خزش
Creep

منابع درسی

1- Fundamental of creep in metals and alloys
By M. Kassner

2- Mechanical metallurgy
By G. Dieter

3-Mechanical properties and working of metals and alloys
By A. Bhaduri

4-Creep-resistant steels
By F. Abe, T.-Ulf Kern and R. Viswanathan

بارم بندی کلاس

۱- فعالیت کلاسی

۲- تمارین کلاسی

۳- سمینار

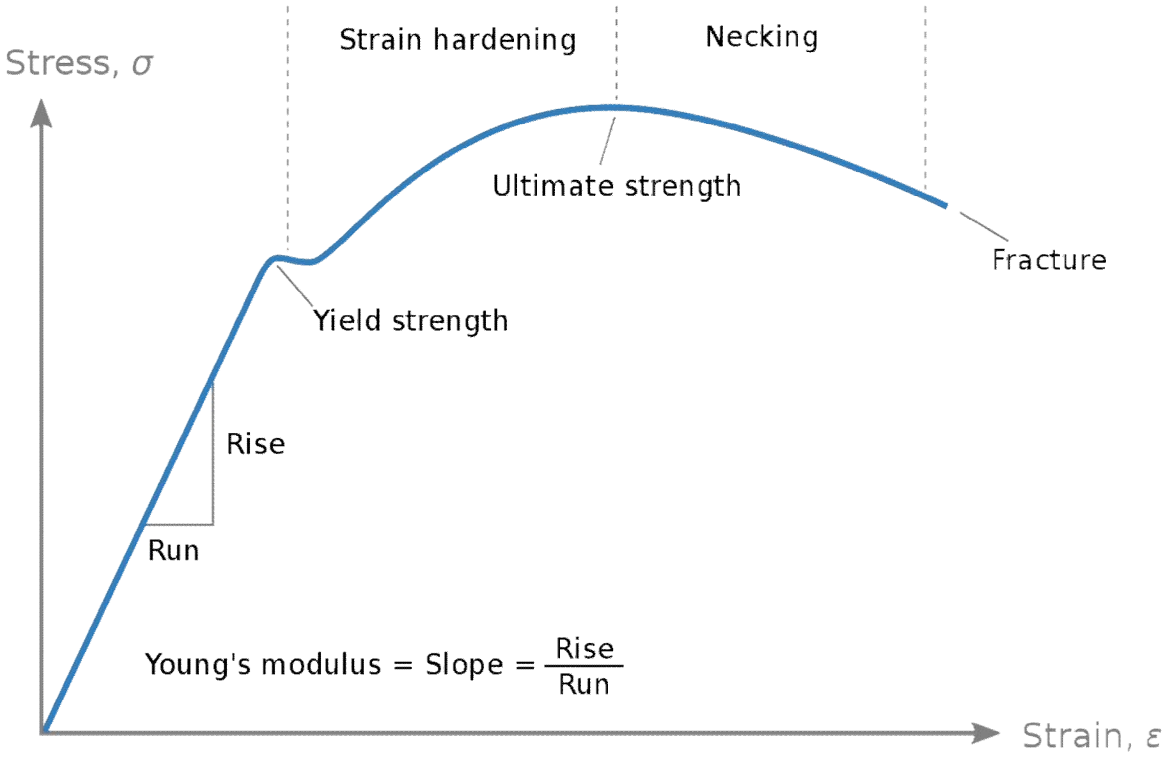
۴- پایان ترم

مباحث مقدماتی

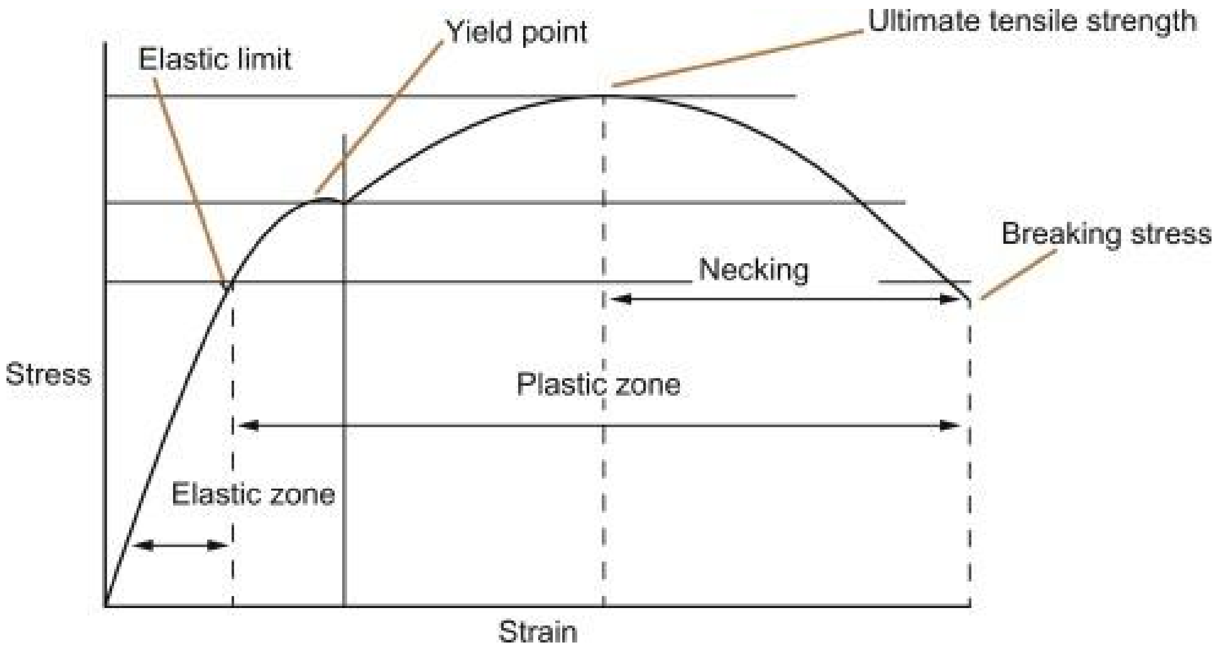
- استحکام
- مدول یانگ
- چقرمگی
- برجهندگی

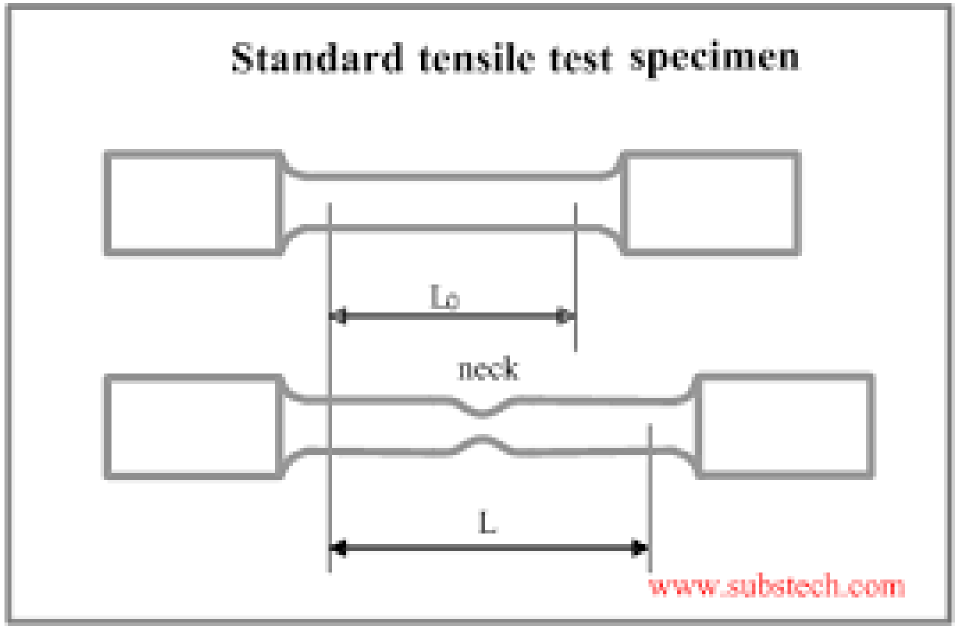
استحکام در دمای پایین و تست کشش

آزمون کشش در دمای محیط



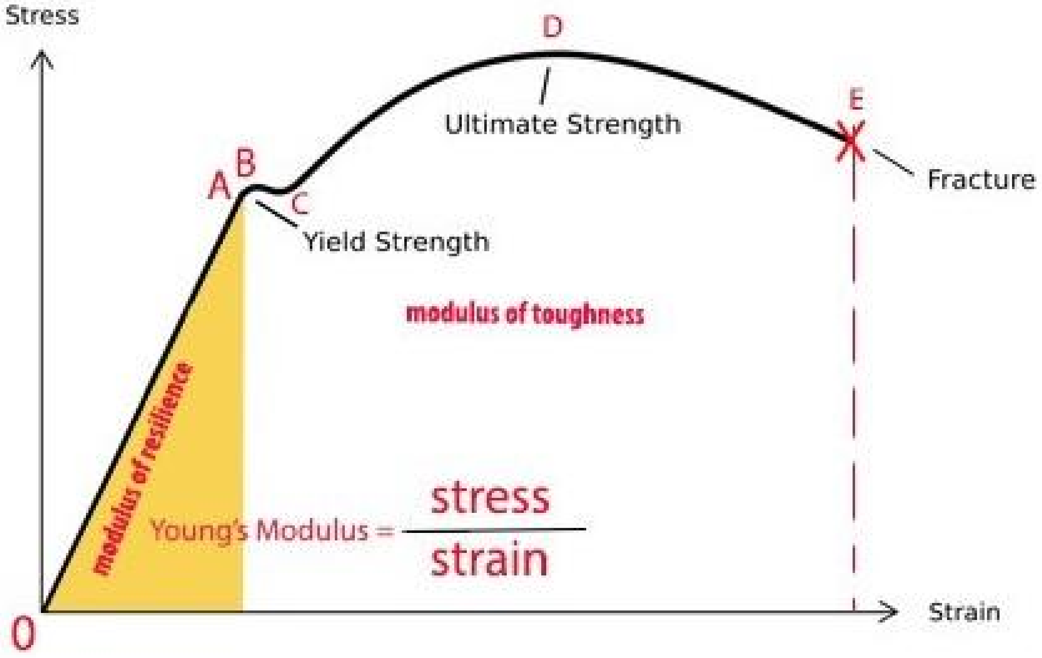
آزمون کشش در دمای محیط



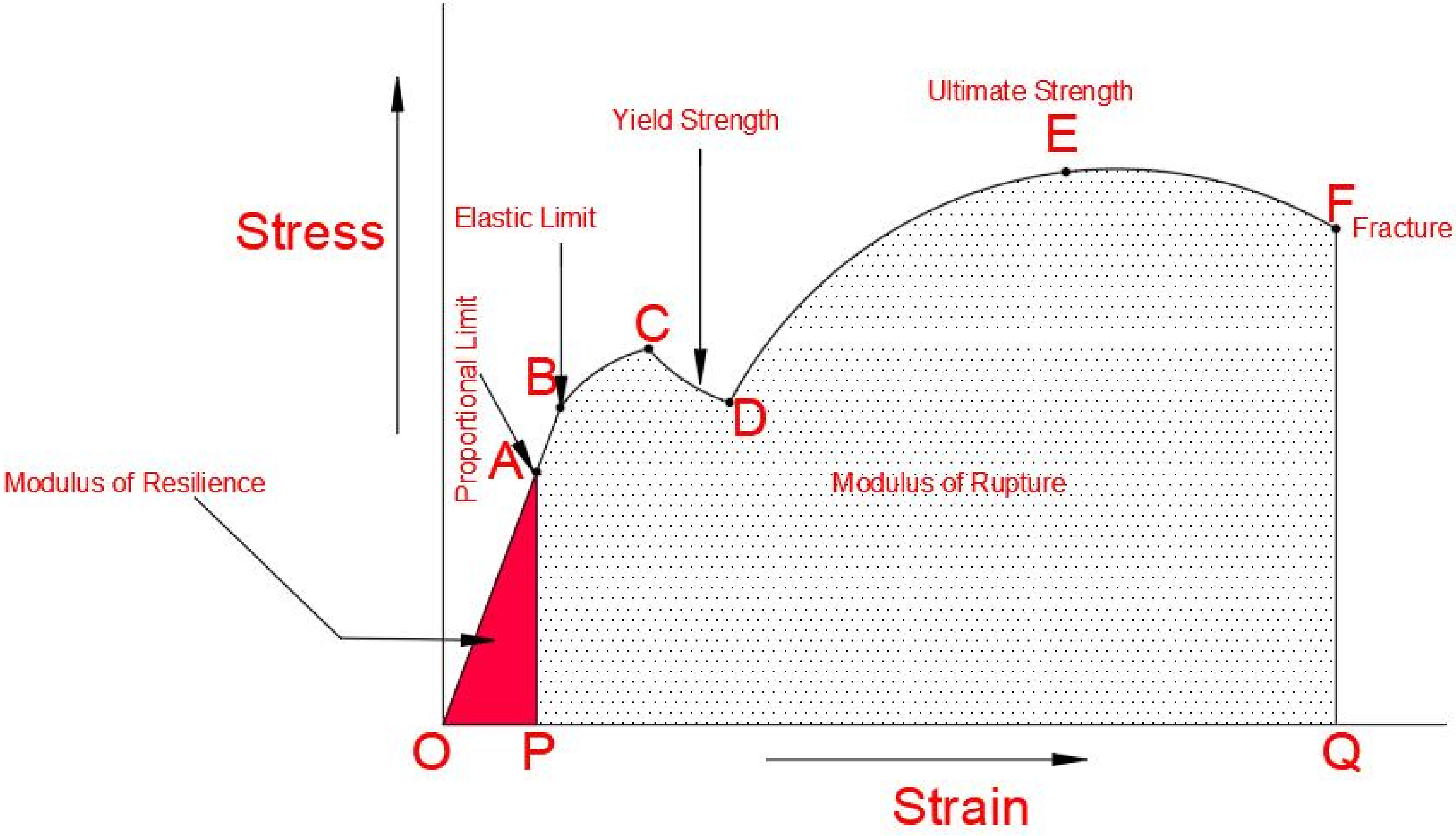


Standard tensile test specimen

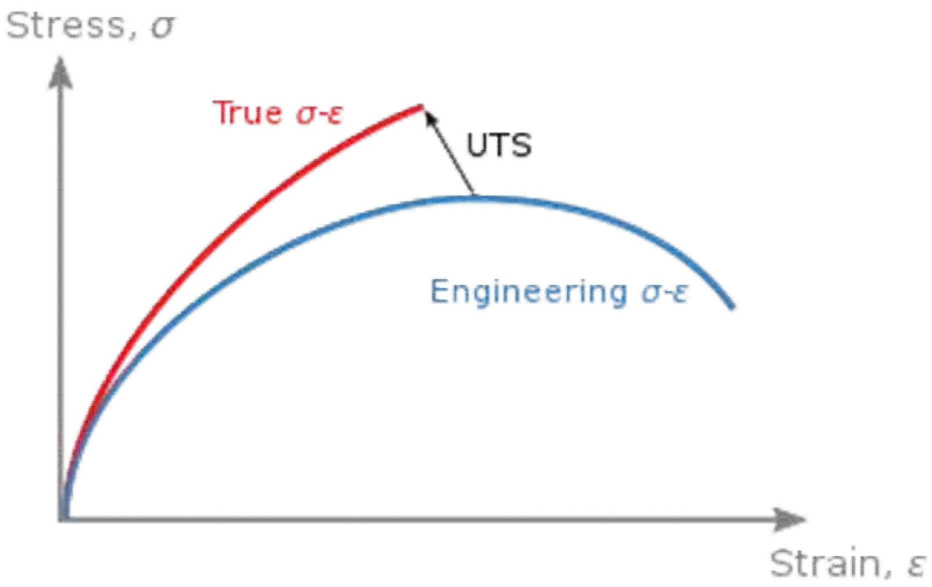
آزمون کشش در دمای محیط



آزمون کشش در دمای محیط



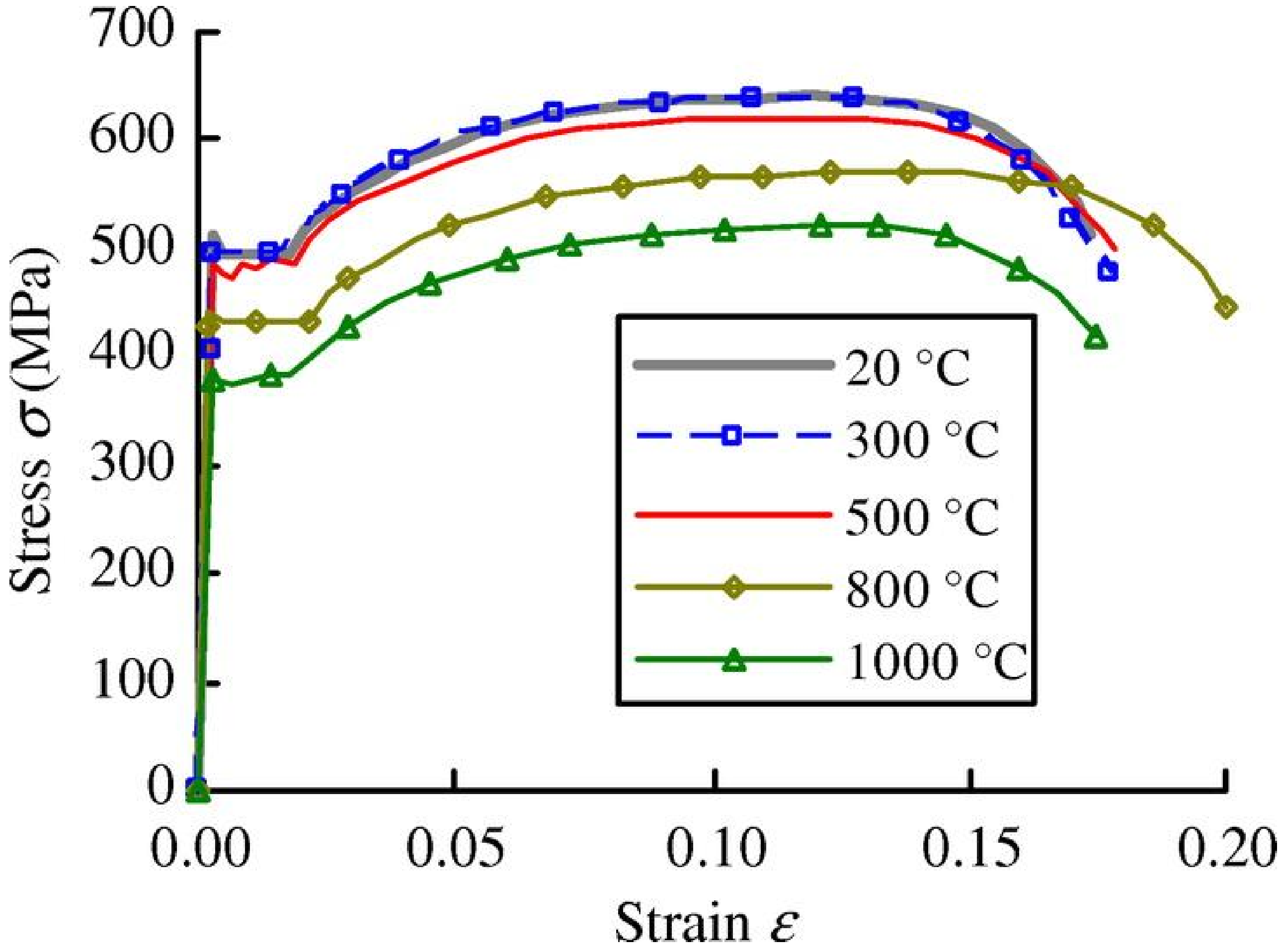
نمودار تنش-کرنش مهندسی
و تنش-کرنش واقعی



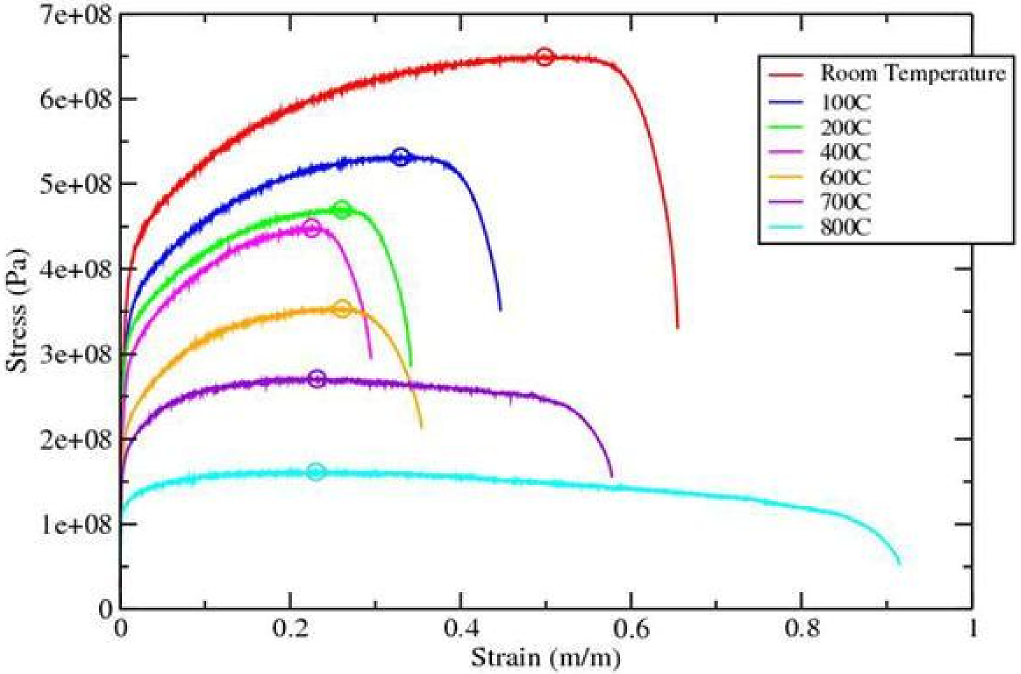
$\sigma=S(1+e)$

$\epsilon=\ln(1+e)$

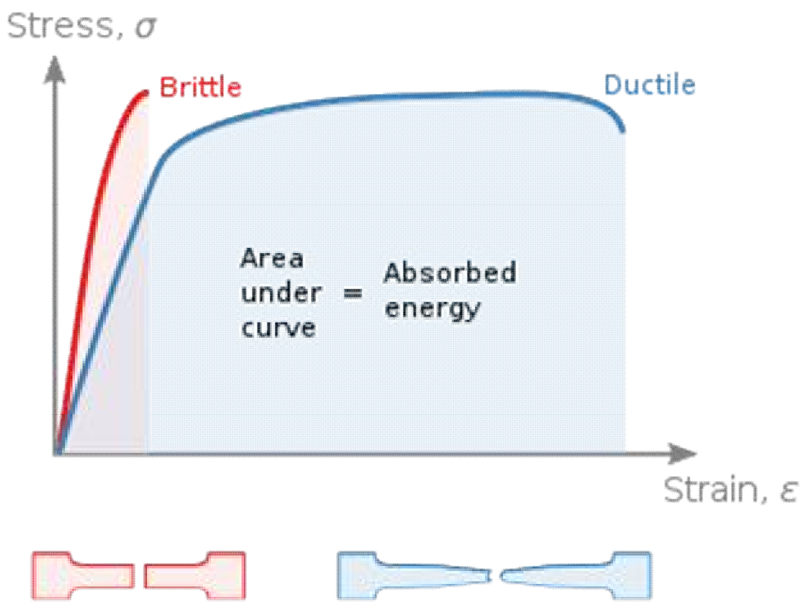
تاثیر دما بر رفتار ماده درآزمون کشش



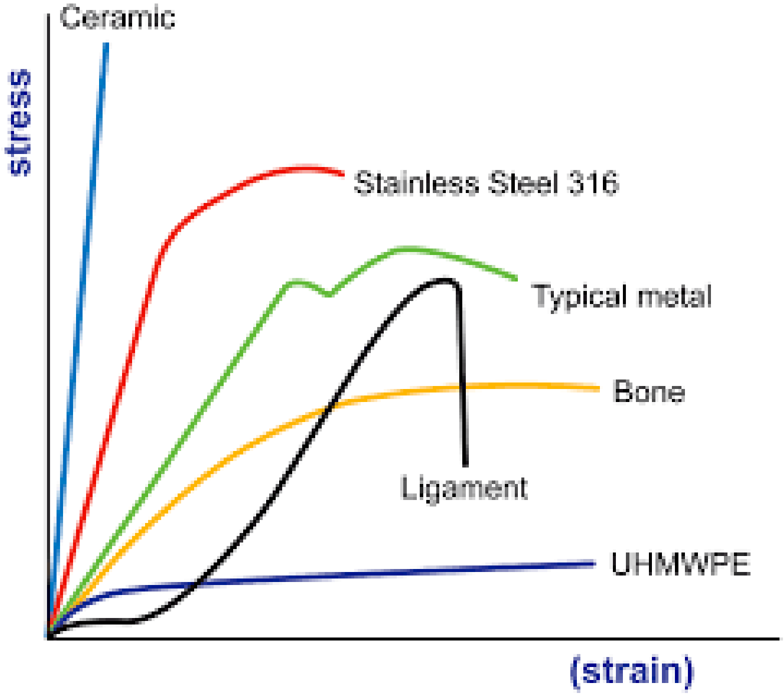
تاثیر دما بر رفتار ماده درآزمون کشش



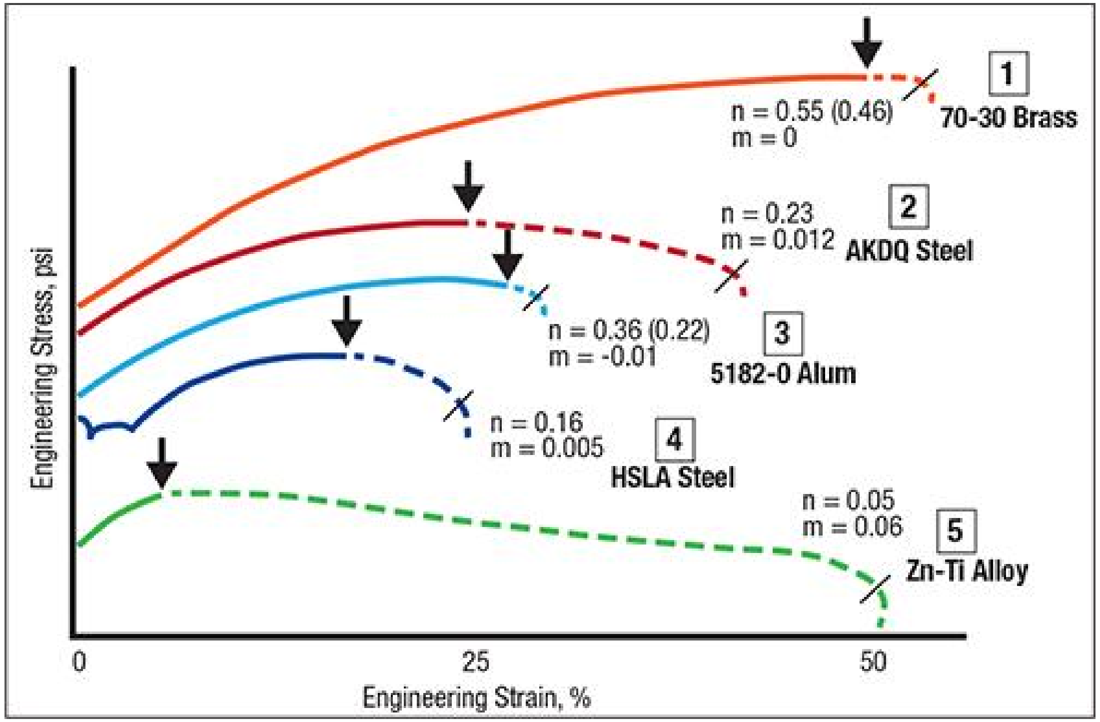
رفتار متفاوت مواد در برابر یک نیروی کششی



رفتار متفاوت مواد در برابر یک نیروی کششی



رفتار متفاوت مواد در برابر یک نیروی کششی



تغییر شکل در مواد

۱- تغییر شکل مستقل از زمان

-تغییر شکل دائم (Plastic)

-تغییرشکل موقت (Elastic)

۲- تغییرشکل وابسته به زمان

- -تغییر شکل دائم (Creep)

-تغییرشکل موقت (Inelastic)

Creep

- Materials in service are often exposed to elevated temperatures or static loads for long duration of time.
- Deformation under such circumstances may be termed as creep.
- **Time-dependent deformation** of a material while under an applied load that is below its **yield strength.**
- Mostly occurs at elevated temperature though some materials creep at room temperature.
- Creep is a deformation mechanism that may or may not constitute a failure mode.

Creep

- Creep is a time-dependent process where a material under an applied stress exhibits a dimensional change at high temprature.
- High temperature progressive deformation of a material at constant stress is called creep.
- The process is also temperature-dependent
- Creep always increases with temperature.

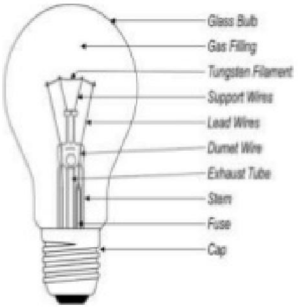
کاربرد خزش

High Temperature Applications

Components exposed to high temperature.

Oil Refinery

Steam turbine used in power plant



Sagging of the filament coil increases with time due to creep deformation caused by the weight of the filament.
Too much deformation--the adjacent turns of the coil touch one another--causing an electrical short and local overheating, which quickly leads to failure of the filament

Chapter 1

Introduction

DESCRIPTION OF CREEP

Creep of materials is classically associated with time-dependent plasticity under a fixed stress at an elevated temperature, often greater than roughly 0.5 Tm, where Tm is the absolute melting temperature

At what temperature the material will creep?

- Different metals have different melting temperatures. e.g. Pb 327°C, W 3407°C.
- Material will creep when the temperature will be **> 0.5Tm (Tm = absolute melting temperature).**

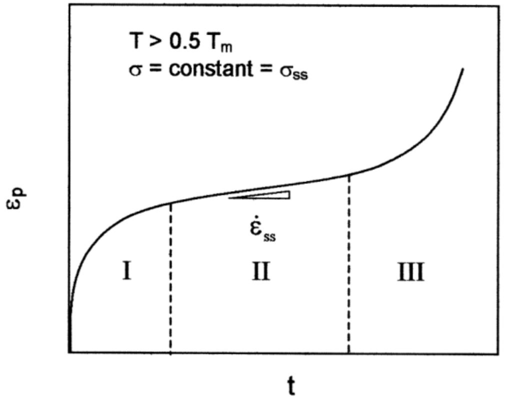
Metal	Melting temp.		0.5xMelting Temp	
Lead	327^{o}C	600^{o}K	327^{o}K	27^{o}C
W	3407	3680	1840	1567

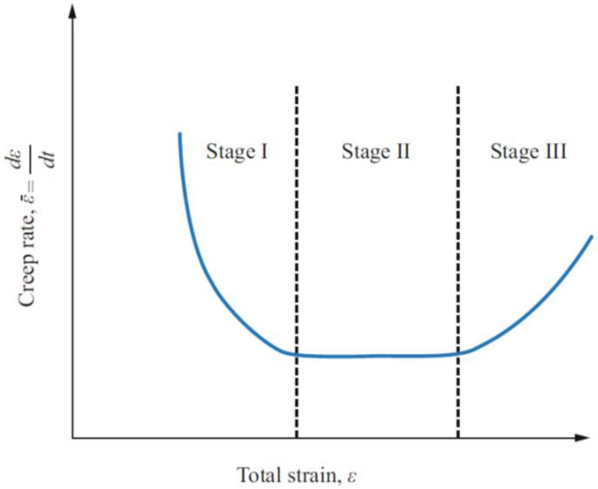
three regions are delineated:

1- Stage I, or primary creep, which denotes that portion where the creep-rate (plastic strain-rate), dϵ/dt is changing with increasing plastic strain or time.

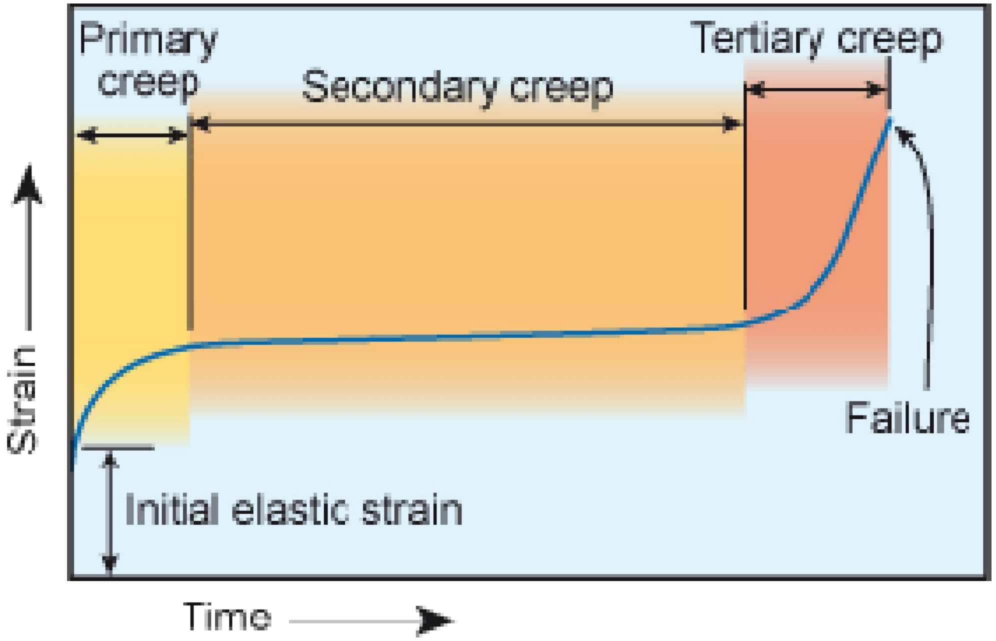
2- constant strain-rate conditions (Stage II, secondary, or steady-state creep), dϵ/dt is constant with increasing plastic strain or time hardens.

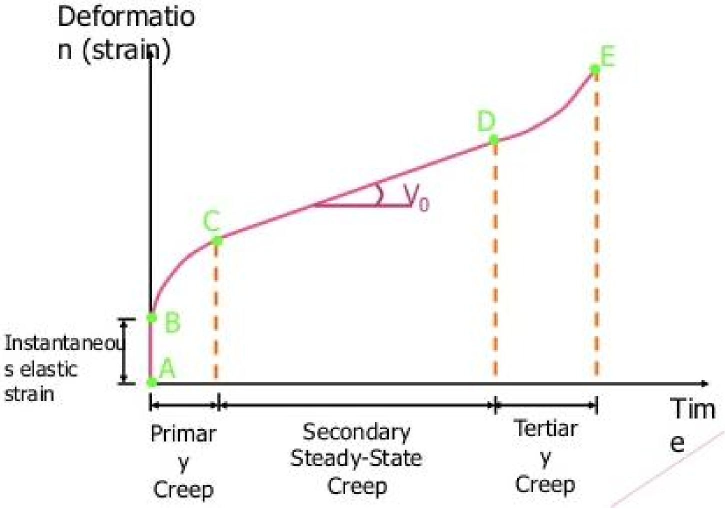
3- cavitation and/or cracking increase the apparent strain-rate or decrease the flow stress. This regime is termed Stage III, or tertiary creep, and leads to fracture.

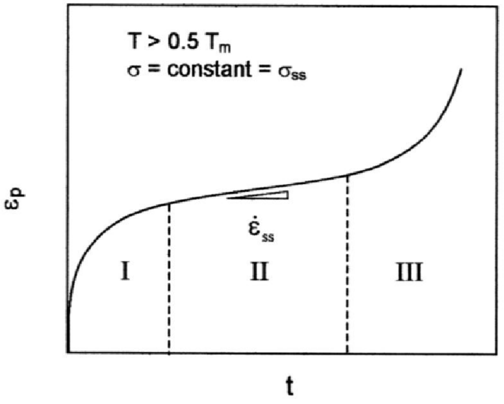


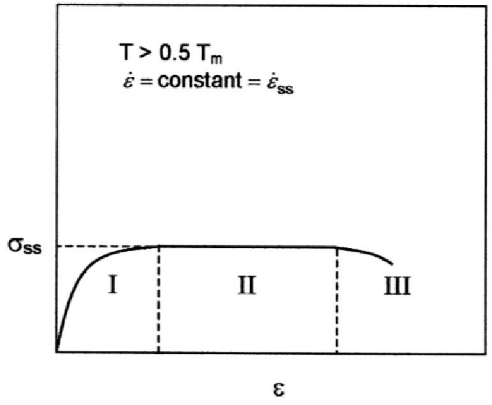


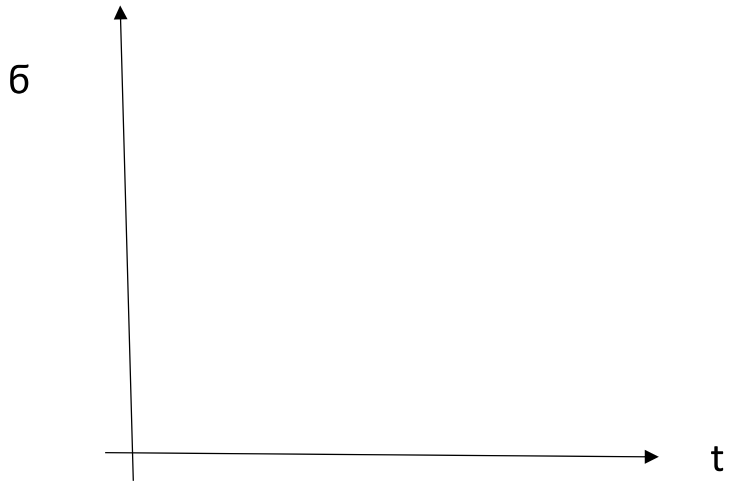
خزش


Primary creep
Secondary creep
Tertiary creep
Failure
Initial elastic strain
Strain
Time


Deformation (strain)
E
D
C
V_0
B
A
Instantaneous elastic strain
Time
Primary Creep
Secondary Steady-State Creep
Tertiary Creep


T > 0.5 T$_m$
σ = constant = σ$_{ss}$
ε$_p$
$\dot{\varepsilon}_{ss}$
I
II
III
t


T > 0.5 T$_m$
$\dot{\varepsilon}$ = constant = $\dot{\varepsilon}_{ss}$
σ$_{ss}$
I
II
III
ε


б
t

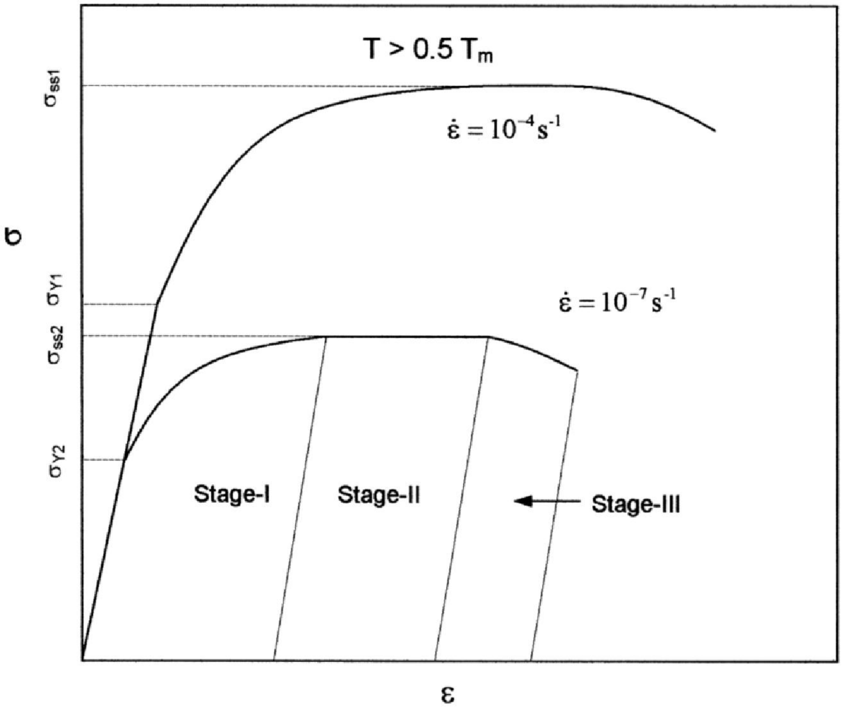


Figure 2. Creep behavior at two different constant strain-rates.

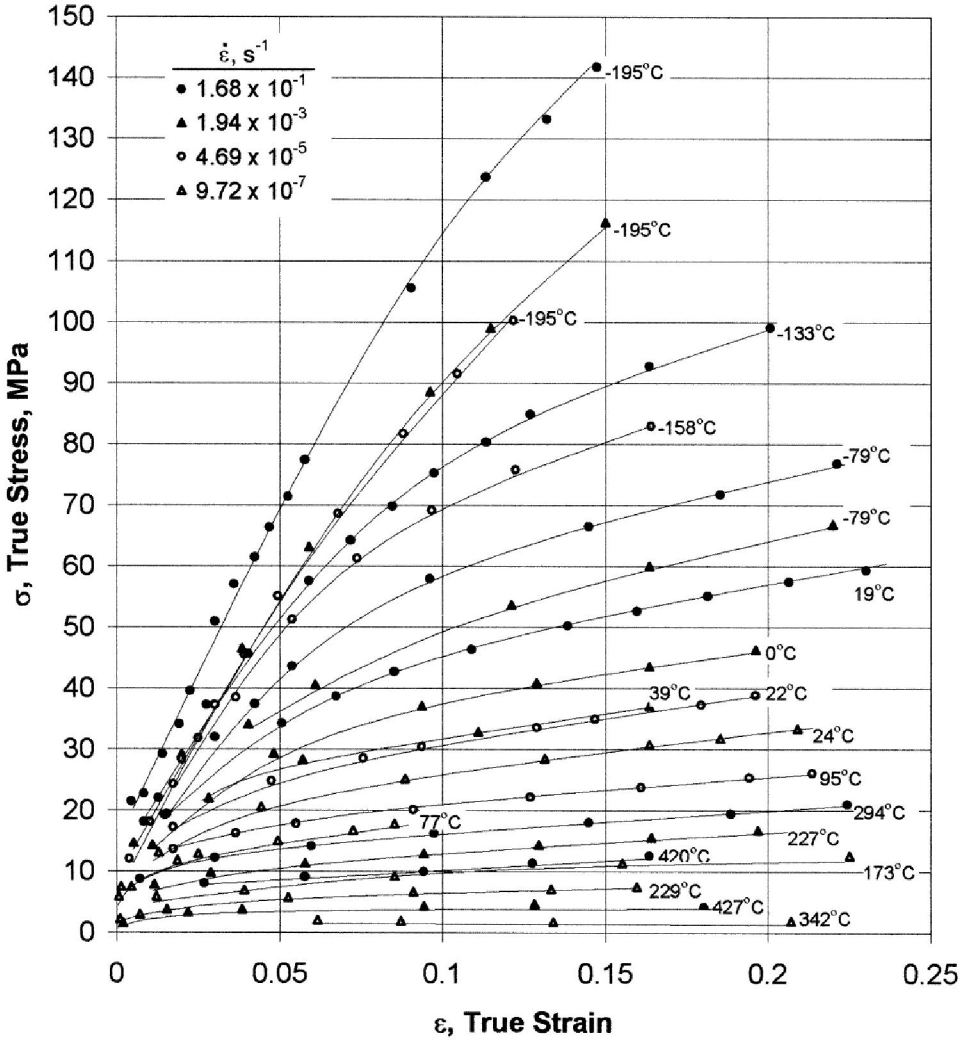


Figure 5. The stress versus strain behavior of high-purity aluminum. Data from Ref. [15].

