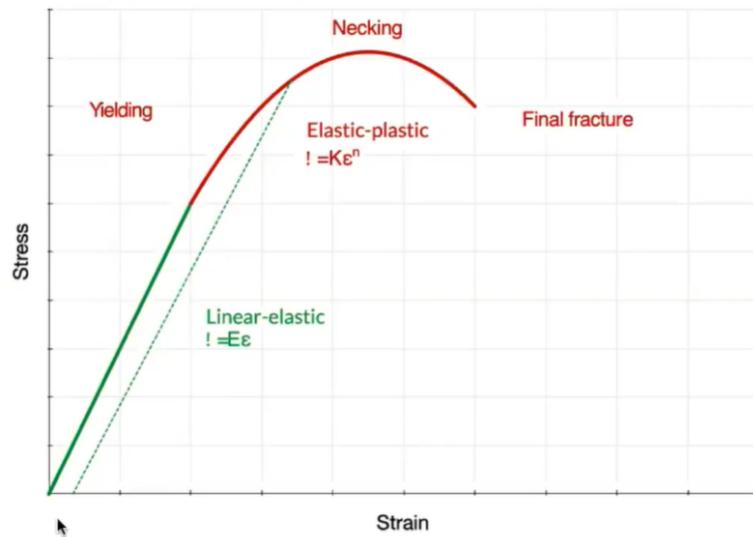


Macroscopic approximation

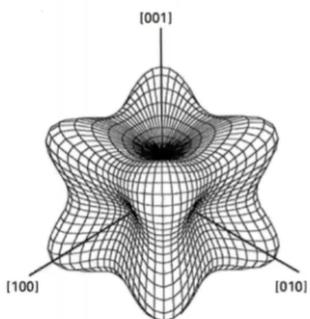


Macroscopic ≠ microscopic

**Homogenous
Isotropic
Elastic**

**Inhomogenous
Anisotropic
Inelastic**

Elastic anisotropy

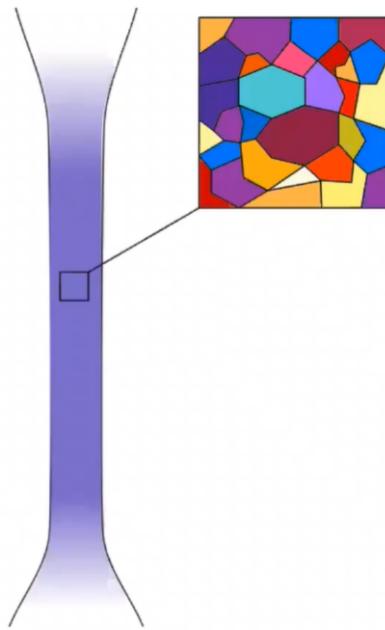


Material	$2(s_{11} - s_{12}) / s_{44}$	E_{111} (Gpa)	E_{100} (Gpa)	ratio
Al	1.219	76.1	63.7	1.19
Cu	3.203	191.1	66.7	2.87
Au	2.857	116.7	42.9	2.72
Fe	2.512	272.7	125.0	2.18
MgO	1.534	350.1	249.4	1.404
spinel	2.425	364.5	170.0	2.133
TiC	0.877	429.2	476.2	0.901
W	1	384.6	384.6	1

Micro level

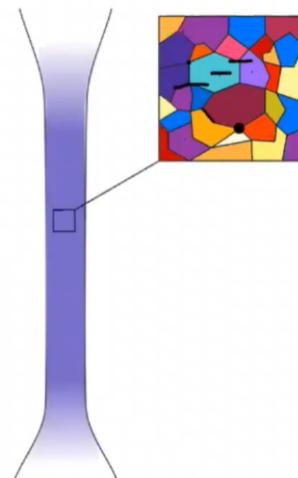
... grain orientation affects the effective stress

... grain geometry caused stress concentrations within grains

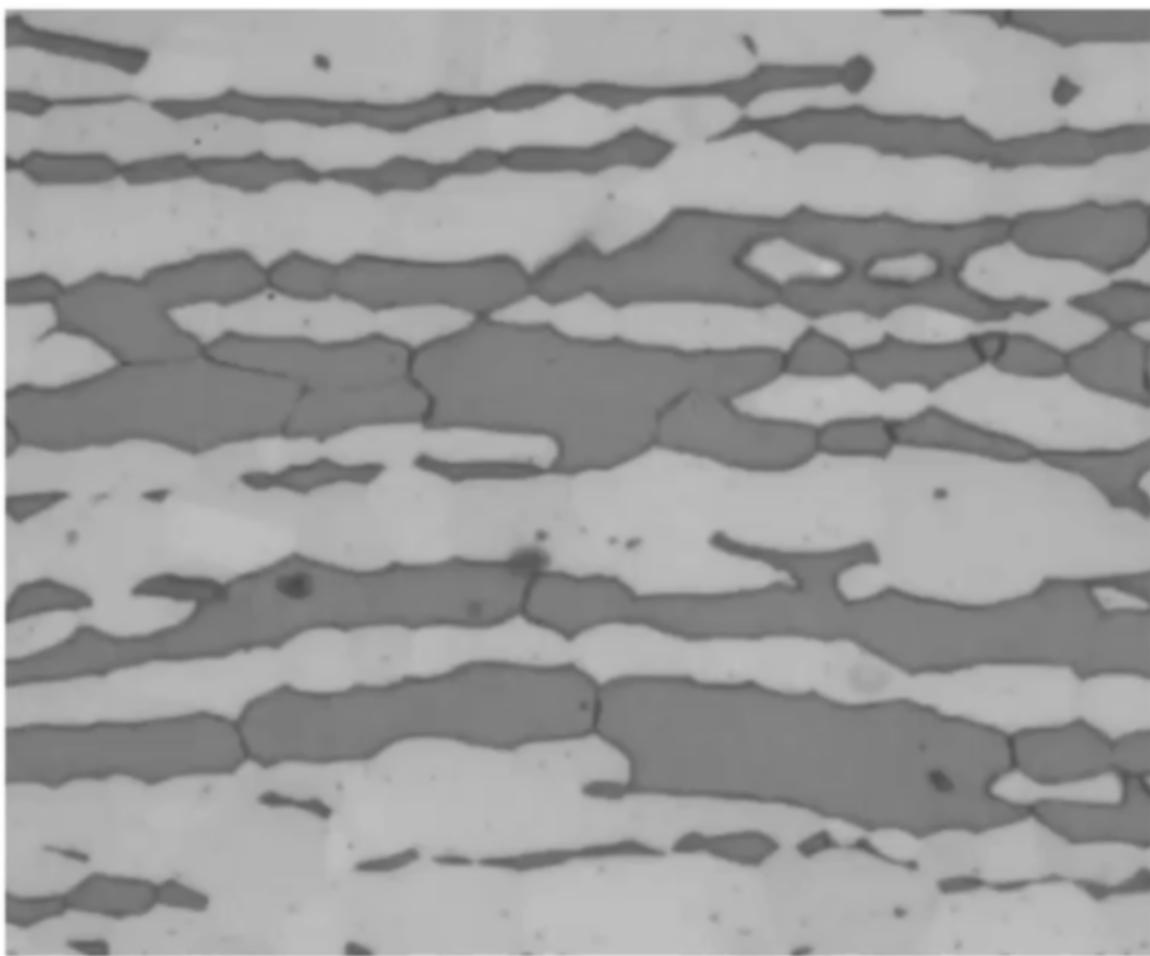
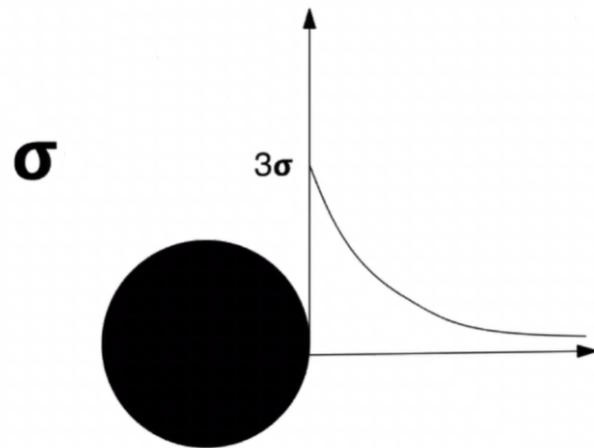


