both Eqs. 12a and 12b reduce to just this term. In order to arrive at the observed stress power law for creep it follows that ρ_s must increase approximately as τ^4, if the reasonable assumption is made that ℓ_j is independent of the stress. But it might be expected that ρ_s is a fraction of ρ of Eq. 11 and therefore increases only at τ^2[67]. This reveals that the actual creep rate at high stresses is much greater than that suggested by either Eq. 12a or Eq. 12b. For example analyses based on the creep data of Cheng et al.[27] suggest that the high secondary creep rate observed in Fe - 3% Si alloy would require an untenably high density of screw dislocations approaching a value of 10^{18} per cm^2 at the higher stress levels. Furthermore the preferred relationship given by Eq. 12b suggests that the creep rate should increase less rapidly with stress at the higher stress levels whereas the opposite is observed experimentally. These deviations of the interpretation based on Eq. 12a and Eq. 12b from reality are too great to hope that any minor refinements to the theory might provide a satisfactory agreement with the observed facts.

If the motion of jogged screw dislocations is not rate-controlling

some other _slower_ mechanism must be sought. The difficulty here centers around the fact that, according to the interpretation given, Eqs. 12a and 12b predict creep rates that are much slower than those obtained experimentally. Since the theoretical formulations are good this demands that the previous interpretations must be modified. Such modification is demanded by the recent observations by Clauer, Wilcox and Hirth[53] that most of the dislocations in crept single crystals of Mo are in edge orientation. Obviously screw dislocations must move rapidly and easily annihilate each other by cross-slip. The high velocity of screw dislocations can be accounted for by either the dislocation core diffusion model, Eq. 12c, or by a volume diffusion model in which ℓ_j is very large.

(b) _Nodal Theory for Creep_. Li[68] noted, from electron-transmission micrographs of crept stainless steels, that some dislocation networks consist of a cross grid of two sets of dislocations which are joined by short nodal segments of the combined dislocations. One set of dislocations bows out between the nodal segments in the plane of the net while the other set seems to remain straight. This suggests that the Burgers vector of the first lies in the slip plane which is also the plane of the net whereas the Burgers vector of the second lies in some alternate plane. Consequently the Burgers vector of the short nodal sections have a component out of the slip plane. If creep arises from glide of the mobile set of dislocations, such glide can only take place at a rate dictated by the climb of the nodal segments. Li tacitly assumed that the climb of the nodes was limited by the frequency of thermal activation and not by the volume diffusion of vacancies. If n are the number atoms along a node, he suggests on the basis of Eyring's reaction rate theory that

$$\dot{\gamma}_s = \rho \, 2^n \, \frac{kTb^2}{h} \, \sin\phi \, e^{-\frac{g_f + ng_m}{kT}} \left(\sinh \frac{\tau \ell b^2 \sin\phi}{2nkT}\right)^n \quad (13)$$

where ρ is the density of the mobile dislocations, h is Planck's constant, g_f is the free energy of formation of vacancies, g_m is the free energy of motion of an atom in the node, ℓ is the length of the dislocation segment between the nodes, and ϕ is the angle between the two sets of dislocations. The power n enters Eq. 13 because it was assumed that the forward motion of the node depended on some cooperative action among n atoms of the node, it is not immediately evident that the apparent activation energy for creep need necessarily always approximate that for self diffusion. The tacit assumption made in interpretation that ρ is independent of the applied stress seems rather unlikely. Furthermore, as presently formulated, the model does not permit comparisons of the creep behavior of different metals in terms of the known effects of their shear moduli of elasticity. The repairs that must be made on this model before it can be accepted are major but do not appear to be nearly as drastic or as difficult as those for the jogged screw dislocation model.

3. Climb from Piled-up Arrays

The first serious formulation dislocation-climb theory for steady-state

creep was presented by Weertman in 1955. It constituted a major advance in the rationalization of the mechanical behavior of metals at high temperatures. The first version[70], which postulated dislocation arrays piled up against Lomer-Cottrell dislocations, was limited to fcc metals. Only the second and more generally applicable model will be discussed here. Weertman assumed that dislocation loops were emitted from fixed and more or less uniformly distributed Frank-Read sources. Screw dislocations of opposite sign from adjacent sources cross-slipped with ease and thus annihilated each other. In contrast, however, the leading edge dislocations from adjacent sources would block each others motion as a result of their mutual interactions. Additional dislocations, however, would continue to issue from the sources until the back stress on the source equaled the applied stress. It was further tacitly assumed that the piled-up arrays were completely analogous to arrays piled-up against rigid barriers. Further deformation, then, could only take place when the two lead dislocations of opposite sign climbed together and annihilated each other, thus relieving the back stresses on the two sources and permitting another dislocation to be generated and move out from each source. The climb rate was assumed to be diffusion-controlled with distances of about the separation of the arrays between the vacancy source and sinks. Every atomic position on the climbing dislocation line was assumed to be an efficient absorber or emitter of vacancies. On this basis Weertman estimated that

$$\dot{\gamma}_s = \frac{\pi^2 \tau^2 D}{G^2 b^2} \sinh \left\{ \frac{\sqrt{3}\, \tau^{2.5} b^{1.5}}{8\, G^{1.5} N^{0.5} kT} \right\} \tag{14}$$

where N are the number of dislocation sources per unit volume. The bracketed term is small over a wide range of stresses and therefore, except for the very highest levels of stress, the secondary creep rate is predicted to increase as $\tau^{4.5}$.