Creep Design

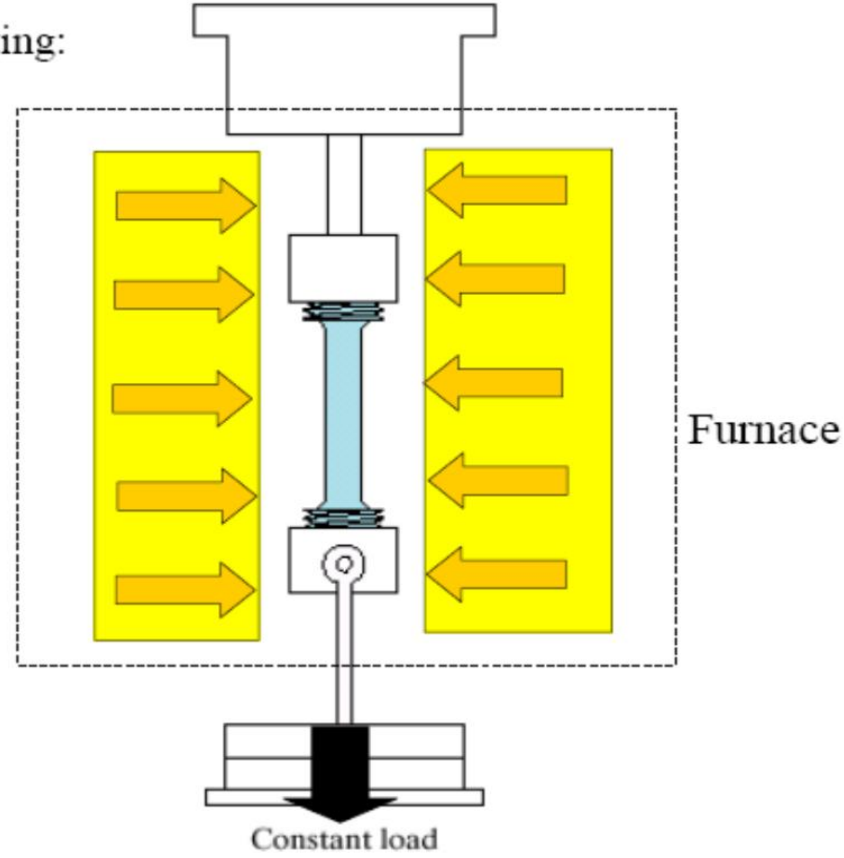
In high-temperature design it is important to make sure:

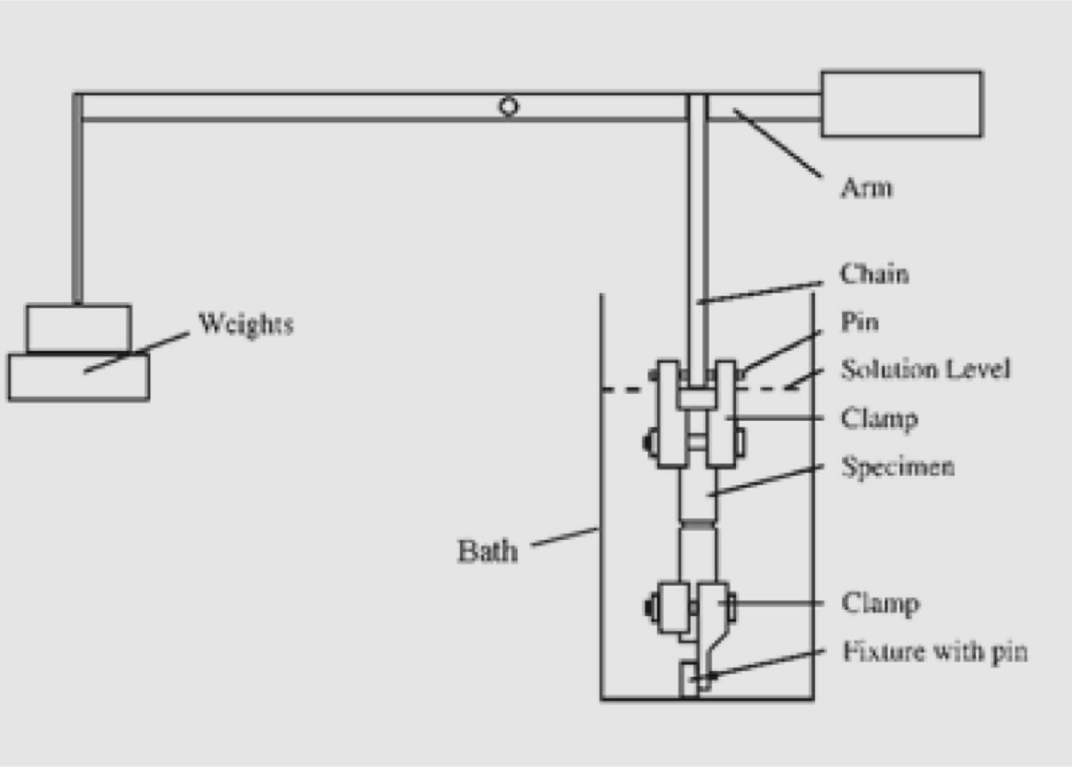
(a) that the creep strain ε^{cr} during the design life is acceptable;
(b) that the creep ductility ε_f^{cr} (strain to failure) is adequate to cope with the acceptable creep strain;
(c) that the time-to-failure, t_f, at the design loads and temperatures is longer (by a suitable safety factor) than the design life.

Creep

Creep is a **time-dependent and permanent** deformation of materials when subjected to a constant load at a **high temperature** (> 0.4 T_m). Examples: turbine blades, steam generators.

Creep testing:




Arm
Chain
Weights
Pin
Solution Level
Clamp
Specimen
Bath
Clamp
Fixture with pin

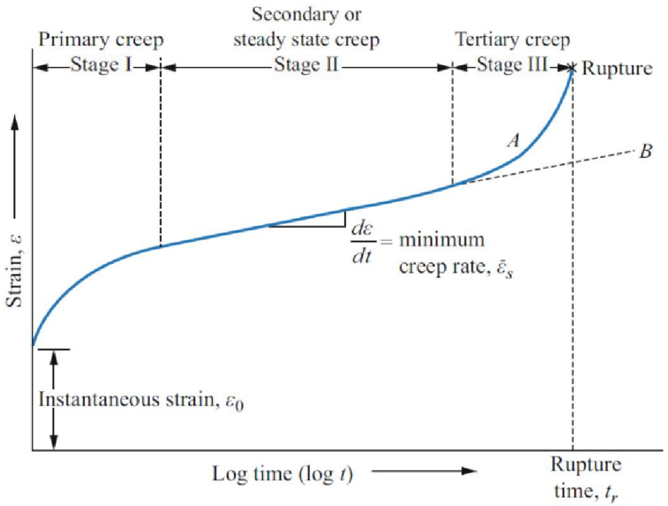


Fig. 7.2 Constant-temperature typical creep curve showing the three distinct stages of creep. Curve '*A*' for constant-load test and curve '*B*' for constant-stress test

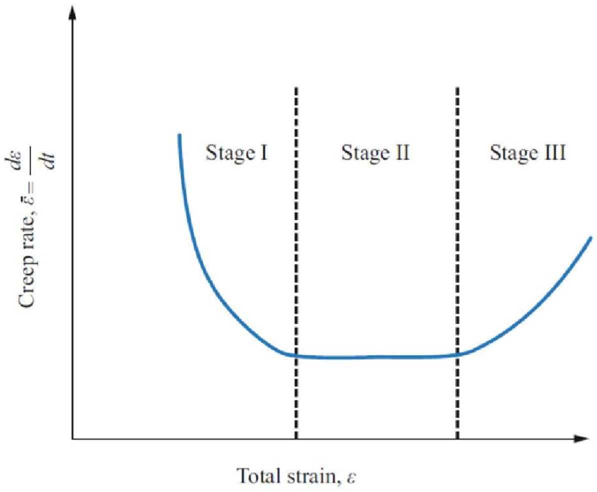
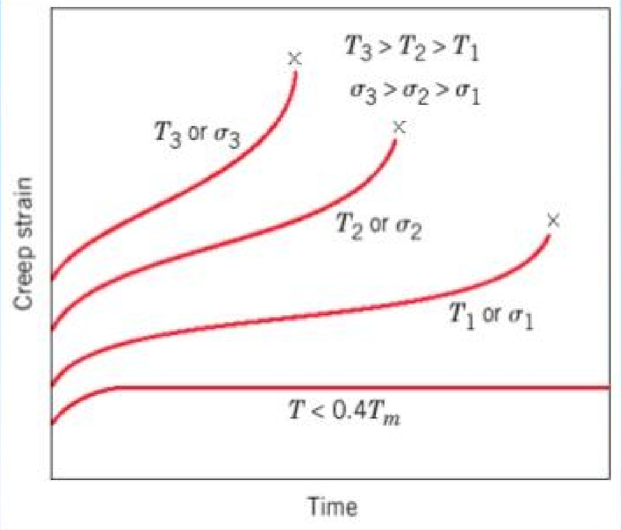


Fig. 7.3 Creep rate as a function of total strain

Sample deformation at a constant stress (σ) vs. time

1. **Instantaneous deformation:** Mainly elastic.
2. **Primary/transient creep:** Slope of strain vs. time decreases with time: work-hardening
3. **Secondary/steady-state creep:** Rate of straining is constant: balance of work-hardening and recovery.
4. **Tertiary/Rapidly accelerating strain rate up to failure:** Formation of internal cracks, voids, grain boundary, separation, necking, etc.

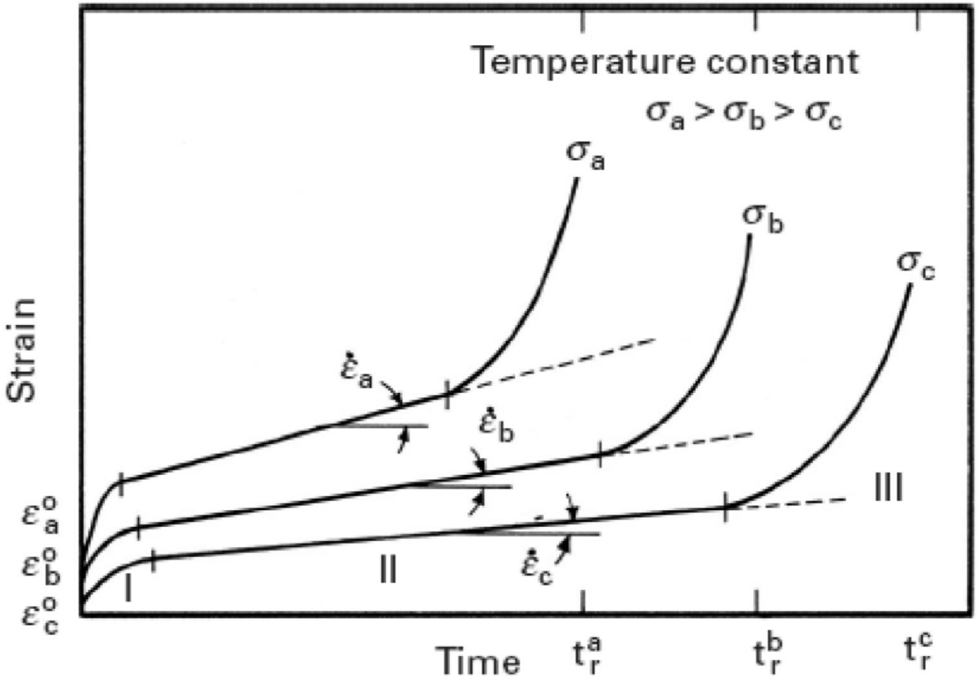
Creep: stress and temperature effects



With Increasing stress or temperature:

- The instantaneous strain increases
- The steady-state creep rate increases
- The time to rupture decreases

Effects of Temperature on Creep



- The instantaneous strain increases
- The steady-state creep rate increases
- The time to rupture decreases

Mechanisms of Creep

- Different mechanisms are responsible for creep in different materials and under different loading and temperature conditions. The mechanisms include

- Stress-assisted vacancy diffusion
- Grain boundary diffusion (diffusion creep)
- Grain boundary sliding
- Dislocation Glide
- Dislocation creep

Homologous temp = $\frac{\text{Testing temperature}}{\text{Melting temperature}}$

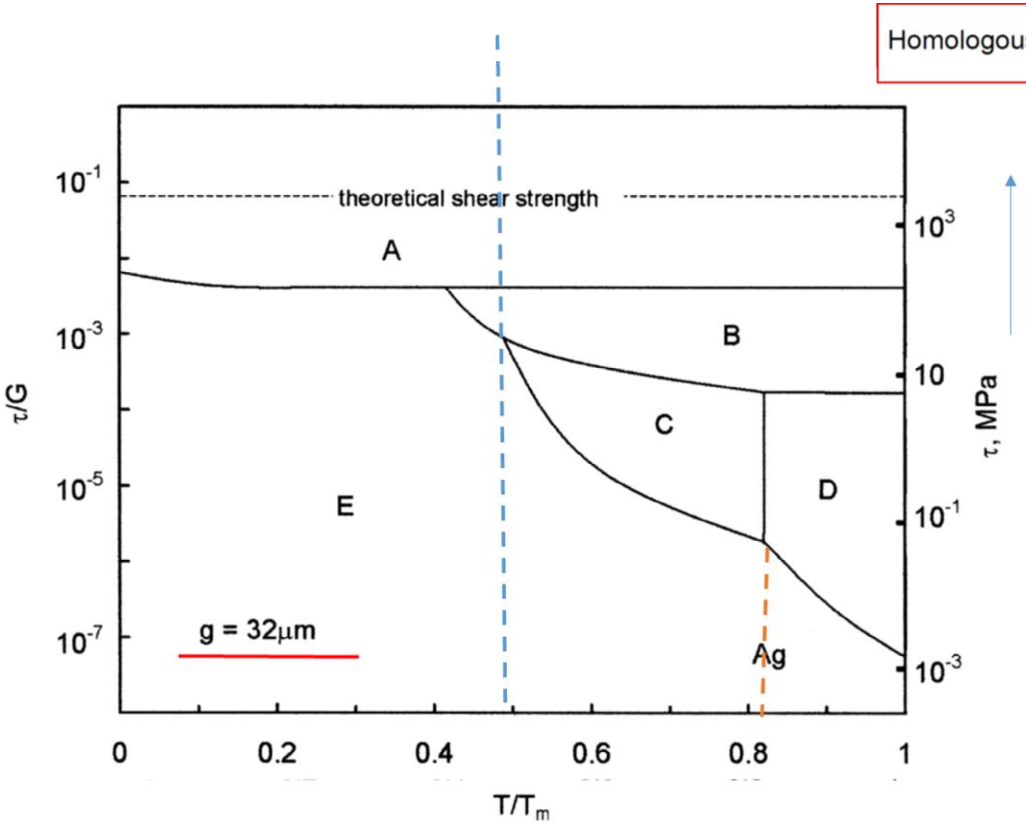


Figure 6. Ashby deformation map of silver from [33]. grain sizes 32 and 100 μm, $\dot{\varepsilon} = 10^{-8}\,\mathrm{s}^{-1}$, A – dislocation glide, B – Five-Power-Law Creep, C – Coble creep, D – Nabarro-Herring creep, E – elastic deformation.

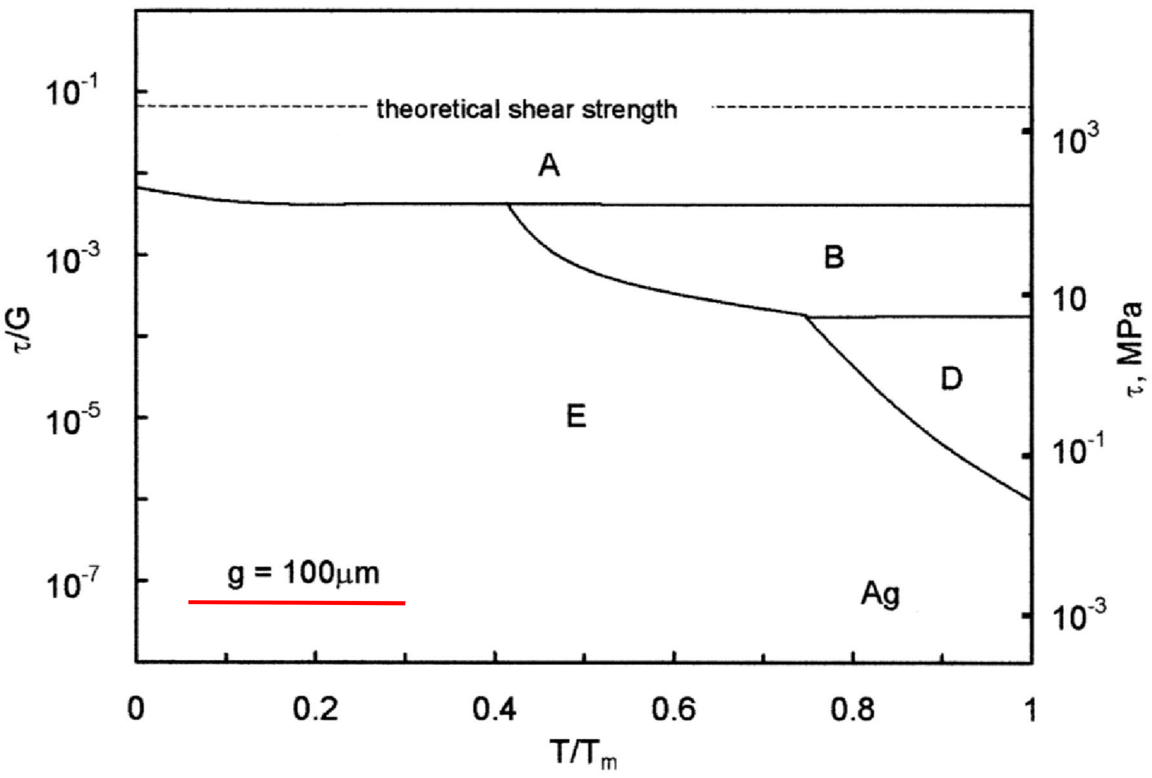
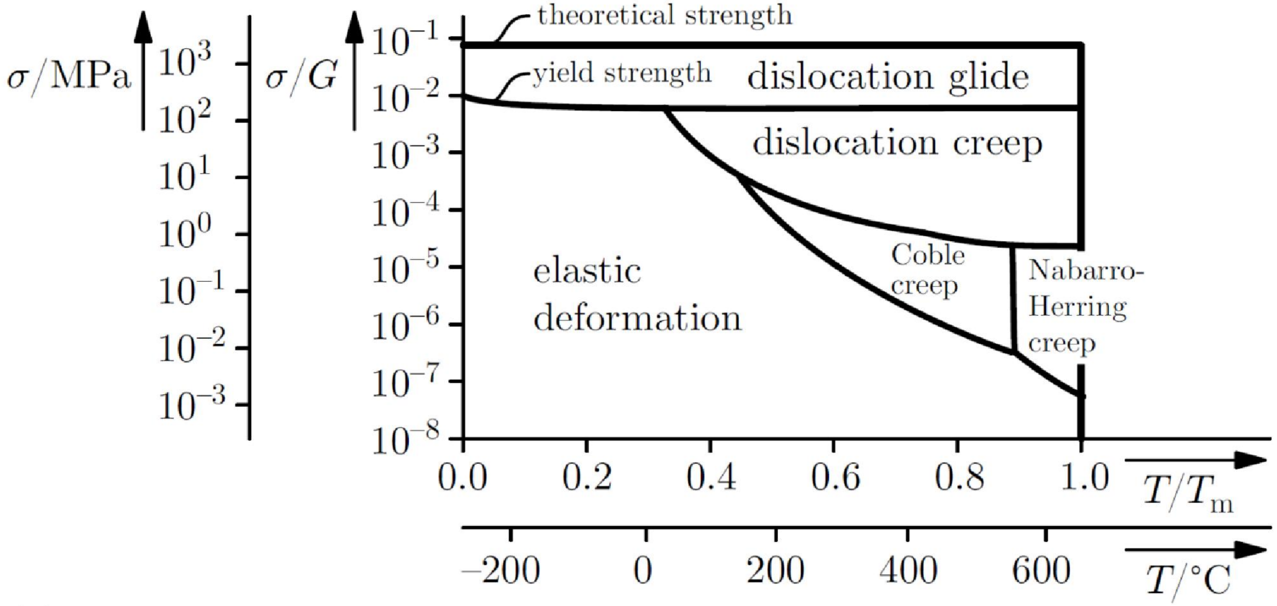
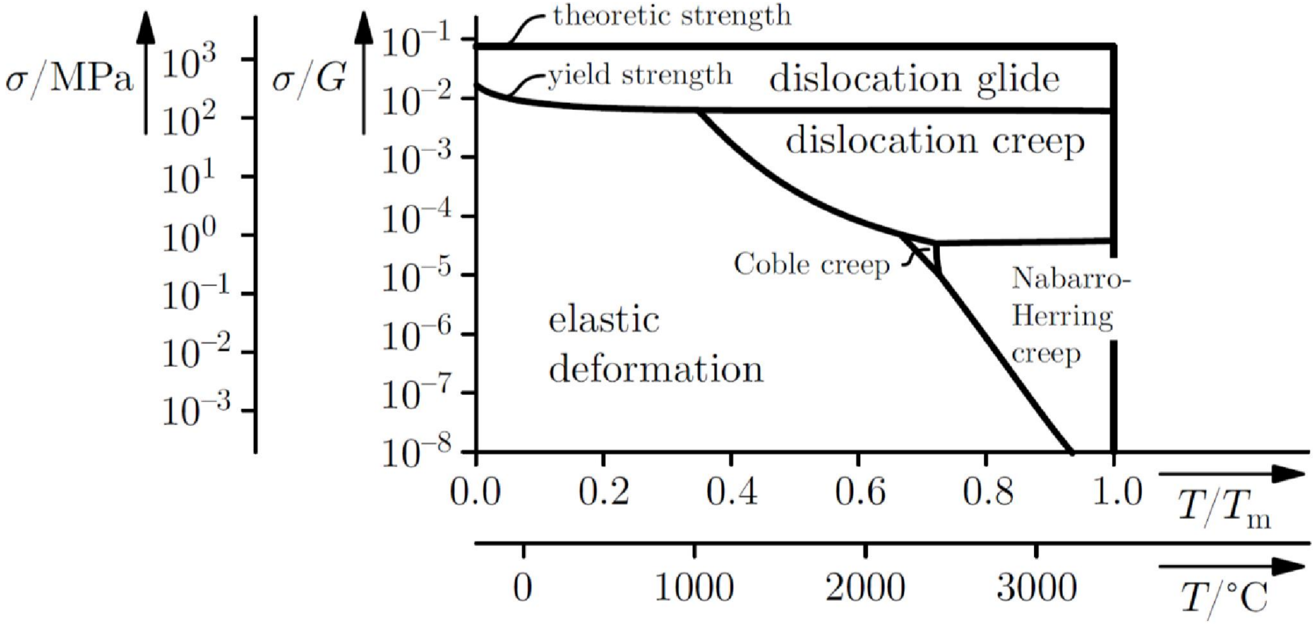


Figure 6. Ashby deformation map of silver from [33]. grain sizes 32 and 100 µm, $\dot{\varepsilon} = 10^{-8}\,s^{-1}$, A – dislocation glide, B – Five-Power-Law Creep, C – Coble creep, D – Nabarro-Herring creep, E – elastic deformation.



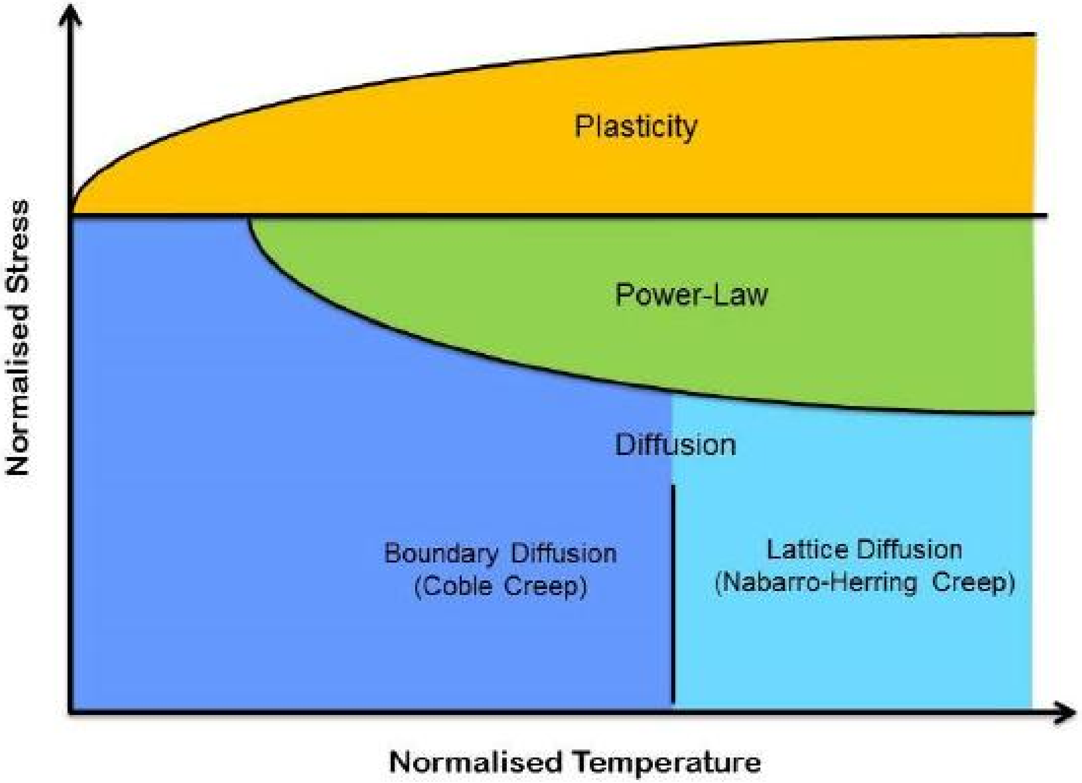
(a) Aluminium

grain size is 32 µm

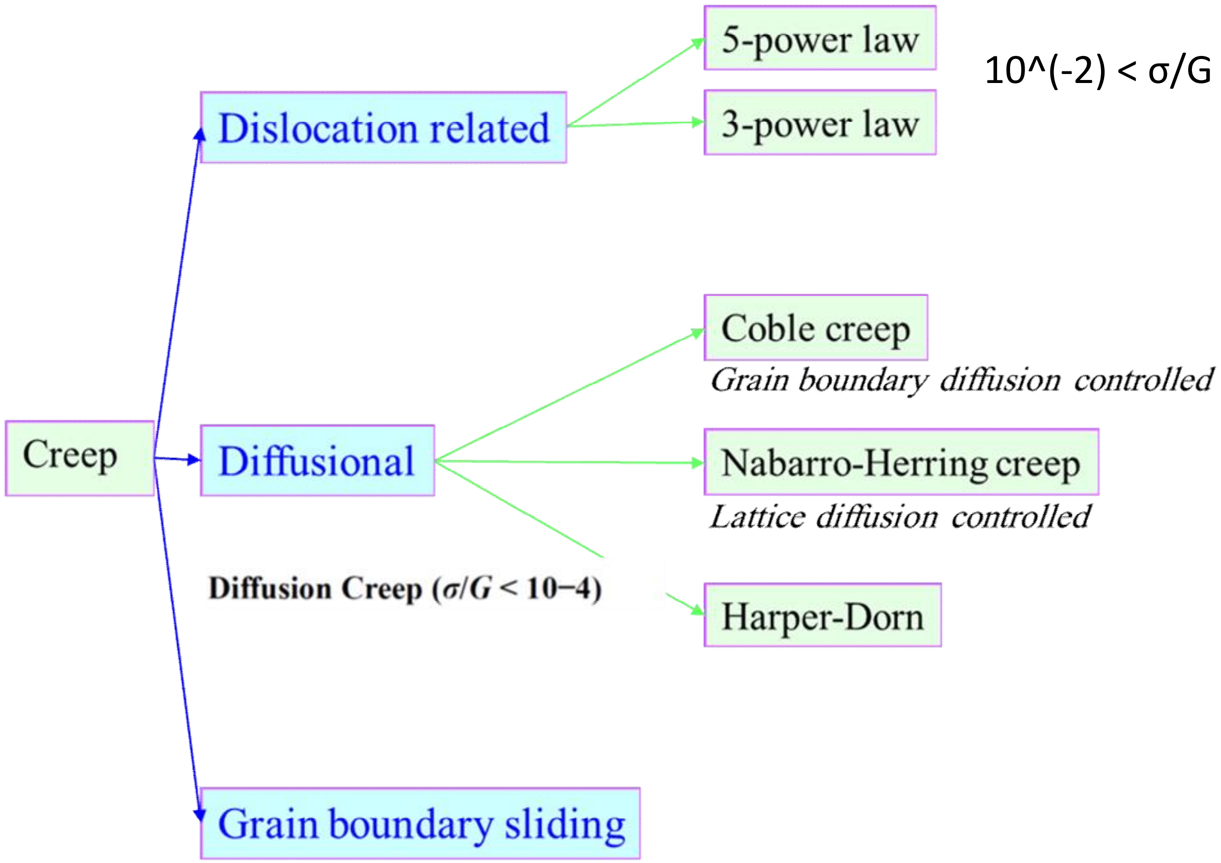


(b) Tungsten

grain size is 32 µm


Plasticity
Power-Law
Diffusion
Boundary Diffusion
(Coble Creep)
Lattice Diffusion
(Nabarro-Herring Creep)
Normalised Stress
Normalised Temperature

Creep Mechanisms *of crystalline materials*



Creep

Dislocation related

5-power law

3-power law

$10^{(-2)} < \sigma/G < 10^{(-4)}$

Diffusional

Coble creep

Grain boundary diffusion controlled

Nabarro-Herring creep

Lattice diffusion controlled

Harper-Dorn

Diffusion Creep ($\sigma/G < 10-4$)

Grain boundary sliding

Chapter 3

Diffusional-Creep

Creep at high temperatures ($T \approx T_{\mathrm{m}}$) and very low stresses in fine-grained materials was attributed 50 years ago by Nabarro [237] and Herring [51] to the mass transport of vacancies through the grains from one grain boundary to another. Excess vacancies are created at grain boundaries perpendicular to the tensile axis with a uniaxial tensile stress. The concentration may be calculated using [23]

$$c = c_v\left[\exp\left(\frac{\sigma b^3}{\mathrm{k}T}\right) - 1\right] \tag{76}$$

where c_v is the equilibrium concentration of vacancies. Usually $(\sigma b^3/\mathrm{k}T) \propto 1$, and therefore equation (76) can be approximated by

$$c = \left[c_v\left(\frac{\sigma b^3}{\mathrm{k}T}\right)\right] \tag{77}$$

Coble creep:
a form of diffusion creep, is a mechanism for deformation of crystalline solids. Coble creep occurs through the diffusion of atoms in a material along the grain boundaries, which produces a net flow of material and a sliding of the grain boundaries.

Coble creep is named after Robert L. Coble, who first reported his theory of how materials creep over time in 1962 in the Journal of Applied Physics.

0.5 Tm<T<0.8 Tm

Coble mechanism

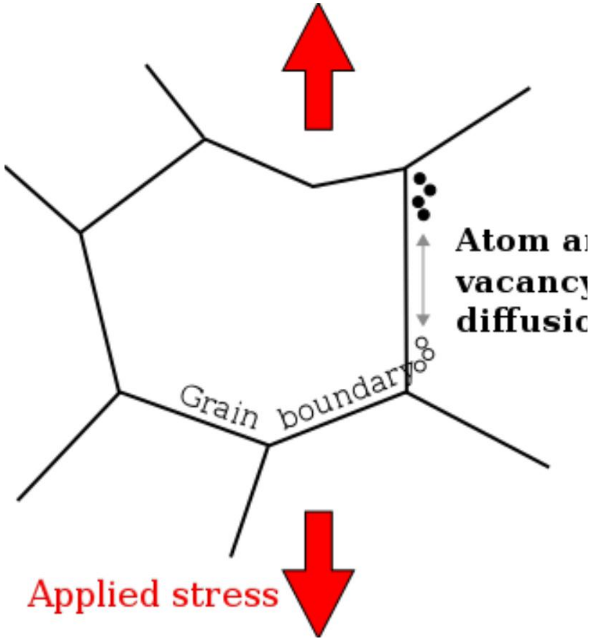
The strain-rate suggested by Coble is

$$\dot{\varepsilon}_{ss} = \frac{\alpha_3 D_{gb} \sigma b^4}{kTg^3}$$

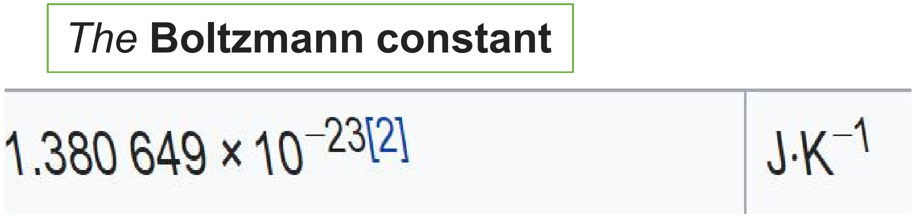
$$D_\nu = D_{0\nu} \exp\left(\frac{-Q_m}{kT}\right)$$

$D\ ,\ \sigma,\ T \uparrow \qquad \epsilon \uparrow$

$g \downarrow \qquad \epsilon \uparrow$

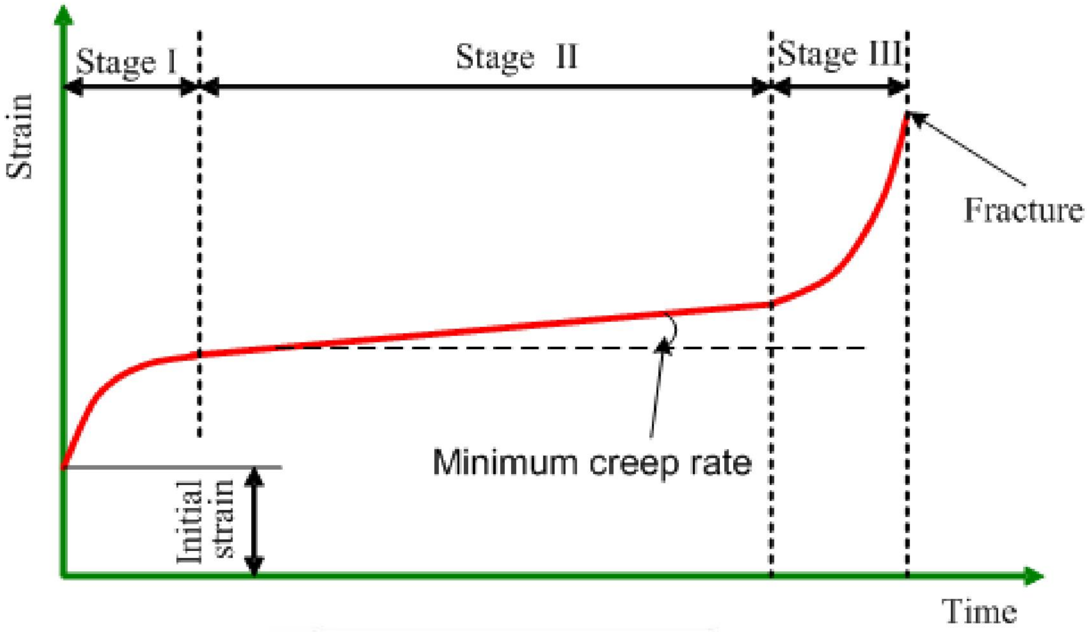


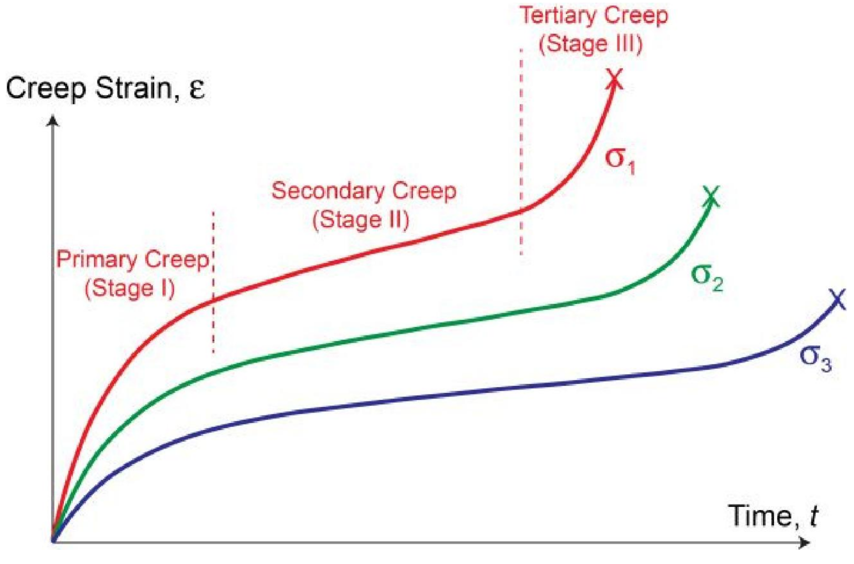
$$\dot{\epsilon} = \frac{ADGb}{kT}\left(\frac{\sigma}{G}\right)^n\left(\frac{b}{d}\right)^p$$

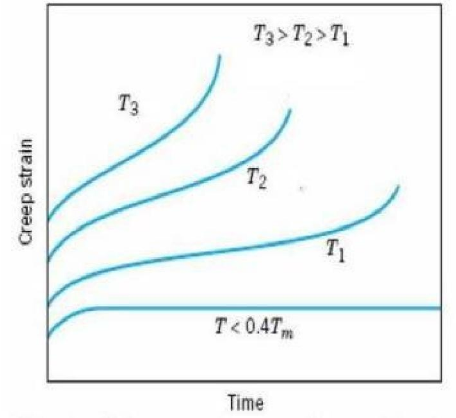

The **Boltzmann constant**

$1.380\,649 \times 10^{-23}$[2]	$J{\cdot}K^{-1}$

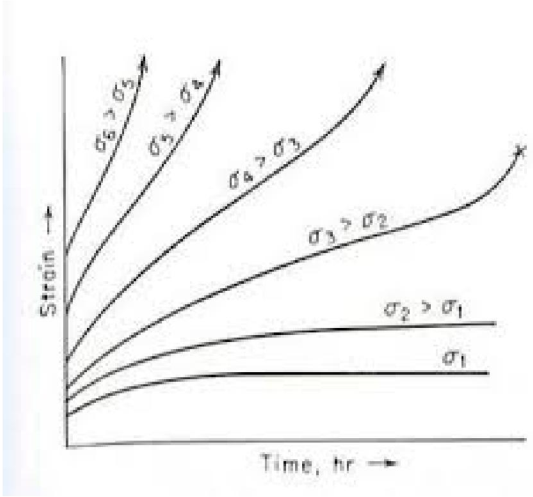
Creep behavior

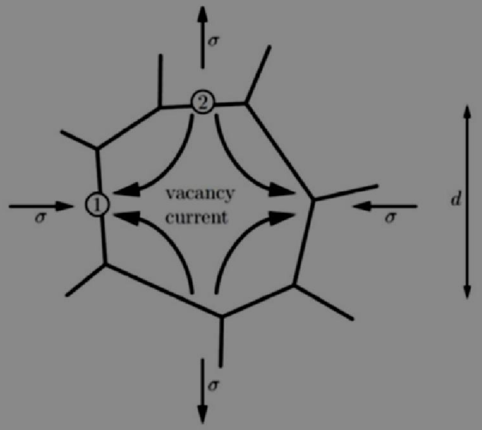






Effect of Temperature on Creep Strain





$$C_V^+ = C_0 \exp -\left(\frac{H_V - \sigma\Omega}{kT}\right) = C_V \cdot \exp\left(\frac{\sigma\Omega}{kT}\right)$$

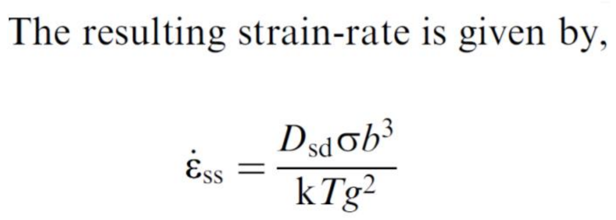
$$C_V^- = C_0 \exp -\left(\frac{H_V + \sigma\Omega}{kT}\right) = C_V \cdot \exp\left(\frac{-\sigma\Omega}{kT}\right)$$

$$J_V = -D_V \nabla C_V = D_V \frac{\Delta C_V}{\Delta x} = D_V \frac{\Delta C_V}{d}$$

$$\Delta C_V = C^+ - C^- = C_V\left(e^{\frac{\sigma\Omega}{kT}} - e^{\frac{-\sigma\Omega}{kT}}\right) \approx C_V \cdot \frac{2\sigma\Omega}{kT}$$

$$J_V = \frac{D_V}{d} \cdot C_V \cdot \frac{2\sigma\Omega}{kT}$$

Nabarro and Herring mechanism



The resulting strain-rate is given by,

$$\dot{\varepsilon}_{\mathrm{ss}} = \frac{D_{\mathrm{sd}}\sigma b^3}{\mathrm{k}Tg^2}$$

$$\dot{\epsilon}_{NH} = A_{NH}\left(\frac{D_L}{d^2}\right)\left(\frac{\sigma\Omega}{kT}\right)$$

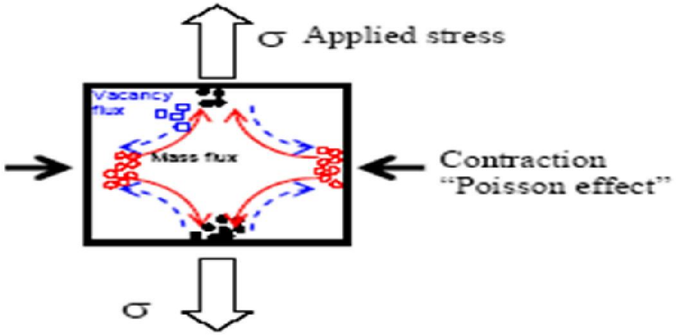
where A_{NH} is a constant that absorbs the approximations in the derivation.

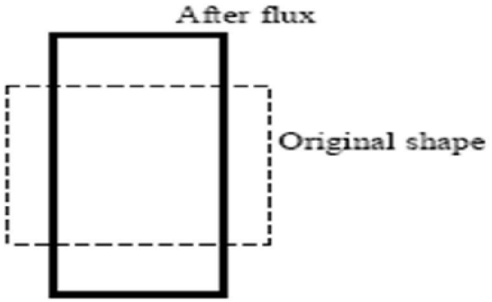
Nabarro-Herring Creep

- Occurs solely by diffusional mass transport.
- Is important for much higher T's and lower σ's than was the case for dislocation glide creep.
- Can occur in crystalline and amorphous materials.

★ Applied stress creates tensile and compressive regions.