Inhomogenous material

Material has inclusions, precipitates, impurities and lattice defects. These cause stress concentrations in microscopic level.



Impurities cause stress concentrations

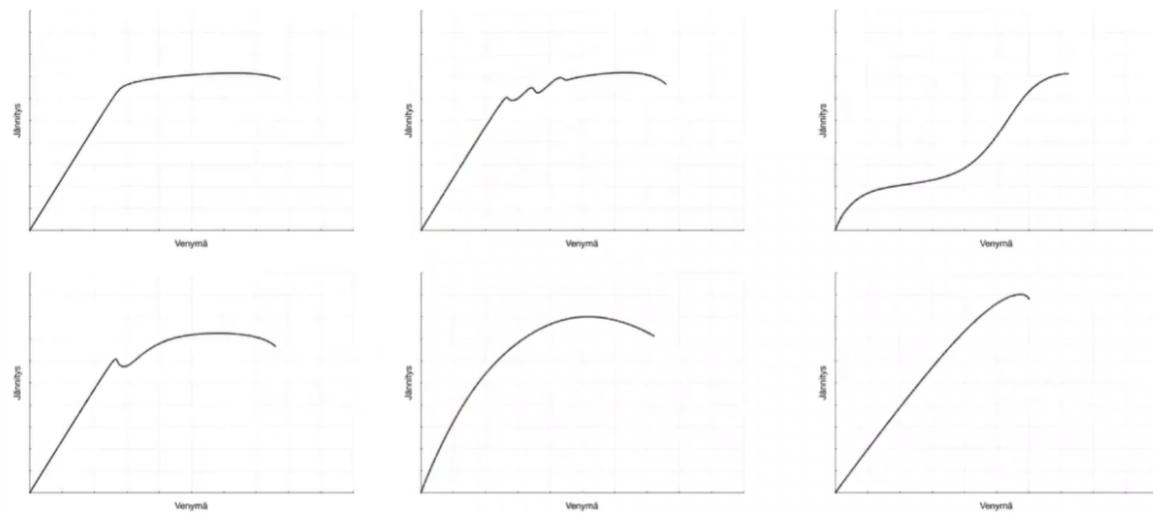


So:

In Linear-elastic region, behaviour is *macroscopically* reversible and even

On the microscopic level, stress varies with microstructure and local plastic yielding occurs in stress concentrations.

Local yielding evens stress concentrations

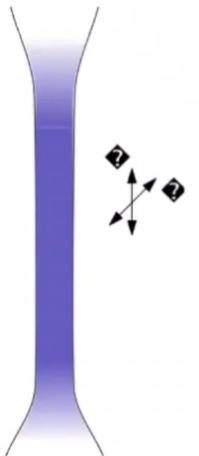


Yields in practice (but not in theory)

Theoretical strength can be calculated from Youngs modulus:
 $\tau = G/2\pi$

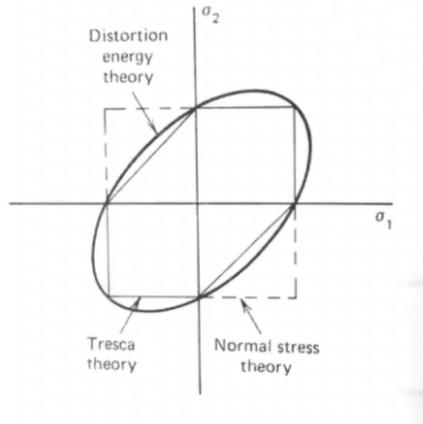
Real measured stresses 1 / 10000 of this
=> not all bonds can break at the same time

Plastic deformation



Greatest shear stress at 45°

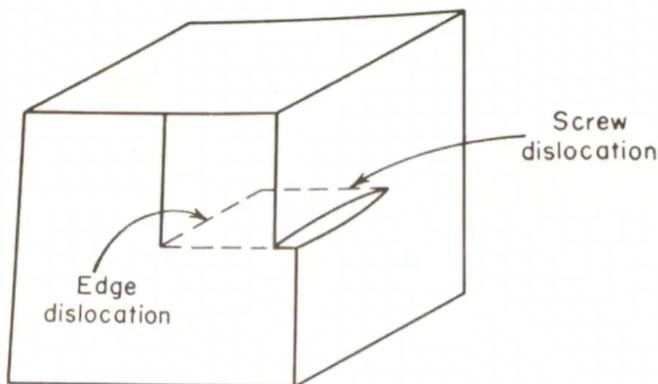
Shear stress activates sliding



"High enough" stress makes dislocation movement and generation possible.