The major virtue of Weertman's theory is that it correlates quite well with most of the mechanical data on high-temperature creep. It accounts for the observed apparent activation energies, the effect of the modulus of elasticity, and gives almost the observed stress law.

Despite its excellence, however, several assumptions require discussion:
1) It is not evident why the number of effective Frank-Read sources should be constant despite theoretical attempts to justify this assumption. For example, if segments of dislocations in the subgrains serve as F-R sources, N should increase as about τ^2. 2) Piled up arrays of dislocations are not seen in metals that are in the secondary stage of creep. This is expected in terms of their dispersal due to the extra degree of freedom of dislocation motion permitted by climb. Calculations by Head[72], Li[73] and Hazzeldine[74] have shown that even in the absence of climb, piled-up arrays of dislocations of opposite signs on nearly parallel slip planes are unstable; they decompose into widely spaced dipoles. 3) The assumption of piled-up arrays of dislocations seems to have been invoked in order to obtain a predicted creep rate that increases with a sufficiently high power of the stress to approach the experimental values. Consequently the need to relax this assumption has a significant effect on qualifying the validity of the predictions by this model.

Christy[75] presented a modification of Weertman's theory which assumed that the driving force for climb arose from the reduction in strain energy upon annihi-

lation of edge dislocations in piled-up arrays. His assumptions that vacancies can be emitted and absorbed only at jogs and that the number of jogs approaches the thermal equilibrium number, led to theoretical activation energies for creep that are at variance with observations on metals. Christy's analysis therefore must be rejected as being somewhat inferior to Weertman's earlier approach.

4. Annihilation of Dipoles by Climb.

Whereas piled-up arrays of edge dislocations are not observed during the secondary stage of creep, several investigators have reported the presence of edge dipoles in the substructure of crept specimens. For this reason Chang[76] suggested that the rate-controlling step in high temperature creep might arise from the recovery of dipoles. Dipoles are the result of a number of dislocation processes. It is known that elongated dislocation dipoles decompose upon annealing into circular loops which, in the absence of a supersaturation of vacancies, then shrink in size. Since the formation of loops takes place by core diffusion and is therefore relatively rapid, this step might be neglected in the present context. Thus the time, Θ, of the removal of a dipole is substantially equal to the time it takes for a loop to vanish. Silcox and Whelan[77] have shown that , in terms of Friedel's [78] climb equation, Θ can be approximated by

$$\Theta = \frac{4\pi(1-\mu)kT\,r^2}{Z\,DGb^3\,\ln\left(\frac{r}{b_o}\right)} \tag{15}$$

where r is the radius of the loop, u is Poissons ratio, Z is the coordination number and b_o is the core radius.

Chang made the reasonable assumptions that the rate of annihilation of dipoles followed first order kinetics and their rate of formation was proportional to the strain rate according to

$$- \frac{dn}{dt} = \frac{n}{\Theta} \tag{16a}$$

and

$$\frac{dn}{dt} = A \dot{\gamma} \tag{16b}$$

respectively, where m is the number of dipoles per unit volume and A is their rate of production per unit strain rate. At the steady-state therefore

$$\dot{\gamma}_s = \frac{n Z D G b^3 \ln\left(\frac{r}{b_o}\right)}{4\pi A (1-\mu) k T r^2} \tag{17}$$

Beyond this point the present authors suggest minor variations on the analysis given by Chang. They suggest that the average value of $2r$ might be a fraction f, of the maximum permissible height of a dipole, namely

$$2r = \frac{f_1 G b}{8\pi (1-\mu) \tau} \tag{18}$$

in lieu of Chang's estimate that $f_1 = 1$, and they suggest that the number of dipoles per unit volume be given by

$$n = \frac{f_2 \rho}{2L} = \frac{f_2 \tau^2}{2\alpha^2 L G^2 b^2} \qquad (19)$$

where f_2 is the fraction of the dislocation density ρ in the form of dipoles of average length L during steady-state creep. The second equality of Eq. 19 follows directly from substructural observations as given by the empirical relationship of Eq. 11. On this basis

$$\dot{\gamma}_s = \frac{32\pi(1-\mu)f_2 Z \tau^4 D}{f_1^2 \alpha^2 A L kT G^3 b} \ln\left\{\frac{f_1 Gb}{16\pi(1-\mu)\tau}\right\} \qquad (20)$$

which gives the same effects of τ, G, D and T on the secondary creep rate as deduced originally by Chang.