  - Concentration of vacancies in tensile region > compressive
  - Vacancy concentration gradient → diffusion
  - Diffusion leads to shape change



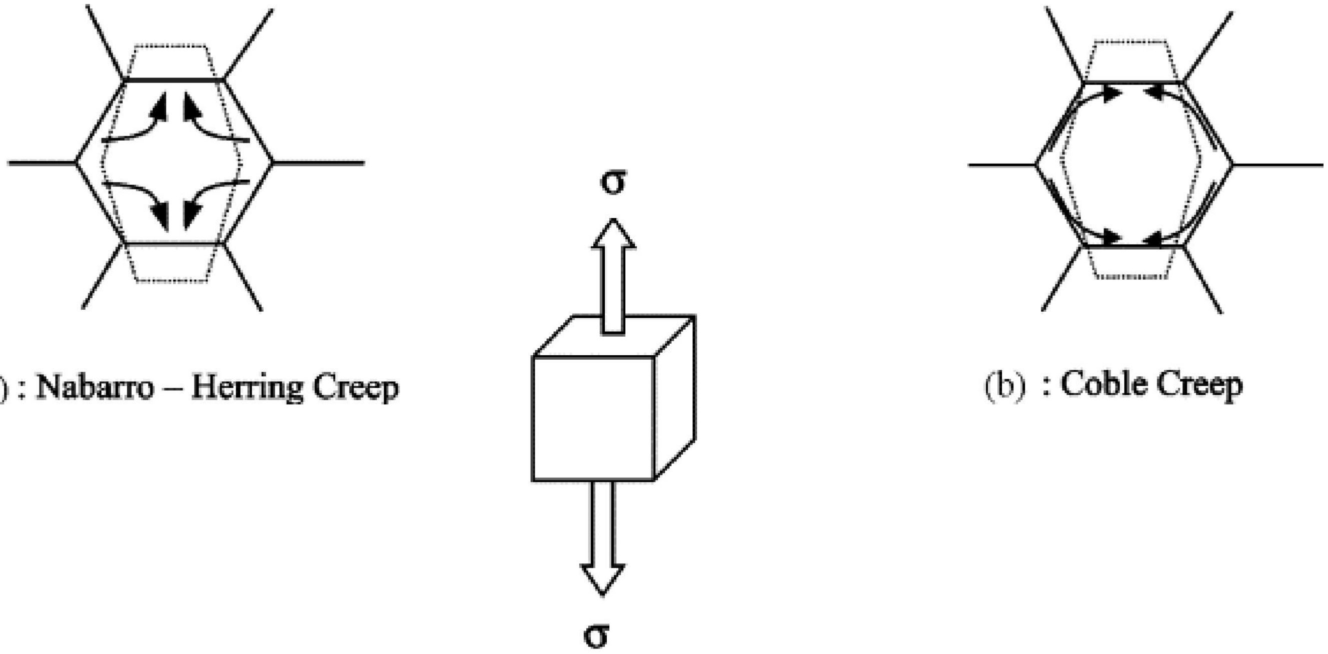


$$\boxed{\dot{\varepsilon}_{NH} \approx \frac{A_{NH} D_L G b}{kT}\left(\frac{b}{d}\right)^2\left(\frac{\sigma}{G}\right)}$$

$A_{NH} \cong 10\text{-}15$

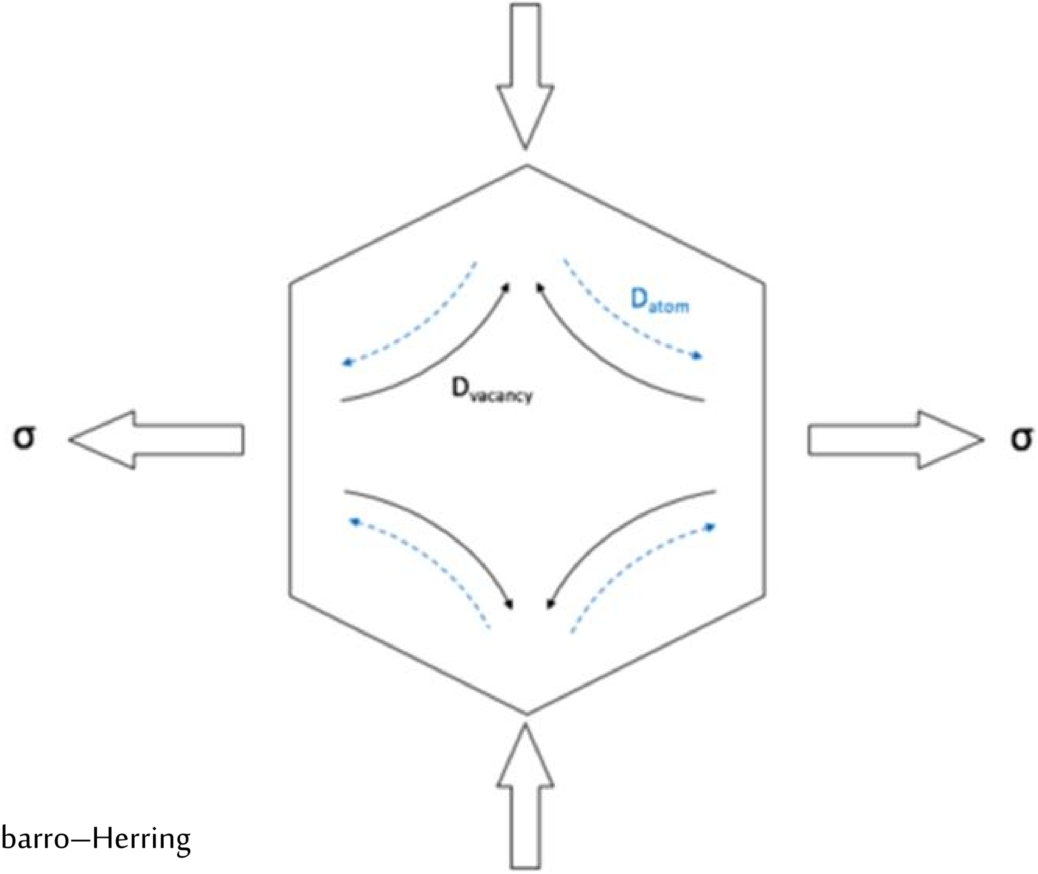
D_L = lattice diffusion coefficient

$$\begin{aligned}\dot{\varepsilon} &= A_{\mathrm{NH}} \frac{\sigma \Omega}{kT} \frac{D_0}{d^2} \exp\left(-\frac{Q_{\mathrm{V}} + Q_{\mathrm{ex}}}{kT}\right) \\ &= A_{\mathrm{NH}} \frac{\sigma \Omega}{kT} \frac{D_{\mathrm{V}}}{d^2} \, .\end{aligned}$$

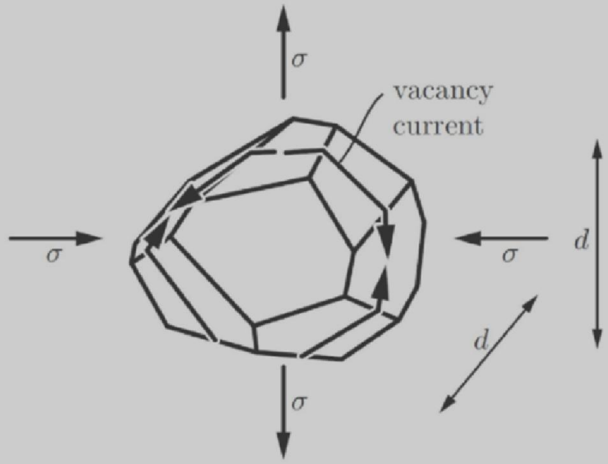

(a) : Nabarro – Herring Creep

(b) : Coble Creep

Mechanism	favorable conditions	Description	A	n	p
Nabarro–Herring creep	High temperature, low stress and small grain size	Vacancy diffusion through the crystal lattice	10–15	1	2
Coble creep	Low stress, fine grain sizes and temperature less than those for which NH creep dominates	Vacancy diffusion along grain boundaries	30–50	1	3



The strain-rate is proportional to g2 in the Nabarro–Herring model whereas it is proportional to g3 in the Coble model



Movement of vacancies along grain boundaries in diffusion creep

Chapter 4

Harper–Dorn Creep

Harper-Dorn creep mechanism observed at high temp. and low stress
in aluminum single crystals and in polycrystalline specimens with a 3.3mm grain size.
With high purity (99.999%)

$$\dot{\varepsilon}_{ss} = A_{HD}\left(\frac{D_{sd}Gb}{kT}\right)\left(\frac{\sigma}{G}\right)^{1}$$

where A_{HD} is a constant.

$$\dot{\varepsilon}_{\mathrm{ss}} = \mathrm{A}_{\mathrm{HD}}\left(\frac{D_{\mathrm{sd}}Gb}{\mathrm{k}T}\right)\left(\frac{\sigma}{G}\right)^{1}$$

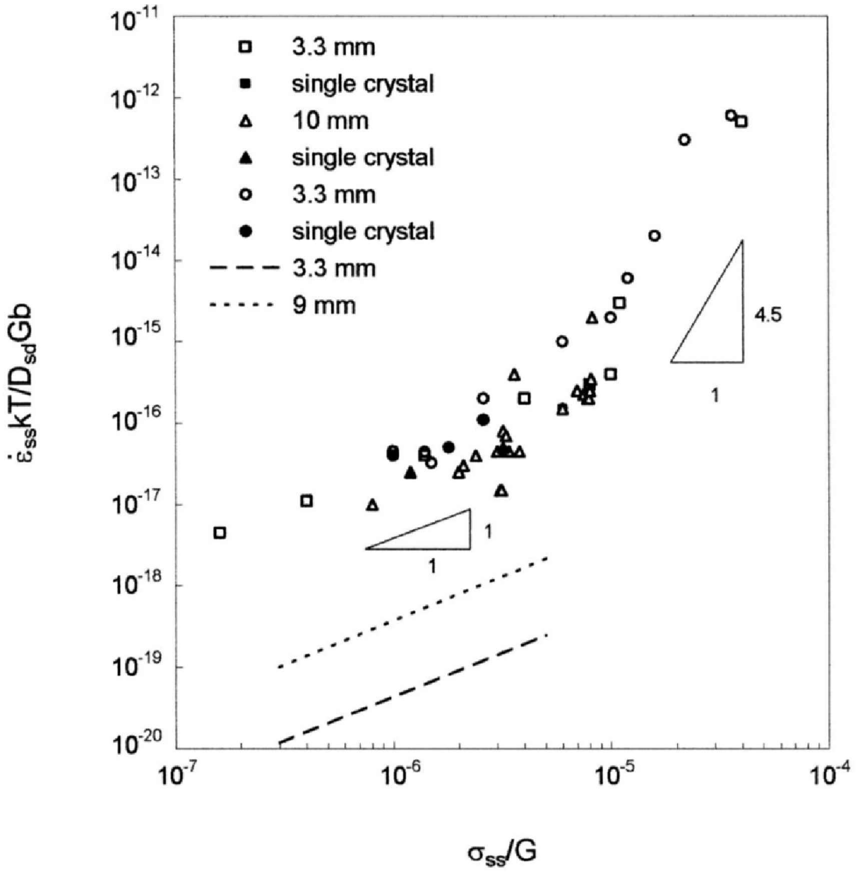
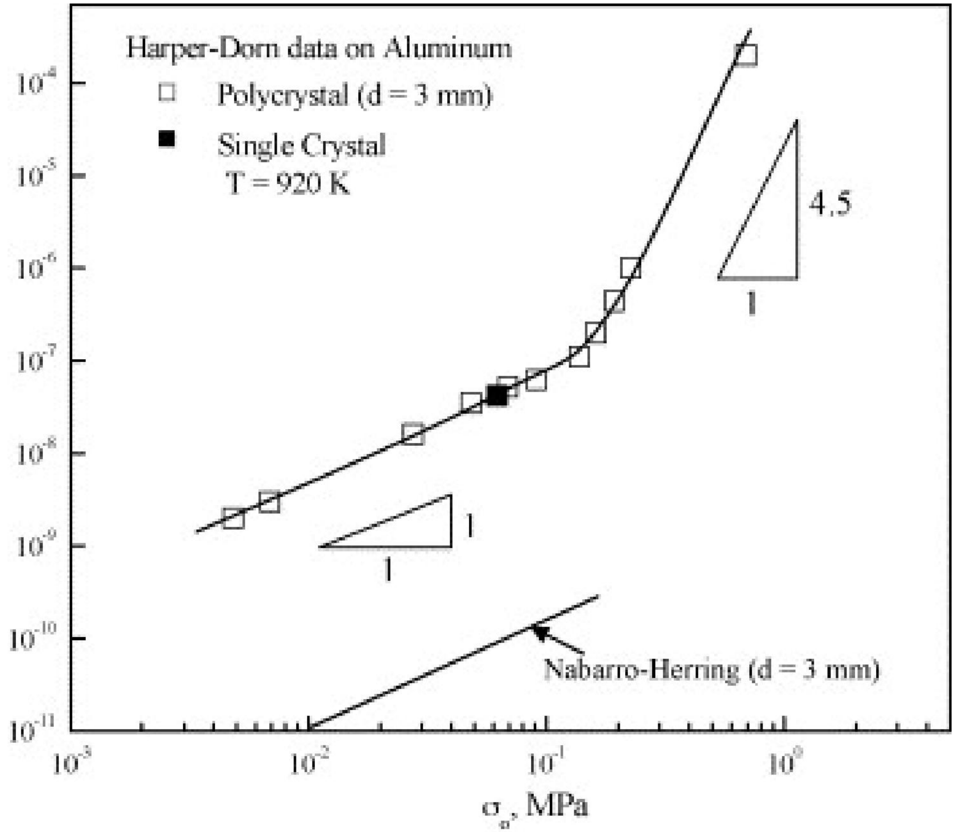


Figure 50. Comparison between the diffusion-coefficient compensated strain-rate versus modulus-compensated stress for pure aluminum based on [50,269,294], with theoretical predictions for Nabarro–Herring creep [295].



$$\dot{\varepsilon}_{ss} = A_{10} \frac{D_{\text{eff}}}{b^2} \left(\frac{\sigma}{E}\right)^n$$

Chapter 6

Superplasticity

Superplasticity

- ***Superplasticity*** is the ability to withstand very large deformation in tension without necking.

- Give elongation > 1000%.

- Materials with high ***strain rate sensitivity*** (m) at high temperature ($T>0.5T_m$) → ***superplasticity***

- Materials characteristics: ***fine grain size*** (<10 μm) with the presence of ***second phase of similar strength to the matrix*** to inhibit grain growth and to avoid extensive internal cavity formation.

- Grain boundary should be ***high angle*** and ***mobile*** to promote grain boundary sliding and to avoid the formation of local stress concentration respectively.

- Superplastic deformation can be utilized to help shape complex geometry at high temperatures

Superplasticity

Superplastic tensile deformation in Pb–62% Sn eutectic alloy tested at 415 K and a strain rate of 1.33 × 10−4 s−1; total strain of 48.5.

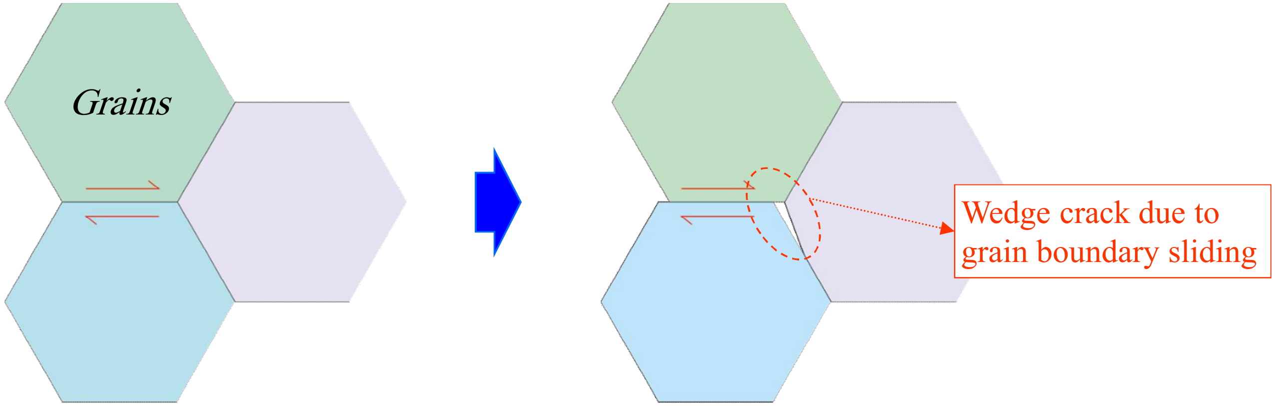
(From M. M. I. Ahmed and T. G. Langdon, Met. Trans. A, 8 (1977) 1832.)

Grain-boundary sliding. It is a shear process occurring in the direction of grain boundary, causing the movement of grains relative to each other in polycrystals. Grain boundaries lying at about 45° to the applied tensile stress will experience the maximum shear stress and slide the most. It is encouraged by decreasing the strain rate and/or increasing the temperature.

$$\sigma \geq 10^{-2} G.$$

Grain boundary sliding

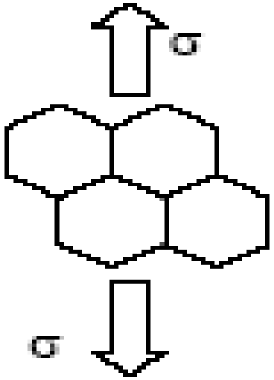
- At low temperatures the grain boundaries are '*stronger*' than the crystal interior and impede the motion of dislocations.
- Being a higher energy region, the grain boundaries *pre-melt* before the crystal interior.
- Above the *equicohesive* temperature, due to shear stress at the 'local scale', grain boundaries slide past one another to cause plastic deformation.
- The relative motion of grain boundaries can lead to wedge cracks at triple lines (junction of three grains). If these wedge cracks are not healed by diffusion (or slip), microstructural damage will accumulate and will lead to failure of the specimen.

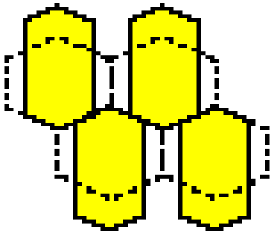


Wedge crack due to grain boundary sliding

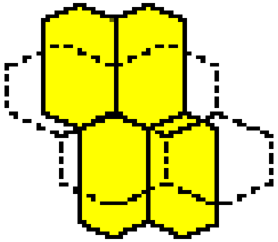
Grain Boundary Sliding

One of the processes to accommodate grain-boundary strain at elevated temperature is grain-boundary migration.





Voids are created during diffusion-assisted flow.



Grain boundary sliding occurs to keep material from falling apart

GB sliding occurs in conjunction with the NH & Coble creep mechanisms

GB sliding is thought to be the mechanisms that allows a material to extend in length with no net change in grain size during superplastic forming operations

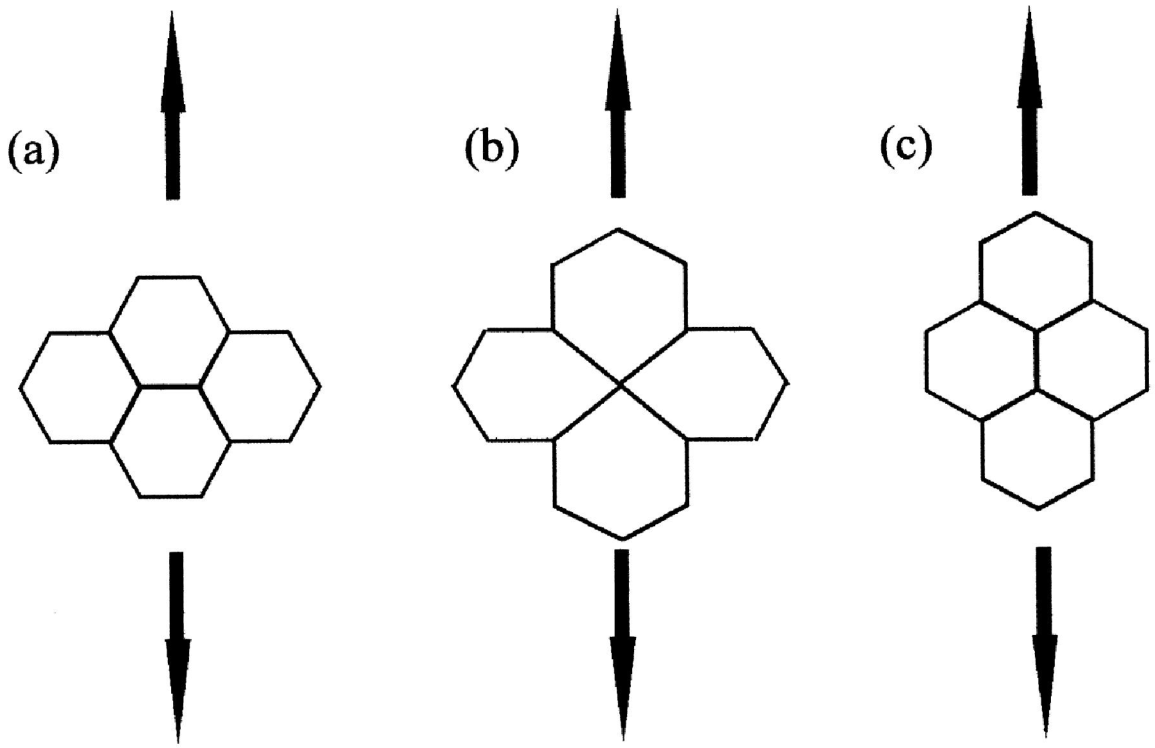
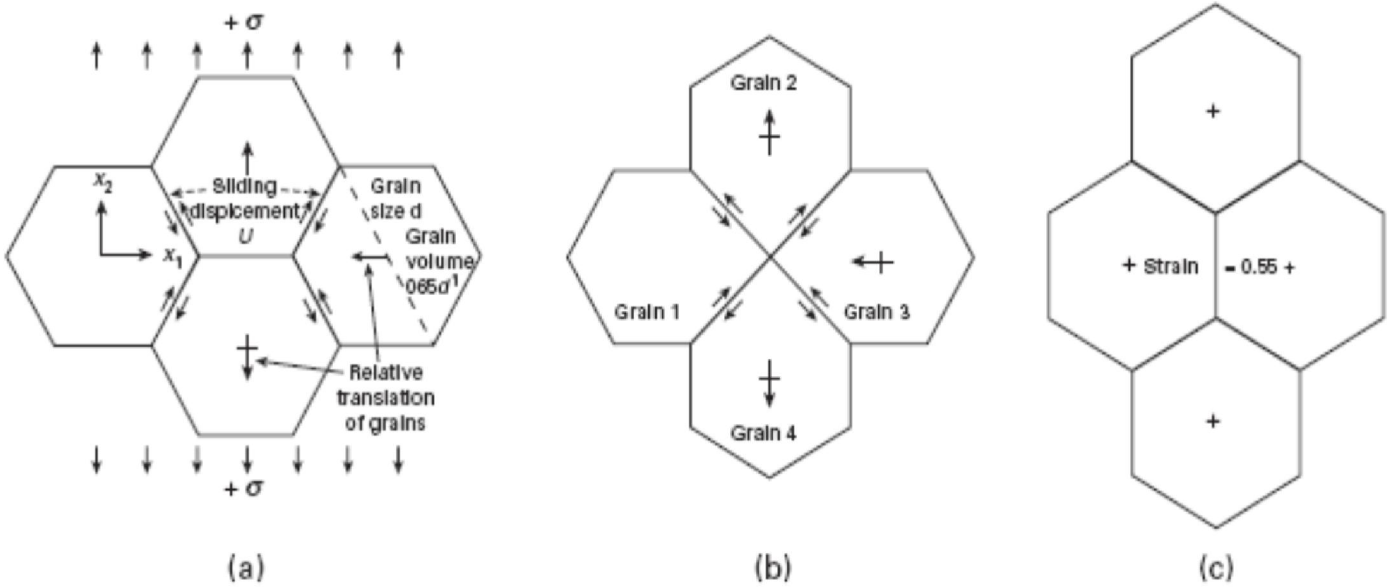


Figure 58. Ashby–Verral model of GBS accommodated by diffusional flow [436].

دانه ها نباید کروی باشند و حتما باید گوشه ای باشند

Ashby-Verrall’s Model



Grain-boundary sliding assisted by diffusion in Ashby–Verrall’s model.

(Reprinted with permission from M. F. Ashby and R. A. Verrall, Acta Met., 21 (1973) 149.)

$$\dot{\varepsilon}_{\mathrm{ss}} = \mathrm{K}_2(b/g)^{p'} D(\sigma/E)^2$$

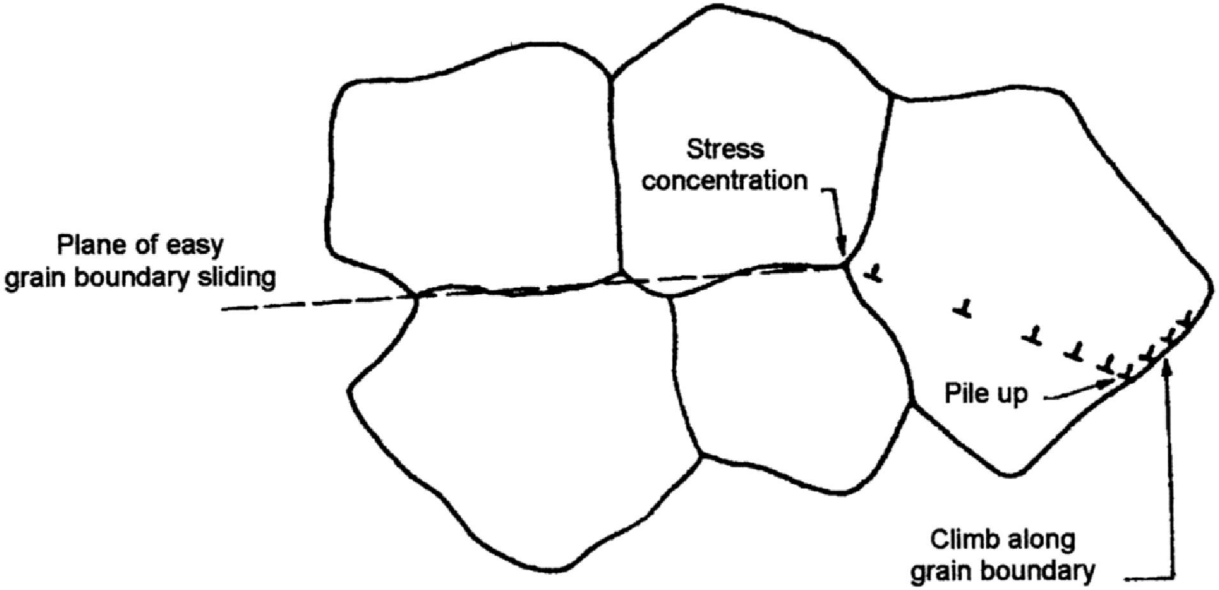


Figure 59. Ball–Hutchinson model of GBS accommodated by dislocation movement [443].

$$\dot{\varepsilon}_{\mathrm{ss}} = \mathrm{K}_3(b/g)^2 D_{\mathrm{sd}}(\sigma/E)^2$$

$$\dot{\varepsilon}_{\mathrm{ss}} = \mathrm{K}_4(b/g)^3 D_{\mathrm{gb}}(\sigma/E)^2$$

لغزش مرزدانه‌ای یکی از مکانیزم‌های تغییرشکل مواد در اثر حرکت دانه‌ها نسبت به یکدیگر در دمای همولوگ بالا و سرعت کرنش پایین است. این مکانیزم نقش عمده‌ای در خزش و سوپرپلاستیسیتهٔ ریز-ساختار دارد.

شبکهٔ مرزدانه‌ای در فرآیندهایی از قبیل مهاجرت مرزدانه، مهاجرت فصل مشترک سه‌تایی، چرخش دانه‌ها و فعالیت نابجایی در خود مرزدانه یا به کمک نابجایی‌های شبکه به لغزش مرزدانه نیاز دارد. مکانیزم اصلی لغزش مرزدانه‌ای حرکت نابجایی‌ها در اثر لغزش و صعود آنهاست.

Superplastic flow

The *superplastic flow* is given by

$$\dot{\varepsilon} = 10^{8}\left(\frac{\sigma}{E}\right)^{2}\frac{bD_{gb}}{\overline{L}^{3}}$$

Eq.3 For grain boundary diffusion

$$\dot{\varepsilon} = 2\times10^{9}\left(\frac{\sigma}{E}\right)^{2}\frac{D_{o}}{\overline{L}^{2}}$$

Eq.3 For lattice self-diffusion

Where $\overline{L}$ is the mean linear intercept measure of grain size.
in this case ***n*** = 2, → ***m*** = 0.5

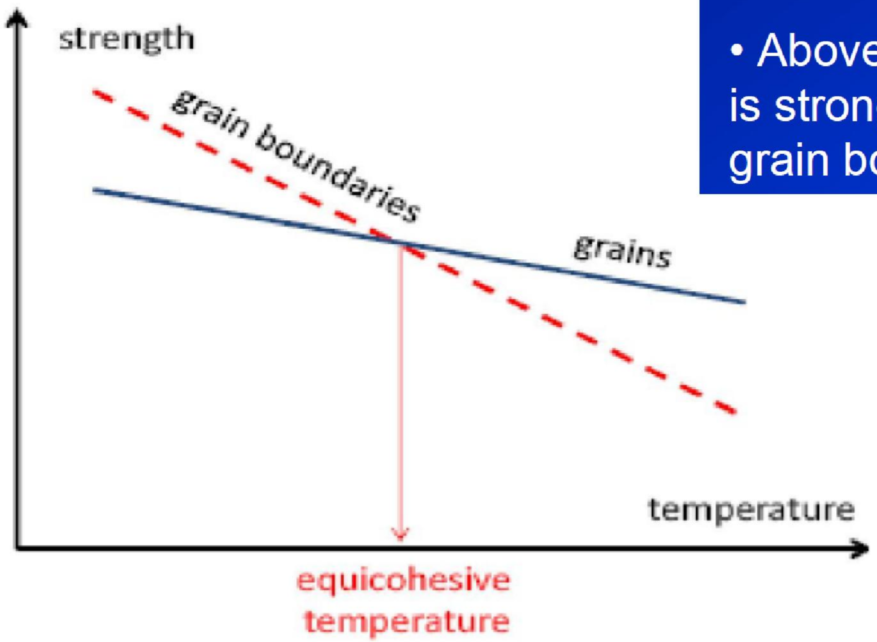
The predominant mechanism for superplasticity deformation is ***grain-boundary sliding*** accommodated by slip.

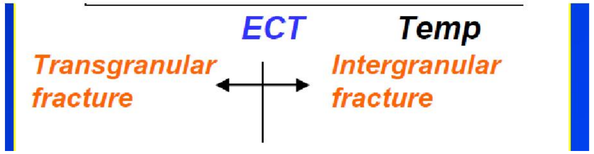
در دما های پایین مرز های دانه مستحکم اند . با افزایش دما ، مرز های دانه شروع به ذوب شدن کرده و حالت نرم پیدا می کنند . به دلیل وجود تنش برشی اعمالی بر روی جسم و نرم شدن مرز های دانه ، مرز های دانه در کنار هم می لغزند و موجب تغییر شکل در ساختار کریستالی ماده می گردند.

Equicohesive temperature

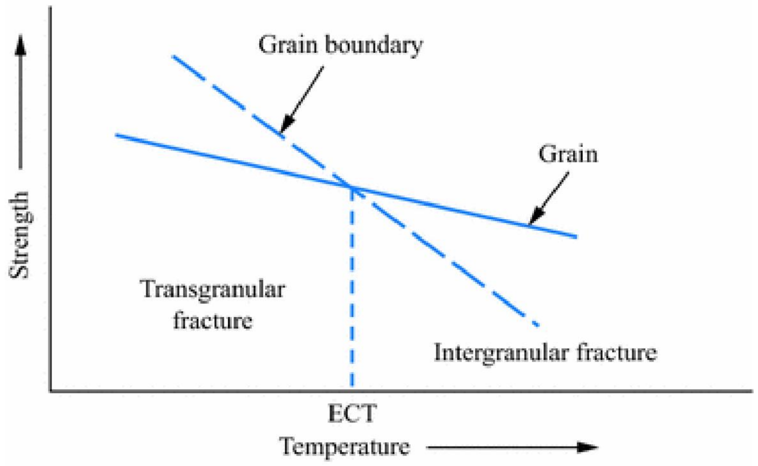
- *Strength of GB = grain at the equicohesive temperature (ECT).*

- Below ***ECT*** small grain sized material is stronger due to high density of grain boundaries to improve strength.
- Above ***ECT*** large grain sized material is stronger due to less tendency for grain boundary sliding.

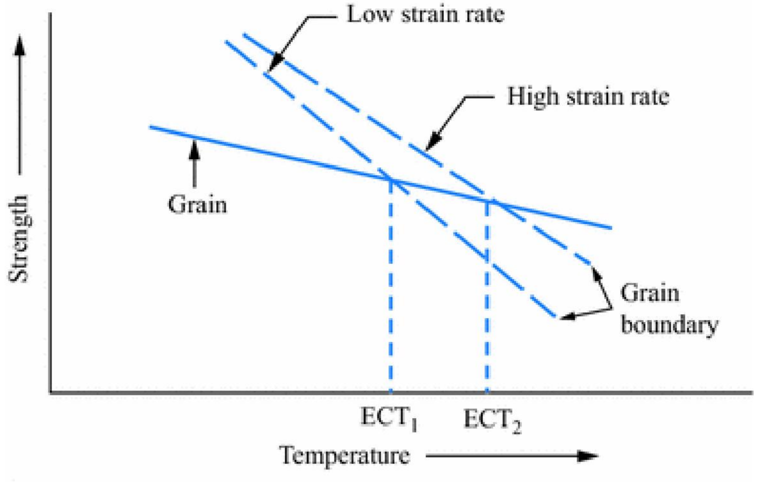


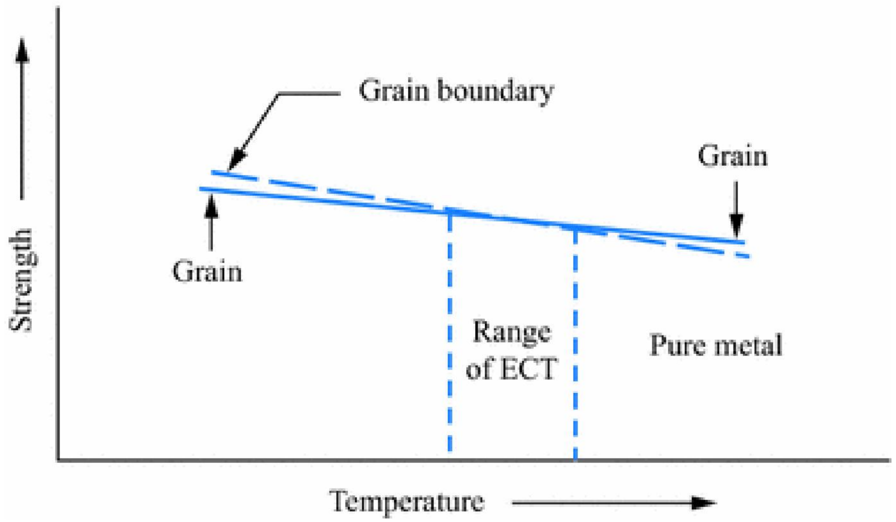


(a)



(b)





Strain rate ⬇ **ECT** ⬇

Increasing the tendency for intergranular failure

Chapter 2

Five-Power-Law Creep

$0.5–0.6\ T_m$,

2.1.1 *Activation Energy and Stress Exponents*

In pure metals and Class M alloys (similar creep behavior similar to pure metals), there is an established, largely phenomenological, relationship between the steady-state strain-rate, $\dot{\varepsilon}_{ss}$, (or creep rate) and stress, σ_{ss}, for steady-state 5-power-law (PL) creep:

$$\dot{\varepsilon}_{ss} = A_0 \exp[-Q_c/kT](\sigma_{ss}/E)^n \quad (3)$$

where A_0 is a constant, k is Boltzmann's constant, and E is Young's modulus (although, as will be discussed subsequently, the shear modulus, G, can also be used). This is consistent with Norton's Law [34]. The activation energy for creep, Q_c, has been found to often be about that of lattice self-diffusion, Q_{sd}. The exponent n is constant and is about 5 over a relatively wide range of temperatures and strain-rates (hence "five-power-law" behavior) power-law-breakdown (PLB) occurs, and n increases

the temperature decreases below roughly 0.5–0.6 Tm

Between this two laws:

$$\dot{\varepsilon}_{\mathrm{ss}} = \mathrm{A}_1 \exp[-Q_{\mathrm{c}}/\mathrm{k}T][\sinh \alpha_1(\sigma_{\mathrm{ss}}/E)]^5$$

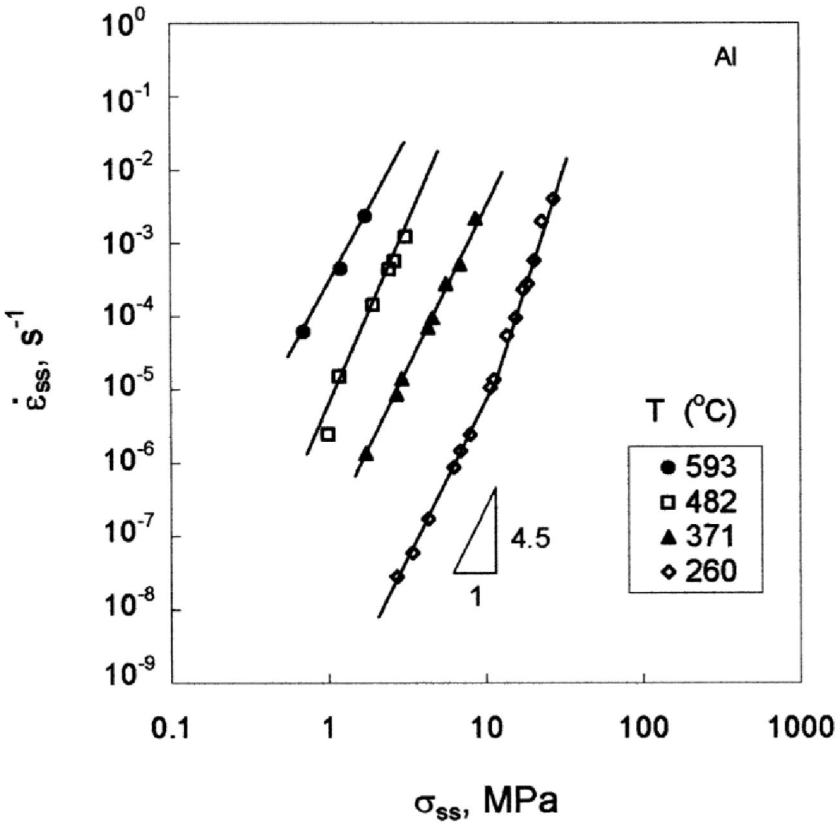


Figure 7. The steady-state stress versus strain-rate for high-purity aluminum at four temperatures, from Ref. [136].

σ_{ss}/E or σ_{ss}/G.

The stress exponent is about 4.5 for aluminum. Although this is not precisely five, it is constant over a range of temperature, stress, and strain-rate, and falls within the range of 4–7 observed in pure metals and class M alloys

$$Q_c = -k\left[\delta(\ln \dot{\varepsilon}_{ss})/\delta(1/T)\right]_{\sigma_{ss}/E,s}$$

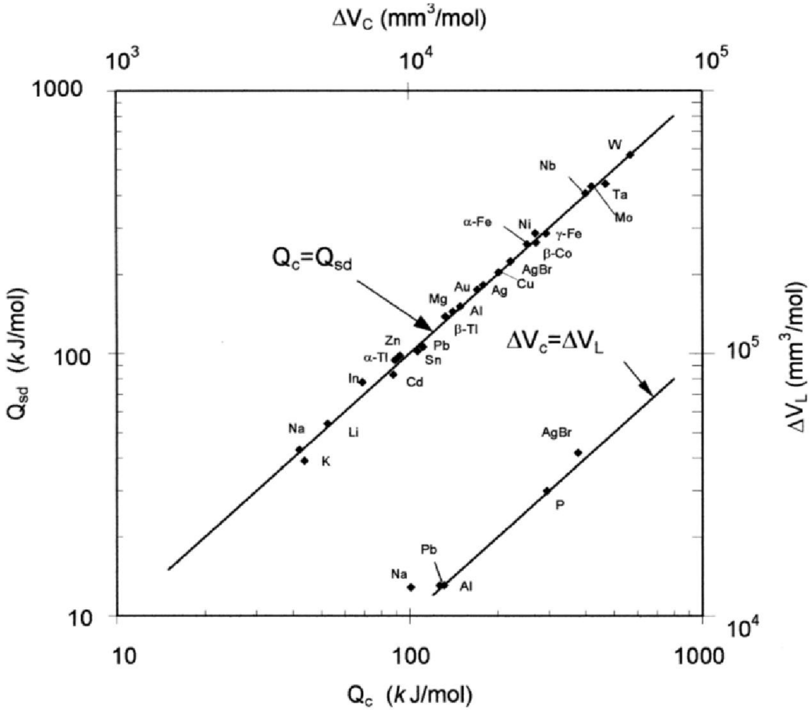
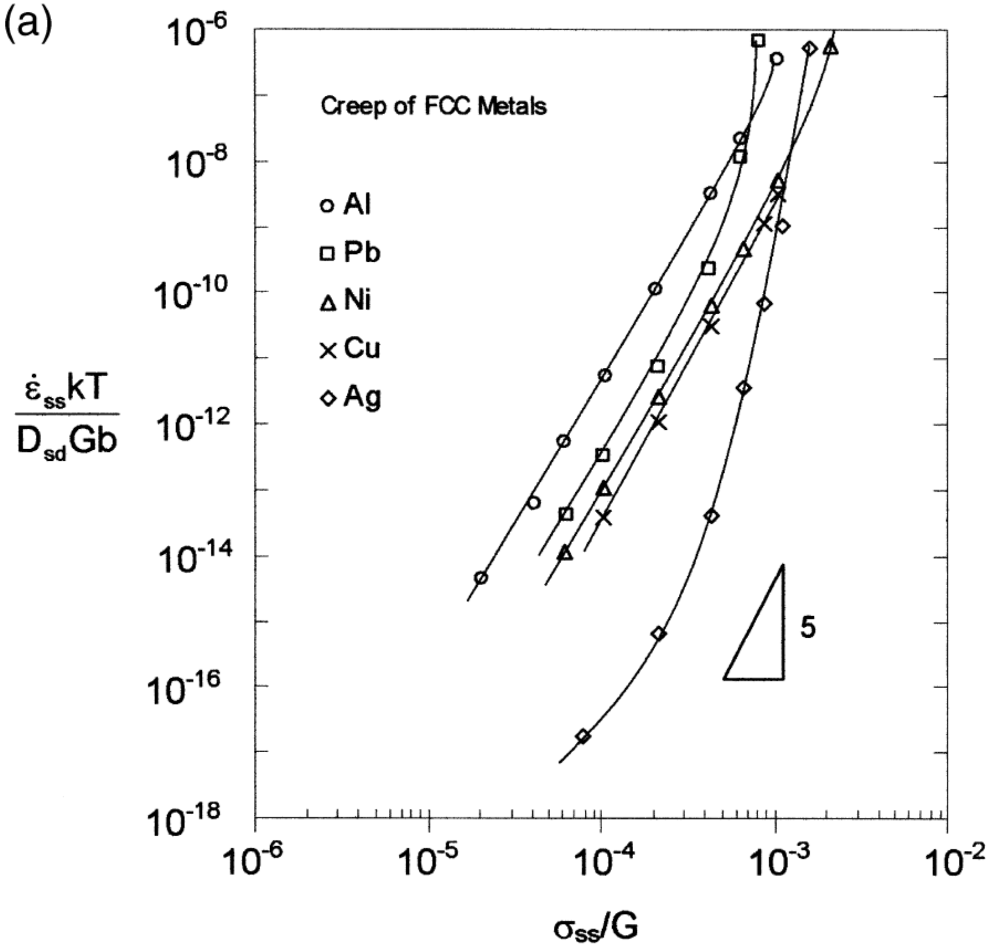


Figure 8. The activation energy and volume for lattice self-diffusion versus the activation energy and volume for creep. Data from Ref. [26].

$$\dot{\varepsilon}_{ss} = A_2 \exp[-Q_{sd}/kT](\sigma_{ss})^{n(\cong 5)}$$


(a)
Creep of FCC Metals
○ Al
□ Pb
△ Ni
× Cu
◇ Ag
5
$\frac{\dot{\varepsilon}_{ss}kT}{D_{sd}Gb}$
σ_{ss}/G
10^{-6}
10^{-8}
10^{-10}
10^{-12}
10^{-14}
10^{-16}
10^{-18}
10^{-6}
10^{-5}
10^{-4}
10^{-3}
10^{-2}

Stacking Fault Energy

$$\dot{\varepsilon}_{ss} = A_6(\chi/Gb)^3(D_{sd}Gb/kT)(\sigma_{ss}/G)^5$$

where χ is the stacking fault energy.