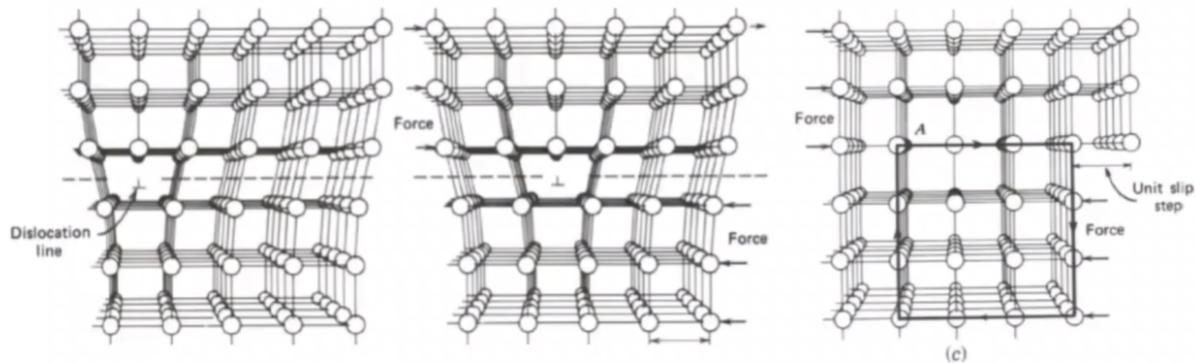
Slip begins when shear stress in the slip system is sufficient

Dislocations



Linear lattice defect
Edge and screw
dislocations
Dislocation movement
causes deformation

Edge dislocation



Screw dislocation

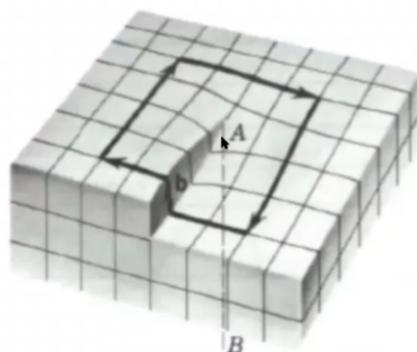


FIGURE 2.8 Screw dislocation AB resulting from displacement of one part of crystal relative to the other. Note that AB is parallel to b .

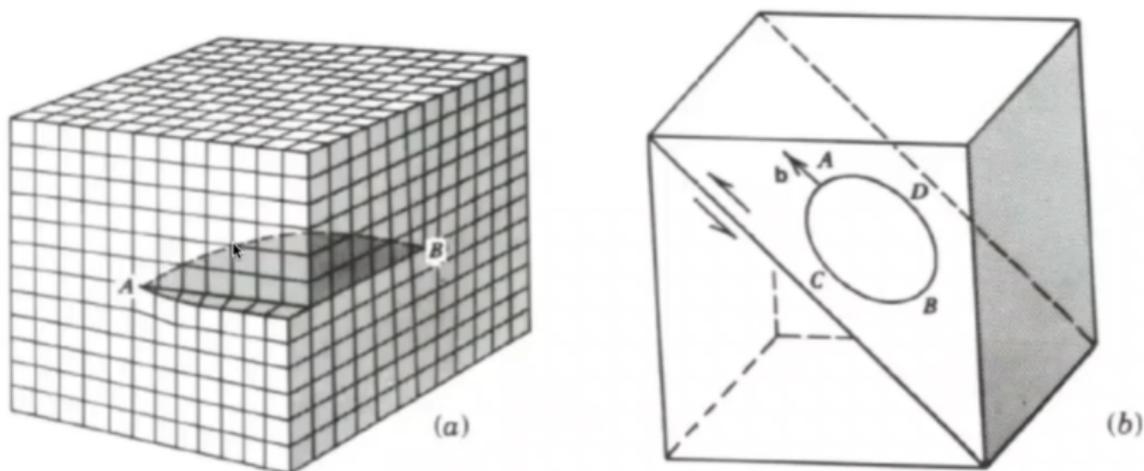


FIGURE 2.10 Curved dislocations containing edge and screw components. (a) Dislocation AB is pure screw at A and pure edge at B ; (b) dislocation loop that grows out radially with shear stress applied parallel to b .

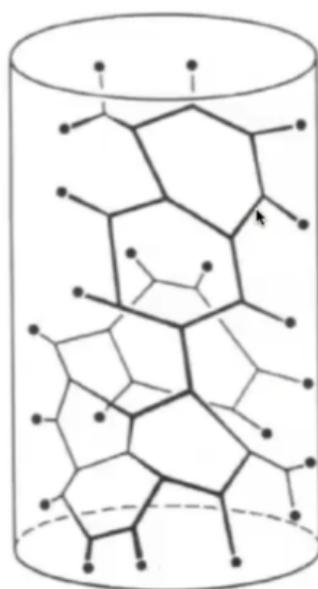


FIGURE 2.11 Network arrangement of dislocations in crystal. Dislocations can terminate only at a node, in a loop, or at a grain boundary or free surface.⁷ (Reproduced by courtesy of the Council of the Institution of Mechanical Engineers from *The Properties of Materials at High Rates of Strain*, by A. H. Cottrell.)

Dislocations in single crystal

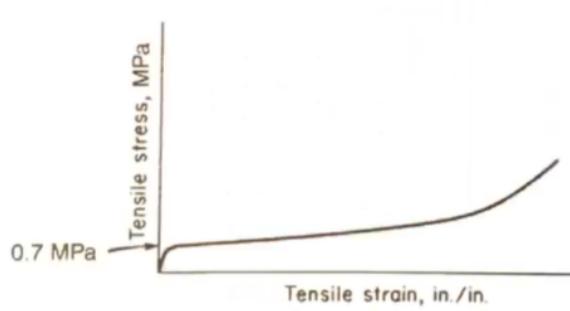


Fig. 4.1 Tensile stress-strain curve for a magnesium single crystal.



Fig. 4.2. Slip lines on magnesium crystal.

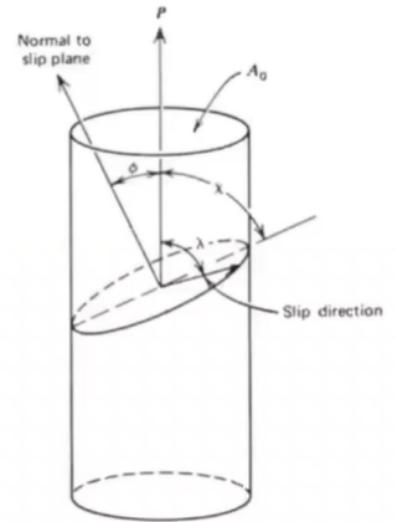
Peiers-Nabarro Stress

- In close packed planes Peiers-stress is negligible
- Slip concentrates to close packed levels (where possible)
- slip plane + direction= slip system
- BCC lattice has significant Peiers-stress