As might be expected the trends predicted by the dipole recovery model are almost the same as those suggested by recovery due to climb from piled-up arrays of edge dislocations. The significant issue is that it deviates from reality more drastically than the piled-up array model in predicting that the creep rate increases slightly less rapidly than τ^4, in lieu of $\tau^{4.5}$ and in contrast with about τ^5 usually obtained experimentally. On the other hand the dipole recovery model seems to be based on much more acceptable assumptions regarding substructural details. As pointed out by Chang, however, a weakness of the dipole model seems to center about the tacit assumption that the dipoles are the

effective features of the substructure that hinder creep despite their short-range stress fields. Of course, during steady-state creep it is necessary that every substructural detail also be in the steady state. Consequently it should be possible at least in principle, to develop a theory for the mechanical behavior during high-temperature steady-state creep in terms of recovery of any one of the several significant substructural details. It is also interesting to note that the pre-exponential term in the creep rate might be expected to increase with decreasing stacking-fault energy, at least for such dipoles as might be produced by multiple cross-slip.

5. Nabarro-Bardeen-Herring Creep.

Recently Nabarro[79] proposed a theory for high-temperature creep as a result of climb of dislocations from Bardeen-Herring sources. It is not in competition with the previously discussed recovery plus glide theories but rather deals with a parallel mechanism somewhat analogous to the Herring-Nabarro creep in polycrystalline aggregates insofar as the mass transfer arises from vacancy fluxes. It differs from the Herring-Nabarro model since the vacancy flux arises between climbing Bardeen-Herring sources of opposite signs. For this mechanism Nabarro estimates a high-temperature creep rate given by

$$\dot{\gamma}_s = \frac{D b \tau^3}{\pi k T G^2 \ln\left(\frac{4G}{\pi \tau}\right)} \qquad (21)$$

A possible example of NBH creep creep is noted in the high-temperature low-stress creep of high purity Al reported by Harper and Dorn[16]. For stresses less than about 30 psi at 920°K, Al exhibited no initial straining and no large primary stage but entered into steady-state creep almost immediately.

Simultaneously there was a rapid reduction in grain-boundary shearing. Whereas

above 30 psi the creep rate depended on a high power of the stress (about $n \simeq 4.5$),

the creep rate seemed to be linear with stress up to about 10 psi in the low stress

range. Although it was evident that a new mechanism was involved, the activation

energy for creep was equal to that for self-diffusion over the lower as well as

the higher stress range. Furthermore the secondary creep rate was over three

orders of magnitude greater than expected from the Nabarro-Herring mechanism

for polycrystalline metals. A review of these data suggests that the creep rate

over the low stress range increases as about the third power of the stress. At

a shear stress of $\tau = 15.4 \times 10^5$ dynes/cm^2 the calculated creep rate based on

the NBH model is 1.5×10^{-8} per second whereas the experimentally determined

value (which could well be slightly high because of the incidence of some glide)

was 1.2×10^{-7} per second. The agreement between these data and the NBH model

appears to be reasonably good.

C. Reflections on the Climb Process in Creep

The most significant single fact about high-temperature creep is that its activation energy is in excellent agreement with that for self diffusion. Although it is possible to account for this fact in terms of special thermally-activated mechanisms, it is much more likely that the controlling mechanisms are diffusion-limited.

Current analysis based on the motion of jogged screw dislocations indicate that this model is not easily adjusted to agree well with all of the mechanical facts. A major reason for disqualifying this model, however, arises from recently reported observations that in one otherwise typical example[53], most of the dislocations within the subgrains and in the subgrain boundaries are in edge orientation. This fact strongly supports the suggestion, first made by Weertman[70], that dynamic recovery by climb of edge dislocations is the most probable rate-controlling mechanism for high temperature creep.

The excellent agreement between the activation energy for creep and for self diffusion further indicates that bulk diffusion of vacancies to or from dislocations is the slowest step in the climb process. It has not yet been established theoretically why this should be the rate-limiting step for the climb of dislocations during creep.

Early theoretical considerations suggested that the emission and absorption of vacancies during climb was facile only at jogs along a dislocation line. During climb, however, existing jogs are continuously eliminated. Jog production is not the rate-controlling step during climb-controlled creep. This means that either: 1) the vacancies are emitted or absorbed at all atomic sites along the core of an edge dislocation, not only at jog

sites, or else 2) vacancies are emitted and absorbed only at jogs, and the
rate of production of these jogs during creep is at least as rapid as the
rate at which they are eliminated.

The first suggestion has received little theoretical consideration to date.
It will not be considered further here but should be considered as a possibility,
since it is in accord with experimental observations on creep. The implications
of the second suggestion will be examined in some detail.

There are several features of diffusion-controlled climb which should be
considered when vacancy emission and absorption occurs only at jogs. If the rate
of diffusion of vacancies along dislocation cores is approximately the same as
for diffusion in the bulk lattice, vacancy flow occurs only to or from jogs. This
corresponds to the case of "emission-controlled" climb which is discussed by
Friedel[78]. The climb rate is directly proportional to the density of jogs c_j
present on the climbing dislocation. During steady-state creep this density is
maintained at a constant level by formation of new jogs to replace those destroyed
by climb.

It was once suggested that jogs were nucleated thermally and that $c_j \simeq e^{-(U_j+U_c)/kT}$.
In this event, however, the activation energy for creep should have equalled that
for self diffusion plus the jog and constriction energies, $U_j + U_c$. If emission-
controlled climb is the rate-controlling mechanism in creep, jogs must be produced
athermally.