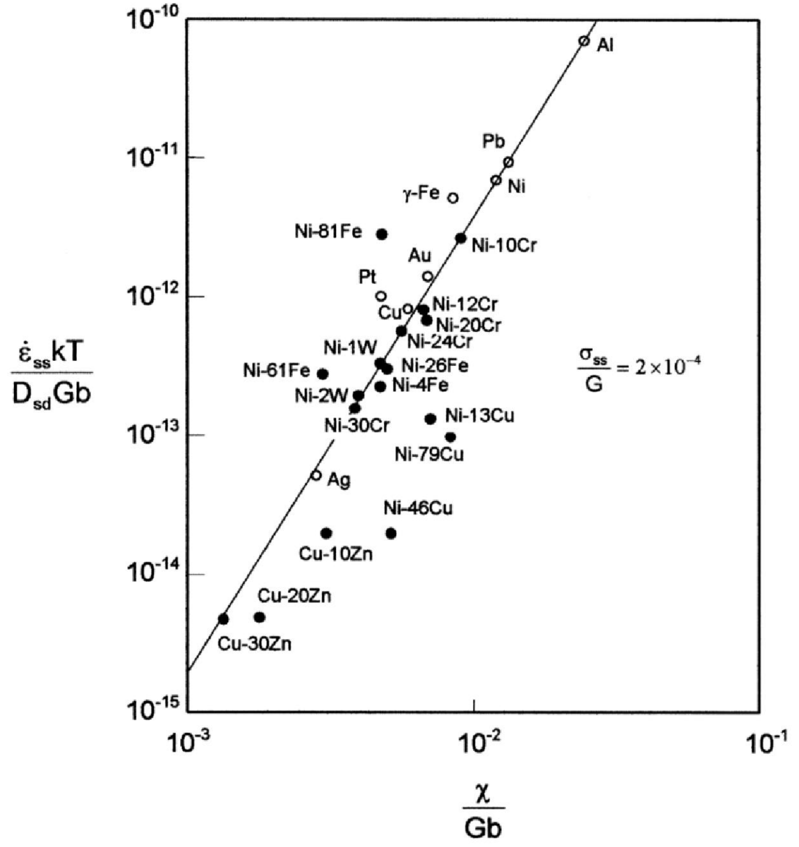


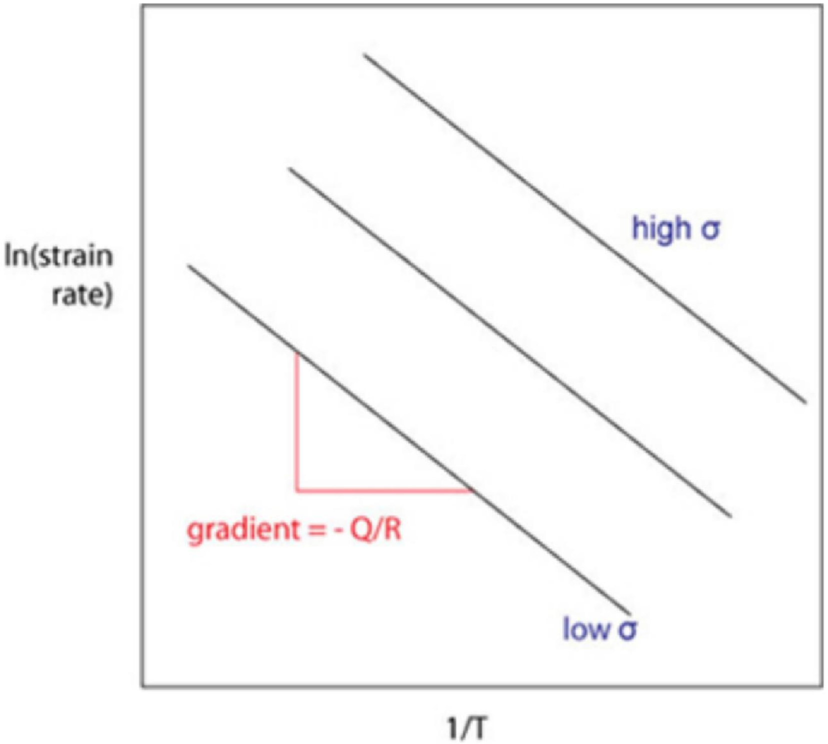
Figure 14. The effect of stacking fault energy on the (compensated) steady-state strain-rate for a variety of metals and Class M alloys based on Ref. [73].

Lower SFE materials display wider stacking faults and have more difficulties for cross-slip.

The width of stacking fault is a consequence of the balance between the repulsive force between two partial dislocations

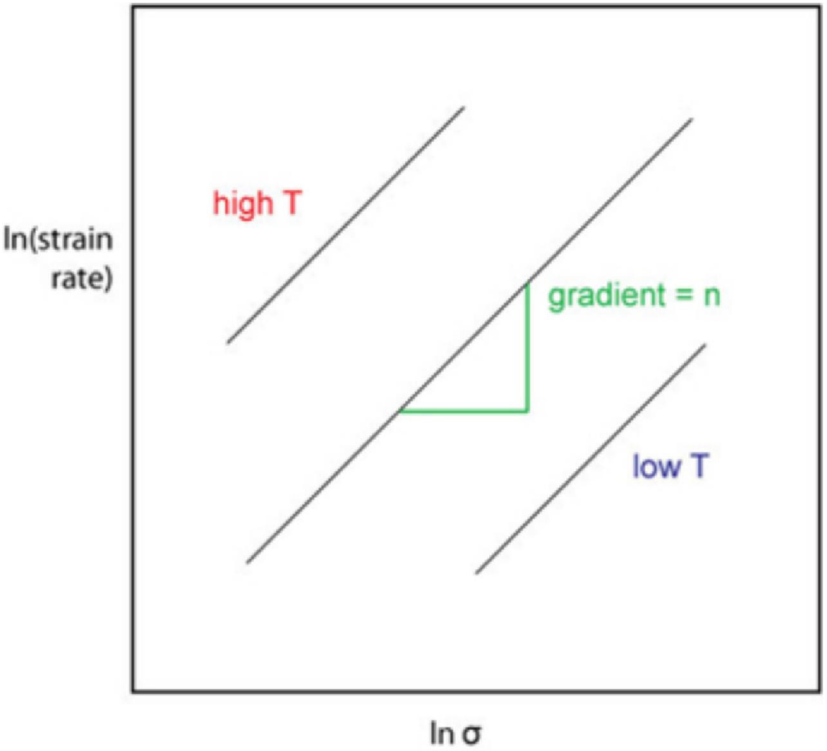
A stacking fault is created by the dissociation of a perfect dislocation into two partial dislocations.

$$\dot{\varepsilon}_{ss} = A_2 \exp[-Q_{sd}/kT](\sigma_{ss})^{n(\cong 5)}$$



The stress exponent *n* can be determined by plotting the strain rate as a function of stress.

$$\dot{\varepsilon}_{ss} = A_2 \exp[-Q_{sd}/kT](\sigma_{ss})^{n(\cong 5)}$$



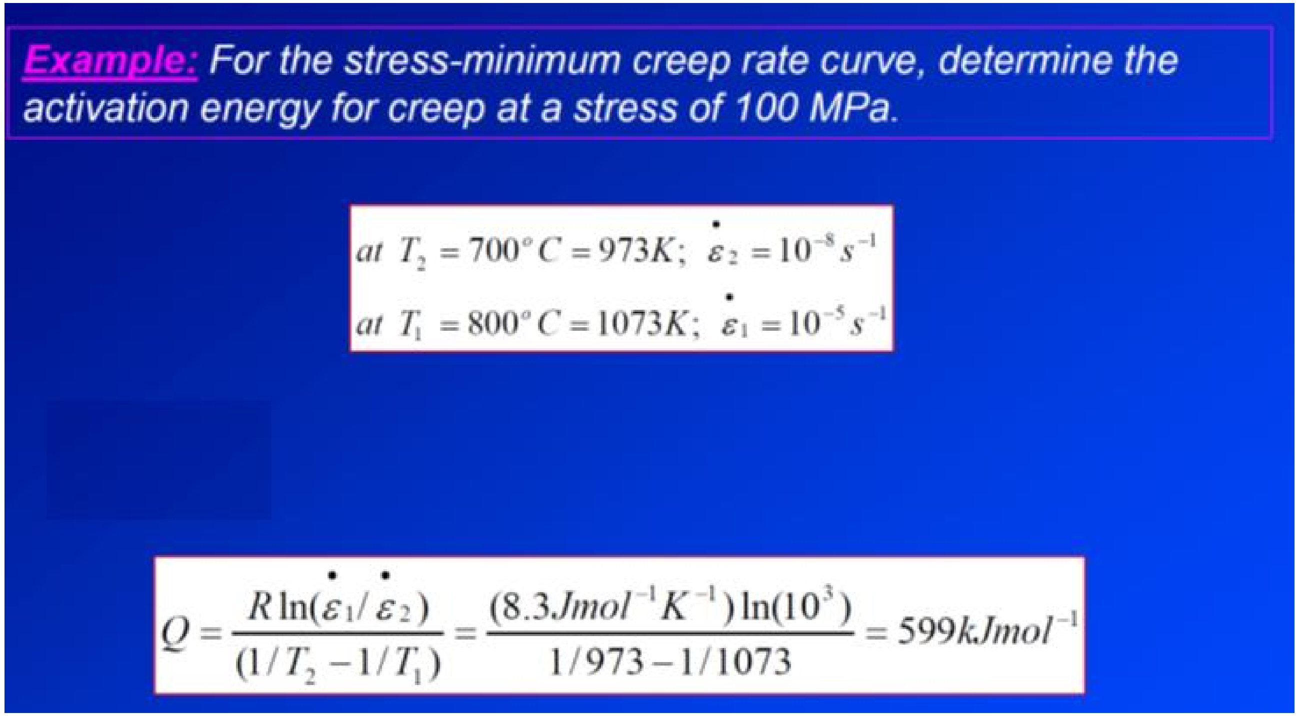
The creeping coil experiment - variable stresses in a single specimen

$$\dot{\varepsilon}_{ss} = A_2 \exp[-Q_{sd}/kT](\sigma_{ss})^{n(\cong 5)}$$

$$Q_c = R\frac{\ln\left(\dot{\varepsilon}_1/\dot{\varepsilon}_2\right)}{\left(\frac{1}{T_2}-\frac{1}{T_1}\right)}$$

- *Correlates with the activation energy for the mechanisms causing creep. We'll address the mechanisms shortly (in great detail).*

$$Q = R\left(\ln\frac{t_2}{t_1}\right)\left(\frac{T_1 \cdot T_2}{T_1 - T_2}\right)$$



Example: For the stress-minimum creep rate curve, determine the activation energy for creep at a stress of 100 MPa.

$$\text{at } T_2 = 700^\circ C = 973K; \ \dot{\varepsilon}_2 = 10^{-8} s^{-1}$$

$$\text{at } T_1 = 800^\circ C = 1073K; \ \dot{\varepsilon}_1 = 10^{-5} s^{-1}$$

$$Q = \frac{R \ln(\dot{\varepsilon}_1 / \dot{\varepsilon}_2)}{(1/T_2 - 1/T_1)} = \frac{(8.3 Jmol^{-1} K^{-1}) \ln(10^3)}{1/973 - 1/1073} = 599 kJmol^{-1}$$

Chapter 5

Three-Power-Law Viscous Glide Creep

Creep of solid solution alloys (designated Class I [16] or class A alloys [338]) at intermediate stresses and under certain combinations of materials parameters, which will be discussed later, can often be described by three regions [36,339,340]. This is illustrated in Figure 52. With increasing stress, the stress exponent, *n*, changes in value from 5 to 3 and again to 5 in regions I, II, and III, respectively. This section will focus on region II, the so-called Three-Power-Law regime.

The mechanism of deformation in region II is viscous glide of dislocations [36]. This is due to the fact that the dislocations interact in several possible ways with the solute atoms, and their movement is impeded [343]. There are two competing mechanisms over this stress range, dislocation climb and glide, and glide is slower and thus rate controlling.

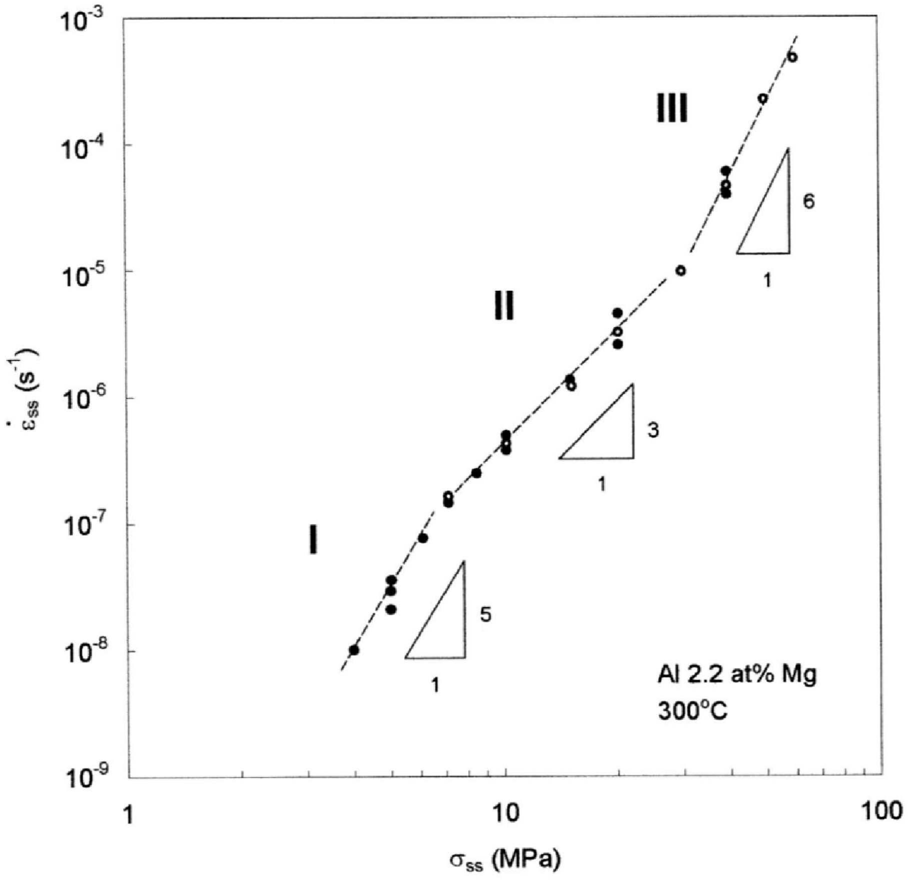


Figure 52. steady-state creep rate vs. applied stress for an Al-2.2 at%Mg alloy at 300°C. Three different creep regimes, I, II, and III, are evident. Based on Refs. [341,342].

Three-Power-Law mechanism

$\dot{\varepsilon} = 1/2\, \bar{v}\, b\, \rho_m$

$\bar{v}$ is proportional to σ

ρ_m is proportional to σ^2

$$\dot{\varepsilon}_{ss} \cong \frac{0.35}{A} G \left(\frac{\sigma}{G}\right)^3$$

Cottrell–Jaswon eq.

$$\boxed{\dot{\varepsilon}_{ss} \cong \frac{\pi(1-\nu)kT\tilde{D}}{6e^2 C b^5 G}\left(\frac{\sigma}{G}\right)^3}$$

where e is the solute–solvent size difference, C is the concentration of solute atoms and $\tilde{D}$ is the diffusion coefficient for the solute atoms, calculated using Darken's analysis.

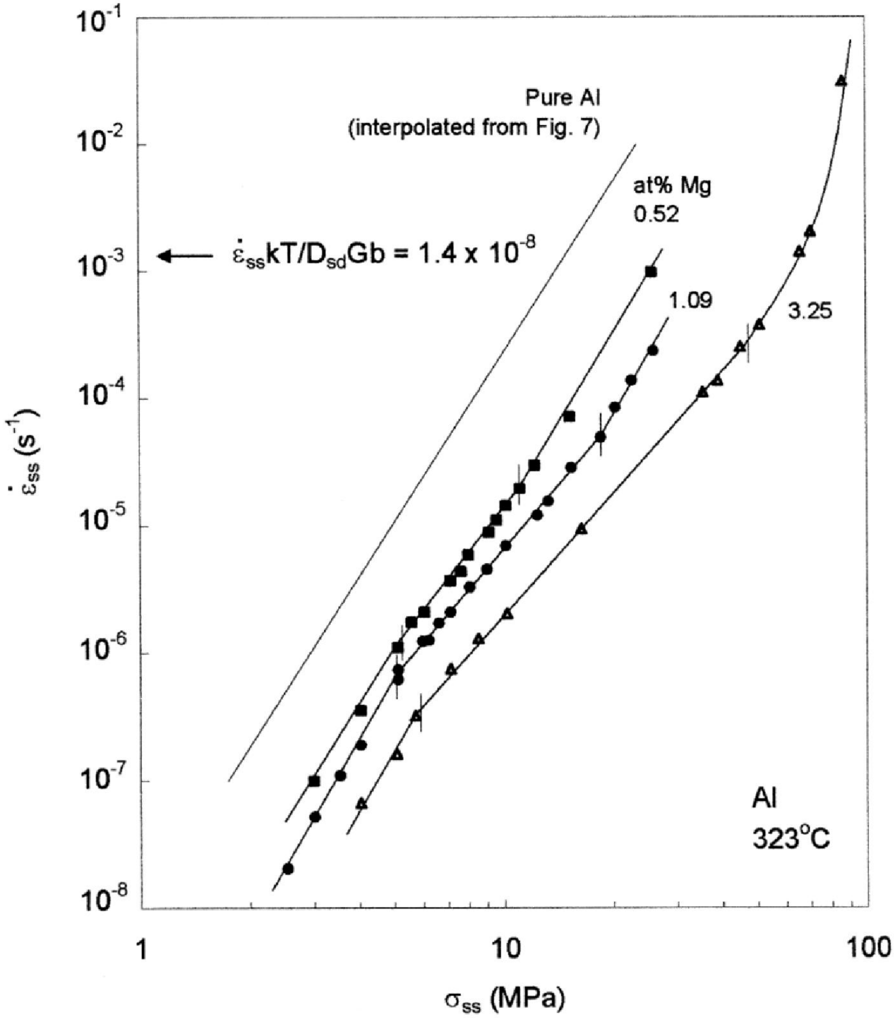


Figure 53. steady-state creep-rate vs. applied stress for three Al–Mg alloys (Al-0.52 at.%Mg, ν; Al-1.09 at.%Mg, λAl-3.25 at.%Mg, σ) at 323°C [356].

Solute drag creep

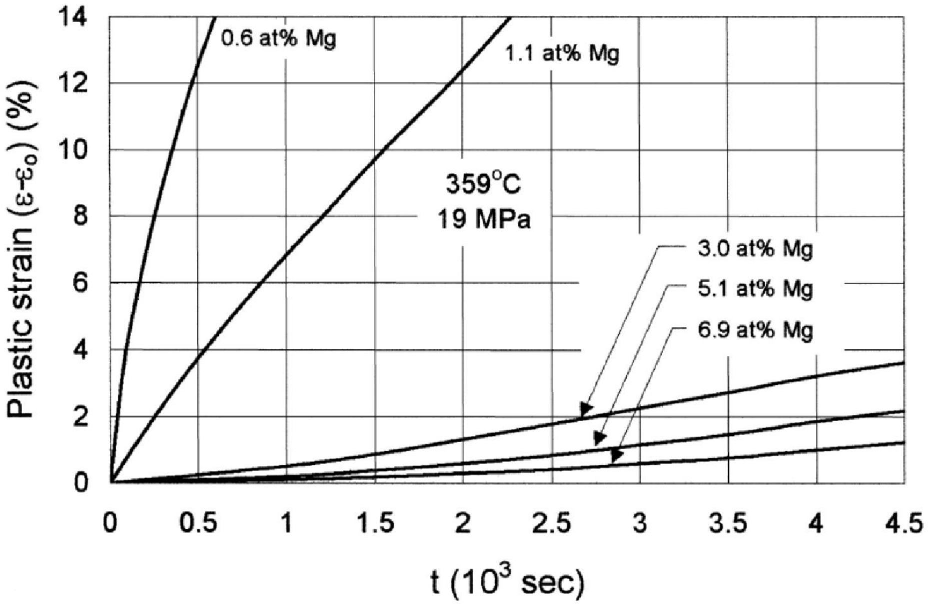


Figure 54. Creep behavior of several aluminum alloys with different magnesium concentrations: 0.6 at.% and 1.1 at.% (class II (M)) and 3.0 at.%, 5.1 at.%, and 6.9 at.% (class I (A)). The tests were performed at 359°C and at a constant stress of 19 MPa [349].

Chapter 8

Creep Behavior of Particle-Strengthened Alloys

The creep behavior of second phase in the matrix:

1- Shape of precipitates

2- The coherency

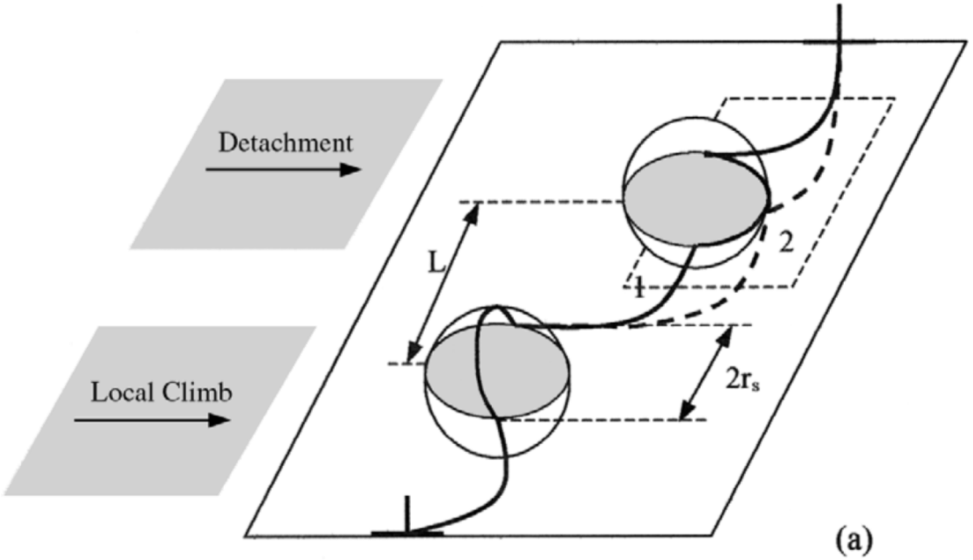
3- Volume content

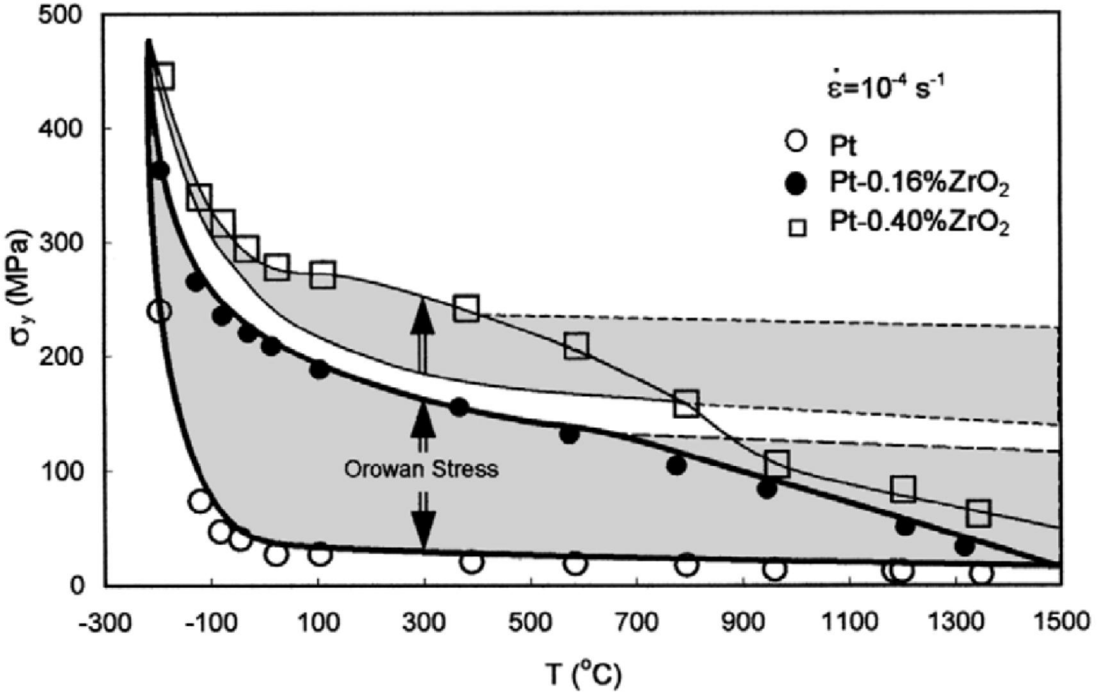
4- Particle size
5- Distribution of particles

It is well known that second-phase particles provide enhanced strength at lower temperatures

Friedel cutting or Orowan by passing

The Orowan stress is determined by the bypass stress based on an Orowan loop mechanism
Strengthening from Coherent particles can occur in a variety of ways that usually involves particle cutting





500
400
300
200
100
0
σy (MPa)
-300
-100
100
300
500
700
900
1100
1300
1500
T (°C)
ε̇=10⁻⁴ s⁻¹
Pt
Pt-0.16%ZrO2
Pt-0.40%ZrO2
Orowan Stress

(b)

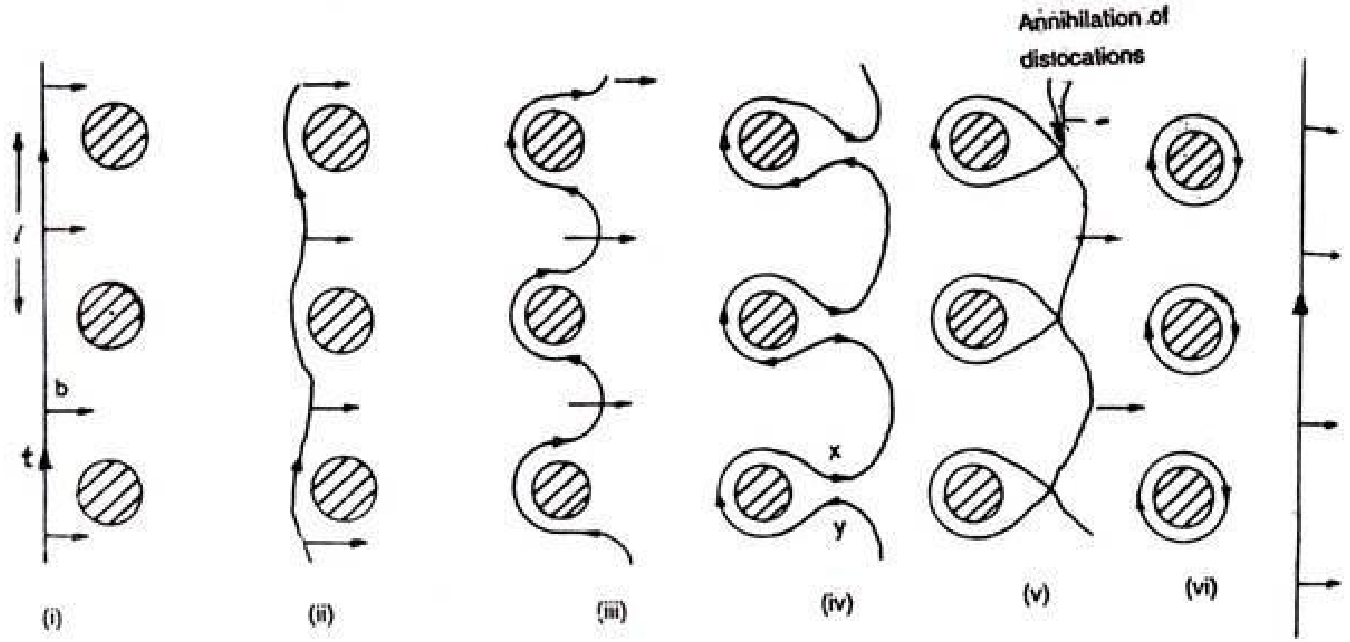
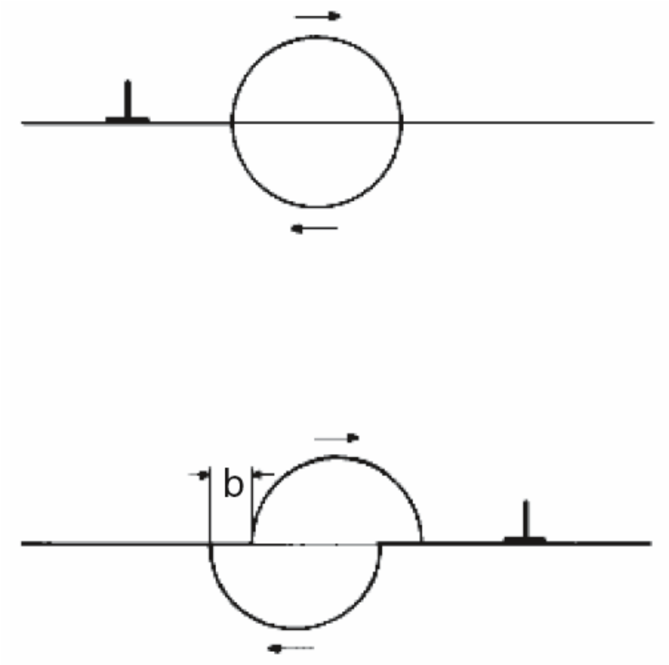
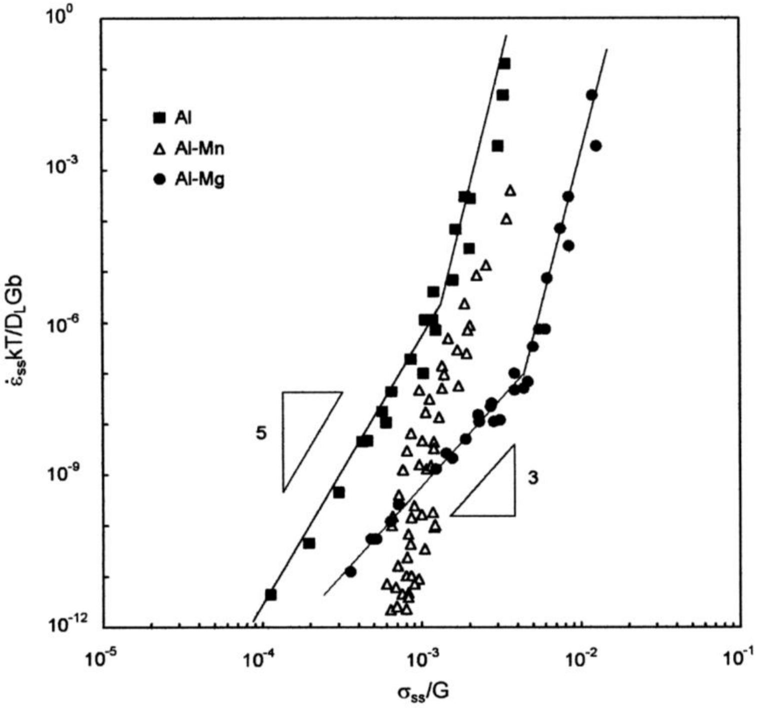


Fig. 13.10. Interaction of a dislocation with a row of widely spaced precipitates to bypass them

Friedel cutting: Dislocation is passing through particle and shifts its upper part with respect to the lower part. Extra surface energy has to be generated which limits this process to small particles.

Friedel cutting is possible for small and coherent particles only due to the extra surface energy which has to be raised when cutting the particle.



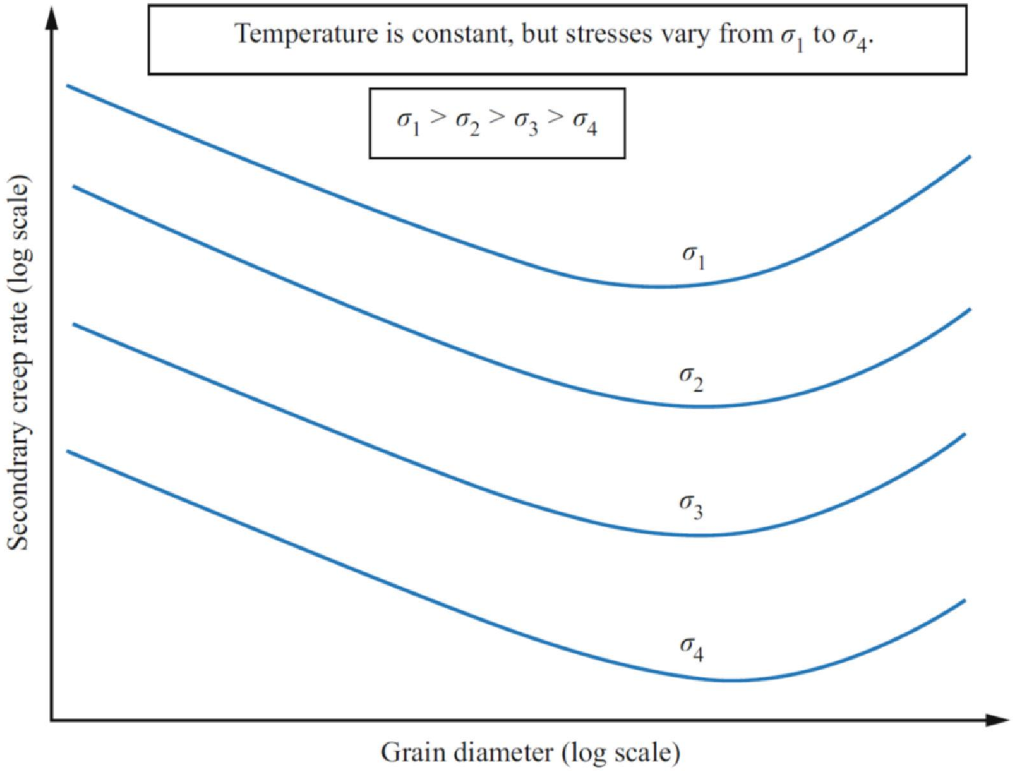


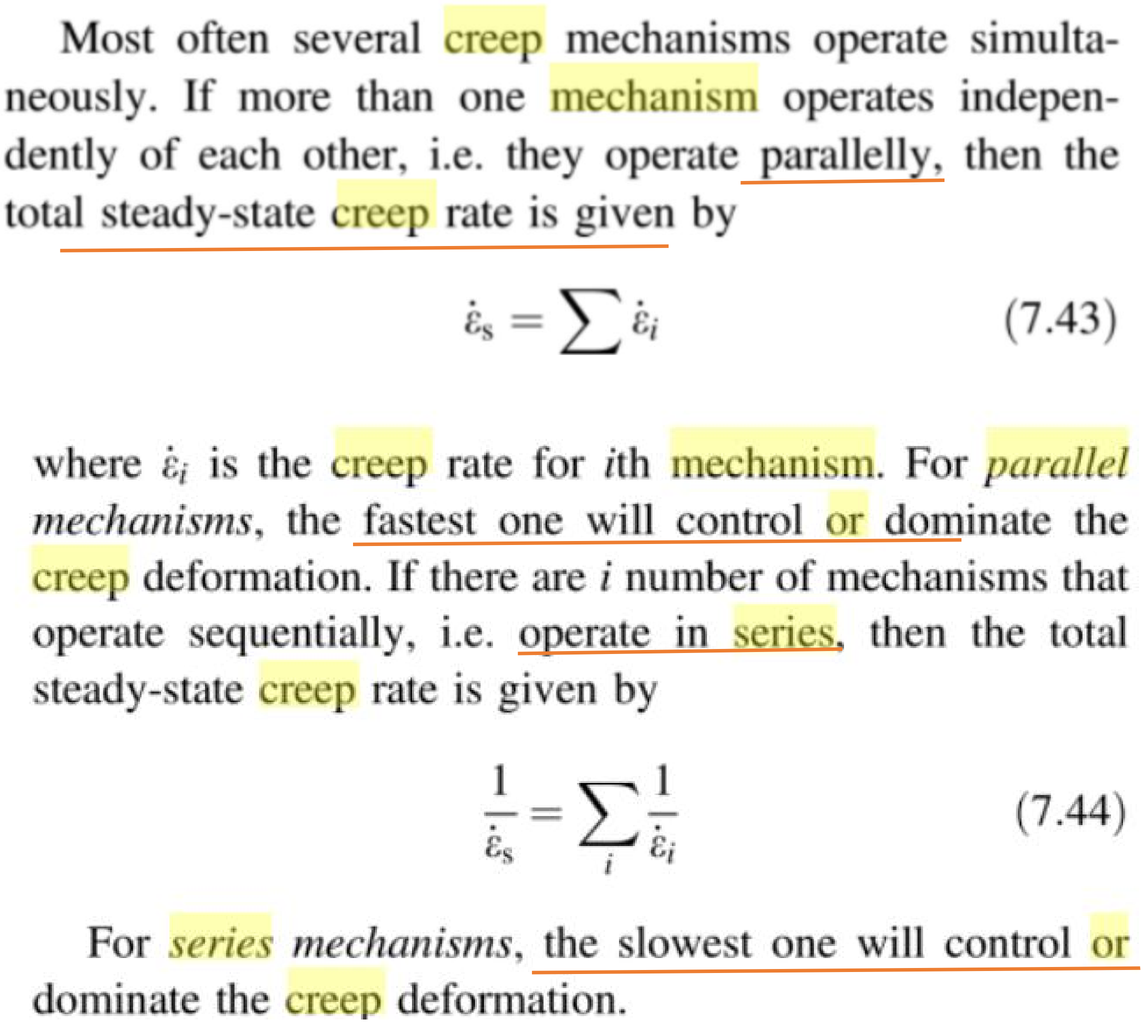
9. Steady-state relation between strain-rate $\dot{\varepsilon}$ and flow stress for the alloys of this work compared to literature data from slow tests (Al, Al–Mg, Al–Mn). Adapted from Ref. [556].

	Mechanism	D	n	A
	Climb of edge dislocations	D_L	5	$6x10^7$
	(Pure Metals and class-M alloys)	(n function of Xal structure & Γ)*		
PLB	Low-temperature climb	$D_\perp$	7	$2x10^8$
	Viscous glide (Class-I alloys - microcreep)	D_s	3	6
	Nabarro-Herring	D_L	1	$14 (\frac{b}{d})^2$
	Coble	D_b	1	$100 (\frac{b}{d})^3$
	Harper-Dorn	D_L	1	$3x10^{-10}$
	GBS (superplasticity)	D_b	2	$200 (\frac{b}{d})^2$

D_L = lattice diffusivity; D_s = solute diffusivity; $D_\perp$ = core diffusivity;
D_b = Grain-Boundary Diffusivity; b = Burgers vector; d = grain size;
δ = subgrain size = $10 \frac{Gb}{\tau}$ and $\rho = \frac{\sigma^2}{G^2 b^2}$ where G is the shear modulus

*n increases with decreasing Γ (stacking-fault energy)


Temperature is constant, but stresses vary from σ_1 to σ_4.
$\sigma_1 > \sigma_2 > \sigma_3 > \sigma_4$
σ_1
σ_2
σ_3
σ_4
Secondrary creep rate (log scale)
Grain diameter (log scale)



Most often several creep mechanisms operate simultaneously. If more than one mechanism operates independently of each other, i.e. they operate parallelly, then the total steady-state creep rate is given by

$$\dot{\varepsilon}_s = \sum \dot{\varepsilon}_i \qquad (7.43)$$

where $\dot{\varepsilon}_i$ is the creep rate for ith mechanism. For *parallel mechanisms*, the fastest one will control or dominate the creep deformation. If there are i number of mechanisms that operate sequentially, i.e. operate in series, then the total steady-state creep rate is given by

$$\frac{1}{\dot{\varepsilon}_s} = \sum_i \frac{1}{\dot{\varepsilon}_i} \qquad (7.44)$$

For *series mechanisms*, the slowest one will control or dominate the creep deformation.

t total= t1=t2=t3

ϵ total= ϵ 1+ ϵ 2+ ϵ3

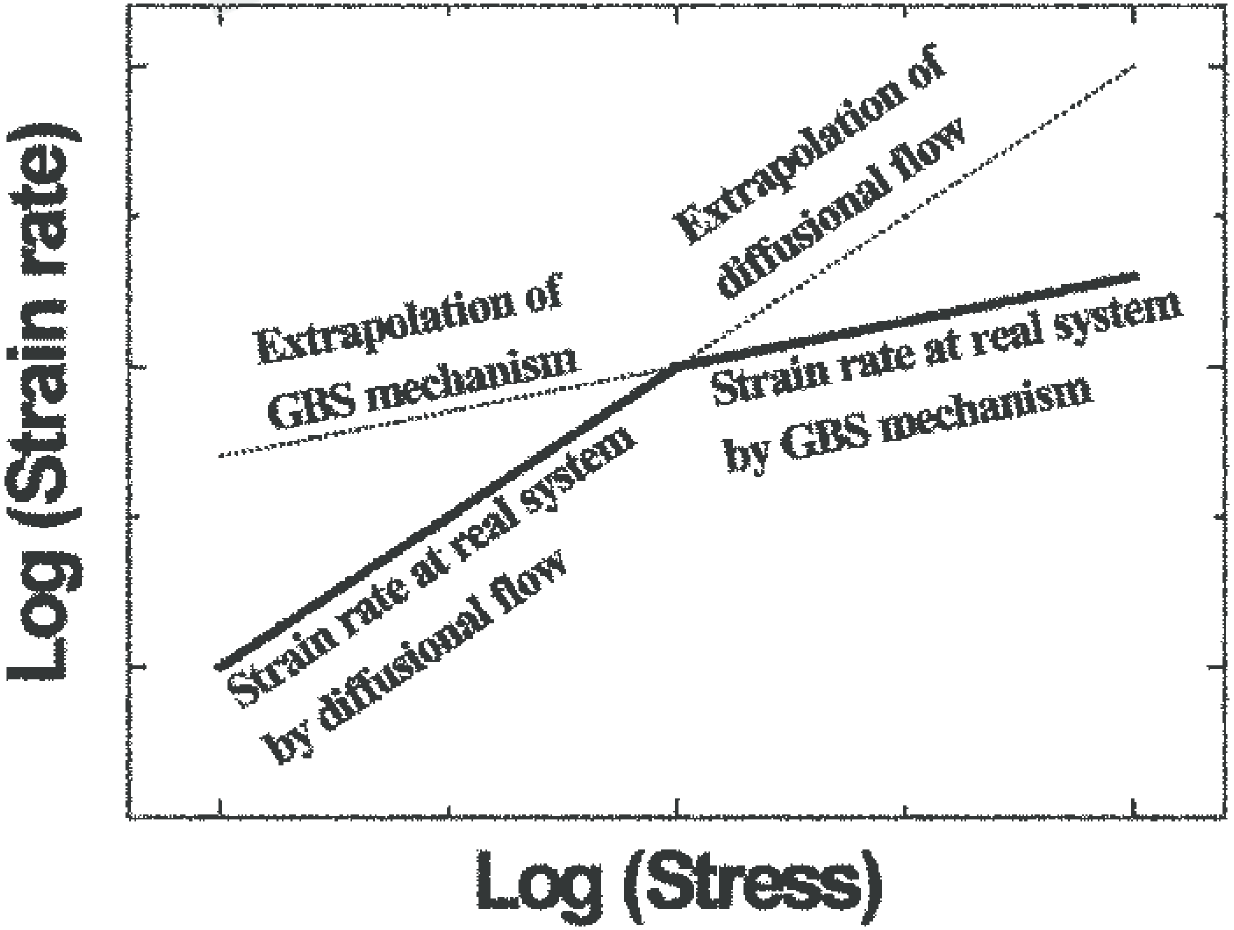
- Many different mechanisms may contribute and the total strain-rate :

parallel mechanism
(fastest controls / dominates)