$$\tau = \frac{2G}{1-\nu} \exp\left(-\frac{2\pi\xi}{b}\right)$$

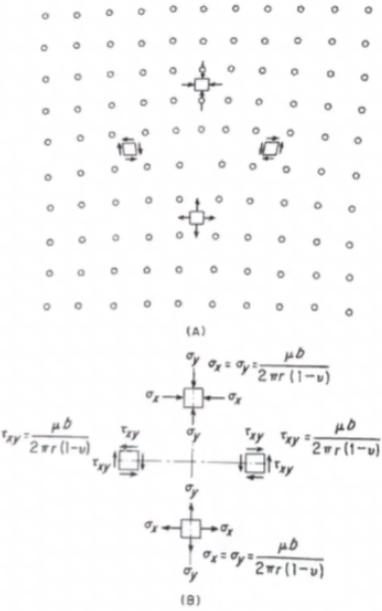
Schmid factor: $\cos\phi \cos\lambda$

$$\tau = P/A \cos\phi \cos\lambda$$



Energy of dislocations

Dislocations cause lattice distortion => energy
 $E = \alpha G b^2$ ($\alpha \approx 0.5 \dots 1$)



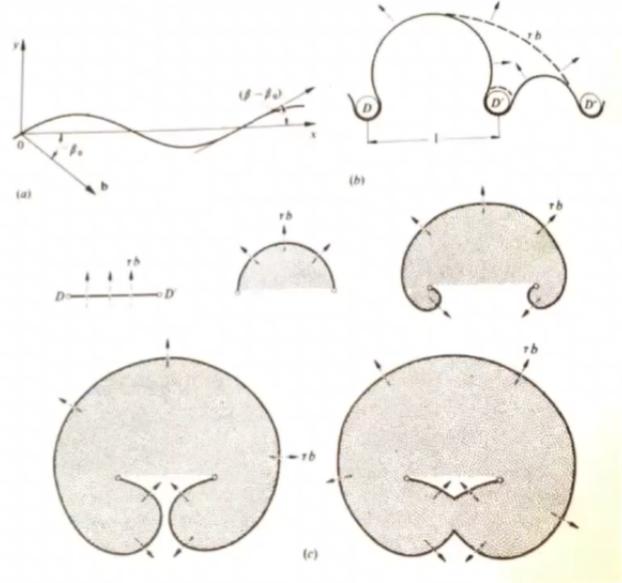
Energy of plastic deformation

Energy is changed to dislocation energy and work done to move dislocations.

The energy does not released once the load is removed

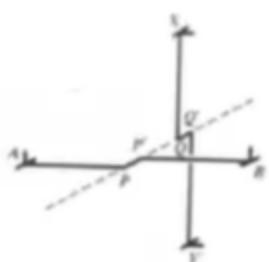
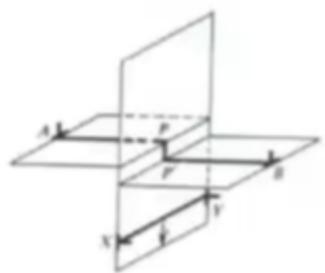
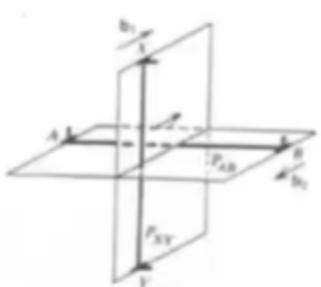
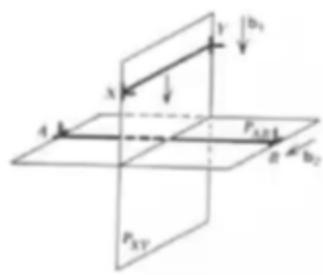
Dislocation generation

Frank-Read generator
Orowan stress $\tau = Gb/l$



Dislocation interaction

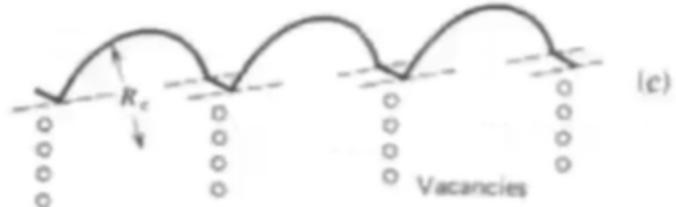
- **Bypassing edge dislocations, the stress-fields meet and increase others energy; bypassing requires stress $\tau \sim Gb/\sqrt{N}$**
- **Crossing dislocations "cut" each other and cause kinks to each other, which decrease mobility. Maximal shear stress $\tau \sim Gb\sqrt{N}$**
- **Dislocations on the same slip plane pack against each other when they meet an obstacle (like grain boundary)**
- **Needed stress increases with increasing dislocation density
=> strain hardening**



(a)

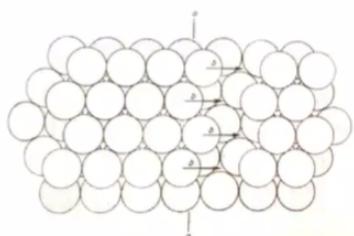
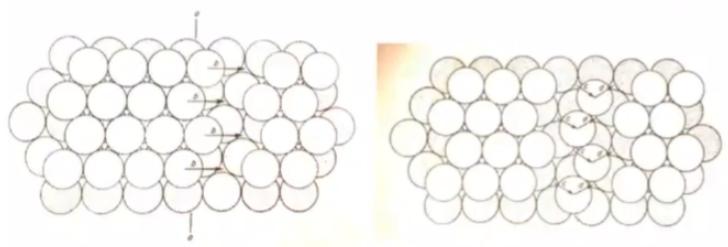
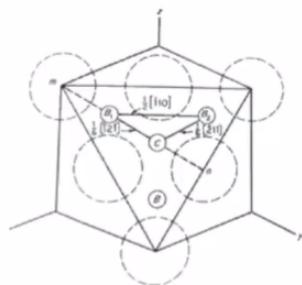
(b)

FIGURE 2.23 Intersection of two edge dislocations. (a) Burgers vectors are at right angles and produce an edge jog PP' in dislocation AB . (From Read,³ *Dislocations in Crystals*; copyright McGraw-Hill Book Co., New York. © 1953. Used with permission of McGraw-Hill Book Company.) (b) Burgers vectors are parallel and produce two screw jogs PP' and QQ' . (From Hull;¹¹ reprinted with permission from Hull, *Introduction to Dislocations*, Pergamon Press, Elmsford, NY, 1965.)