The suggestion has been made that jogs are produced mechanically as a
result of intersection of glide dislocations with screw dislocations. Although
this concept had some support in the observation [80] that Cu and
Ag do not polygonize unless they are deformed concurrently, it is unlikely

that during slow creep intersection can produce a jog formation rate which is as large as that of the destruction rate. Thus, this suggestion too must be discarded.

Several other ways have been suggested by which jogs can be placed athermally on a dislocation line. It was shown by DeWit[81] and by Weertman[82] that in a stressed crystal any straight dislocation is in unstable equilibrium and there is no energy barrier to prevent it from assuming the shape of a helix, provided that diffusion processes can occur. A dislocation in a helical form obviously contains a density of geometrical jogs. Hence an athermal concentration of jogs can be obtained on any dislocation line in a stressed crystal. This process is more difficult in metals of low stacking-fault energy, for as Weertman pointed out it is difficult for a widely extended dislocation to assume a helical form.

Lothe[83] suggested another mechanism where the nucleation of great supersaturations of jogs at sharp, constricted dislocation bends may provide the necessary jogs. At sharp bends, the reduction in line energy upon the formation of a jog may be so great that a jog begins to form as soon as a vacancy jumps to the bend. In fcc metals cross-slip past barriers cannot go on indefinitely unless the edge components left behind are removed by climb. The edge components left behind in cross-slip will always end at bends of the type required for nucleation. It is not known whether these processes can produce jogs with the rapidity needed to maintain diffusion-controlled kinetics during creep.

The observation that the activation energy for creep equals that for self diffusion means that, for emission-controlled climb, jog nucleation and motion must not require constriction of the jog as earlier theory

suggested. Several mechanisms for direct nucleation of an extended double jog which do not require constriction of a dislocation have been proposed both by Thomson and Balluffi[84] and by Escaig[85]. Available experimental evidence does not suggest any means of distinguishing between the various possibilities of double jog nucleation on extended dislocations, and energy calculations are difficult to perform because the critical nucleus is of the order of core dimensions. Hirth and Lothe[86] have proposed a mechanism whereby extended jogs can climb directly by the absorption of a vacancy without prior constriction. The mechanism is applicable to the climb of extended superjogs provided that a multiple number of vacancies is absorbed before the entire superjog advances by one interatomic distance. Thus it appears possible that extended dislocations can climb without prior constriction.

The preceding shows that emission-controlled climb cannot possibly be the rate-controlling process during creep unless jogs are produced athermally, without thermally-activated constriction, and at a rate not slower than their removal by climb. Theoretical mechanisms which meet the first two of these requirements have been proposed. Further consideration must be given to kinetics of jog production before it can be established theoretically that emission-controlled climb is a possible mechanism for high-temperature creep.

It now seems likely that vacancies can be absorbed or emitted from the core of a climbing dislocation at positions other than jogs, even if the actual creation or destruction of the vacancies occurs only at the jogs. When vacancy core diffusion is rapid in comparison to bulk diffusion, the quasi-equilibrium level of vacancies that is maintained at a jog by climb stresses can also be maintained along portions of the core near the jogs. If core diffusivity is sufficiently high, the quasi-equilibrium

concentration can be maintained along the whole of the dislocation and the condition of "saturation" that is discussed by Lothe[83] is achieved. This condition corresponds to the situation in vacancy-absorption climb where the average impinging vacancy is able to travel far along the dislocation line so that it is destroyed at a jog before jumping off. In vacancy-emission climb it corresponds to the case where vacancies that are created at jogs have sufficient mobility along the core to make vacancy emission equally probable at all sites along the dislocation.

When bulk diffusion of vacancies is rate-controlling, the fastest rate of dislocation climb is achieved when the dislocation is saturated, and the slowest rate of climb achieved when emission or absorption occurs only at jogs (the emission-controlled case). Intermediate rates of climb are obtained when neither condition is completely fulfilled. Which case is actually achieved during high-temperature creep depends upon the relative magnitudes of vacancy core diffusivity and jog spacing along the core.

Thomson and Balluffi[84] have considered this problem in the context of climb of an unconstrained pure edge dislocation in a crystal which is supersaturated with vacancies. This case is not exactly the same situation as that encountered during creep but many of the results should still be applicable. In the case considered by Thomson and Balluffi vacancy flow arises because of a gradient in vacancy chemical potential between jogs on edge dislocations and the rest of the lattice. At jogs the vacancy concentration is maintained at its equilibrium level, while other regions of the crystal are supersaturated. During creep the situation is reversed. The bulk lattice contains an equilibrium concentration of vacancies[87], while at jogs the concentration is enhanced by $e^{\frac{\sigma_{xx}\Omega}{kT}}$, where σ_{xx} is the

net climb stress which acts on the dislocation at the jog. When the climb rate is diffusion-controlled, however, these two cases are analogous since the flux of vacancies to the dislocation depends only on the gradient of the vacancy chemical potential and not on its absolute value.