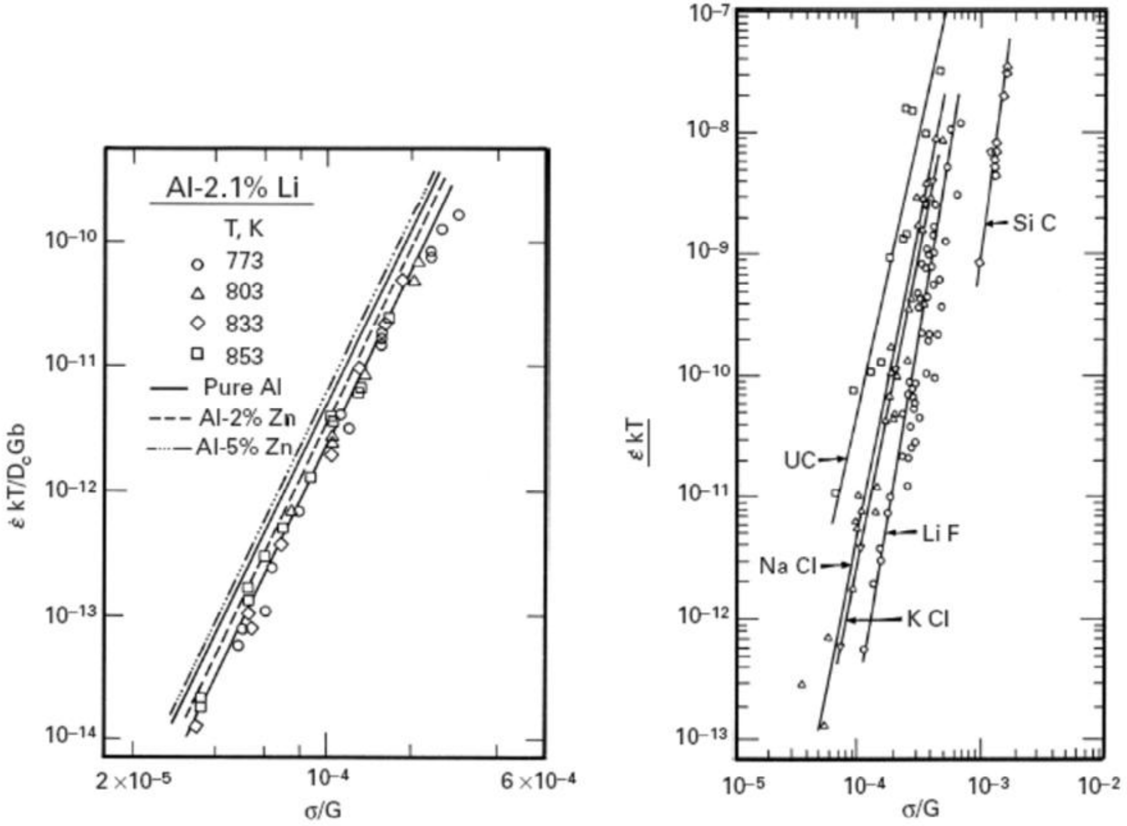
$$\dot{\varepsilon} = \sum_{\mathrm{i}} \dot{\varepsilon}_{\mathrm{i}}$$

series mechanisms
(slower controls / dominates)

$$\dot{\varepsilon} = \sum \left(\frac{1}{\dot{\varepsilon}_i} \right)^{-1}$$

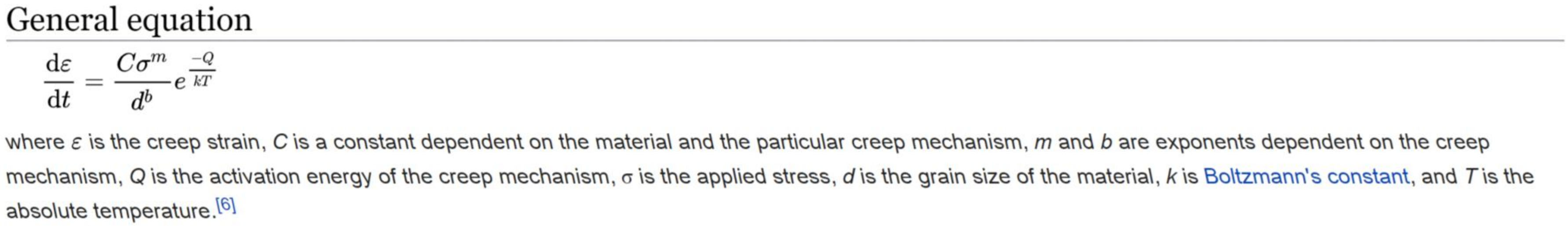

Extrapolation of diffusional flow
Extrapolation of GRS mechanism
Strain rate at real system by GBS mechanism
Strain rate at real system by diffusional flow
Log (Strain rate)
Log (Stress)

Mukherjee-Bird-Dorn Equation



(a) (b)

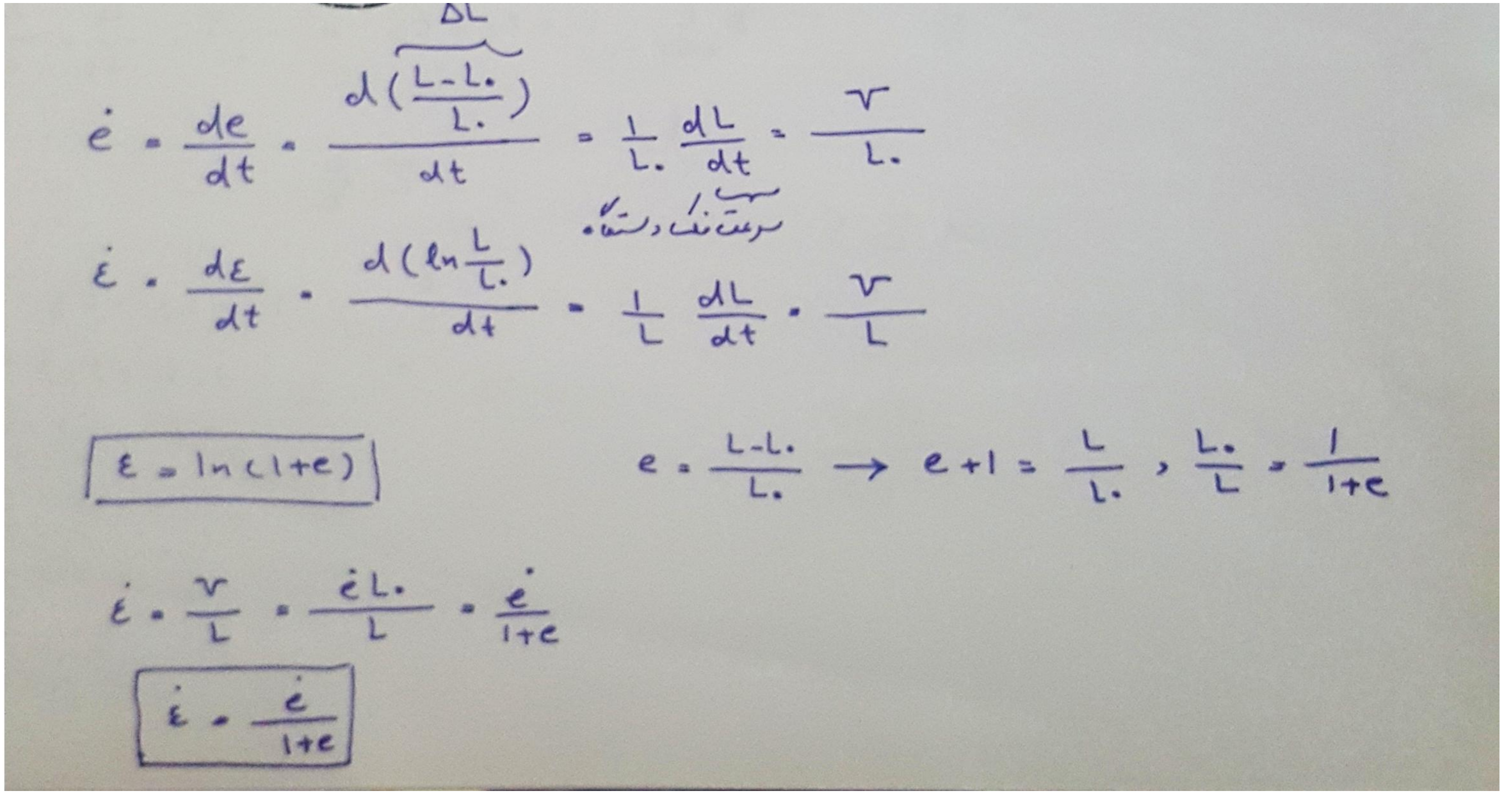
$$\dot{\varepsilon}_s = \frac{AGb}{kT} D_0 \exp(-Q_c/RT) \left(\frac{b}{d}\right)^p \left(\frac{\sigma}{G}\right)^n$$



General equation

$$\frac{\mathrm{d}\varepsilon}{\mathrm{d}t} = \frac{C\sigma^m}{d^b} e^{\frac{-Q}{kT}}$$

where ε is the creep strain, C is a constant dependent on the material and the particular creep mechanism, m and b are exponents dependent on the creep mechanism, Q is the activation energy of the creep mechanism, σ is the applied stress, d is the grain size of the material, k is Boltzmann's constant, and T is the absolute temperature.[6]



$$\dot{e} = \frac{de}{dt} = \frac{d\left(\overbrace{\frac{L-L_0}{L_0}}^{\Delta L}\right)}{dt} = \frac{1}{L_0}\underbrace{\frac{dL}{dt}}_{\text{سرعت فک دستگاه}} = \frac{v}{L_0}$$

$$\dot{\varepsilon} = \frac{d\varepsilon}{dt} = \frac{d\left(\ln\frac{L}{L_0}\right)}{dt} = \frac{1}{L}\frac{dL}{dt} = \frac{v}{L}$$

$$\boxed{\varepsilon = \ln(1+e)}$$

$$e = \frac{L-L_0}{L_0} \rightarrow e+1 = \frac{L}{L_0} , \frac{L_0}{L} = \frac{1}{1+e}$$

$$\dot{\varepsilon} = \frac{v}{L} = \frac{\dot{e}L_0}{L} = \frac{\dot{e}}{1+e}$$

$$\boxed{\dot{\varepsilon} = \frac{\dot{e}}{1+e}}$$

Chapter 10

Creep Fracture

10.1 BACKGROUND

Creep plasticity can lead to tertiary or Stage III creep and failure. It has been suggested that Creep Fracture can occur by *w* or Wedge-type cracking, illustrated in Figure 101(a), at grain-boundary triple points. Some have suggested that *w*-type cracks form most easily at higher stresses (lower temperatures) and larger grain sizes [786] when grain-boundary sliding is not accommodated. Some have suggested that the Wedge-type cracks nucleate as a consequence of grain-boundary sliding. Another mode of fracture has been associated with *r*-type irregularities or cavities illustrated in Figure 102. The Wedges may be brittle in origin or simply an accumulation of *r*-type voids [Figure 101(b)] [787]. These Wedge cracks may propagate only by *r*-type void formation [788,789]. Inasmuch as *w*-type cracks are related to *r*-type voids, it is sensible to devote this short summary of Creep Fracture to cavitation.

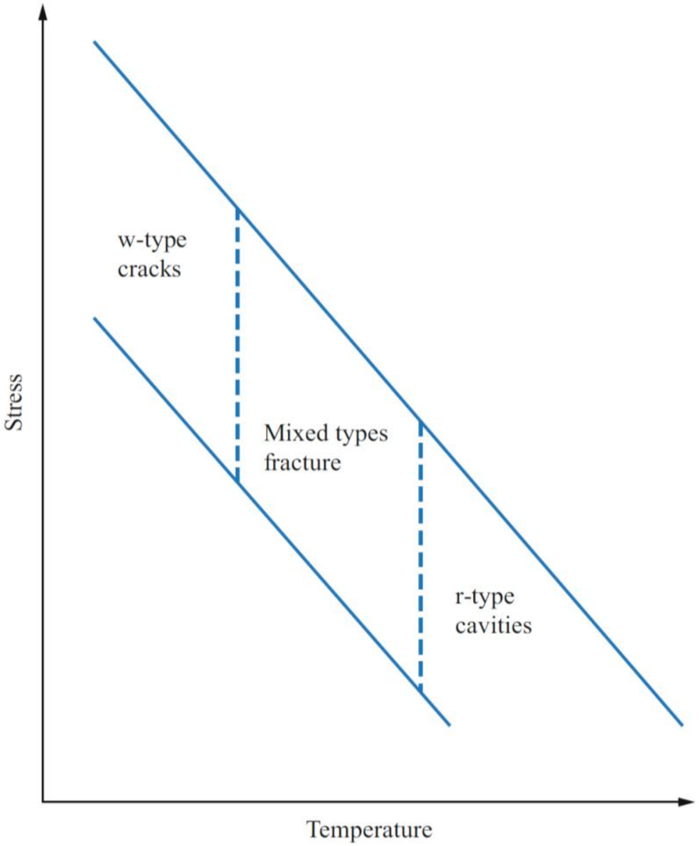
عوامل موثر بر نوع شکست

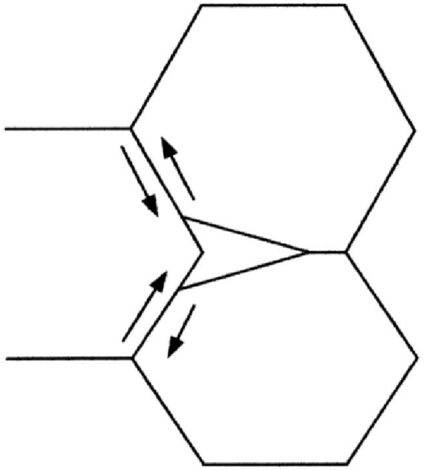
۱– دما

۲– تنش

۳– کار سرد (مرزدانه های سه تایی)

۴– تبلور مجدد



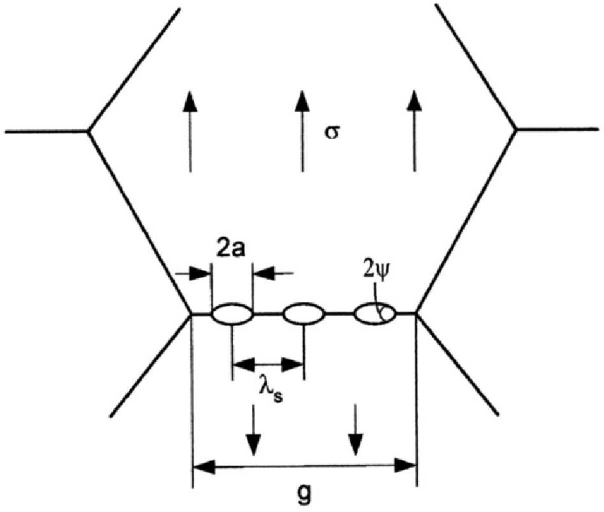


w or Wedge-type cracking

$$\sigma r_0 = 2\gamma_s, \quad \text{or}, \quad \sigma = \frac{2\gamma_s}{r_0}$$

where

γ_s the surface energy per unit area, and
r_0 the interatomic distance



Cavitation (r-type) or voids

$$\sigma_{\max} = \sqrt{\frac{L}{2\rho_t}} \cdot \tau \quad \text{for } L \gg \rho_t$$

where

τ shear stress along grain boundary, say mn,
L length of the sliding boundary and
ρ_t radius of curvature at the tip of the boundary

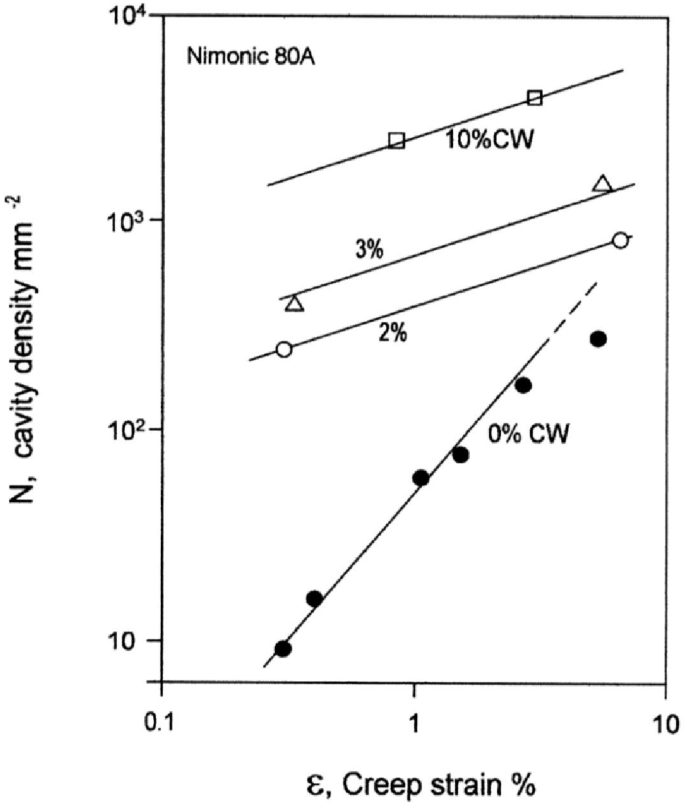
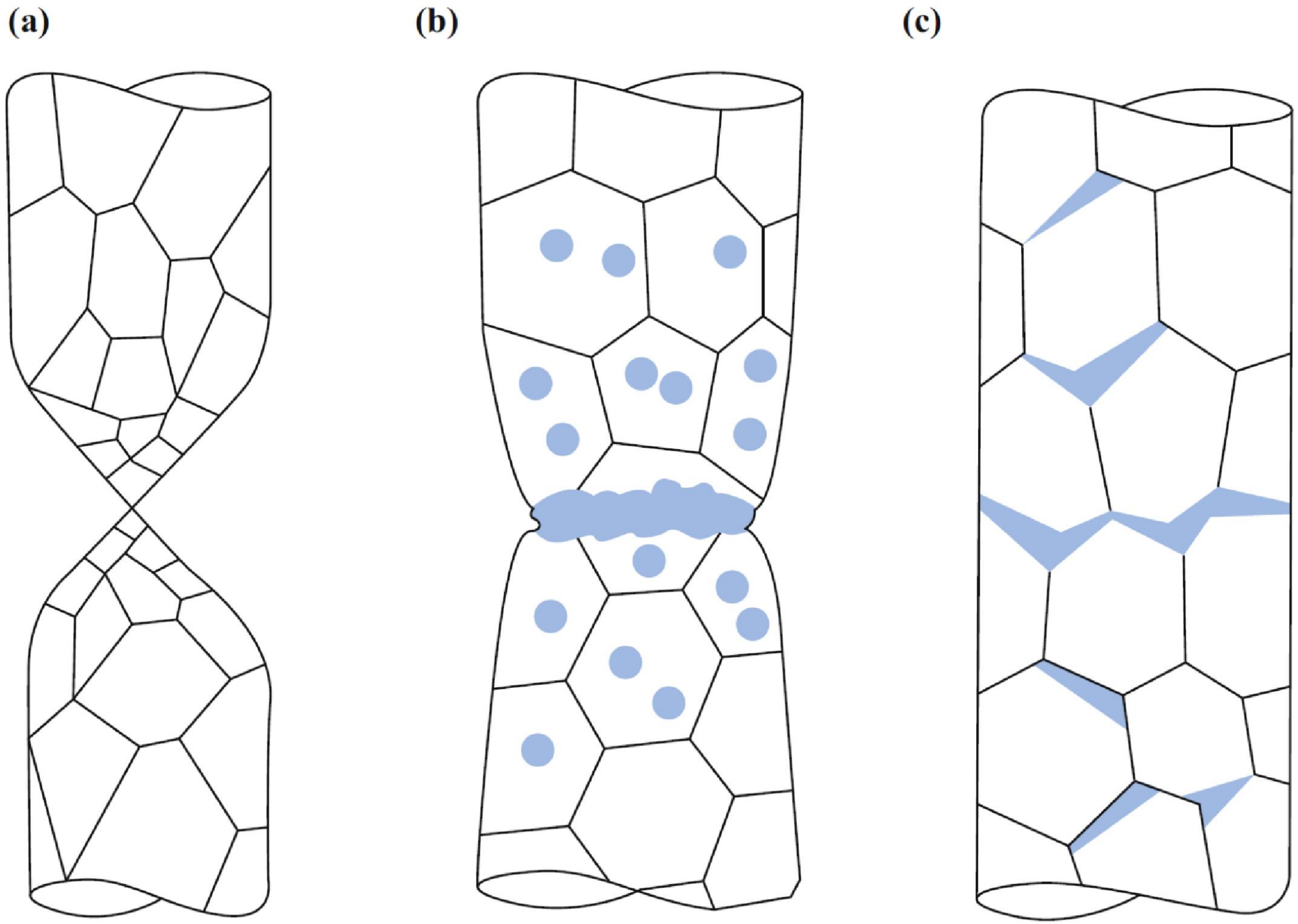
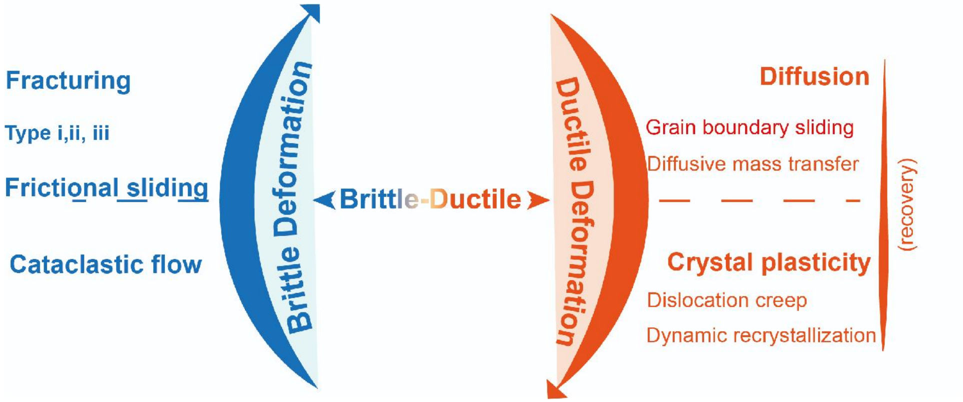


Figure 106. The variation of the cavity concentration versus creep strain in Nimonic 80A (Ni–Cr alloy with Ti and Al) for annealed and pre-strained (cold-worked) alloy. Adapted from Dyson [611]. Cavities were suggested to undergo unconstrained growth.

Fig. 7.22 Three modes of high temperature failure: **a** rupture; **b** transcrystalline creep fracture, in which the coloured circles represent intragranular voids that form, grow and coalescence leading to failure; **c** intercrystalline creep fracture, in which the coloured shaded regions at grain boundaries are intercrystalline voids or cracks that nucleate, grow and coalescence to some degree followed by fracture (Courtney 1990)



Deformation Mechanism Processes



Fracturing

Type i,ii, iii

Frictional sliding

Cataclastic flow

Brittle Deformation

Brittle-Ductile

Ductile Deformation

Diffusion

Grain boundary sliding

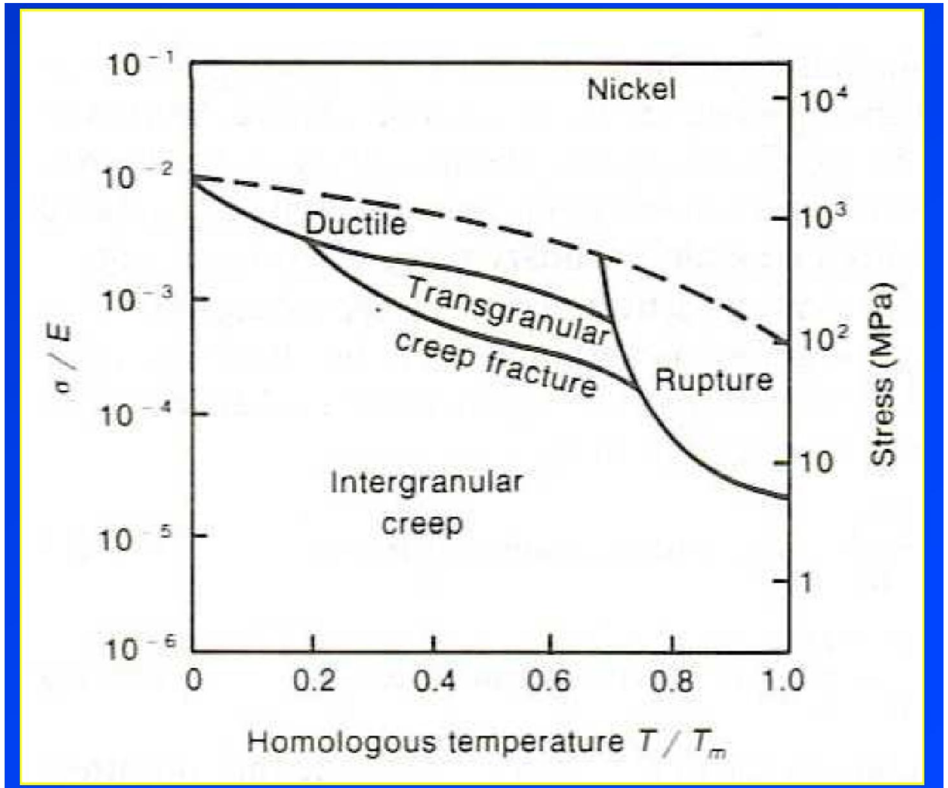
Diffusive mass transfer

(recovery)

Crystal plasticity

Dislocation creep

Dynamic recrystallization



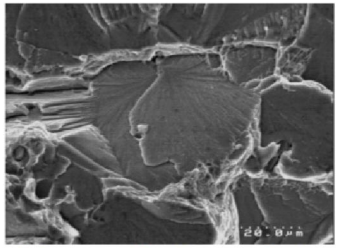
Fracture mechanism map for nickel

Fracture at elevated temperature

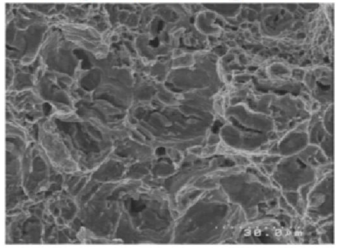
Temp

Transgranular fracture

Slip planes are weaker than grain boundaries



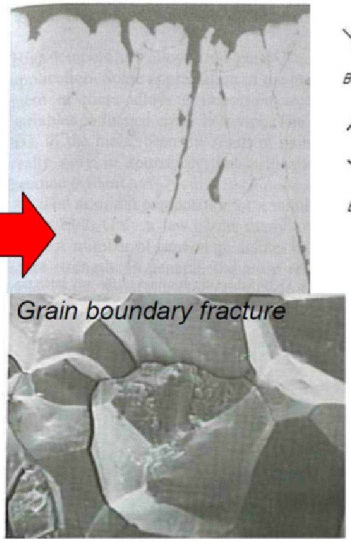
Transgranular cleavage fracture



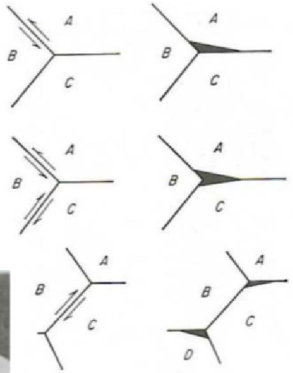
Transgranular microvoid coalescence

Intergranular fracture

Grain boundaries are weaker than slip planes.



Grain boundary fracture



The formation of intergranular crack by grain boundary sliding

Note: at T just below T_{recrys}, ductility drops due to grain boundary sliding → intergranular failure.

Ductile fracture:

Ductile fracture is characterized by extensive plastic deformation and absorbs significant energy before fracture. A crack, formed as a result of the ductile fracture, propagates slowly and when the stress is increased.

Brittle fracture:

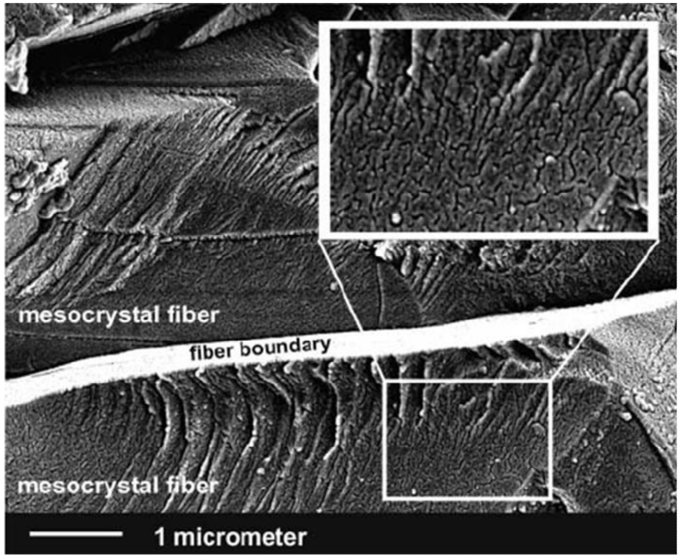
Brittle fracture is characterized by very low plastic deformation and low energy absorption prior to breaking. A crack, formed as a result of the brittle fracture, propagates fast and without increase of the stress applied to the material.

Generally fracture can be divided into 2 types such

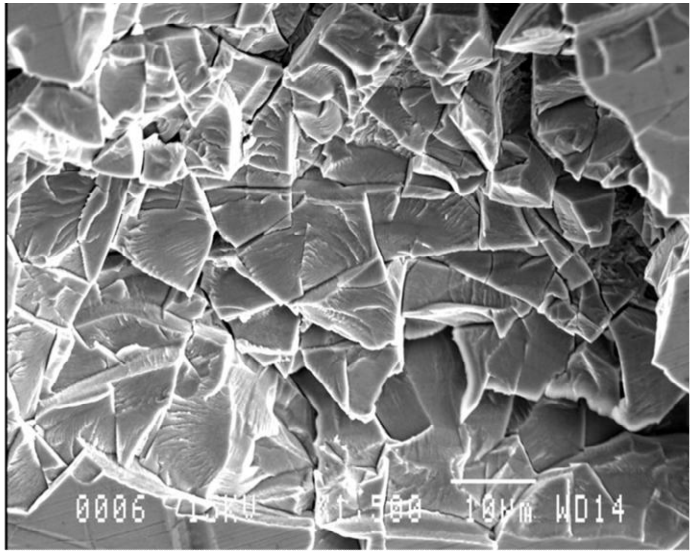
- Brittle fracture (eg. Cast iron)
- Ductile fracture (eg. Mild steel)

Further it can e classified,

Depends on the appearance as 1.shearing fracture and 2.cleavage fracture and crystallographic nature as1. fibrous and 2.granular fracture



Fibrous fracture



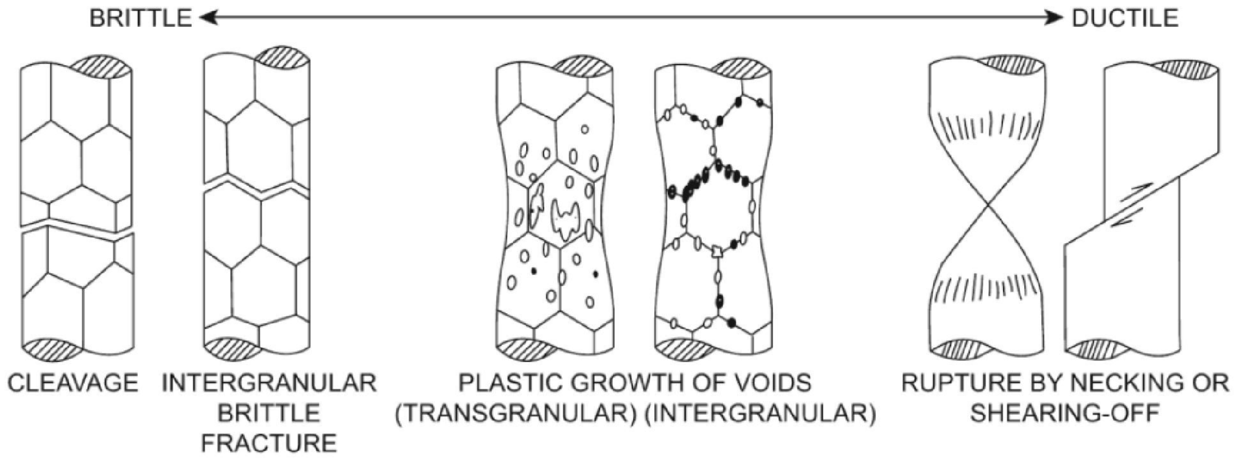
cleavage fracture

Define cleavage fracture.

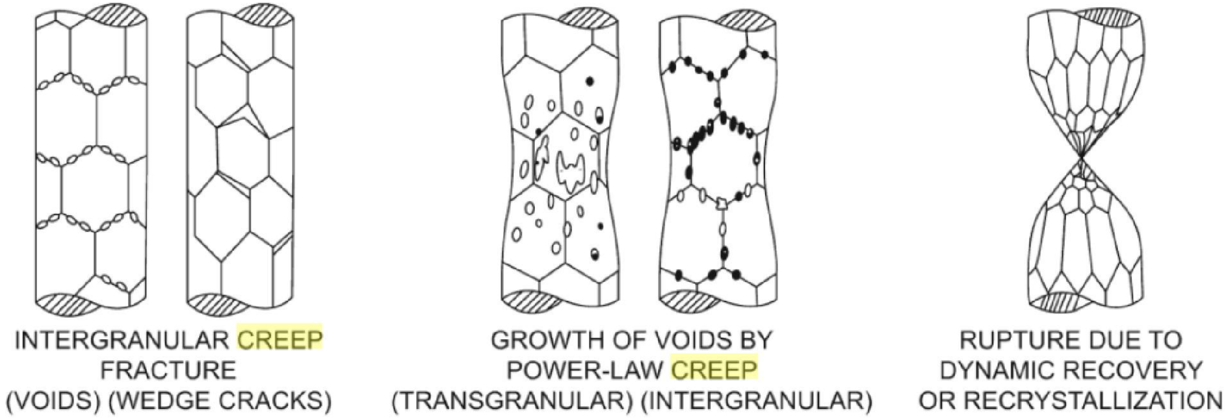
In brittle crystalline materials, fracture can occur by *cleavage* as the result of tensile stress acting normal to crystallographic planes with low bonding (cleavage planes). After the formation of micro crack described above, if the crack propagates along a weak crystallographic plane it is known as cleavage fracture.

7. Distinguish between ductile fracture and brittle fracture?

Ductile fracture	***Brittle fracture***
Materials fractures after plastic deformation and slow propagation of crack.	Materials fractures with very little or no plastic deformation, e.g. in a china clay, glass etc
Fractured surfaces are dull or fibrous in appearance.	Fractured surfaces are crystalline in appearance
Percentage elongation is about 30% prior to fracture occurs.	Percentage elongation is about 0.5% or almost nil prior to fracture occurs.
There is reduction in cross-sectional area of the specimen.	There is virtually no change in the cross sectional area.
Fracture takes place after necking with little sound.	Fracture occurs rapidly often accompanied by a Loud noise.



(a) Low-temperature ($T<0.3T_m$) fracture



(b) Creep temperature ($T>0.3T_m$)

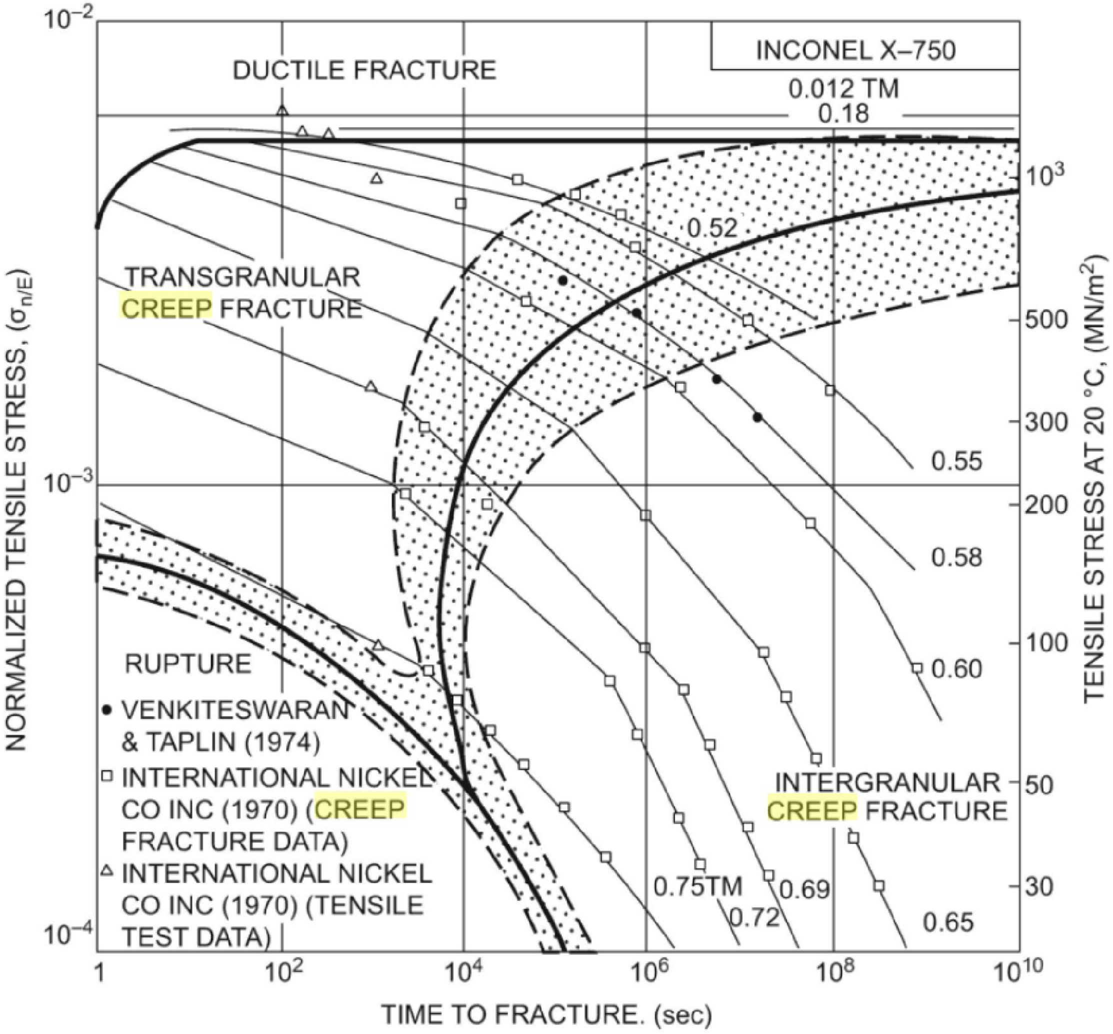
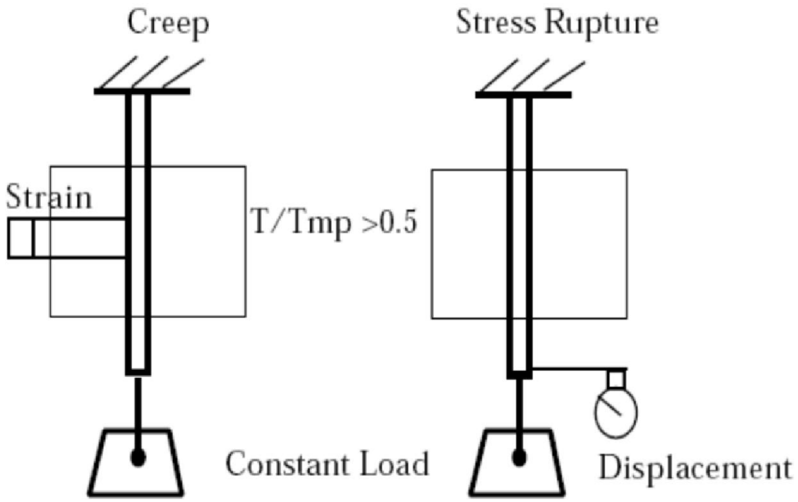


Fig. 2.3 Fracture mechanism map for Inconel X-750. Source: Ref 2.1

Creep vs. Stress Rupture Test

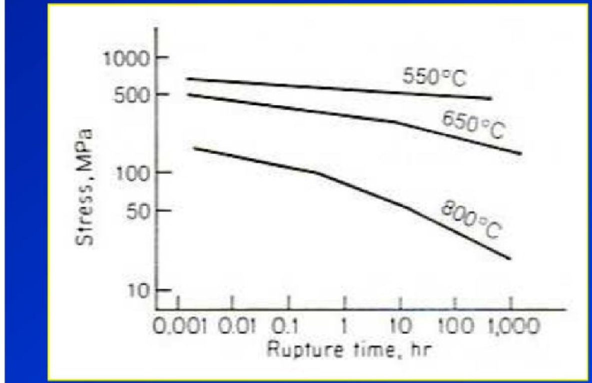


- Low Loads	- High Loads
- Precision Strain Measurement (ε_f<0.5%)	- Gross Strain Measurement (ε_f up to 50%)
- Long term (2000-10,000 h)	- Short term (<1000 h)
- Expensive equipment	- Less expensive equipment
Emphasis on minimum strain rate at stress and temperature	Emphasis on time to failure at at stress and temperature

Stress Rupture Tests

- Determines the time necessary for material to result in failure under a overload.
- Useful in materials selection where dimensional tolerances are acceptable, but rupture cannot be tolerated.
- Generally performed at elevated temperatures.
- Smooth, notched, flat specimens or samples of any combination can be tested.

The ***rupture test*** in carried out in a similar manner to the ***creep test*** but at a ***higher stress level*** until the specimen fails and the ***time at failure*** is measured.



Stress rupture- time data on log-log scale

- ***Rupture strength*** and ***failure time*** are plotted, normally showing a ***straight line***.
- ***Changing of the slope*** indicates ***structural changes in the material***, i.e., transgranular → intergranular fracture, oxidation, recrystallisation, grain growth, spheroidization, precipitation.
- Direct application in ***design***.

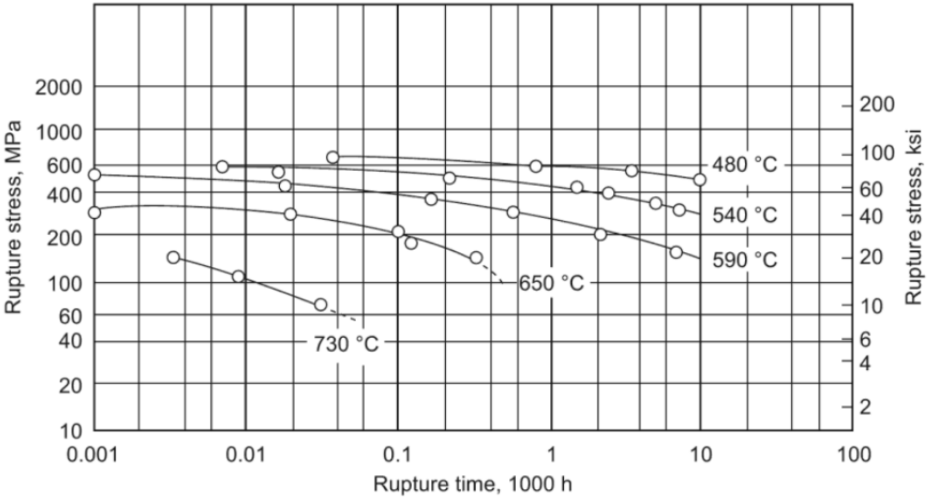


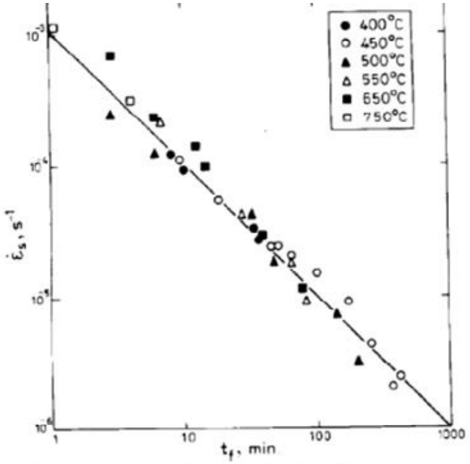
Fig. 2.2 Example of time-temperature-rupture data of 1Cr-1Mo-0.25V steel

Table 2.1 Some stress-rupture time-temperature parameters developed in the 1950s

Authors' names and year	Reference	Form of parameter
Larson and Miller (1952)	Ref 2.3	$T(C_{LM} + \log t)$
Manson and Haferd (1952)	Ref 2.5	$(\log t - \log t_a)/(T - T_a)$
Manson and Brown (1953)	Ref 2.8	$(\log t - \log t_a)/(T - T_a)^n$
Orr et al. (1954)	Ref 2.6	$\log t - \Delta H/RT$
Manson and Succop (1956)	Ref 2.9	$(\log t + C_{MS}T)$

- Monkman-Grant : $\dot{\varepsilon}_s t_r = \mathrm{K}$

  Eq. 13-24



Demonstration of Monkman-Grant Relationship in Cu (Feltham and Meakin 1959)

Monkman-Grant relationship predicts time of failure due to creep mechanisms.