Stacking faults

In FCC-lattice dislocations may separate into Shockley partial dislocations and a stacking fault



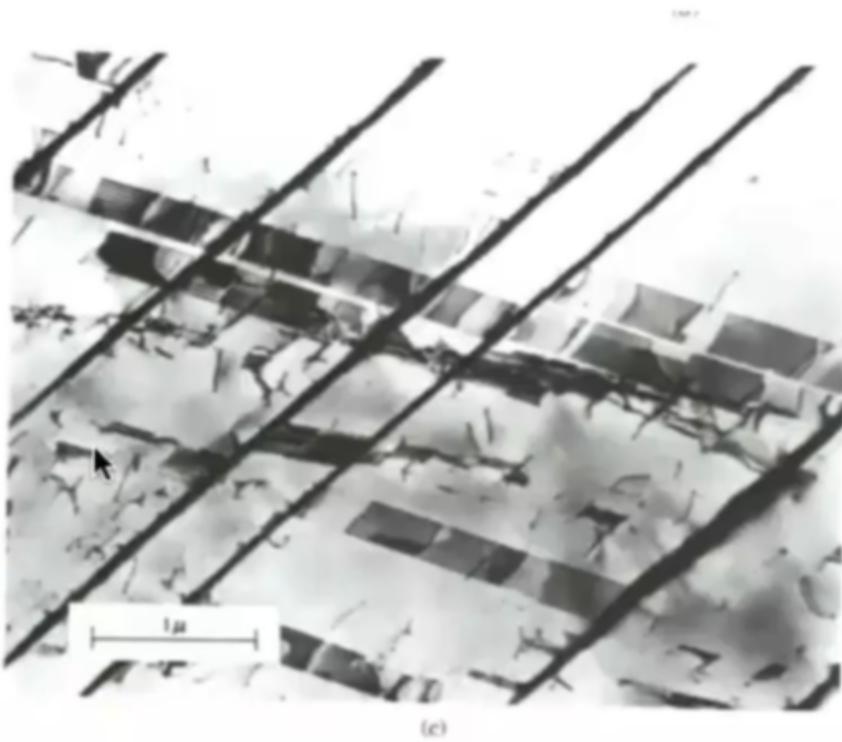


FIGURE 2.17 (a) Path of whole and partial (Shockley) dislocations; (b) Shockley b_2 and b_3 surrounding stacking fault region A; (c) Long stacking fault ribbons (bands of closely spaced lines) in low SFE 18Cr-8Ni stainless steel. Faults are bounded at ends by partial dislocations. Thin black bands are mechanical twins. (After Michalak,¹⁰ reprinted with permission from *Metals Handbook*, Vol. 8, American Society for Metals, Metals Park, OH, © 1973.)

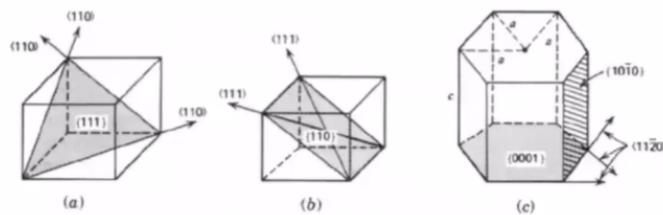
Multigrain material

Arbitrary deformation requires 5 independent slip systems

FCC: $\{111\}/<110>$; 12 slip systems, 5 independent

BCC: $\{110\}/<111>$ ($<112>$, $<123>$), 48 slip systems, 5 independent

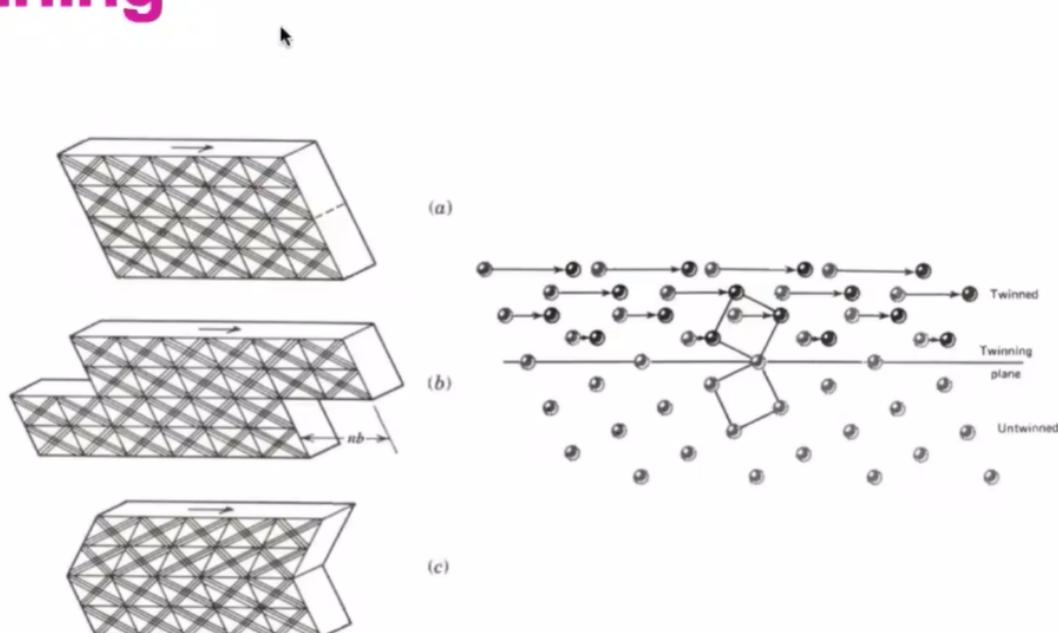
CPH: $\{10\bar{1}0\}$ / $<11\bar{2}0>$ or $\{0001\}/<11\bar{2}0>$; not enough independent slip systems



Planes in Hexagonal Crystals

Metal	c/a Ratio	Observed Slip Plane
Be	1.568	{0001}
Ti	1.587	{10\bar{1}0}
Zr	1.593	{10\bar{1}0}
Mg	1.623	{0001}
Co	1.623	{0001}
Zn	1.856	{0001}
Cd	1.886	{0001}

Twinning



Summary

FCC lattice has negligible Peiers stress

- Dislocations are easily created and moved
- Dislocations entangle and pile against grain boundaries

BCC-lattice

- Dislocation creation and movement is more difficult
- Less dislocations

CPH-lattice

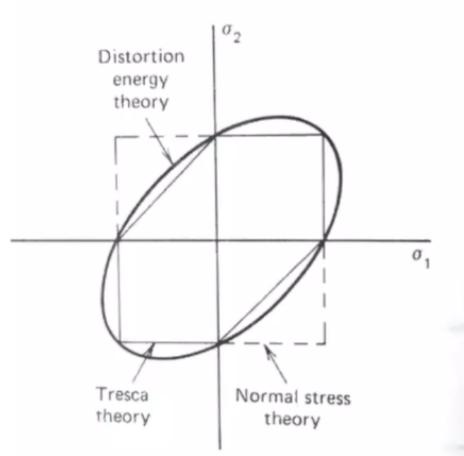
- Deformation requires non-conservative dislocation movement or twinning