The critical parameters in the model of Thomson and Balluffi are the average spacing between jogs, ℓ_j, and the average distance a vacancy travels along the core before jumping off, α^{-1}. This latter parameter is, of course, proportional to the rate of core diffusivity. The model shows that when $\alpha\ell_j \leq 3$ the condition of vacancy saturation is achieved along the entire dislocation line. As long as this condition is satisfied, it does not matter whether the process of jog formation is thermally activated, the activation energy for climb will be simply that for self-diffusion.

When $\alpha\ell_j > 3$ the climb rate is considerably more difficult to calculate. Only a portion of the dislocation line is saturated so that not all vacancies which diffuse to the dislocation will be absorbed, and not all portions of the dislocation can emit vacancies. The climb rate will depend upon the concentration of jogs along the core of a dislocation in a manner analogous to the case of emission-controlled climb (for which $\alpha\ell_j$ is infinite). For climb under these conditions to yield an activation energy equal to that of self-diffusion, jogs must be produced athermally, without thermally-activated constriction, and at a rate not slower than their removal by climb. It is necessary to consider whether this third condition can be achieved during creep before dislocation climb can be considered as a possible rate-controlling mechanism for creep when $\alpha\ell_j > 3$. This problem has not yet been solved theoretically.

Seidman and Balluffi[88,89] have investigated how well the condition
of saturation is fulfilled for dislocation climb in the presence of
vacancy supersaturations which are produced by quenching. Their results
are most easily expressed in terms of the efficiency of dislocation climb.
This quantity is defined as the ratio of the observed climb rate to that
calculated for diffusion-controlled climb of saturated dislocations. They
studied the rate of generation of vacancies in temperature-pulsed pure
metals where the vacancy source structure was known. From a knowledge of
various source densities (e.g., dislocation density, subgrain size and
grain size), the maximum possible generation rates at each of the possible
sources were calculated and then compared with experiments. Their results
revealed that in the presence of both large and moderate vacancy super-
saturations dislocations climb with an efficiency of from 0.1 to 1.0, even
in low-stacking-fault-energy gold. Thus, in quenching experiments, at
least, the condition of saturation is often achieved experimentally.

There are no experimental results available which yield direct estimates
of the climb efficiency of dislocations during high-temperature creep.
The results of Seidman and Balluffi are applicable to creep only if the pro-
cess of jog formation is the same in both cases, and there is no a priori
reason for believing this should be true. On the other hand, these are
the only existing measurements of dislocation climb efficiency under any
conditions, so that it may well be that climbing dislocations are saturated
with vacancies during creep. Certainly the equivalence between activation
energies for creep and self diffusion suggests that this is the case.

Weertman[90] has made the suggestion that the degree of saturation

of vacancies along dislocations might depend upon the width of extended

dislocations. If the rate of core diffusion for a widely dissociated

dislocation is sufficiently slow, the condition of saturation is not likely

to be achieved. The climb rate will then be decreased below that possible

under saturation. This could provide an explanation for the difference

in creep rates in fcc metals with stacking-fault energy which was discussed

previously.

The critical assumption in this proposal is that diffusion of vacan-

cies along extended dislocations is slow in comparison to bulk diffusion.

Recent tracer diffusion studies of Baker, Wutting, and Birnbaum[91]

clearly show, however, that core diffusion is very rapid in comparison

to bulk diffisuion, even in a 40% Ni-60% Co alloy where the dislocations

are separated by approximately 15 atomic distances. The activation energy

for core diffusion in the alloy is less than half of that of bulk diffusion,

so that the rate of core diffusion is several orders of magnitude greater

than for bulk diffusion. Earlier observations on enhanced diffusion in

Ag[92] suggested the same result, but in these experiments it was not so

evident that the enhanced diffusivity developed solely from core diffusion

along extended dislocations. It now seems that undersaturation of extended

dislocations cannot arise from a low rate of core diffusion. An explana-

tion for the apparent effect of stacking-fault energy on steady-state creep

rates must be sought elsewhere.

Stacking fault energy may conceivable influence the length of path

over which dislocations must climb during creep, and so affect the average

rate of climb. Its influence may lie in its effects on other details of

the substructure developed during creep. This problem has not received nearly as much experimental and theoretical attention as it deserves.

Other features of substructure certainly influence high-temperature creep behavior. The creep rates of metals are highly sensitive to the substructural changes that take place. Such effects seem to be reasonably consistent; different metals exhibit similar substructural details for steady-state creep at identical values of σ/G. The formulation of any realistic theory must at least consider if not predict these significant features in order to arrive at an appropriate description of high-temperature creep.

VI. SUMMARY

Despite the substantial progress that has been made in unraveling the nature of creep of metals at high-temperatures (above about one-half of the melting temperature) many significant details yet require clarification. The following generalizations are now fairly well established:

1. The apparent activation energies for high-temperature creep of metals and dilute solid solutions are independent of the creep strain and insensitive to the applied stress; they are in close agreement with the activation energies for self-diffusion. The agreement becomes even better when the temperature dependence of modulus is taken into account and the true enthalpy of activation of creep is estimated.

2. The changes in the creep rate with time under constant conditions of the independent variables of stress and temperature are reflections of substructural changes that accompany creep; such substructural changes are insensitive to the temperature and depend almost exclusively on the time and the stress divided by the modulus of elasticity.