Monkman-Grant relationship relates minimum strain rate and time to failure

$$\dot{\varepsilon}_{min} t_f = C \approx \varepsilon_f$$

$$\varepsilon^{\cdot} = \frac{d\varepsilon}{dt} = Aexp\left(-\frac{Q}{RT}\right)\sigma^{n} = A^{*}$$

$\varepsilon^{\cdot}\ dt$=A^{*}

$\varepsilon^{\cdot}t$=MGP

Modified Monkman-Grant relation (MMGR)

$$\dot{\varepsilon}_{m} \cdot \frac{t_{r}}{\varepsilon_{f}} = \mathit{constant} = C_{MMG}\,.$$

where ε_f is the strain to failure

- Larson-Miller Parameter : $\boxed{P_{L\text{-}M} = T\,(\log t + C)}$

- Sherby-Dorn Parameter : $\boxed{P_{S\text{-}D} = t\, e^{-Q/RT}}$

- Manson-Haferd Parameter : $\boxed{P_{M\text{-}H} = \dfrac{T - T_a}{\log t - \log t_a}}$

Material Parameters

			Manson–Haferd	
Material	Sherby–Dorn Q, kJ/mol	Larson–Miller C	$T_{a'}$, K	log t_a
Various steels and stainless steels	≈400	≈20	–	–
Pure aluminum and dilute alloys	≈150	–	–	–
S-590 alloy (Fe base)	350	17	172	20
A-286 stainless steel	380	20	367	16
Nimonic 81A (Ni base)	380	18	311	16
1% Cr–1% Mo–0.25%V steel	460	22	311	18

Larson Miller Parameter

- Model based on Arrhenius rate equation.

$$LMP = T(C + \log t_r)$$

Where T = temperature (K)

t_r = time before failure (hours)

C = material specific constant

- Predicts rupture lives given certain temperature and stress.
- First used by General Electric in the 50's to perform research on turbine blades.

$$d\epsilon/dt = A \exp(-Q/PT)$$

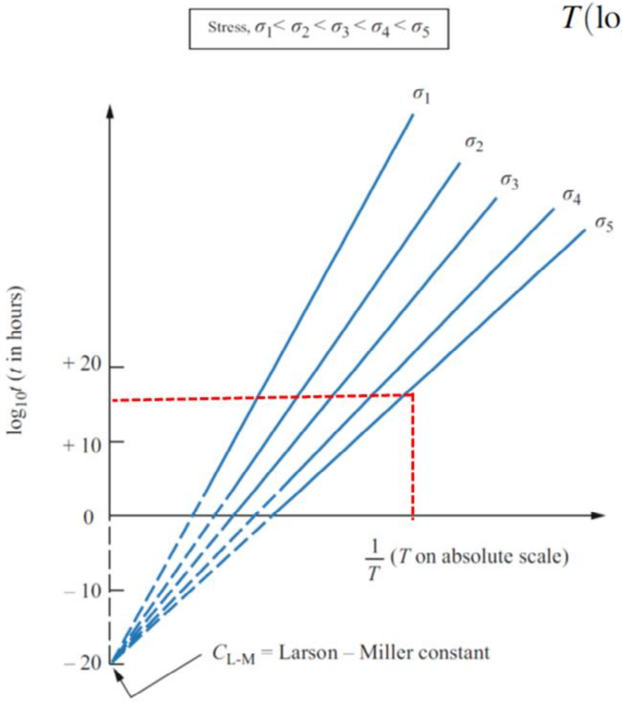
$$\int_0^{\varepsilon} \mathrm{d}\varepsilon = A_1 \mathrm{e}^{-Q/RT} \int_0^{t} \mathrm{d}t$$

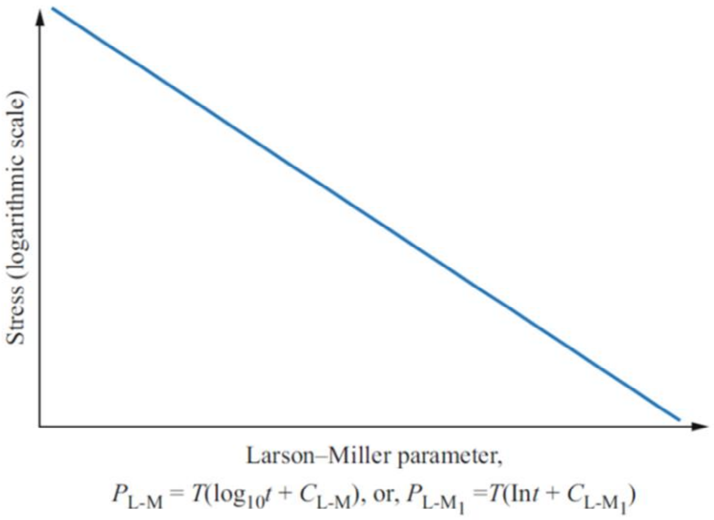
$$\therefore t = \frac{\varepsilon}{A_1} \mathrm{e}^{Q/RT} = \theta \mathrm{e}^{Q/RT}$$

$$\log_{10} t + C_{\mathrm{L-M}} = m\left(\frac{1}{T}\right)$$

$$T(\log_{10} t + C_{\mathrm{L-M}}) = m = \text{function of stress } (\sigma)$$

$$T(\log_{10} t + C_{\mathrm{L-M}}) = m = \text{function of stress } (\sigma)$$





3) Hertzberg, 5.8

If the Larson-Miller parameter for a given elevated temperature alloy was found to be 26,000, by how much would the rupture life of a sample be estimated to decrease if the absolute temperature of the test were increased from 1100 to 1250K? Assume that the Larson-Miller constant is equal to 20.

Solution:

1100K:

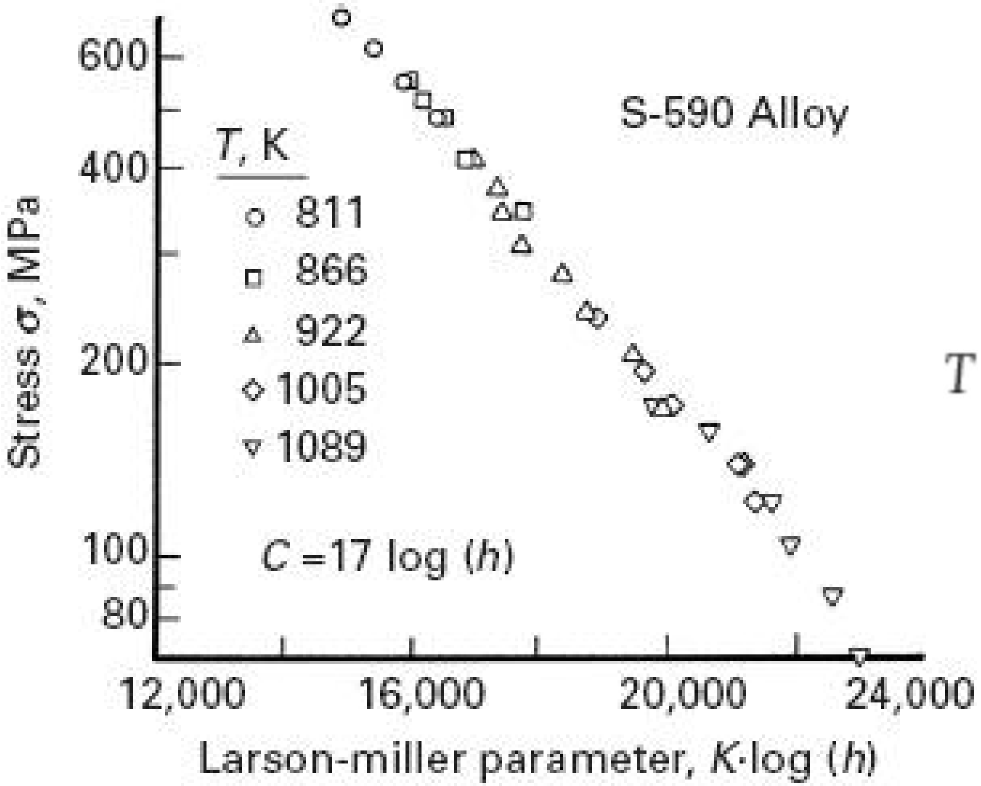
$$26000 = 1100(20 + \log t_1)$$
$$t_1 = 4329hr$$

1250K:

$$26000 = 1250(20 + \log t_2)$$
$$t_2 = 6.3hr$$

The rupture life will decrease from 4329*hr* to 6.3 *hr*.

Larson-Miller Parameter



$$T(\log t_r + C) = m.$$

Master plot for Larson–Miller parameter for S-590 alloy (an Fe-based alloy) (*C* = 17).

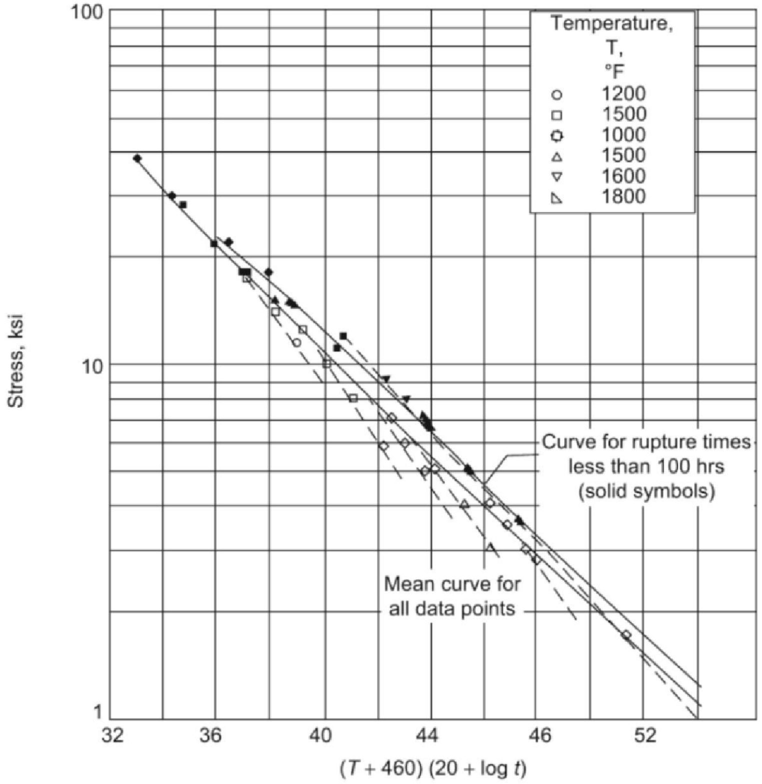
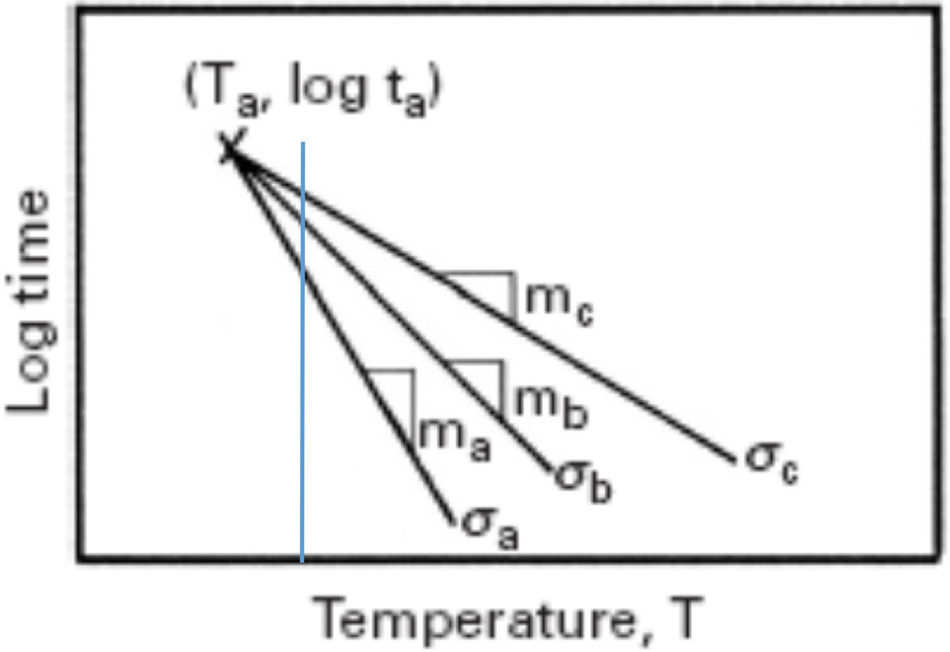


Fig. 2.4 "Peeling onion-skin" appearance of master rupture curves for 18-8 stainless steel using the Larson-Miller parameter $(T + 460) \times (20 + \log t)$

Manson-Hafered Parameter

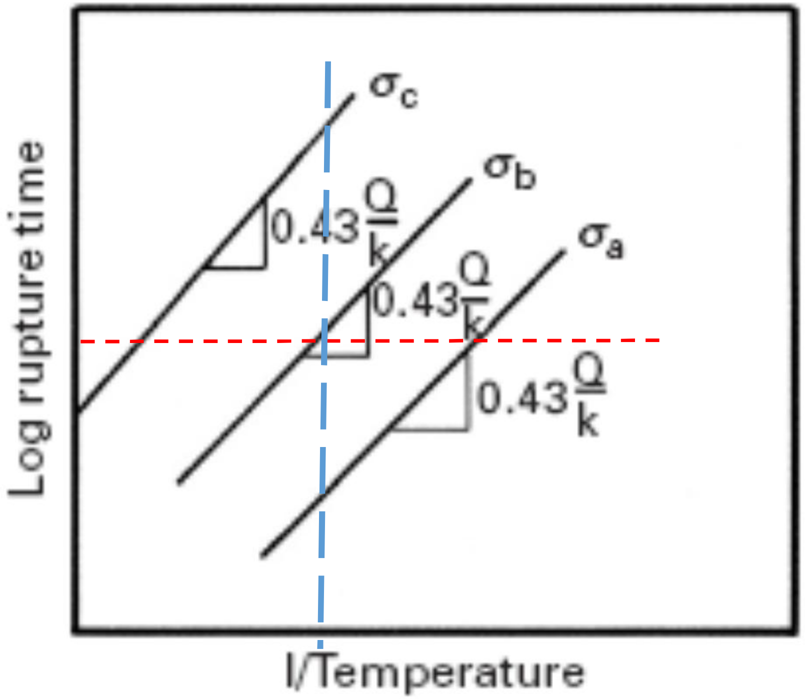


$$\frac{\log t_r - \log t_a}{T - T_a} = m.$$

Relationship between time rupture and temperature at three levels of stress, *σa*, *σb*, and *σc*, using Manson–Haferd parameter (*σa* > *σb* > *σc*).

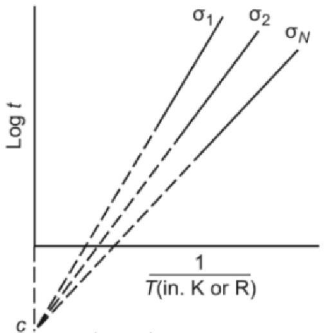
Sherby-Dorn Parameter

- Sherby-Dorn Parameter : $\boxed{P_{S\text{-}D} = t\, e^{-Q/RT}}$



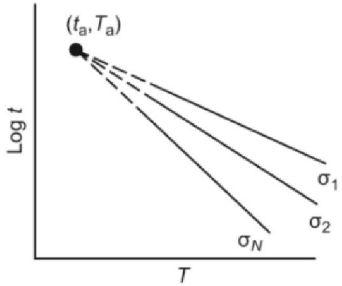
$$\log t_r - m = 0.43\frac{Q_c}{kT}$$

Relationship between time to rupture and temperature at three levels of stress, *σa* > *σb* > σc, using Sherby–Dorn parameter.



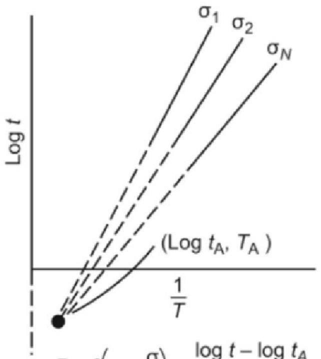
$P = f\left(\sigma \text{ or } \frac{\sigma}{E}\right) = (T)(\log t + C_{\mathrm{LM}})$

(a) **Larson-Miller**



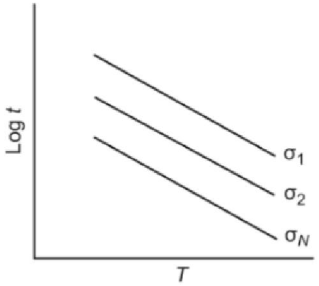
$P = f\left(\sigma \text{ or } \frac{\sigma}{E}\right) = \frac{\log t - \log t_a}{T - T_a}$

(b) **Manson-Haferd**



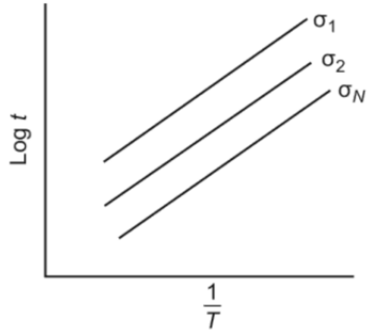
$P = f\left(\sigma \text{ or } \frac{\sigma}{E}\right) = \frac{\log t - \log t_A}{1/T - 1/T_A}$

(c) **Goldhoff-Sherby**



$P = f\left(\sigma \text{ or } \frac{\sigma}{E}\right) = \log t - TC_{\mathrm{MS}}$

(d) **Manson-Succop**



$P = f\left(\sigma \text{ or } \frac{\sigma}{E}\right) = \log t - \Delta H/\mathrm{RT})$

(e) **Orr-Sherby-Dorn**

Example :

Calculate the time to rupture at 650°C and 100MPa stress for a 1%Cr-1% Mo-0.25%V steel, according to the Larson-Miller , Sherby--Dorn, and Manson--Haferd methods, if this alloy underwent rupture in 20hrs when tested in tension at the same stress level at a temperature of 750°C.

Solution:

The Larson-Miller equation is $T(\log t_r + C) = m$.

At 750°C, $T = 750 + 273 = 1{,}023^0$ K and $t_r = 20$ hours. Therefore,

$m = 1023 \times (\log 20 + 22) \approx 2.4 \times 10^4$

At 650°C, $T = 650 + 273 = 923^0$K, and we have

$923(\log t_r + 22) = 2.4 \times 10^4$, so that $\log t_r = (2.4 \times 10^4/923) - 22$

$t_r = 6.7 \times 10^3$ hours.

The Sherby-Dorn equation is $\log t_r - Q/(kT) = m$. From Table 1, $Q = 460$ kJ/mol. Because Q here involves moles, we must use R instead of k. At 750°C, $T = 1{,}023^0$ K and $t_r = 20$ hours. Thus, $m = \log 20 - (460 \times 10^3/8.314 \times 1023)$

At 650°C, $T = 923^0$ K, and we obtain $\log t_r = m + 0.43(Q/kT)$

so that **$t_r = 6 \times 10^3$ hours.**

The Manson-Haferd equation is $(\log t_r - \log t_a)/(T - T_a) = m$. From Table 1, $Ta = 311$ K, and $\log t_a = 18$. At 750°C, $T = 1{,}023^0$ K, and it follows that $t_r = 20$ hours. Therefore, $m = (\log 20 - 18)/(1{,}023 - 311) = -0.023$.

At 650°C, $T = 923^0$ K, and we have $(\log t_r - \log t_a)/(T - T_a) = m$

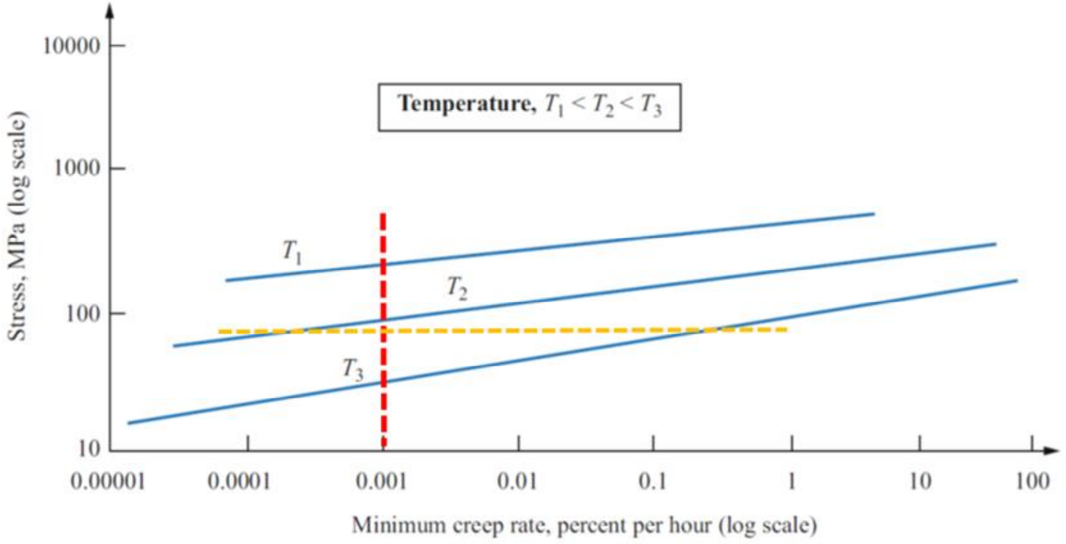
$(\log t_r - 18)/(923 - 311) = -0.023$,

$\log t_r = 3.924$,

$t_r = 8.4 \times 103$ hours.

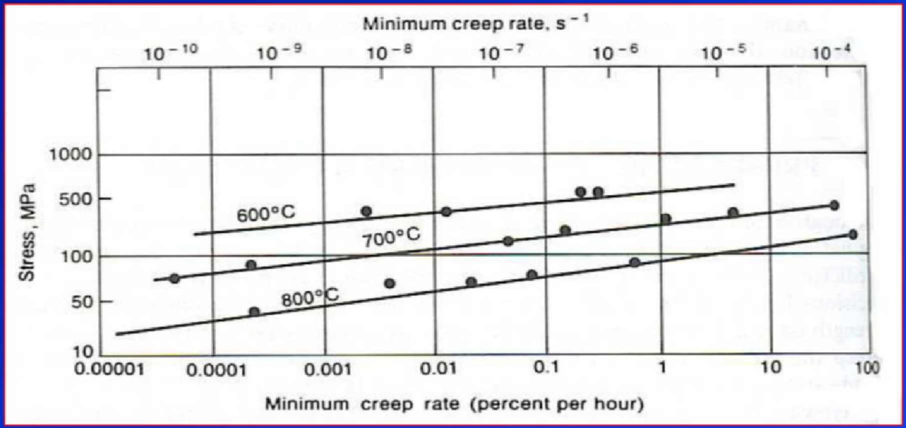
Prediction of Creep Strength

Fig. 7.30 Variation of minimum creep rate with stress at various temperatures for a given material (schematic)



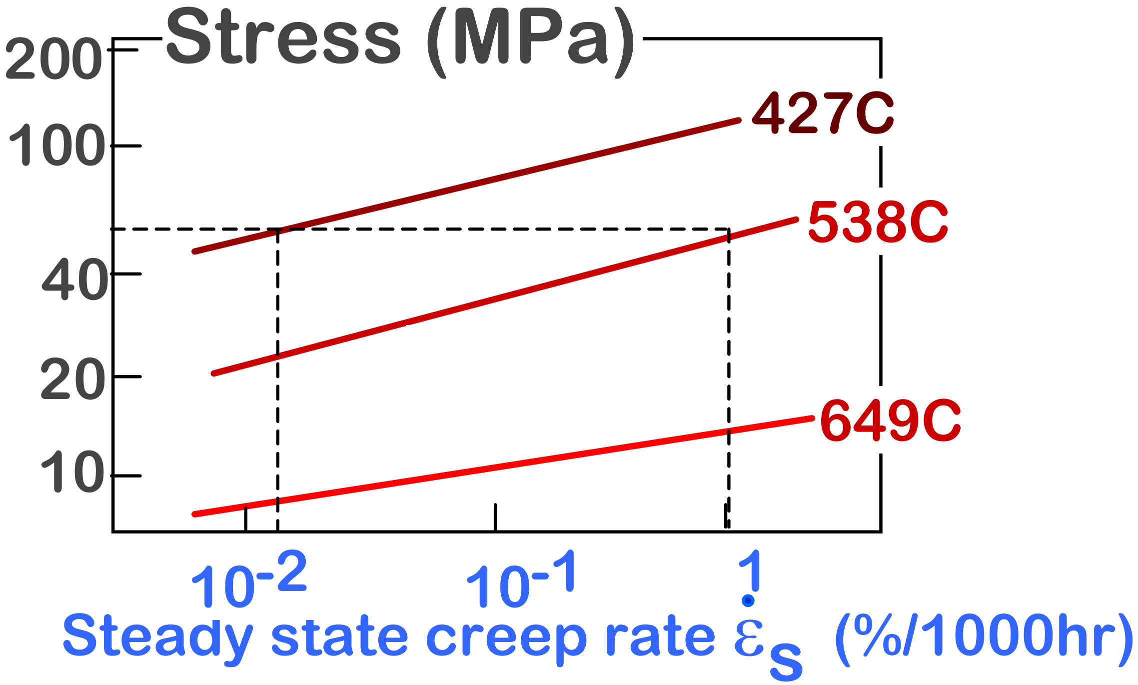
Presentation of engineering creep data

Creep strength is defined as the stress at a given temperature, which produces a ***steady-state creep rate*** (10^{-11} to 10^{-8} s^{-1}.)



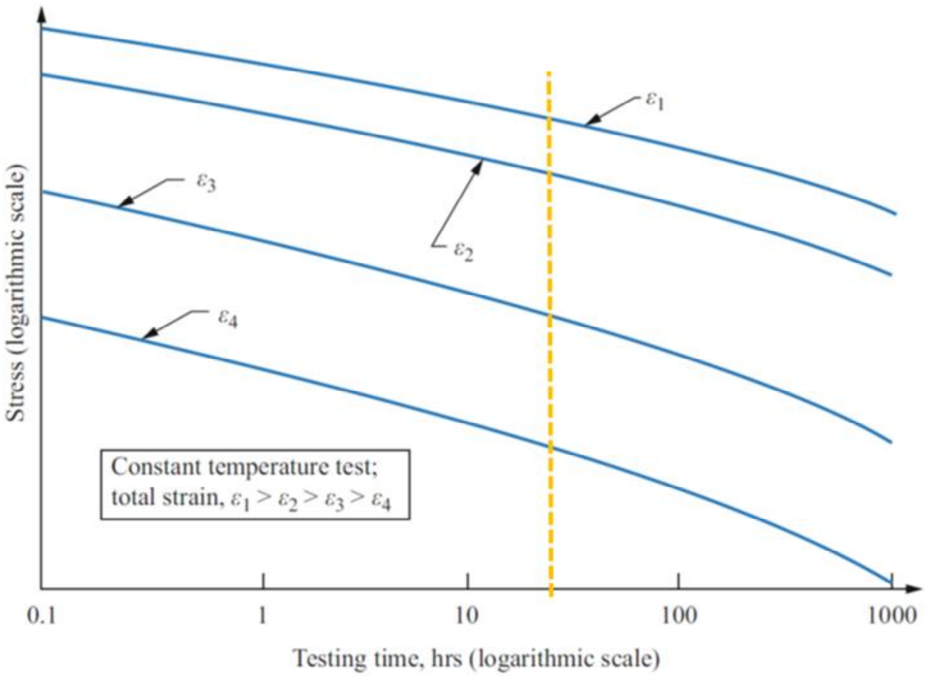
Stress vs minimum creep rate

- ***Log-log plot*** is used so that the extrapolation of one log-cycle represents a ***tenfold change***.


Stress (MPa)
200
100
40
20
10
427C
538C
649C
10-2
10-1
1
Steady state creep rate ε̇s (%/1000hr)

Prediction of Creep Strength

Fig. 7.28 Stress versus testing time on log-log plot for different amounts of total strain for a given material at a constant temperature (schematic)



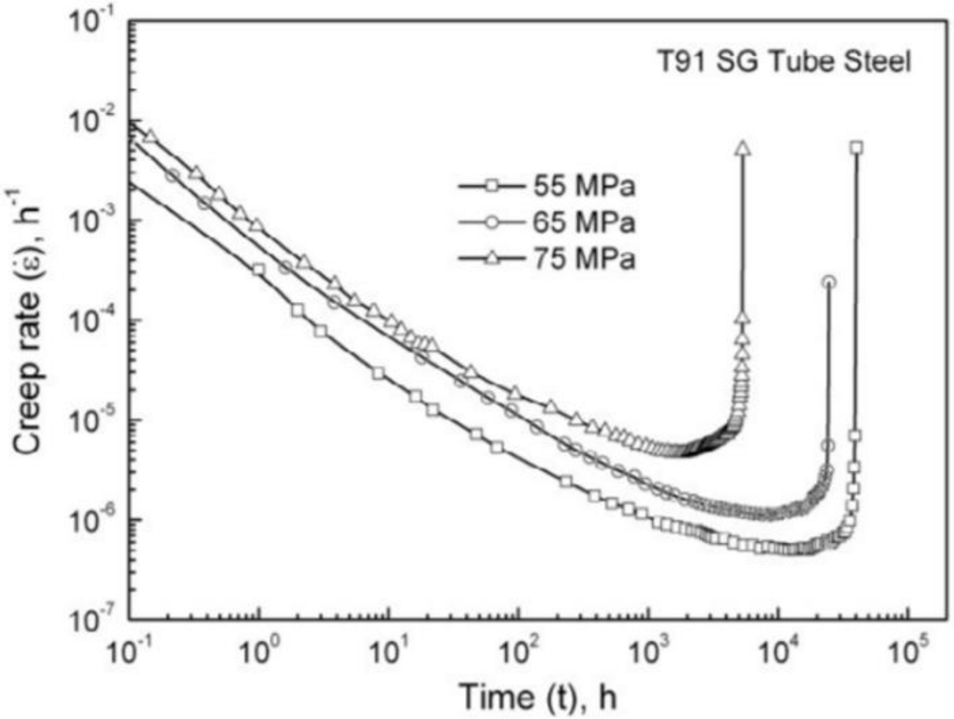


Fig. 4. Typical creep curves showing creep rate-time plots for stress levels 55, 65 and 75 MPa at 923 K.

Prediction of Creep Strength

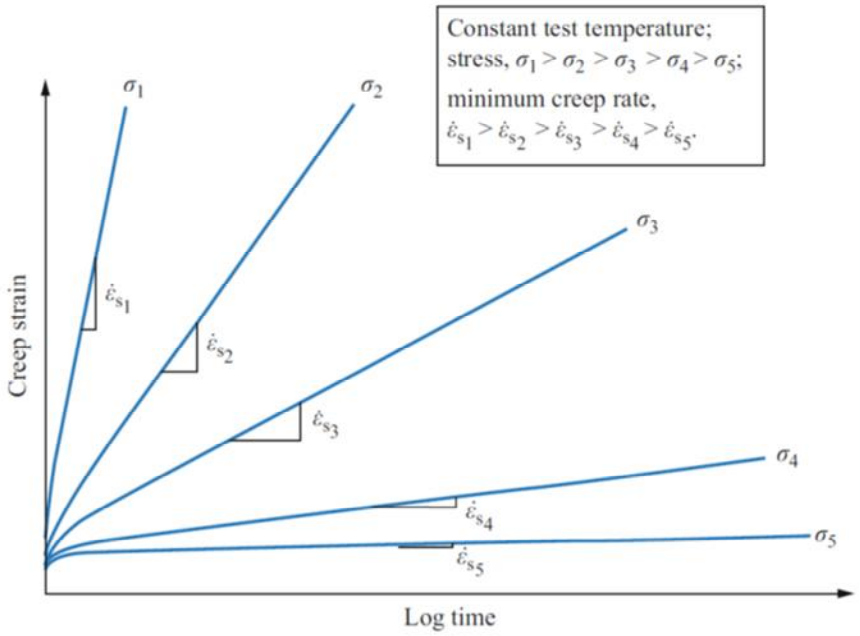


Fig. 7.29 Family of schematic creep strain–time curves at various stress levels for a given material tested at the same temperature, showing minimum creep rates

Creep deformation

Low temperature Creep $T < 0.3T_m$

Medium temperature Creep $0.3T_m < T < 0.8T_m$

High temperature Creep $T > 0.8T_m$

LOW TEMPERATURE CREEP

Creep at low temperatures (< 0.25Tmp) is generally transient, achieving creep exhaustion during stage 1 (primary creep) and never reaching stage 2 (secondary creep). Thermal energy required for dynamic recovery is not available in the temperature ranges of interest; therefore, the material experiences creep exhaustion or creep saturation.

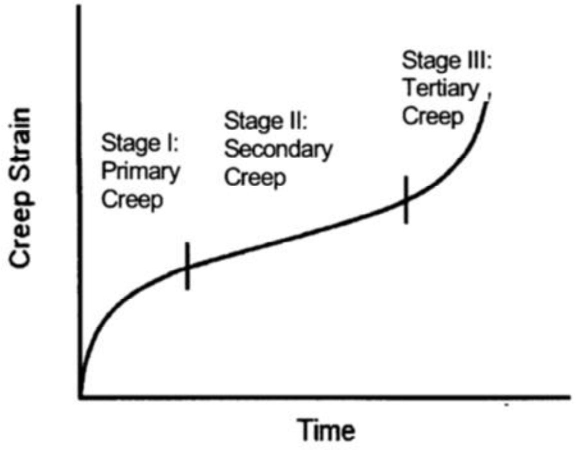
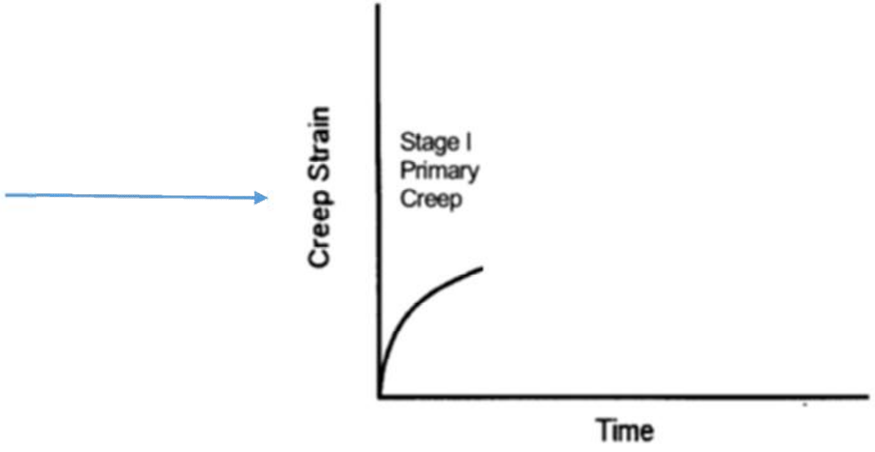
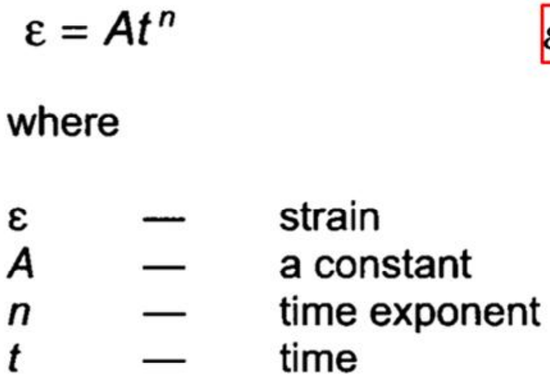


Figure 2-1. Typical Creep Curve Showing the Three Stages of Creep



Primary or transient creep is often described by the following empirical power law equation



$$\varepsilon = At^n$$

where

ε — strain
A — a constant
n — time exponent
t — time

$$\boxed{\varepsilon^{\cdot} = A(n-1)t^{n-1} = Bt^{n-1}}$$

The constant A is reportedly dependent on microstructure and temperature

Andrade Law

$$\boxed{\varepsilon = \varepsilon_0 + \beta t^{1/3}}$$

$$\boxed{\varepsilon^{\cdot} = At^{-2/3}}$$

where $\boldsymbol{\varepsilon}_0$ is the instantaneous strain, β isa constant and t is time. Equation is in accordance with the time law of creep proposed by Andrade,[1] known as Andrade's β-flow.

Creep mechanisms can be visualized by using superposition of various strain-time curves as shown in Fig. 8.5. An empirical relation which describes the strain-time relation is:

$$\boxed{\varepsilon = \varepsilon_i\left(1 + \beta t^{1/3}\right)\exp(kt)} \qquad (8.2)$$

$$\varepsilon = \varepsilon_0 (1 + \beta t^{1/3}) e^{kt}$$

Transient creep may also follow a logarithmic fit of the type

$$\varepsilon = A' + B\ln(t)$$

where

ε — instantaneous plastic and creep strain
A' and B — creep constants
t — time measured in hours

In comparison to the normal creep equation, the logarithmic creep behavior is usually described by

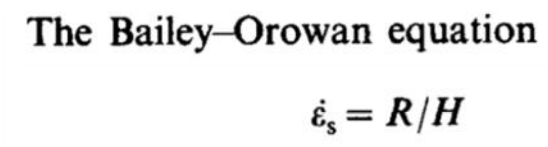
$$\boxed{\varepsilon = \varepsilon_0 + \alpha \ln(1 + \gamma t),}$$

where $\boldsymbol{\alpha}$ and $\boldsymbol{\gamma}$ are constants. This equation indicates that over a long period of time, the strain rate of deformation tends to become zero. Such an equation, as discussed in the previous section, would be useful for describing exhaustion creep.

primary or transient creep :

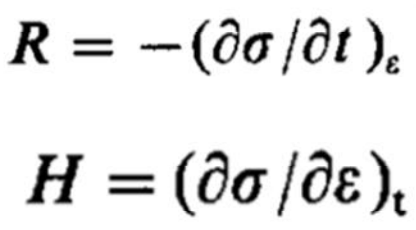
- Andrade-β flow (or 1/3 rd law) : $\varepsilon_p = \beta t^{1/3}$ ⇔ problem as $t \rightarrow 0$
- Garofalo / Dorn Equation : $\varepsilon_p = \varepsilon_t (1 - e^{-rt})$, r is related to $\frac{\dot{\varepsilon}_i}{\dot{\varepsilon}_s}$

Medium TEMPERATURE CREEP



The Bailey–Orowan equation

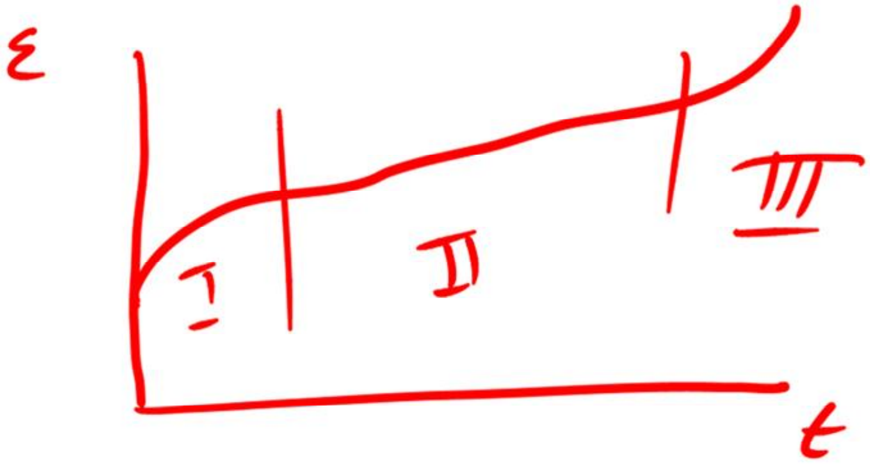
$$\dot{\varepsilon}_s = R/H$$



$$R = -(\partial\sigma/\partial t)_\varepsilon$$

$$H = (\partial\sigma/\partial\varepsilon)_t$$

Medium TEMPERATURE CREEP



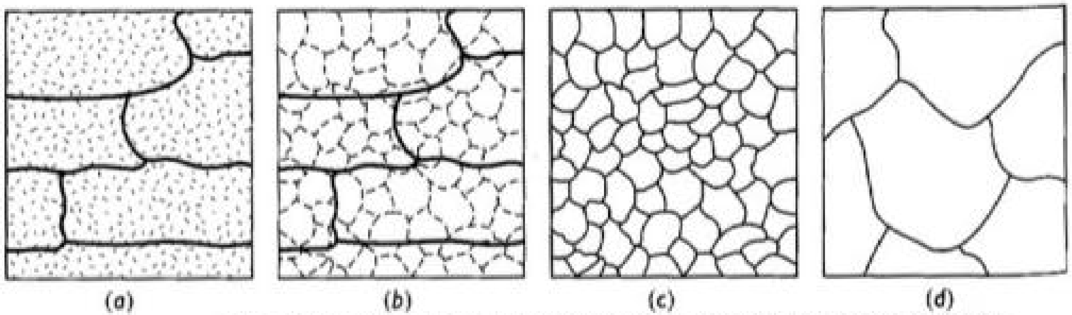
Chapter 7

Recrystallization

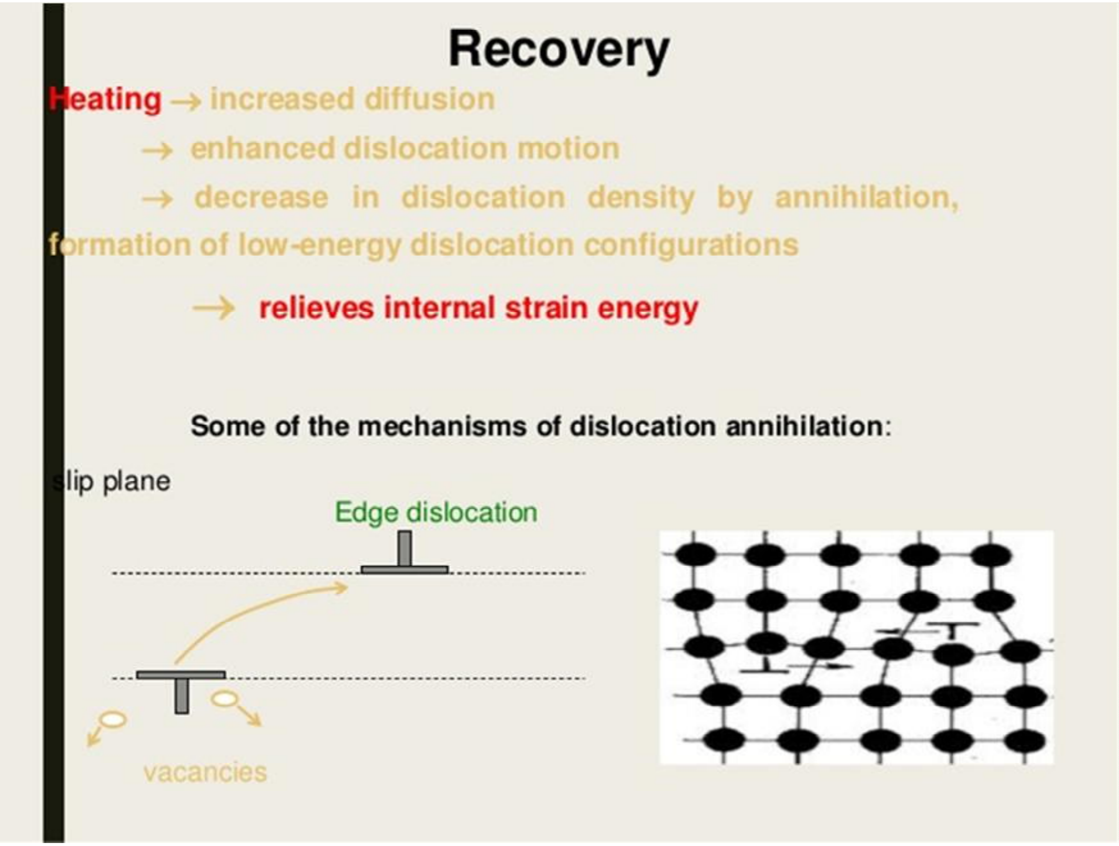
Dynamic recrystallization

Annealing is a heat treatment designed to eliminate the effects of cold working. The properties of a metal may revert back to the precold-worked states by Annealing, through recovery, recrystallization and grain growth.

Recovery: the relief of some of the internal strain energy of a previously cold-worked material.
Recrystallization: the formation of a new set of strain-free grains within a previously cold-worked material.
Grain Growth: the increase in average grain size of a polycrystalline material. An elevated temperature heat treatment (annealing) is needed for these 3-processes.



The effect of annealing temperature on the microstructure of cold-worked metals: (a) cold worked, (b) after recovery, (c) after recrystallization, and (d) after grain growth.