... and it follows (usually)

FCC materials have low yield strength and high strain hardening

BCC materials have high yield strength and low strain hardening

Strain hardening



So:

- Plastic deformation increases dislocation density
- Increased dislocation density makes dislocation movement more difficult
- => Hardening

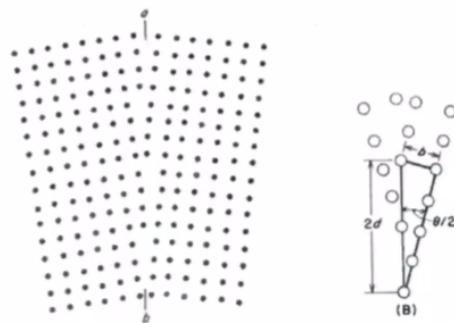
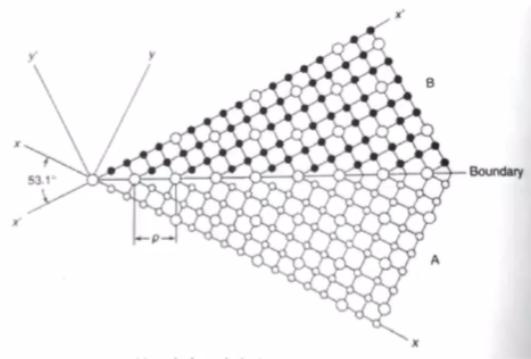
But:

- Kinematic, isotropic, or what?

Grain boundaries

Necessary dislocations mediate grain boundary mismatch

Dislocations may form small angle boundaries and twins

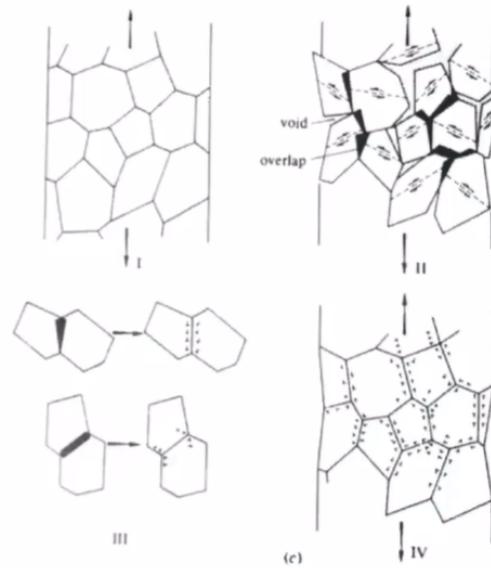


Grain boundary effects on dislocations

Dislocations may not cross grain boundaries

Dislocations pile on to grain boundaries

Pile-up causes stress on the next grain



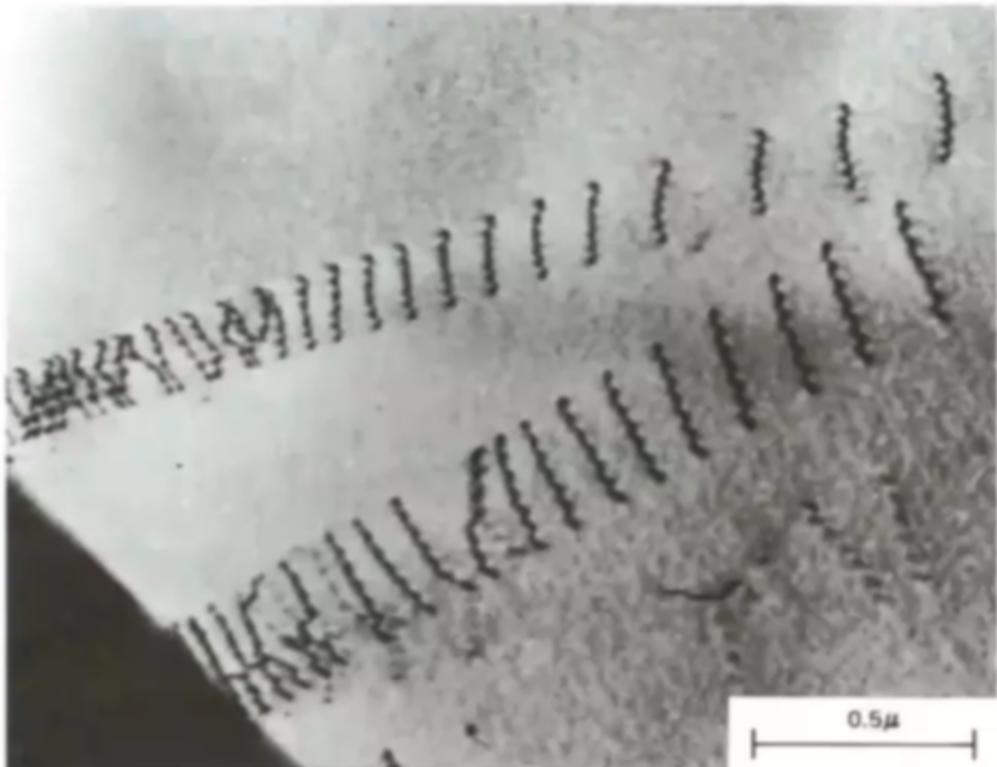


FIGURE 2.22 Dislocation pileups on two systems against a grain boundary in 309 stainless steel ($\gamma = 35 \text{ mJ/m}^2$). (Courtesy of Anthony Thompson, Carnegie-Mellon University.)