3. Under steady-state creep the significant substructural details remain constant. More creep resistant substructures are produced by higher stresses. Thus the steady-state creep rate depends not only on the stress but also on the steady-state substructure produced in the test metal by that stress.

4. Although many exceptions are known, the secondary creep rate usually increases with approximately the fifth power of the stress. Attempts to separate the structural effects from the instantaneous stress effect have not been conclusive.

5. Attempts to present a unified picture of high-temperature creep have been made difficult by the effects of other factors such as modulus, grain size, subgrain size and stacking-fault energies. Perhaps the most significant of these is the modulus effect which can be incorporated into the analysis on a quasi-theoretical basis. In this way a more universal comparison of the secondary creep rates of different metals can be realized. Nevertheless, even when this is done, the secondary creep rates yet differ from metal to metal by several orders of magnitude due to the effects of other factors.

6. Although early investigations suggested that grain size effects may be large, the more recent evidence suggests that they are small.

7. Recent experimental evidence suggests that the creep rate increases with an increase in the stacking-fault energy.

8. Although the subgrain size decreases with increasing ratios of stress to modulus it may also depend upon the stacking-fault energies or other factors. Its effect on the creep rate has not yet been isolated.

9. No theoretical models for primary creep have yet been announced due undoubtedly to the complexity of the problem.

10. Two major types of models have been suggested for the somewhat simpler case of steady-state creep, namely the motion of jogged screw dislocations and the climb of edge dislocations. The screw dislocation models suffer from various handicaps and require major repairs in order to account correctly for the activation energies and stress laws that are obtained. Some climb models appear to be able to account more readily for the observed activation energies and stress laws for creep. Many of the assumptions on which they were based, however, are now questionable. None

take into consideration the possible effects of jog energies or stacking-fault energies on the rate of climb. The earlier models, despite their virtues, assumed the presence of piled-up arrays of edge dislocations. Such arrays are unstable under conditions where climb can take place and would disperse. In fact, no such arrays are seen in electron micrographs of crept metals. Other assumptions involving fixed numbers of Frank-Read sources regardless of the stress level, etc., are also disconcerting.

11. Several approaches to a better understanding of high-temperature creep might be suggested:

a) More complete empirical data on the creep behavior of metals having different moduli, different stacking-fault energies, etc., to provide an acceptable correlation of their creep rates, i.e., a single universal relationship.

b) A more thorough investigation on the substructures that are developed with specific emphasis on their significance to creep mechanisms.

c) The eventual building of more realistic models for creep including primary as well as steady-state creep.

REFERENCES

1. Creep and Fracture of Metals at Elevated Temperatures, Symposium, National Physical Laboratory, 1954, London: H. M. Stationery Office (1956)

2. A. H. Sully, Progress in Metal Physics, 6 (1955) 135.

3. Creep and Recovery, Seminar, Am. Soc. Metals, Cleveland: ASM (1957).

4. D. McLean, "Point Defects and the Mechanical Properties of Metals and Alloys at High-Temperatures," Vacancies and Other Point Defects in Metals and Alloys, London: Institute of Metals, (1957) 159-198.

5. Mechanical Behavior of Materials at Elevated Temperature, J. E. Dorn (ed.), New York: McGraw Hill (1961).

6. Structural Processes in Creep, London: Iron & Steel Institute (1961).

7. J. E. Dorn, "Progress in Understanding High-Temperature Creep," Horace W. Gillett Memorial Lecture, Am. Soc. Test. Matls. (1962).

8. D. McLean, "Deformation at High Temperatures," Met. Reviews, 7 (1962) 481.

9. A. J. Kennedy, Processes of Creep and Fatigue in Metals, New York: John Wiley and Sons (1963).

10. J. E. Dorn and J. Mote, "Physical Aspects of Creep," High-Temperature Structures and Materials, New York: Pergamon Press, (1963) 95-168.

11. F. Garafalo, Fundamentals of Creep and Creep Rupture in Metals, New York: Macmillan (1965).

12. D. McLean, "The Physics of High-Temperature Creep in Metals," Reports on Progress in Physics, 29, part 1 (1966) 1.

13. O. D. Sherby and P. M. Burke, Progress in Mat. Sci., 13 (1967) 325.

14. F. R. Nabarro, Report of a Conference on the Strength of Solids, London Phy. Soc. (1948) 75-90.

15. C. Herring, J. Appl. Phy., 21 (1950) 437.

16. J. Harper and J. E. Dorn, Acta Met., 5 (1957) 654.

17. R. Lagneborg and R. Attermo, "Creep of 20%Cr-35%Ni Stainless Steels at Low Stresses" to be published in J. Matl. Sci.

18. J. G. Harper, L. A. Shepard, and J. E. Dorn, Acta Met., 6 (1958) 509.

19. O. D. Sherby, Acta Met., 10 (1962) 135.

20. H. Laks, "Effect of Low Stresses on the Creep Rates of a Polycrystalline Aluminum Alloy under Constant Structure," M. S. Thesis, University of California, Berkeley, (1953). Data presented in Ref. 1, p. 125-131.