Recovery

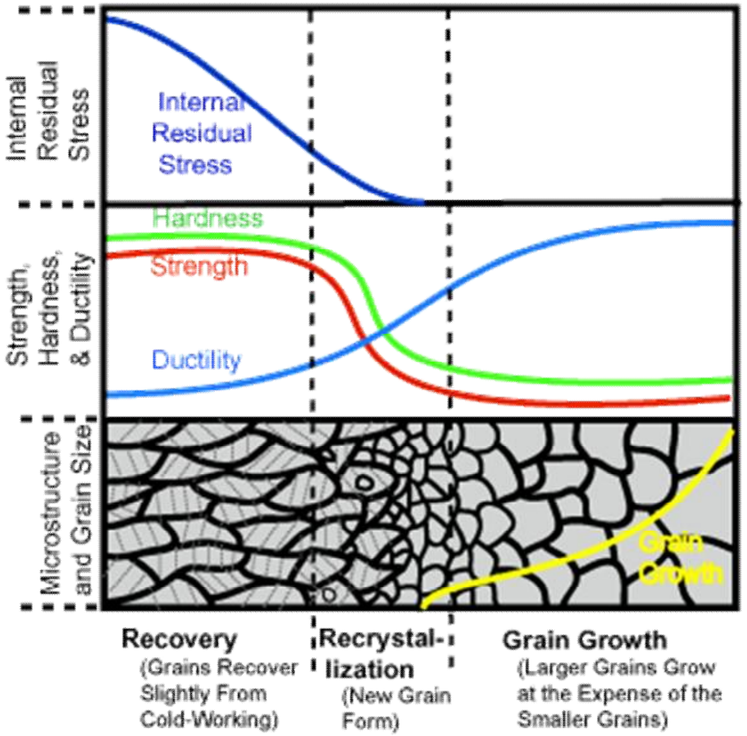
Heating → increased diffusion

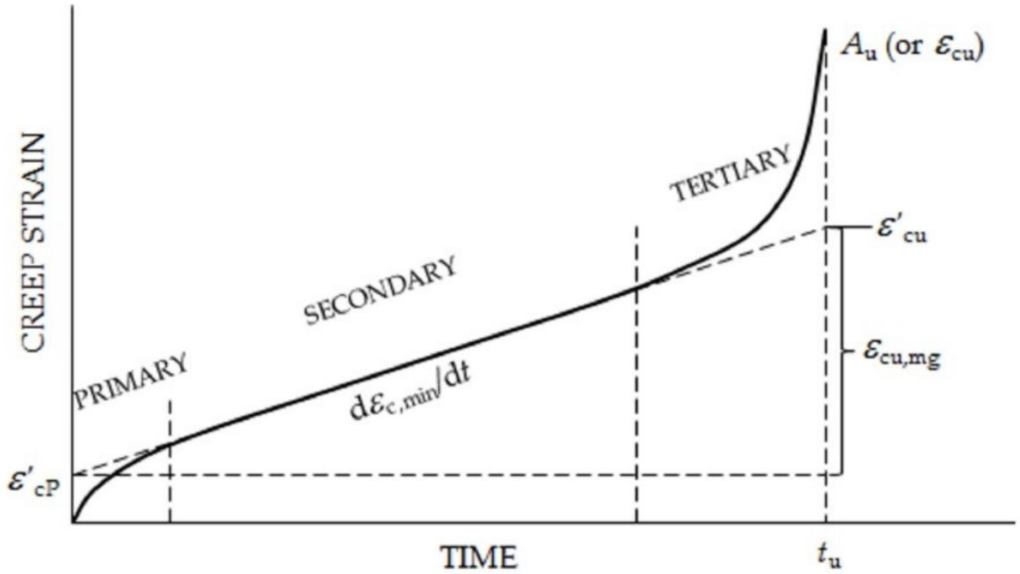
→ enhanced dislocation motion

→ decrease in dislocation density by annihilation, formation of low-energy dislocation configurations

→ relieves internal strain energy

Some of the mechanisms of dislocation annihilation:




CREEP STRAIN
TIME
PRIMARY
SECONDARY
TERTIARY
$d\varepsilon_{c,min}/dt$
A_u (or ε_{cu})
ε'_{cu}
$\varepsilon_{cu,mg}$
ε'_{cP}
t_u

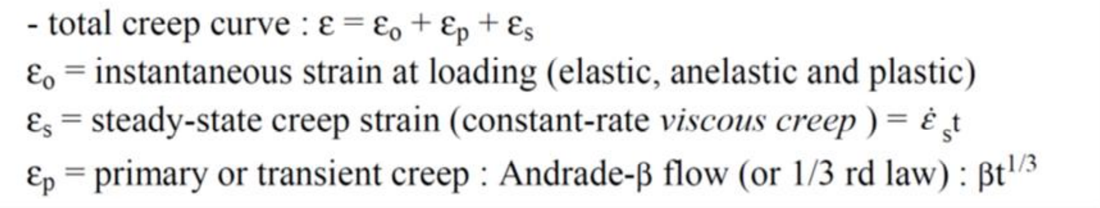
During this stage, the steady state is achieved because of an approximate balance between two opposing factors: the strain hardening that tends to reduce the creep rate and the softening or recovery process that tends to increase it

σ=cte , σ=f(t, ϵ) →

$$\mathrm{d}\sigma = \frac{\partial\sigma}{\partial t}\partial t + \frac{\partial\sigma}{\partial\varepsilon}\partial\varepsilon = 0, \quad \text{or,} \quad \frac{\partial\sigma}{\partial\varepsilon}\partial\varepsilon = -\frac{\partial\sigma}{\partial t}\partial t;$$

$$\therefore \dot{\varepsilon}_{\mathrm{s}} = \frac{\mathrm{d}\varepsilon}{\mathrm{d}t} = -\frac{\partial\sigma/\partial t}{\partial\sigma/\partial\varepsilon} = \frac{r}{h}$$

Microstructural softening due to recovery process in the low-temperature region includes cross-slip of screw dislocations while that at a homologous temperature above 0.5 includes rearrangement and annihilation of dislocations and climbing of edge dislocations.



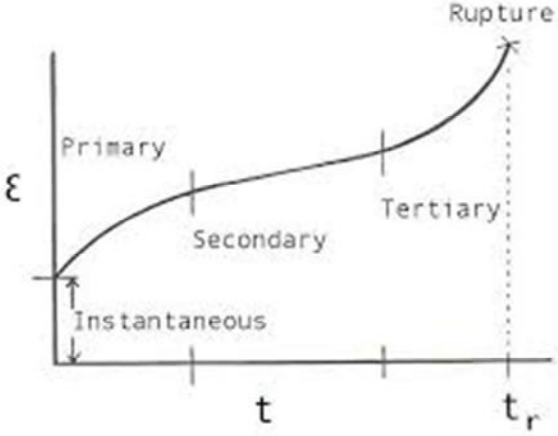
- total creep curve : $\varepsilon = \varepsilon_o + \varepsilon_p + \varepsilon_s$

ε_o = instantaneous strain at loading (elastic, anelastic and plastic)

ε_s = steady-state creep strain (constant-rate *viscous creep*) = $\dot{\varepsilon}_s t$

ε_p = primary or transient creep : Andrade-β flow (or 1/3 rd law) : $\beta t^{1/3}$

$$\boxed{\varepsilon = \varepsilon_0 + \beta t^{1/3} + \varepsilon^{\cdot} t}$$



Garofalo equation:

$$\boxed{\varepsilon = \varepsilon_0 + \varepsilon_t(1 - e^{-rt}) + \dot{\varepsilon}_s t}$$

$$\varepsilon = \varepsilon_i + \varepsilon_t(1 - \exp(rt)) + t\dot{\varepsilon}_{ss}$$

$$\boxed{\varepsilon = \varepsilon_0 + \beta t^{1/3} + \dot{\varepsilon}_s t + \gamma t^3,}$$

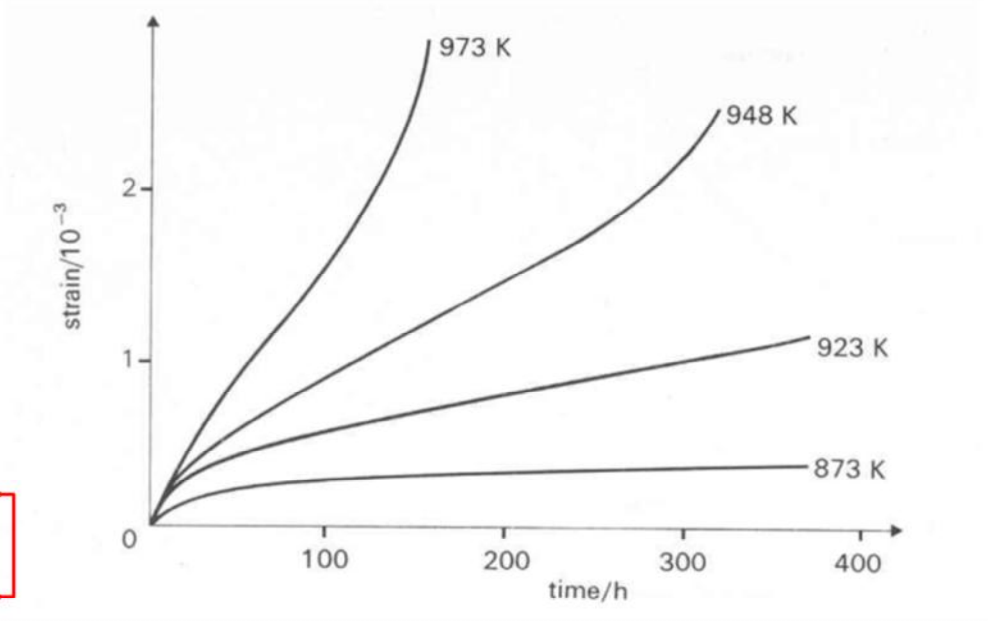
where $\boldsymbol{\gamma} t^3$ describes the tertiary component of the creep curve

$$\boxed{\varepsilon = \varepsilon_i + B\sigma^m t + D\sigma^\alpha(1 - \exp(\beta t))}$$

High TEMPERATURE CREEP

$$\boxed{\varepsilon = \varepsilon_0 + \gamma t^n}$$

$$\boxed{\varepsilon = B + C\exp(\gamma t)}$$

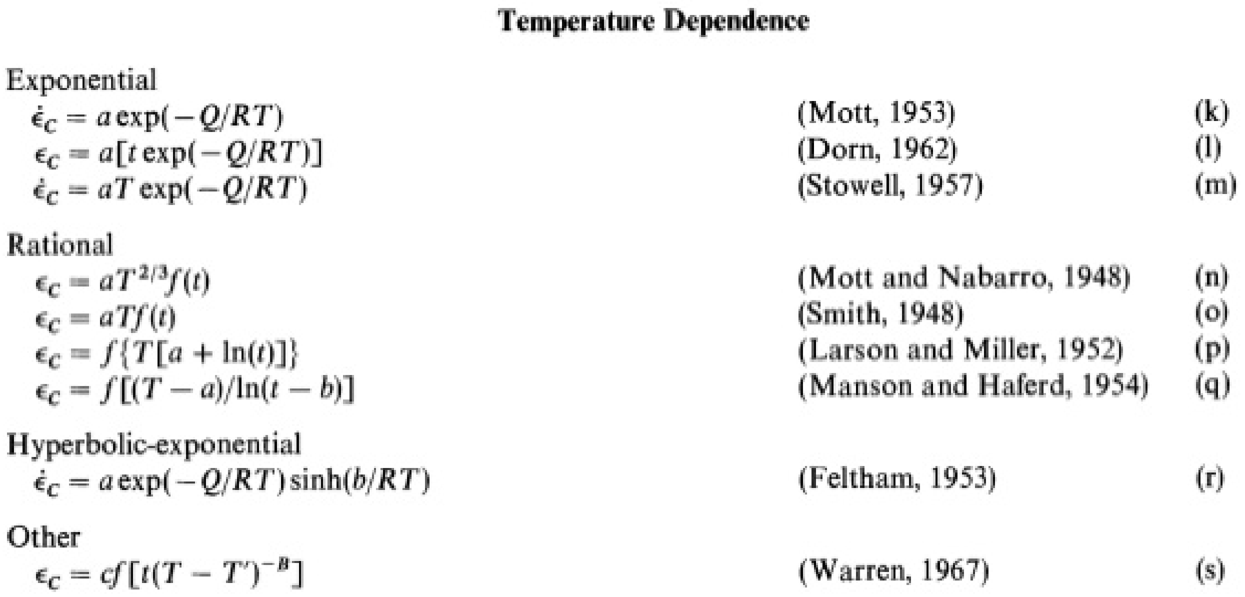


Proposed by	Formula	Comments
Andrade (1910, 1914)	$\varepsilon_c = Bt^{1/\beta}$	Applicable to primary stage; β=3;
Lomnitz (1956, 1957)	$\varepsilon_c = A\ln(1+\alpha t)$	Applicable to primary stage
Modified Lomnitz law	$\varepsilon_c = A + B\log(t) + Ct$	Primary and secondary stages
Norton's law	$\varepsilon_c = A\sigma_a^n t$ or $\varepsilon_c = A\sigma_a^n$	Applicable to secondary stage and n=4-5
Modified Norton's law	$\varepsilon_c = B\left\langle \frac{\sigma_a}{\sigma_{ct}} - 1 \right\rangle^n t$ or $\varepsilon_c = B\left\langle \frac{\sigma_a}{\sigma_{ct}} - 1 \right\rangle^n$	Applicable to secondary stage and σ_{ct} is the stress threshold to induce steady state creep response.
Griggs and Coles (1958)	$\varepsilon_c = A + Bt^2$	Applicable to tertiary stage
Aydan et al. (2003)	$\varepsilon_c = A\left(1 - e^{-t/\tau_1}\right) + B\left(e^{t/\tau_2} - 1\right)$	Applicable to all stages creep leading to failure

$A, B, C, \alpha, \tau_1, \tau_2$, and n are constants to be determined from experimental results. $\sigma_a, \varepsilon_c, \dot{\varepsilon}_c$, and t are the applied stress, creep strain, strain rate, and time, respectively, hereafter

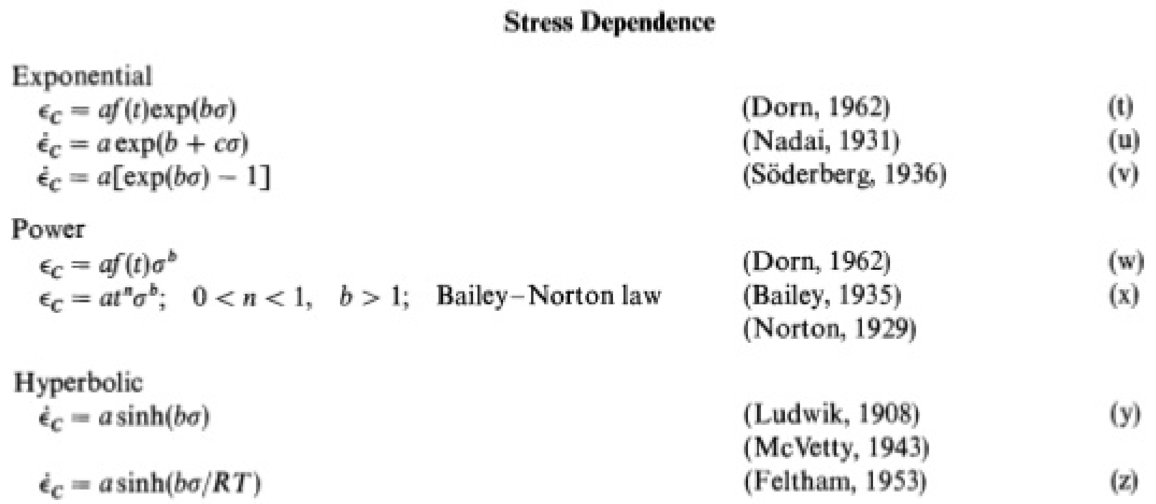
Equation Form	References	Equation
Time Dependence		
Rational		
$\epsilon_C = at/(1 + bt)$	(Freundenthal, 1936)	(a)
Logarithmic		
$\epsilon = a + b\ln(t)$	(Phillips, 1905)	(b)
$\epsilon = a + b\ln(1 + ct)$	Modification of (b)	(c)
Exponential		
$\epsilon = a + bt - c\exp(-dt)$	(McVetty, 1934)	(d)
$\epsilon_C = at + b[1 - \exp(-ct)]$	(McVetty, 1934) (Söderberg, 1936)	(e)
Power		
$\epsilon_C = bt^n; \quad 1/3 < n < 1/2$	(Bailey, 1935)	(f)
Power series		
$\epsilon_C = at^m + bt^n; \quad m > 1, \quad 0 < n < 1$	(de Lacombe, 1939)	(g)
$\epsilon_C = at_m + bt_n + ct_p \ldots$	(Graham, 1953)	(h)

- The temperature dependency of creep is often related to thermodynamics and rate processes of solid-state physics, the temperature dependency is often of exponential form.



Temperature Dependence

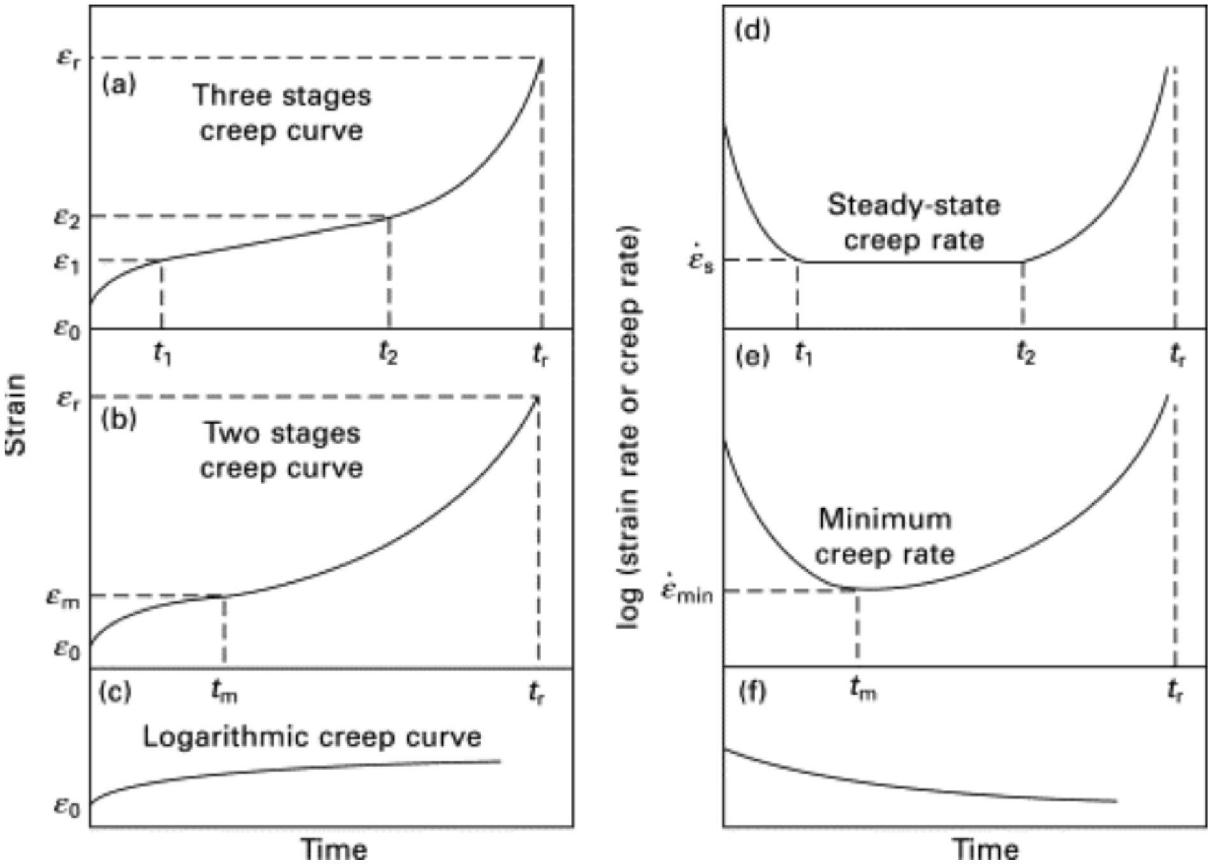
Exponential		
$\dot{\epsilon}_C = a\exp(-Q/RT)$	(Mott, 1953)	(k)
$\epsilon_C = a[t\exp(-Q/RT)]$	(Dorn, 1962)	(l)
$\dot{\epsilon}_C = aT\exp(-Q/RT)$	(Stowell, 1957)	(m)
Rational		
$\epsilon_C = aT^{2/3}f(t)$	(Mott and Nabarro, 1948)	(n)
$\epsilon_C = aTf(t)$	(Smith, 1948)	(o)
$\epsilon_C = f\{T[a + \ln(t)]\}$	(Larson and Miller, 1952)	(p)
$\epsilon_C = f[(T - a)/\ln(t - b)]$	(Manson and Haferd, 1954)	(q)
Hyperbolic-exponential		
$\dot{\epsilon}_C = a\exp(-Q/RT)\sinh(b/RT)$	(Feltham, 1953)	(r)
Other		
$\epsilon_C = cf[t(T - T')^{-B}]$	(Warren, 1967)	(s)



Stress Dependence

Exponential		
$\epsilon_c = af(t)\exp(b\sigma)$	(Dorn, 1962)	(t)
$\dot{\epsilon}_c = a\exp(b + c\sigma)$	(Nadai, 1931)	(u)
$\dot{\epsilon}_c = a[\exp(b\sigma) - 1]$	(Söderberg, 1936)	(v)
Power		
$\epsilon_c = af(t)\sigma^b$	(Dorn, 1962)	(w)
$\epsilon_c = at^n\sigma^b; \quad 0 < n < 1, \quad b > 1;$ Bailey–Norton law	(Bailey, 1935) (Norton, 1929)	(x)
Hyperbolic		
$\dot{\epsilon}_c = a\sinh(b\sigma)$	(Ludwik, 1908) (McVetty, 1943)	(y)
$\dot{\epsilon}_c = a\sinh(b\sigma/RT)$	(Feltham, 1953)	(z)

ϵ denotes total strain, ϵ_c creep strain, $\boldsymbol{\sigma}$ stress, T temperature, t time, ln the natural logarithm, exp the exponential e, and a, b, c,..., A, B, C,... parameters that may be functions of $\boldsymbol{\sigma}$, t, T or they may be constants. Time derivative is denoted by a dot over a symbol (e.g.,$\in^{o}$). The notation f(x) denotes a function of x.


Strain
(a)
Three stages creep curve
ε_r
ε_2
ε_1
ε_0
t_1
t_2
t_r
(b)
Two stages creep curve
ε_r
ε_m
ε_0
(c)
t_m
t_r
Logarithmic creep curve
ε_0
Time
log (strain rate or creep rate)
(d)
Steady-state creep rate
$\dot{\varepsilon}_s$
(e)
t_1
t_2
t_r
Minimum creep rate
$\dot{\varepsilon}_{min}$
(f)
t_m
t_r
Time

Creep Under Multiaxial Loading

(text 14-14)

Use Levy-Mises Equations in plasticity

$$\sigma_{eff} = \frac{1}{\sqrt{2}}\sqrt{(\sigma_1-\sigma_2)^2 + (\sigma_2-\sigma_3)^2 + (\sigma_3-\sigma_1)^2}$$

and $$d\varepsilon_1 = \frac{d\varepsilon_{eff}}{\sigma_{eff}}\left[\sigma_1 - \frac{1}{2}(\sigma_2+\sigma_3)\right],$$

since creep is plastic deformation 1/2 appears as in plasticity.

Similarly, $d\varepsilon_2$ and $d\varepsilon_3$.

Dividing by dt, get the corresponding creep-rates,

$$\dot{\varepsilon}_1 = \frac{\dot{\varepsilon}_{eff}}{\sigma_{eff}}\left[\sigma_1 - \frac{1}{2}(\sigma_2+\sigma_3)\right], \text{ etc.}$$

One first determines the uniaxial creep-rate equation,

$$\dot{\varepsilon}_s = A\,\sigma^n e^{-Q/RT}$$

and assume the same for effective strain-rate : $\dot{\varepsilon}_{eff} = A\,\sigma_{eff}^{n}\, e^{-Q/RT}$

so that $$\boxed{\dot{\varepsilon}_1 = A\,\sigma_{eff}^{n-1}\, e^{-Q/RT}\left[\sigma_1 - \frac{1}{2}(\sigma_2+\sigma_3)\right]} \quad \text{etc.}$$

Stress Relaxation

(1) Elastic strain, ε_e;
(2) Time-independent plastic strain, ε_p, that occurs on loading;
(3) Time-dependent creep strain, ε_c.

The total strain after loading is therefore given as:

$$\varepsilon_t = \varepsilon_e + \varepsilon_p + \varepsilon_c = \text{constant} \tag{7.72}$$

$$0 = \frac{d\varepsilon_e}{dt} + 0 + \frac{d\varepsilon_c}{dt}; \quad \text{or,} \quad \frac{d\varepsilon_e}{dt} = -\frac{d\varepsilon_c}{dt}$$

But for a linear elastic material, $\varepsilon_e = \sigma/E$, where σ is the instantaneous stress, which is a function of time, and E is the elastic modulus. Hence,

$$\frac{d\varepsilon_e}{dt} = \frac{1}{E}\frac{d\sigma}{dt} \tag{7.74}$$

Substituting (7.74) into (7.73) and then integrating with respect to time t, one obtains

$$\int_{\varepsilon_p}^{\varepsilon_c} \frac{d\varepsilon_c}{dt} dt = -\frac{1}{E}\int_{\sigma_i}^{\sigma} \frac{d\sigma}{dt} dt$$

$$\therefore \varepsilon_c - \varepsilon_p = \frac{\sigma_i - \sigma}{E} \tag{7.75}$$

where ε_p is the time-independent plastic strain at time $t = 0$, when relaxation begins, and σ_i is the initial stress at the same instant. Equation (7.75) shows that as σ decreases ε_c increases. Creep therefore occurs under conditions of decreasing stress.

$$\frac{d\varepsilon_c}{dt} = \dot{\varepsilon}_s = A_2'\sigma^n$$

$$\frac{1}{E}\frac{d\sigma}{dt} = -A_2'\sigma^n \tag{7.77}$$

Equation (7.77) is the differential equation for the idealized case of stress relaxation where steady-state creep at a low stress level and a fixed total strain, ε_t, are assumed. Integrating (7.77), we get

$$\begin{aligned} \int \frac{d\sigma}{\sigma^n} &= -A_2'E\int dt \\ -\frac{1}{(n-1)\sigma^{n-1}} &= -A_2'Et + C \end{aligned} \tag{7.78}$$

where C is the integration constant. At the start of testing, i.e. at time $t = 0$, the initial stress $\sigma = \sigma_i$. From this condition, C can be evaluated from (7.78):

$$C = -\frac{1}{(n-1)\sigma_i^{n-1}} \tag{7.79}$$

Substitution of the value of C from (7.79) into (7.78) gives the relation between stress and time in stress relaxation as follows:

$$\frac{1}{\sigma^{n-1}} = \frac{1}{\sigma_i^{n-1}} + A_2'E(n-1)t \tag{7.80}$$

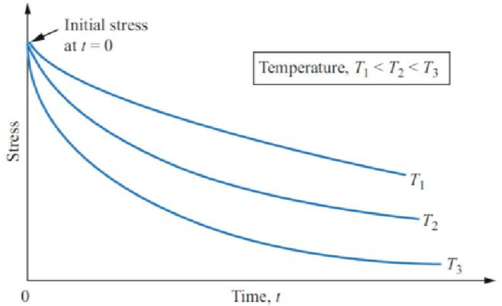
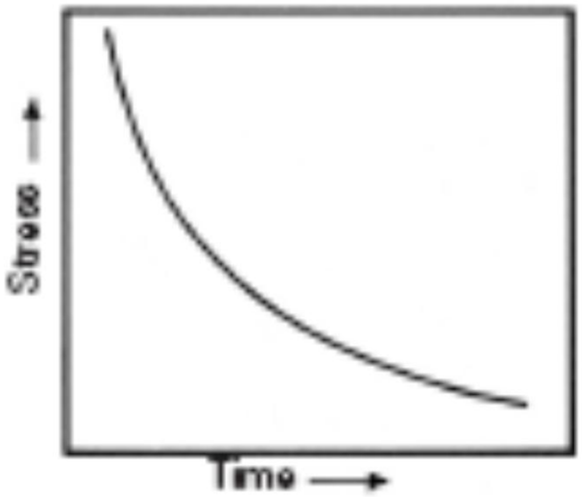


Fig. 7.41 Schematic stress-relaxation curves for a given material at various temperatures for the same initial stress, showing the rate of decrease of stress increases with increasing temperature

Stress Relaxation



7.16.2. A steel bolt clamping two rigid plates together is kept over a period of 5 years at a constant temperature of 650 °C. It is found that the stress (σ in MPa) dependence of steady-state creep rate ($\dot{\varepsilon}_s$ in s^{-1}) for this steel at 650 °C is given by $\dot{\varepsilon}_s = \text{constant}\,(\sigma)^5$. Test of the bolt steel at this temperature indicates that $\dot{\varepsilon}_s = 7 \times 10^{-9}\ s^{-1}$ at a stress of 41 MPa. If Young's modulus of the steel at 650 °C is

124 GPa and the stress in the bolt must not drop below 3 MPa during the 5 years, determine the initial stress to which the bolt must be tightened.

Solution

Given that $\dot{\varepsilon}_s = \text{constant}\,(\sigma)^5$, at 650 °C. Further, it is indicated that when the stress is $\sigma = 41$ MPa, the steady-state creep rate is $\dot{\varepsilon}_s = 7 \times 10^{-9}\ s^{-1}$, therefore

$$\text{Constant, say, } A'_2 = \frac{\dot{\varepsilon}_s}{\sigma^5} = \frac{7 \times 10^{-9}}{(41)^5}\ \text{MPa}^{-5}\ s^{-1}$$
$$= 6.04197 \times 10^{-17}\ \text{MPa}^{-5}\ s^{-1}$$

Since it is given that the dependence of steady-state creep rate ($\dot{\varepsilon}_s$) on stress (σ) is governed by power relation, so the relation between stress and time in stress relaxation will be given by (7.80), which is:

$$\frac{1}{\sigma^{n-1}} = \frac{1}{\sigma_i^{n-1}} + A'_2 E(n-1)t$$

where

σ the stress remaining after 5 years = 3 MPa;
n the power index of stress (σ) in the relation between steady-state creep rate ($\dot{\varepsilon}_s$) and stress (σ) = 5;
σ_i the initial stress in MPa, which is to be determined;
A'_2 the constant in the relation between steady-state creep rate ($\dot{\varepsilon}_s$) and stress (σ) = $6.04197 \times 10^{-17}\ \text{MPa}^{-5}\ s^{-1}$;
E Young's modulus of the steel at 650 °C = 124×10^3 MPa;
t the time in seconds = $(5 \times 365.25 \times 24 \times 3600)\ s = 157.788 \times 10^6$ s;

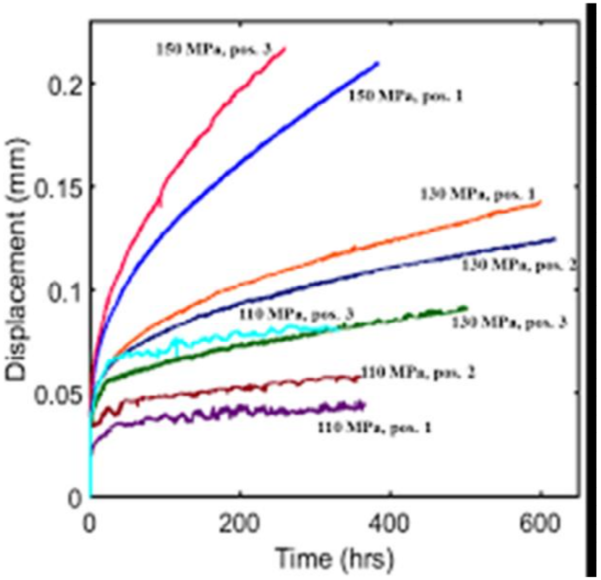
Hence, substituting the above values into (7.80), we get:

$$\frac{1}{\sigma_i^{5-1}} = \frac{1}{(3)^{5-1}} - (6.04197 \times 10^{-17}) \times (124 \times 10^3) \times (5-1) \times (157.788 \times 10^6)$$
$$= 7.617 \times 10^{-3}\ \text{MPa}^{-4}.$$

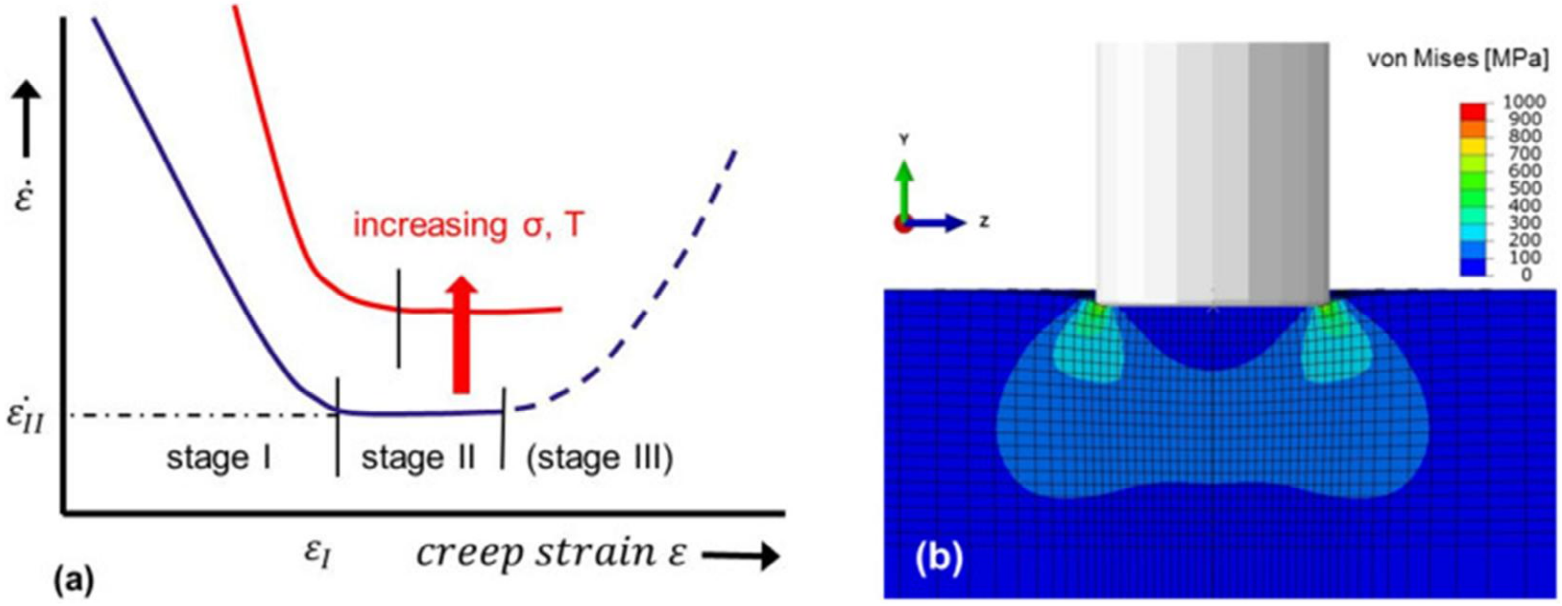
$$\text{Or,}\quad \sigma_i^4 = \frac{1}{7.617 \times 10^{-3}} = 131.285\ \text{MPa}^4;$$

$$\therefore \sigma_i = (131.285)^{\frac{1}{4}}\text{MPa} = \underline{3.385\ \text{MPa}}.$$

Impression Creep–New Creep Test



New flat-punch indentation creep testing



factors that influencing functional life of components at elevated temp?

1) Creep
2) Corrosion
3) High temperature fracture
4) Thermo mechanical Fatigue
5) Micro structural changes
6) Metallurgical ageing and metallurgical stability
7) Interaction of all above with each other

Design Considerations to avoid Creep

- Reduce the effect of grain boundaries:
  - Use single crystal material with large grains.
  - Addition of solid solutions to eliminate vacancies.
- Employ materials of high melting temperatures.
- Consult Creep Test Data during materials Selection
  - Type of service application
  - Set adequate inspection intervals according to life expectancy.

- High Temp Behavior of Materials:

Mechanical degradation

Chemical Degradation

- Gas Turbine and jet Turbine
- Nuclear reactors
- Power plants
- Spacecraft
- Chemical processing

$$\varepsilon_{ss}^{ceramics} < \varepsilon_{ss}^{metals} \ll \varepsilon_{ss}^{polymers}$$

Creep resistance

- High melting point → E.g. Ceramics
- Dispersion hardening → ThO_2 dispersed Ni (~0.9 T_m)
- Solid solution strengthening
- Single crystal / aligned (oriented) grains

Rules for Increasing Creep Resistance

- Large Grain Size
  (directionally solidified superalloys)

- Low Stacking Fault Energy
  (Cu vs Cu-Al alloys)

- Solid Solution Alloying
  (Al vs Al-Mg alloys)

- Dispersion Strengthening
  (Ni vs TD-Ni)

Parameters affects the creep

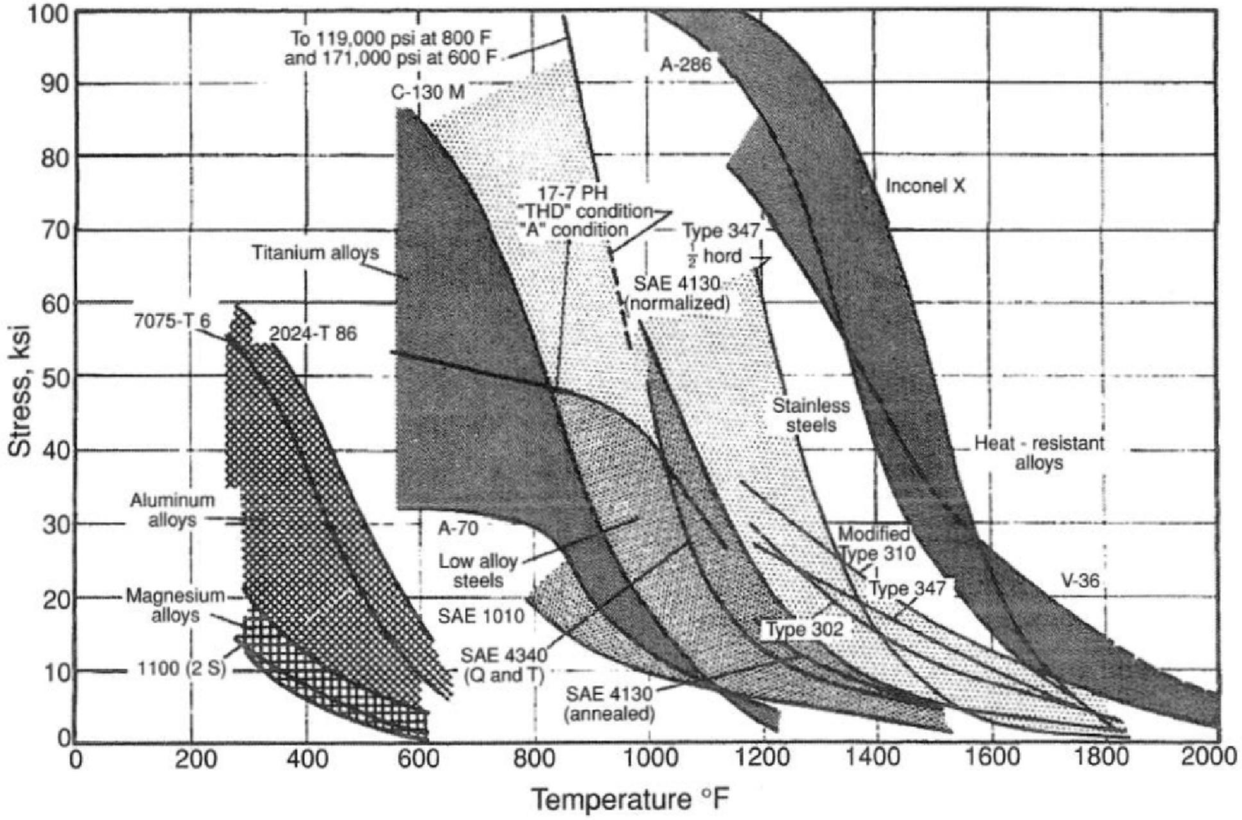
1- Microstructure (grain size, columnar grain)

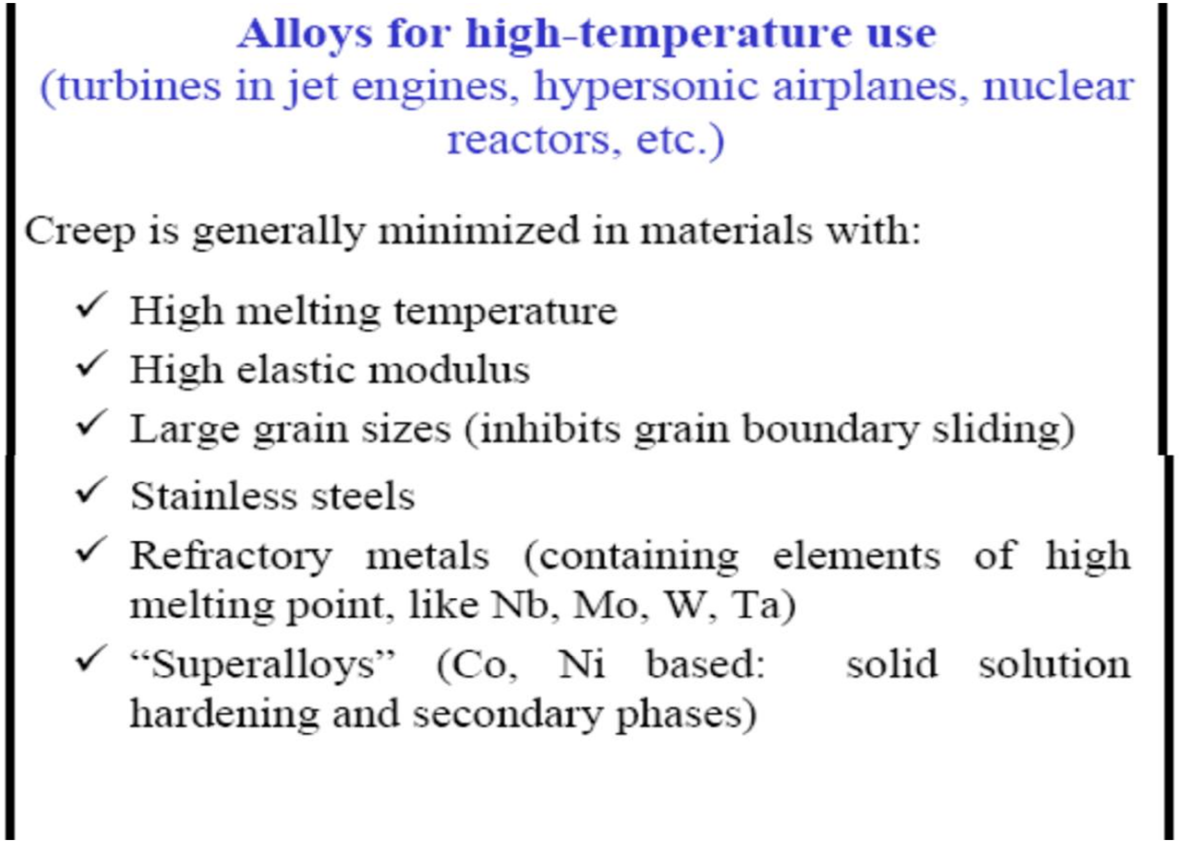
2- Chemical composition (solid solution and precipitates)and purity

3- Melting temperature

4- Production technique

Alloys for high temperature applications





Alloys for high-temperature use

(turbines in jet engines, hypersonic airplanes, nuclear reactors, etc.)

Creep is generally minimized in materials with:

- ✓ High melting temperature
- ✓ High elastic modulus
- ✓ Large grain sizes (inhibits grain boundary sliding)
- ✓ Stainless steels
- ✓ Refractory metals (containing elements of high melting point, like Nb, Mo, W, Ta)
- ✓ "Superalloys" (Co, Ni based: solid solution hardening and secondary phases)

2477 ºC
2896 ºC
3695 ºC
3290 ºC

Effect of structure and properties on creep resistance:

- ε• ↑ with: ↑ applied stress, ↑ diffusivity (↓ activation energy), ↓ grain size
- For a given application T, use of higher T_{mp} material will ↓diffusivity and correspondingly ↓ ε_{ss}
- In general:
  - Diffusivities in bcc metals >> diffusivities in fcc metals
  - if we have a bcc metal and an fcc metal with the same T_{mp}, then

    ε• *bcc* > ε• *fcc*
- ↑ grain size, ↓ ε• for diffusion controlled creep but does not have any influence on the dislocation creep mechanism.