Solute atoms and dislocations

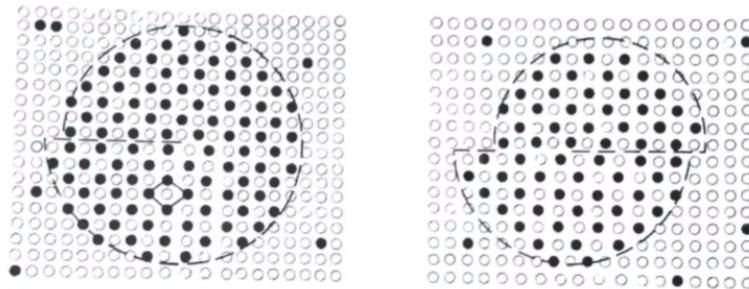
Stress fields of dislocations and solute atoms cause interaction

- Oversized solute atoms concentrate on tensile stress region (and interstitial atoms)
- Subsized concentrate on compressive area
- Dislocations develop a could of solute atoms (Cottrell cloud)
- These "lock in" the dislocation and cause straing-ageing

Coherent precipitates hinder movement

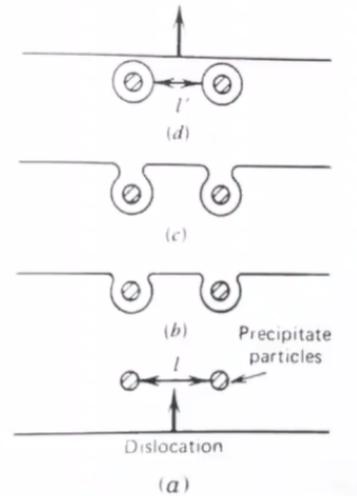
Dislocations may cross coherent precipitates (but this requires high energy)

Precipitates hinder dislocation movement



Incoherent precipitates and dispersions

Dislocations move around the dispersions



Necking

Necking decreases surface area

~
Strain hardening
At necking $n=\epsilon$

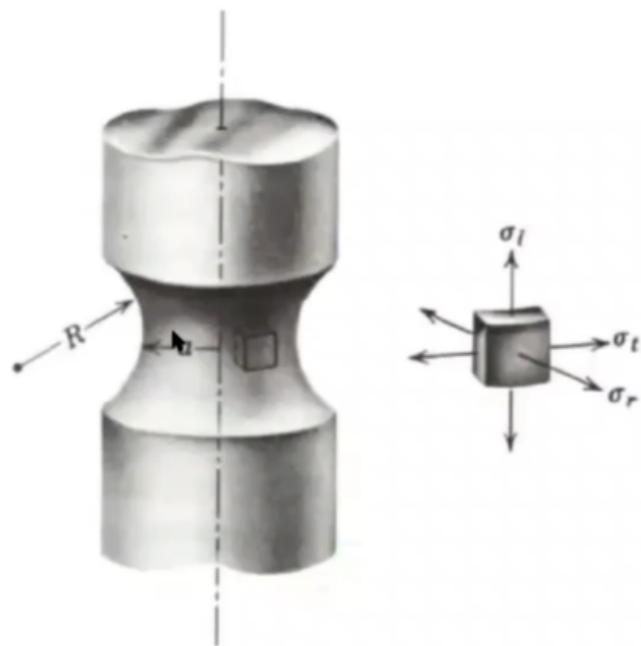
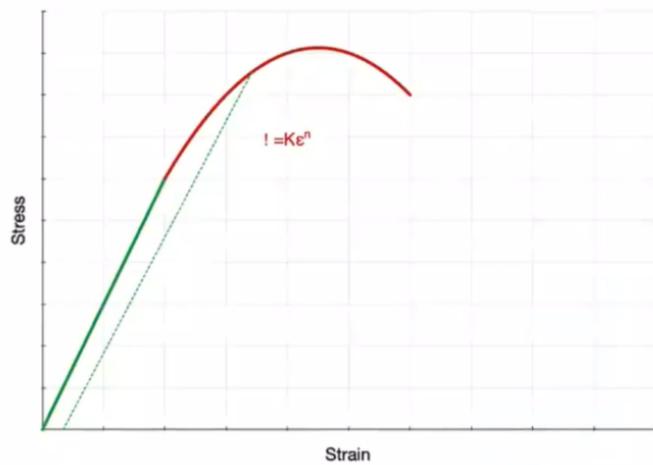


FIGURE 1.8 Triaxial tension stress distribution within necked region which acts to constrain additional deformation in the neck.

typical final fracture (cup&cone)

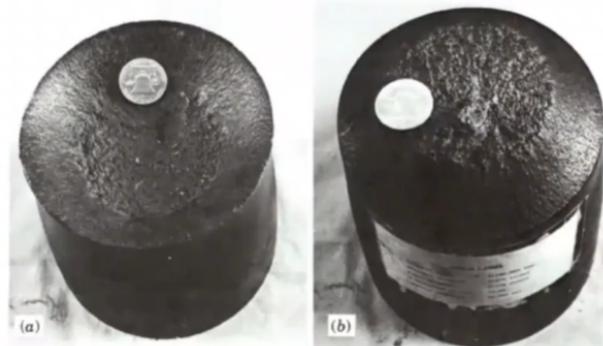


FIGURE 1.9 Typical cup-cone fracture appearance of unnotched tensile bar: (a) cup portion; (b) cone portion. (Courtesy of Richard Sopko, Lehigh University.)

Effect of temperature and strain rate

$\sigma = K\dot{\epsilon}^m$
 m = strain-rate sensitivity factor
 $\dot{\epsilon}$ = strain rate
 K = material constant
 σ = true stress