21. C. M. Sellars and A. G. Quarrell, J. Inst. of Met., 90 (1961-62) 329.

22. R. G. Davies, Trans. AIME, 227 (1963) 665.

23. A. Lawley, J. A. Coll, and R. W. Cahn, Trans. AIME, 218 (1960) 166.

24. D. McLean and H. F. Hale, "The Stress Sensitivity of Creep," Structural Processes in Creep, London: Iron and Steel Institute, (1961) 19.

25. E. Aerts, P. Delavignette, R. Siems, and S. Amelinckx, J. Appl. Phys., 33, (1962) 3078.

26. Y. Ishida, C. Y. Cheng and J. E. Dorn, Trans. AIME, 236 (1966) 964.

27. C. Y. Cheng, A. Karim, T. G. Langdon and J. E. Dorn, Trans. AIME, 242 (1968) 890

28. C. Y. Cheng, "On the Mechanism of Creep in Alpha Iron," Report, Metallurgy Division, Argonne National Laboratory, Argonne, Illinois, (1968).

29. S. Karashima, H. Oikawa and T. Watanabe, Acta Met., 14 (1966) 791.

30. D. McLean, J. Inst. Metals, 81 (1952-53) 293.

31. B. Fazan, O. D. Sherby and J. E. Dorn, Trans. AIME, 200 (1954) 919.

32. H. Brunner and N. J. Grant, J. Inst. Met., 85 (1956-57) 77.

33. F. Weinberg, Trans. AIME, 212 (1958) 808.

34. S. K. Tung and R. Maddin, Trans. AIME, 209 (1957) 905.

35. D. McLean and M. H. Farmer, J. Inst. Met., 85 (1956-57) 41.

36. J. A. Martin, M. Herman and N. Brown, Trans. AIME, 209 (1957) 78.

37. C. R. Barrett, Ph.D. Dissertation, Univ. of Stanford, Stanford, California (1964).

38. T. S. Ke, Phy. Rev., 72 (1947) 41.

39. D. McLean, Grain Boundaries in Metals, Oxford: The Clarendon Press (1957).

40. W. A. Wood and G. R. Wilms, J. Inst. Metals, 75 (1948-49) 693.

41. D. Hanson and M. A. Weeler, J. Inst. Metals, 45 (1931-32) 229.

42. A. I. Blank and H. L. Burghoff, Proc. ASTM, 51 (1951) 981.

43. P. Feltham and J. D. Meakin, Acta Met., 7 (1959) 614.

44. C. Crussard, Comptes Rendus, 219 (1944) 681.

45. C. R. Barrett, J. L. Lytton and O. D. Sherby, Trans. AIME, 239 (1967) 170.

46. C. R. Barrett and O. D. Sherby, Trans. AIME, 233 (1965) 1116.

47. J. L. Lytton, C. R. Barrett and O. D. Sherby, Trans AIME, 233 (1965) 1399.

48. R. M. Bonsesteel and O. D. Sherby, Acta Met., 14 (1966) 385.

49. A. Franks and D. McLean, Phil. Mag., 1 (1956) 101.

50. C. R. Barrett, Acta Met., 13 (1965) 1088.

51. D. Hardwick and W. J. McG. Tegart, J. Inst. Met., 90 (1961-62) 17.

52. C. R. Barrett and W. D. Nix, Acta Met., 13 (1965) 1247.

53. A. H. Claver, B. A. Wilcox, and J. P. Hirth, "Correlation of Dislocation Substructure with Creep Properties in Refractory Metals," Report BMI-X-517, Battelle Memorial Institute, Columbus, Ohio (1968).

54. E. A. Wong, H. J. McQueen and J. J. Jonas, J. Inst. Met., 95 (1967) 129.

55. H. J. McQueen, J. of Metals, 20 (1968) 31.

56. F. H. Hammad and W. D. Nix, Trans. ASM, 59 (1966) 94.

57. L. Raymond and J. E. Dorn, Trans. AIME, 230 (1964) 560.

58. R. W. Bailey, Inst. Met., 35 (1926) 27.

59. E. Orowan, J. West Scotland, Iron and Steel Inst., 54 (1946) 45.

60. A. H. Cottrell and V. Aytekin, J. Inst. Met., 77 (1950) 389.

61. S. K. Mitra, J. Inst. Met., 93, (1964-65) 321.

62. S. K. Mitra and D. McLean, Phil. Mag., 295A (1966) 288.

63. N. F. Mott, "A Discussion of Some Models of the Rate-Determining Process in Creep," Creep and Fracture of Metals at Elevated Temperatures, London: H. M. Stationery Office, (1956) 21.

64. J. Friedel, Phil. Mag., 46 (1956) 1169.

65. P. B. Hirsch and D. H. Warrington, Phil. Mag., 6 (1961) 735.

66. J. B. Gibbs, "Equilibrium Velocities of Defect-Emitting Jogs," RD/B/N884, Berkeley Nuclear Laboratory, England (1967).