- For dislocation glide mechanism: ↑ in grain size will ↑ε•

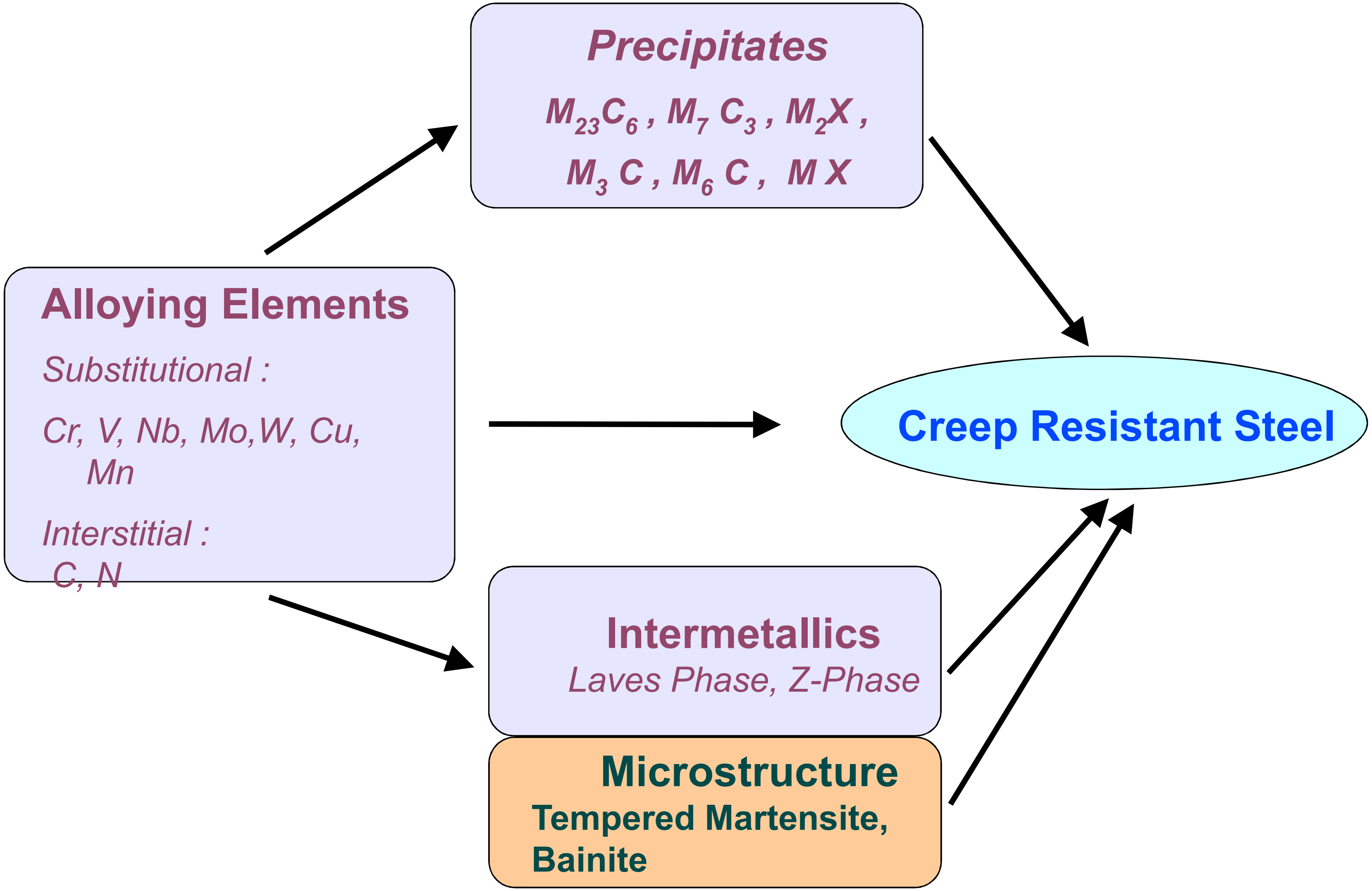
Resistance to Creep

⇒ Solid solution hardening

⇒ Precipitate hardening

⇒ Microstructure

Heat Resisting Steel



Precipitates

$M_{23}C_6$, M_7 C_3 , M_2X ,

M_3 C , M_6 C , M X

Alloying Elements

Substitutional :

Cr, V, Nb, Mo,W, Cu, Mn

Interstitial :

C, N

Creep Resistant Steel

Intermetallics

Laves Phase, Z-Phase

Microstructure

Tempered Martensite, Bainite

Creep resistance alloys

1- superalloy

2- stainless steel (ferritic and austenitic matrix): 9-12 % Cr, Mo, V (9Cr-1Mo-0.2V steel)

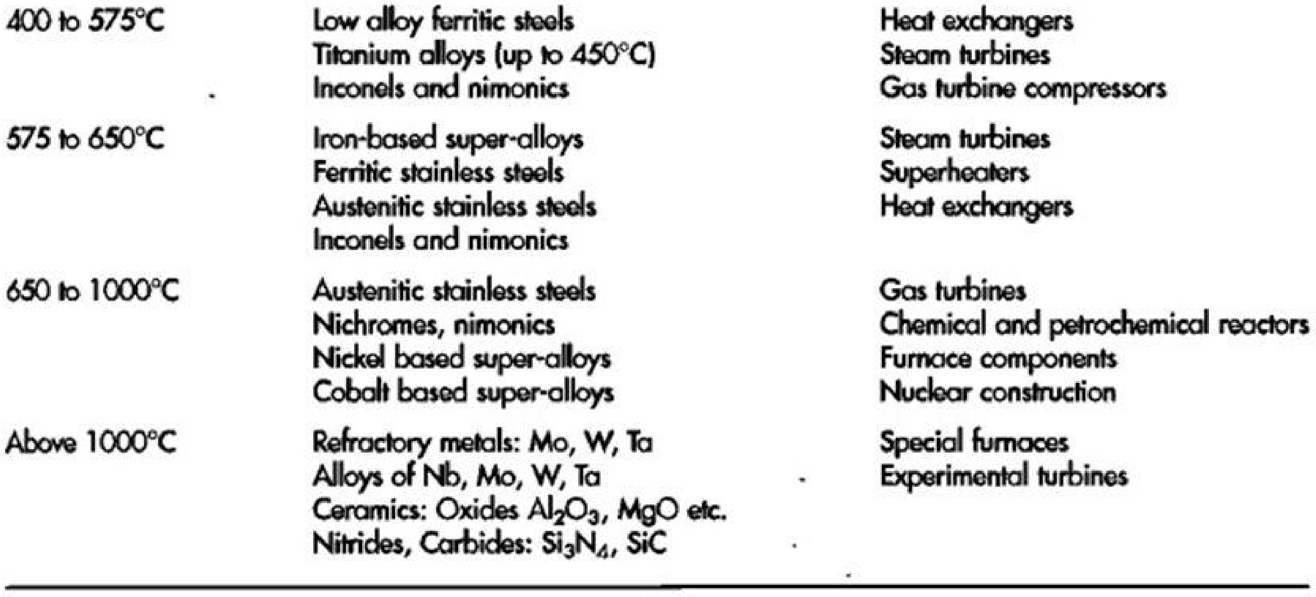
with temperature up to 550 °C and 650 °C, respectively

3- high melting point metals

4- cast-iron

5-aluminum

Creep Resist Materials



400 to 575°C	Low alloy ferritic steels Titanium alloys (up to 450°C) Inconels and nimonics	Heat exchangers Steam turbines Gas turbine compressors
575 to 650°C	Iron-based super-alloys Ferritic stainless steels Austenitic stainless steels Inconels and nimonics	Steam turbines Superheaters Heat exchangers
650 to 1000°C	Austenitic stainless steels Nichromes, nimonics Nickel based super-alloys Cobalt based super-alloys	Gas turbines Chemical and petrochemical reactors Furnace components Nuclear construction
Above 1000°C	Refractory metals: Mo, W, Ta Alloys of Nb, Mo, W, Ta Ceramics: Oxides Al_2O_3, MgO etc. Nitrides, Carbides: Si_3N_4, SiC	Special furnaces Experimental turbines

A ***superalloy****, or* ***high-performance alloy****, is an alloy with the ability to operate at a high fraction of its melting point.*

Several key characteristics of a superalloy are excellent mechanical strength, resistance to thermal creep deformation, and resistance to corrosion or oxidation.

The crystal structure is typically face-centered cubic (FCC) austenitic.

Superalloys develop high temperature strength through solid solution strengthening and precipitation strengthening from secondary phase precipitates such as gamma prime and carbides.

Oxidation or corrosion resistance is provided by elements such as aluminium and chromium.

1- **Ni-based superalloys** have emerged as the material of choice for these applications.

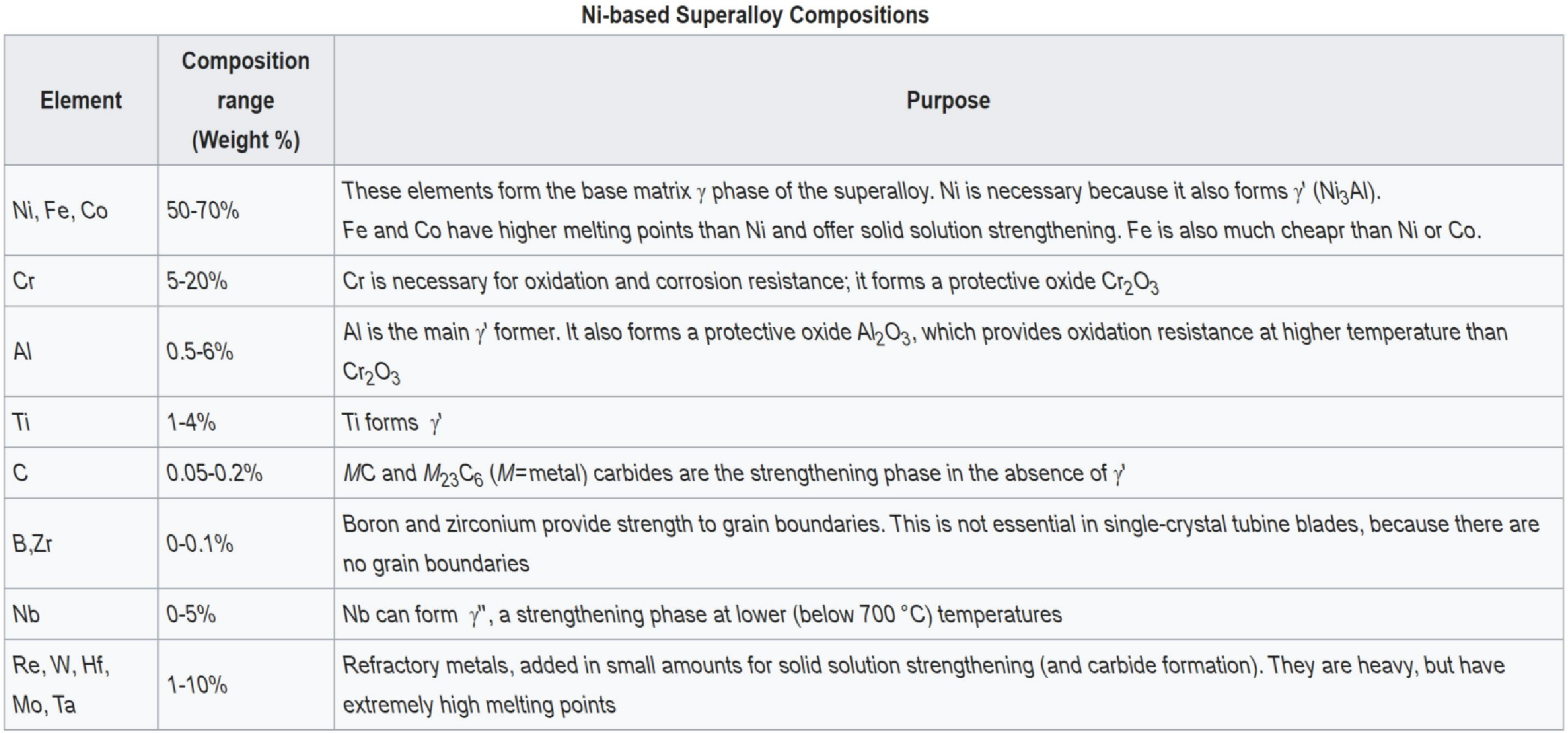
2- **Co-based superalloys** potentially possess superior hot corrosion, oxidation, and wear resistance as compared to Ni-based superalloys.

3-**Fe-based superalloy**

The use of steels in superalloy applications is of interest because certain steel alloys have showed creep and oxidation resistance similar to that of Ni-based superalloys, while being far less expensive to produce.

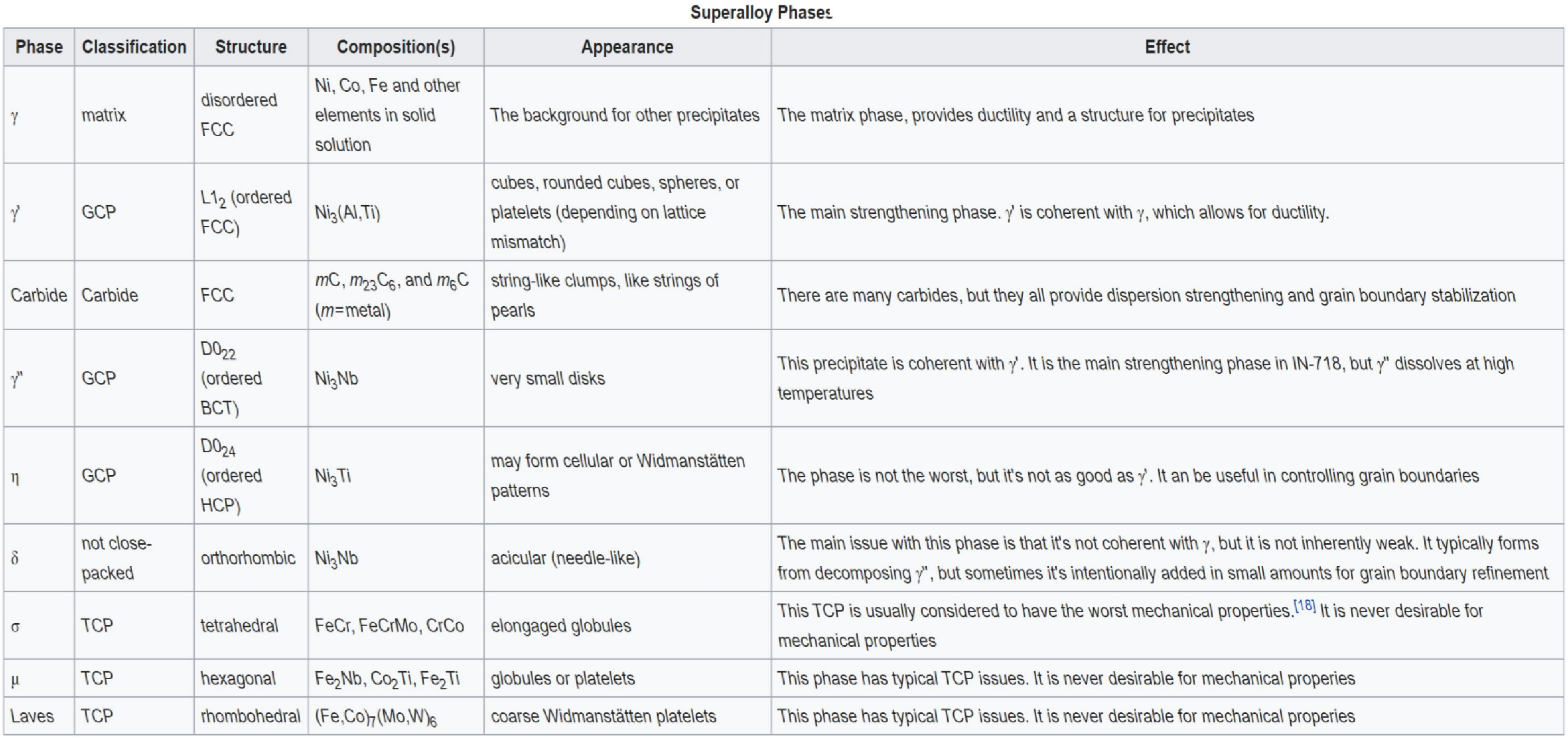
Super alloys including not only metals, but also nonmetals; chromium, iron, cobalt, molybdenum, tungsten, tantalum, aluminium, titanium, zirconium, niobium, rhenium, yttrium, vanadium, carbon, boron or hafnium are some examples of the alloying additions used.

Each of these additions has been chosen to serve a particular purpose in optimizing the properties for high temperature application.



Ni-based Superalloy Compositions

Element	Composition range (Weight %)	Purpose
Ni, Fe, Co	50-70%	These elements form the base matrix γ phase of the superalloy. Ni is necessary because it also forms γ' (Ni_3Al). Fe and Co have higher melting points than Ni and offer solid solution strengthening. Fe is also much cheapr than Ni or Co.
Cr	5-20%	Cr is necessary for oxidation and corrosion resistance; it forms a protective oxide Cr_2O_3
Al	0.5-6%	Al is the main γ' former. It also forms a protective oxide Al_2O_3, which provides oxidation resistance at higher temperature than Cr_2O_3
Ti	1-4%	Ti forms γ'
C	0.05-0.2%	MC and $M_{23}C_6$ (M=metal) carbides are the strengthening phase in the absence of γ'
B,Zr	0-0.1%	Boron and zirconium provide strength to grain boundaries. This is not essential in single-crystal tubine blades, because there are no grain boundaries
Nb	0-5%	Nb can form γ'', a strengthening phase at lower (below 700 °C) temperatures
Re, W, Hf, Mo, Ta	1-10%	Refractory metals, added in small amounts for solid solution strengthening (and carbide formation). They are heavy, but have extremely high melting points



Superalloy Phases

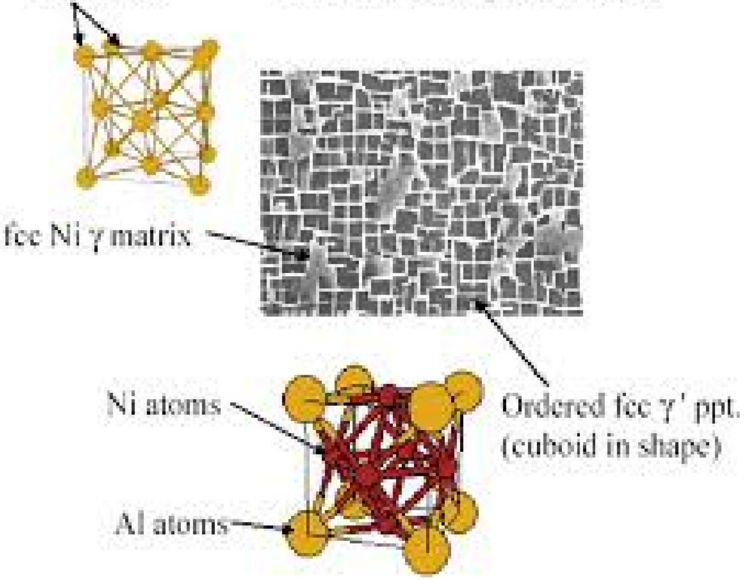
Phase	Classification	Structure	Composition(s)	Appearance	Effect
γ	matrix	disordered FCC	Ni, Co, Fe and other elements in solid solution	The background for other precipitates	The matrix phase, provides ductility and a structure for precipitates
γ'	GCP	$L1_2$ (ordered FCC)	$Ni_3(Al,Ti)$	cubes, rounded cubes, spheres, or platelets (depending on lattice mismatch)	The main strengthening phase. γ' is coherent with γ, which allows for ductility.
Carbide	Carbide	FCC	mC, $m_{23}C_6$, and m_6C (m=metal)	string-like clumps, like strings of pearls	There are many carbides, but they all provide dispersion strengthening and grain boundary stabilization
γ''	GCP	$D0_{22}$ (ordered BCT)	Ni_3Nb	very small disks	This precipitate is coherent with γ'. It is the main strengthening phase in IN-718, but γ'' dissolves at high temperatures
η	GCP	$D0_{24}$ (ordered HCP)	Ni_3Ti	may form cellular or Widmanstätten patterns	The phase is not the worst, but it's not as good as γ'. It an be useful in controlling grain boundaries
δ	not close-packed	orthorhombic	Ni_3Nb	acicular (needle-like)	The main issue with this phase is that it's not coherent with γ, but it is not inherently weak. It typically forms from decomposing γ'', but sometimes it's intentionally added in small amounts for grain boundary refinement
σ	TCP	tetrahedral	FeCr, FeCrMo, CrCo	elongaged globules	This TCP is usually considered to have the worst mechanical properties.[18] It is never desirable for mechanical properties
μ	TCP	hexagonal	Fe_2Nb, Co_2Ti, Fe_2Ti	globules or platelets	This phase has typical TCP issues. It is never desirable for mechanical properies
Laves	TCP	rhombohedral	$(Fe,Co)_7(Mo,W)_6$	coarse Widmanstätten platelets	This phase has typical TCP issues. It is never desirable for mechanical properies

-The γ’ phase is a cubic L12-structure Ni3(Al, Ti, Ta, Nb) phase that produces cuboidal precipitates. Superalloys often have a high (60-75%) volume fraction of γ’ precipitates. γ’ precipitates are coherent with the parent γ phase, and are resistant to shearing due to the development of an anti-phase boundary when the precipitate is sheared.

-The γ” phase is a tetragonal Ni3Nb or Ni3V structure. The γ” phase, however, is unstable above 650 °C, so γ” is less commonly used as a strengthening phase in high temperature applications.

-Carbides are also used in polycrystalline superalloys to inhibit grain boundary sliding.

Ni base superalloys



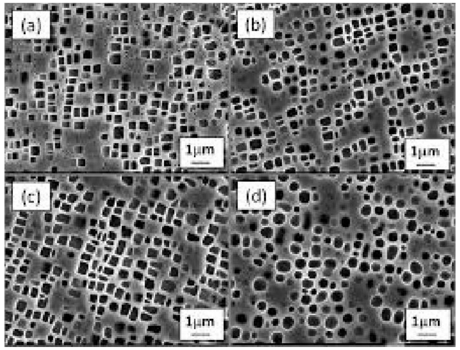
Ni atoms

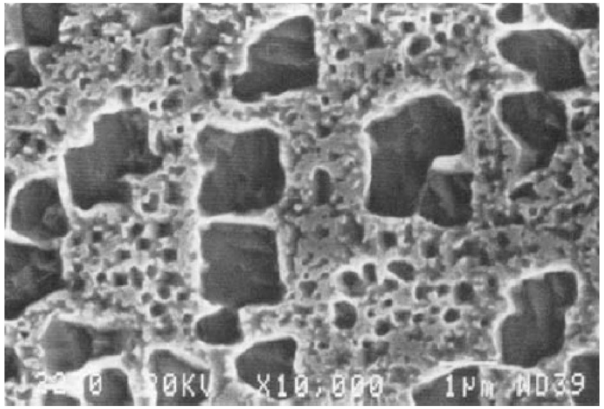
fcc Ni γ matrix

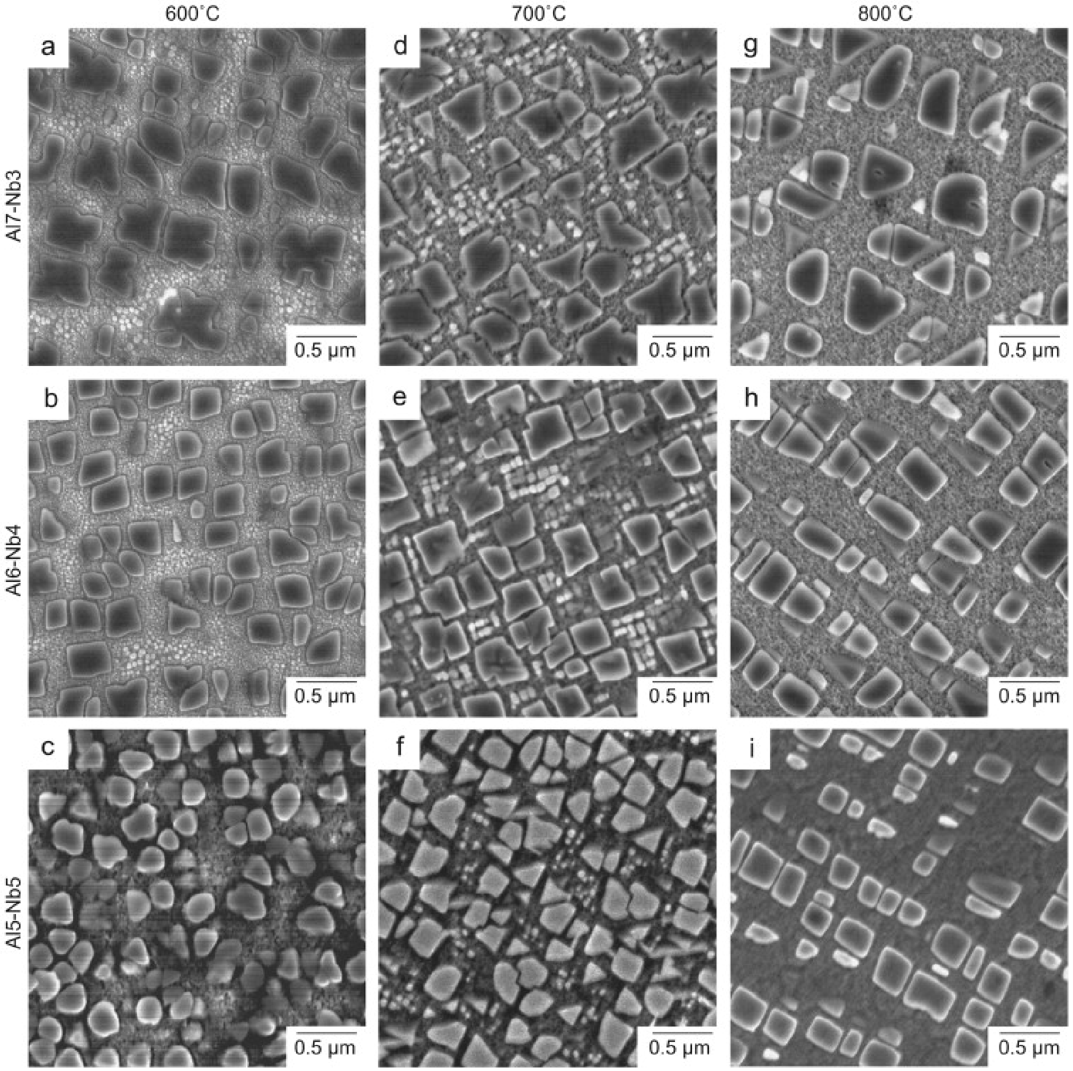
Ni atoms

Al atoms

Ordered fcc γ' ppt.
(cuboid in shape)





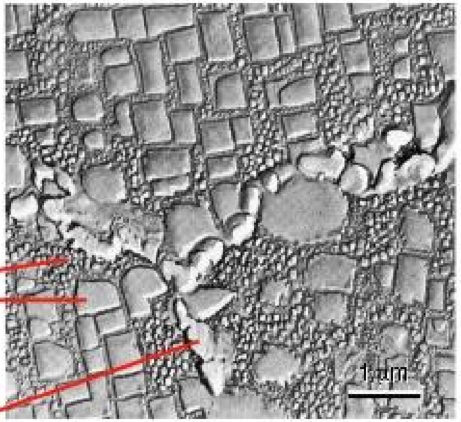

600˚C
700˚C
800˚C
Al7-Nb3
Al6-Nb4
Al5-Nb5
a
b
c
d
e
f
g
h
i
0.5 µm

SUPERALLOYS

STRENGTHENING MECHANISM

Three strengthening mechanisms are used in Ni superalloys:

- **Solid solution hardening**
- **Coherent precipitate hardening**
- **Carbide phases on grain boundaries**



Examples of such alloys are Hastelloy, Inconel, Waspaloy, Rene alloys, Incoloy, MP98T, TMS alloys, and CMSX single crystal alloys.

Composition of some high temperature alloys

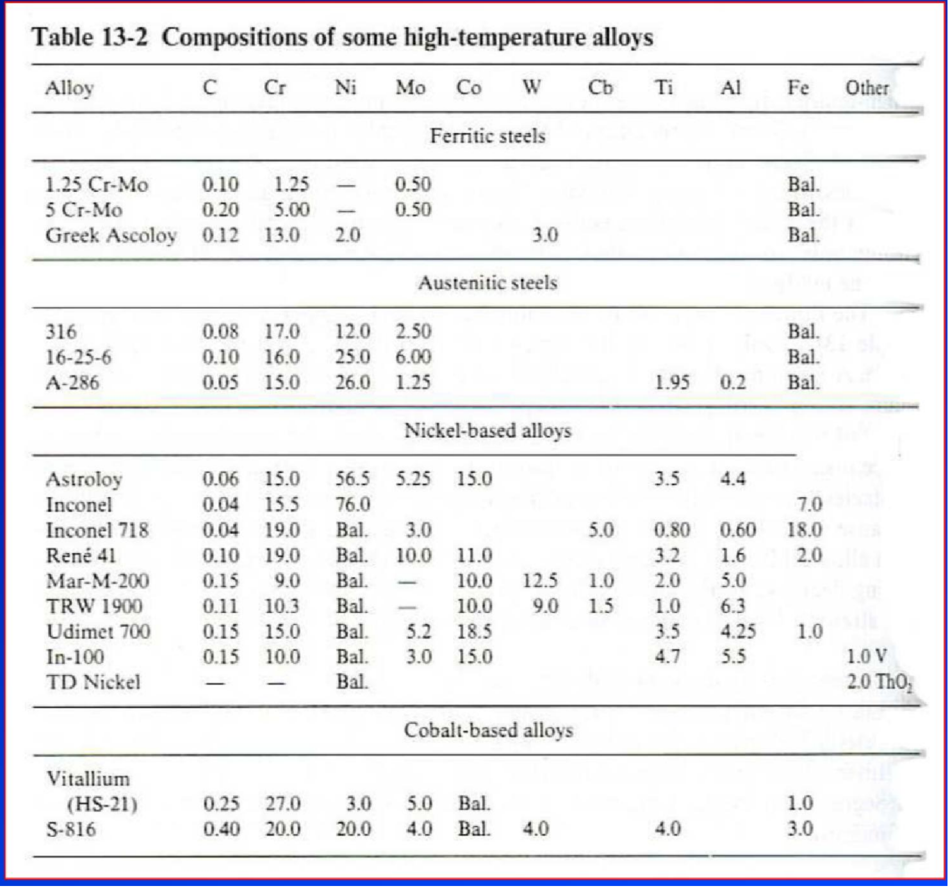


Table 13-2 Compositions of some high-temperature alloys

Alloy	C	Cr	Ni	Mo	Co	W	Cb	Ti	Al	Fe	Other
Ferritic steels											
1.25 Cr-Mo	0.10	1.25	—	0.50						Bal.	
5 Cr-Mo	0.20	5.00	—	0.50						Bal.	
Greek Ascoloy	0.12	13.0	2.0			3.0				Bal.	
Austenitic steels											
316	0.08	17.0	12.0	2.50						Bal.	
16-25-6	0.10	16.0	25.0	6.00						Bal.	
A-286	0.05	15.0	26.0	1.25				1.95	0.2	Bal.	
Nickel-based alloys											
Astroloy	0.06	15.0	56.5	5.25	15.0			3.5	4.4		
Inconel	0.04	15.5	76.0							7.0	
Inconel 718	0.04	19.0	Bal.	3.0			5.0	0.80	0.60	18.0	
René 41	0.10	19.0	Bal.	10.0	11.0			3.2	1.6	2.0	
Mar-M-200	0.15	9.0	Bal.	—	10.0	12.5	1.0	2.0	5.0		
TRW 1900	0.11	10.3	Bal.	—	10.0	9.0	1.5	1.0	6.3		
Udimet 700	0.15	15.0	Bal.	5.2	18.5			3.5	4.25	1.0	
In-100	0.15	10.0	Bal.	3.0	15.0			4.7	5.5		1.0 V
TD Nickel	—	—	Bal.								2.0 ThO_2
Cobalt-based alloys											
Vitallium (HS-21)	0.25	27.0	3.0	5.0	Bal.					1.0	
S-816	0.40	20.0	20.0	4.0	Bal.	4.0		4.0		3.0	

Phase Formation

Adding new elements is usually good because of solid solution strengthening, but engineers need to be careful about which phases precipitate.

Precipitates can be classified as geometrically close-packed (GCP), topologically close-packed (TCP), or carbides. GCP phases are usually good for mechanical properties, but TCP phases are often deleterious. Because TCP phases are not truly close packed, they have few slip systems and are very brittle. They are additionally bad because they "scavenge" elements away from GCP phases.

Many elements that are good for forming γ' or have great solid solution strengthening may precipitate TCPs.

Engineers need to find the balance.

An area of the alloy with TCP phase formation will be weak because

- *the TCP phase has inherently poor mechanical properties*
- *the TCP phase is incoherent with the γ matrix*
- *the TCP phase is surrounded by a "depletion zone" where there is no γ'*
- *the TCP phase usually forms sharp plate or needle-like morphologies which easily nucleate cracks*

187. Comment on the statement 'Failure due to creep occurs in several cases long before rupture occurs'.

Ans. Statement is correct. Failure occurs when the dimensions of the part undergoing creep exceed the permissible tolerance limits such that it stops functioning. Thus failure due to creep occurs long before rupture occurs.

188. Give temperatures at and above which creep becomes pronounced for aluminium alloys, titanium alloys and alloys of refractory metals.

Ans. Aluminium alloys and titanium alloys creep significantly above 200°C and 325°C, respectively. Alloys of refractory metals, such as tungsten, molybdenum, niobium and tantalum, creep significantly at temperatures lying in between 1000°C and 1500°C.

189. Describe the effect of alloying elements on the creep deformation response of metallic materials.

Ans. Almost all alloying elements forming solid solutions with base metal or forming additional hard phases raise the resistance of the metallic materials against creep deformation. The alloying elements forming substitutional solid solution and decrease stacking fault energy of the solid solution are more effective to resist creep deformation. Similarly interstitial solute elements are more effective than substitutional alloying elements in resisting creep deformation.

190. Whether the alloying elements forming solid solution or those forming additional hard phase, will raise the creep resistance of the material to grater extent.

Ans. Alloying element forming additional hard phase will raise the creep resistance of the material to greater extent than the alloying element forming solid solution. Finer the dispersion of hard second phase better will be the resistance against creep.

191. Explain why low alloy steels containing molybdenum, vanadium or tungsten as alloying element are extensively used as creep resisting steels.

Ans. Molybdenum, vanadium and tungsten are strong carbide forming elements. Carbides of these elements are present in the steel in the form of fine dispersoids which resist creep. Thus, these steels offer good resistance against creep.

192. What is the effect of nature of dispersed phase on creep strength of the material?

Ans. Coarser second phase particles provide better creep strength at higher temperatures, whereas finer particles impart enhanced creep strength at lower temperatures.

193. Describe the effect of grain size on the creep strength of a material.

Ans. As a general rule, it can be stated that creep strength is better for coarse grained materials at elevated temperatures (more precisely above the equicohesive temperature) while creep strength is better for fine grained materials below the equicohesive temperature.

194. Explain why equicohesive temperature is of much importance with reference to creep.

Ans. Equicohesive temperature is the temperature at which strength of grain boundary is equal to the strength of grain interior. Below this temperature, grain boundaries are stronger than the grains interior. Reverse is true above this temperature. As creep occurs due to grain boundary sliding above equicohesive temperature so this temperature is of much importance with reference to creep. A coarse grain structure seems to be desirable above this temperature but a fine grain structure is preferred below this temperature.

195. Does strain hardening affect creep strength of a material?

Ans. Yes, strain hardening affects creep strength of a material. Strain hardening by prior cold working will enhance creep strength of the material up to those service temperatures that are lying below the recrystallization temperature.

196. Creep test is usually performed at a constant temperature and under constant load (not at constant stress) conditions. Why?

Ans. Creep test is usually performed at a constant temperature and under constant load conditions as these conditions match with service conditions. In actual service conditions, it is not possible to maintain constant stress condition.

197. Which stage is of much importance to a design engineer dealing with creep and why?

Ans. Secondary stage is of much importance to a design engineer dealing with creep, as creep rate is almost constant during this stage. In primary stage and tertiary stage, material creeps rather quickly. This is why creep life is determined by secondary stage.

198. Describe the effect of enhancing temperature on the creep behaviour of a material.

Ans. Creep rate increases with increase in the temperature. At higher temperatures, thermally activated processes such as recovery, dislocation climb, cross slip, grain boundary sliding, etc., take place more easily. All these factors account for enhanced creep rates at higher temperatures.

199. What is the effect of enhancing stress on the creep behaviour of a material?

Ans. Material will creep at a faster rate on enhancing the stress. Increased stress will offset the effect of work hardening, and so, formation of internal voids will take place much earlier.

200. Is there any well-defined relationship between creep rate of secondary stage and temperature? What will be the effect of increasing stress on this creep rate?

Ans. For a material, creep rate at secondary stage is related to the service (test) temperature by an equation similar to Arrhenius type equation

i.e. $$\dot{\varepsilon}_s = Ae^{-H/RT} \qquad \text{(i)}$$

where $\dot{\varepsilon}_s$ is the creep rate of secondary stage
R is the universal gas constant
T is the temperature in Kelvin
H is the activation energy for creep and
A is a pre-exponential constant whose value depends on the materials nature, applied stress and temperature, the entropy change. Thus, this is not a true constant.

Creep rate of secondary stage increases with increase in the stress (σ) and the two are related by the relation

$$\dot{\varepsilon}_s = B\sigma^n$$

where B and n are constants

By combining Eqs. (i) and (ii), one gets the relation

$$\dot{\varepsilon}_s = K\sigma^n e^{-H/RT}$$

where, K and n are constants. The value of n usually varies between 3 and 8.

201. What is the effect of increasing temperature or stress on secondary stage of creep?

Ans. Creep rate of secondary stage will increase and duration of secondary stage will decrease by increasing the temperature or the stress.

202. Name the factors controlling the shape of creep curve.

Ans. The shape of the creep curve for a material depends mainly on two factors, namely, the temperature and the stress at any time. It is because that these factors control the work hardening and recovery process governing the creep rates at various stages of creep curve.

203. As a design engineer, what information(s) you would like to have while dealing with a problem in which creep is important. Is a typical creep curve of any significance?

Ans. While dealing with a problem in which creep is important, a design engineer is highly interested in knowing about the permissible (tolerable) creep strain during the expected life of the component or about the maximum possible stress that can be applied on the component over an expected period of time provided the actual creep strain is always below the prescribed specified limit for permissible creep strain.

Such information cannot be achieved by typical creep curve for the material under consideration, and hence, a typical creep curve is not of much significance to a design engineer.

در تنش ۲۰ مگاپاسکال، در صورتی که میزان کرنش ایجاد شده (0.002) در یک آلیاژ آلومینیوم در دمای ۲۰۰ درجه سانتی گراد در زمان ۱۰۰ ثانیه ایجاد شده باشد و همین میزان کرنش در دمای ۲۵۰ درجه سانتی گراد در زمان ۳۵ ثانیه انجام شده باشد. الف– مطلوبست محاسبه میزان انرژی فعالسازی این آلیاژ ب– درصورتی که بخواهید این زمان به ۱۰۰۰ ثانیه برسد چه راه حلی را پیشنهاد می دهید (از نظر متالورژیکی و از نظر شرایط تست آزمون)

If stress= cte

$$Q = R\left(\ln\frac{t_1}{t_2}\right)\left(\frac{T_1 \cdot T_2}{T_2 - T_1}\right)$$

Q= 8.314 (ln 100/35)[(523*473)-(523-473)

Q=47.309 Kj/mol

4) Hertzberg, 5.9

Gas turbine component A was originally designed to operate at $700°C$ and exhibited a stress rupture life 800h. Component B in the same section of the turbine was redesigned, thereby allowing its operating temperature to be raised to $725°C$. Could component A be used at that temperature without modification so long as its stress rupture life exceed 100h? (Assume that the Larson-Miller constant for the material is equal to 20.)

Solution:

Using the Larson-Miller relation, we find

$$(273+700)(20+\log 800)=(273+725)(20+\log t)$$

$$\therefore \qquad t=212hr$$

As such, component A could be used at the higher temperature.

Example :

Calculate the time to rupture at 650°C and 100MPa stress for a 1%Cr-1% Mo-0.25%V steel, according to the Larson-Miller , Sherby--Dorn, and Manson--Haferd methods, if this alloy underwent rupture in 20hrs when tested in tension at the same stress level at a temperature of 750°C.

Solution:

The Larson-Miller equation is $T(\log t_r + C) = m$.
At 750°C, $T = 750 + 273 = 1{,}023^0$ K and $t_r = 20$ hours. Therefore,
$m = 1023 \times (\log 20 + 22) \approx 2.4 \times 10^4$
At 650°C, $T = 650 + 273 = 923^0$K, and we have
$923(\log t_r + 22) = 2.4 \times 10^4$, so that $\log t_r = (2.4 \times 10^4/923) - 22$
$t_r = 6.7 \times 10^3$ hours.

The Sherby-Dorn equation is $\log t_r - Q/(kT) = m$. From Table 1, $Q = 460$ kJ/mol. Because Q here involves moles, we must use R instead of k. At 750°C, $T = 1{,}023^0$ K and $t_r = 20$ hours. Thus, $m = \log 20 - (460 \times 10^3/8.314 \times 1023)$
At 650°C, $T = 923^0$ K, and we obtain $\log t_r = m + 0.43(Q/kT)$
so that **$t_r = 6 \times 10^3$ hours.**

The Manson-Haferd equation is $(\log t_r - \log t_a)/(T - T_a) = m$. From Table 1, $Ta = 311$ K, and $\log t_a = 18$. At 750°C, $T = 1{,}023^0$ K, and it follows that $t_r = 20$ hours. Therefore, $m = (\log 20 - 18)/(1{,}023 - 311) = -0.023$.
At 650°C, $T = 923^0$ K, and we have $(\log t_r - \log t_a)/(T - T_a) = m$
$(\log t_r - 18)/(923 - 311) = -0.023$,
$\log t_r = 3.924$,

$t_r = 8.4 \times 103$ hours.

Example 1:

A specimen has been subjected to creep. If the strain measured in stage || is 0.003 at 600h and 0.002 at 200h, calculate the steady state creep rate of the specimen.

Solution:

Since the steady-state creep rate is the slope of the strain versus time straight line the steady-state creep rate can be calculated as:

$$\text{Steady-state creep rate} = \frac{\Delta\epsilon}{\Delta t} = \frac{0.003-0.002}{600h-200h} = \frac{0.001}{400h} = 0.0000025\ \text{h}^{-1}$$

Or $$\dot{\varepsilon}_s = 2.5 \times 10^{-6}\ \text{h}^{-1}$$

Example 2:

A metallic alloy exhibits a steady-state creep rate of 2.5×10^{-4} h^{-1} at a temperature of 550 °C and 3.2×10^{-3} h^{-1} at 650 °C. Calculate the activation energy for creep.

Solution:

Since the steady-state creep rate is given at two different temperatures, and we wish to calculate the activation energy

The temperature need to be expressed in degrees K. therefore 550°C =550+273=823K and 650°C =650+273=923K. Substituting the appropriate values into this equation at those temperature, we obtain

$$\dot{\varepsilon}_s = C \exp(-Q/RT) \ldots\ldots \text{ (Mott, 1953)}$$

$$3.2 \times 10^{-3}\,\mathrm{h}^{-1} = C \exp(- \frac{Q}{8.314\ \mathrm{j\ mol^{-1}\ K^{-1}} \times (923K)})$$

$$2.5 \times 10^{-3}\,\mathrm{h}^{-1} = C \exp(- \frac{Q}{8.314\ \mathrm{j\ mol^{-1}\ K^{-1}} \times (823K)})$$

Dividing the above equations with each other, we get : $\frac{3.2 \times 10^{-3}\,\mathrm{h}^{-1}}{2.5 \times 10^{-3}\,\mathrm{h}^{-1}} = \exp[- \frac{Q}{8.314\ \mathrm{j\ mol^{-1}\ K^{-1}}} (\frac{1}{923} - \frac{1}{823})]$

$$1.28 \times 10\,\mathrm{h}^{-1} = \exp[- \frac{Q}{8.314\ \mathrm{j\ mol^{-1}\ K^{-1}}} (-1.316\ 10^{-4})]$$

Taking natural logarithms on both sides

$$\mathrm{Ln}(1.28 \times 10) = [(- \frac{Q}{8.314})(-1.316 \times 10^{-4})]$$

$$2.55 = Q \times 1.583 \times 10^{-5}$$

$$\mathbf{Q = 1.611 \times 10^5\ J\ mol^{-1}, \text{that is },\ 161\ kJ\ mol^{-1}}$$

درصورتی که میزان انرژی نقص انباشتگی برای دو آلیاژ A و B به ترتیب ۱۰۰ و ۲۰۰ ارگ (erg) بر سانتی متر مربع باشد. کدامیک از آلیاژهای ذیل را در شرایط خزشی انتخاب می کنید؟ قانون خزشی برای این آلیاژها از کدام قانون پیروی می کند؟