67. C. R. Barrett, Trans. AIME, 239 (1967) 1726.

68. J. C. M. Li, Trans. AIME, 227 (1963) 1474.

69. J. Weertman, Trans. AIME, 227 (1963) 1475.

70. J. Weertman , J. Appl. Phys., 26 (1955)1213.

71. J. Weertman, J. Appl. Phys., 28 (1957) 362.

72. A. K. Head, Phil. Mag., 4 (1959) 295.

73. J. C. M. Li, Discussions of Faraday Society, 38 (1964) 138.

74. P. M. Hazzledine, J. de Physique, 27, Supplement C3 (1966) 210.

75. R. W. Christy, J. Appl. Phys., 30 (1959) 760.

76. R. Chang, The Physics and Chemistry of Ceramics, New York: Gordon and Breach, (1963) 275.

77. J. Silcox and M. J. Whelan, Phil. Mag., 5 (1960) 1.

78. J. Friedel, Les Dislocations, Paris: Gauthier-Villars (1956); also Dislocations, Oxford: Pergamon Press (1964).

79. F. R. N. Nabarro, Phil. Mag., 16 (1967) 231.

80. G. Schoeck, Mechanical Behavior of Materials at Elevated Temperatures, J. E. Dorn (ed.), McGraw Hill, (1961) 93.

81. R. Dewit, Trans AIME, 227 (1963) 1443.

82. J. Weertman, Trans. AIME, 227 (1963) 1439.

83. J. Lothe, J. Appl. Phys., 31 (1960) 1077.

84. R. M. Thomson and R. W. Balluffi, J. Appl. Phys., 33 (1962) 803; J Appl. Phys., 33 (1962) 817.

85. B. Escaig, Acta Met., 11 (1963) 595.

86. J. P. Hirth and J. Lothe, Theory of Dislocations, New York: McGraw Hill, (1968)530.

87. A. L. Ruoff, Interactions Between Lattice Defects, R. R. Hasiguti (ed.), New York: Gordon and Breach (1966) 961.

88. D. N. Seidman and R. W. Balluffi, reference 87, p. 913.

89. D. N. Seidman and R. W. Balluffi, Phys. Stat. Sol., 17 (1966) 531.

90. J. Weertman, Trans. AIME, 233 (1965) 2069.

91. C. Baker, M. Wuttig and H. K. Birnbaum, "Diffusion Along Edge Dislocations in Ni-Co Alloys," presented at the International Conference on Strength of Metals and Alloys, Tokyo, Sept. 4-8, (1967)

92. A. Hendrickson and E. S. Machlin, Trans. AIME, (1954) 1035.

93. P. S. Dobson, P. J. Goodhew, and R. E. Smallman, Phil. Mag., 16 (1967) 9.

94. J. P. Tartour and J. Washburn, to be published in Phil. Mag.(1968).

95. V. C. Kannan and G. Thomas, J. Appl. Phys., 38 (1967) 4076.

96. P. Humble, M. H. Lorretto and L. M. Clarebrough, Phil. Mag., 15 (1967) 297.

97. F. Haussermann and M. Wilkens, Phys, Stat. Sol., 18 (1966) 609.

98. T. Jøssang and J. P. Hirth, Phil. Mag., 14 (1966) 657.

99. P. Humble, R. L. Segall, and A. K. Head, Phil. Mag., 15 (1967) 281.

100. R. E. Smallman, I. L. Dillamore, and P. S. Dobson, J. De Physique, 27, Supp. C3, (1966) 86.

101. P. C. J. Gallagher and J. Washburn, Phil. Mag., 14 (1966) 971.

102. L. M. Clarebrough, R. L. Segall, and M. H. Loretto, Canad. J. Phys., 45 (1967) 1135.

103. L. K. Ives, and A. W. Ruff, Jr., Phys. Stat. Sol., 27 (1968) 117.

104. M. Wilkens, M. Rapp, and K. Differt, Z. Metallkunde, 57 (1966) 746.

105. I. S. Servi and N. J. Grant, Trans. AIME, 191 (1951) 917.

106. H. Busboom, Master's Report, Jan., 1963, Dept. of Materials Science, Stanford University, Stanford, California.

107. C. R. Barrett, W. D. Nix., and O. D. Sherby, Trans. ASM 59 (1966) 94.

108. A. Goldberg, J. Iron and Steel Inst., 204 (1966) 268.

109. R. Guard, Creep and Recovery, Seminar, Am. Soc. Metals, Cleveland: ASM (1957) 251.

110. V. P. Gupta and P. R. Strutt, Canad, J. Phys., 45 (1967) 1213.

LIST OF FIGURES

FIGURE 1

XBL 6SS-1494

FIGURE 2

XBL 688-1498

FIGURE 3

FIGURE 4

XBL 688-1491

CODE FOR FCC METALS

- ● Al (a)   ○ Al (b)   ⊕ Al (c)
- ▣ Ni (a)   □ Ni (b)   ⊞ Ni (c)
- ◇ Cu (a)   ◆ Cu (b)   ◈ Cu (c)
- ▲ Ag(a)   △ Ag(b)   ◮ Ag(c)   ◬ Ag(d)
- ▽ Au   ▽ Pt   ◉ Pb

NOTE: Al(b), Ni(a), Ag(a) and Ag(d) tests were not performed at a constant stress.

XBL 6810-6010

FIGURE 5

FIGURE 6

-84-

FIGURE 11

XBL 688-1499