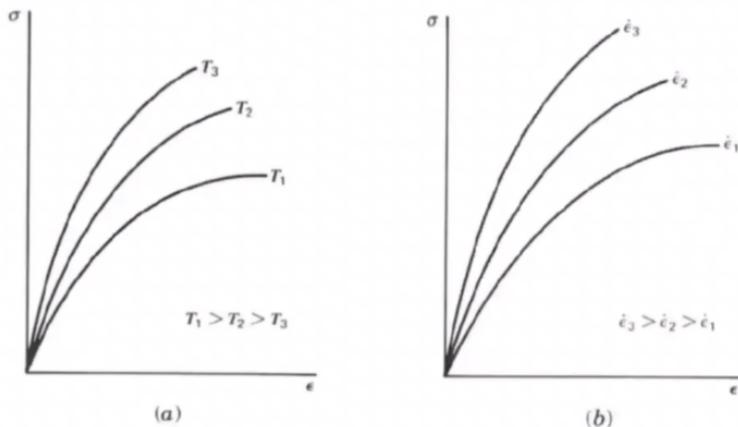
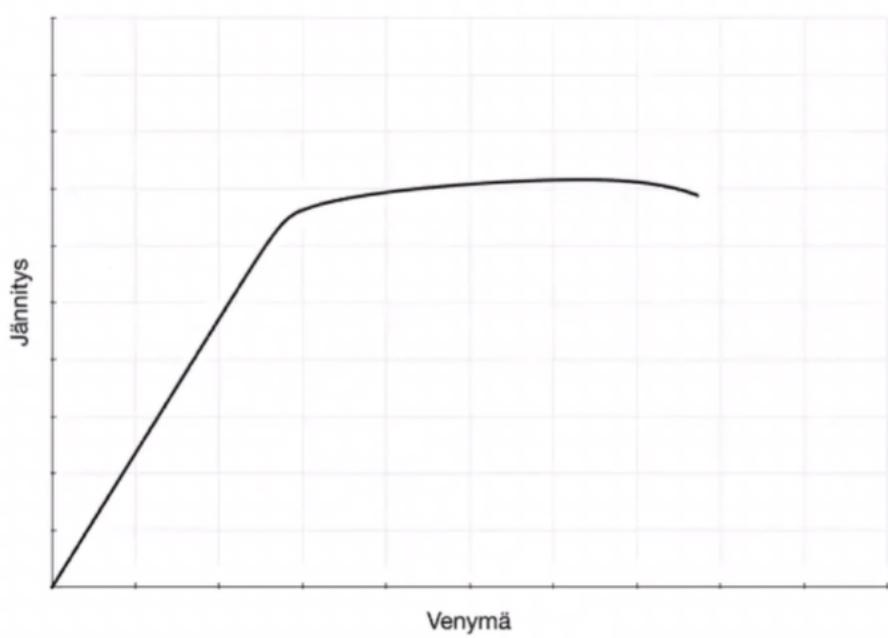
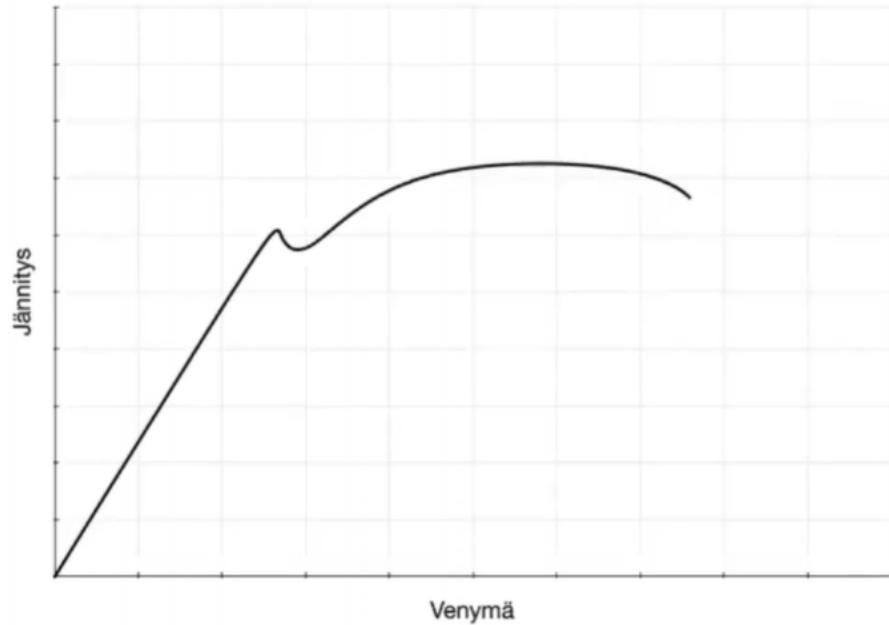


FIGURE 1.27 Yield strength change as a function of (a) temperature and (b) strain rate.

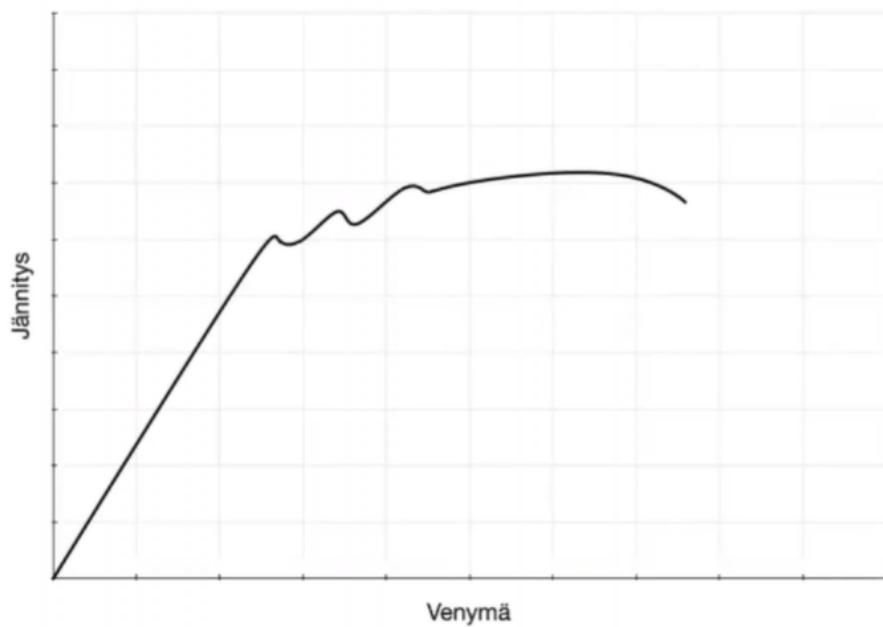
BCC



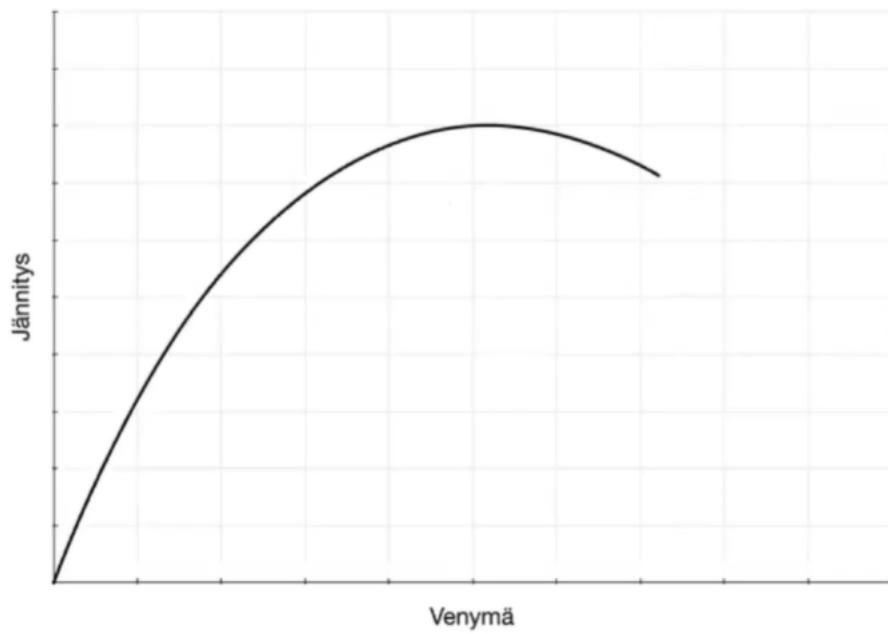
Pronounced yield strength



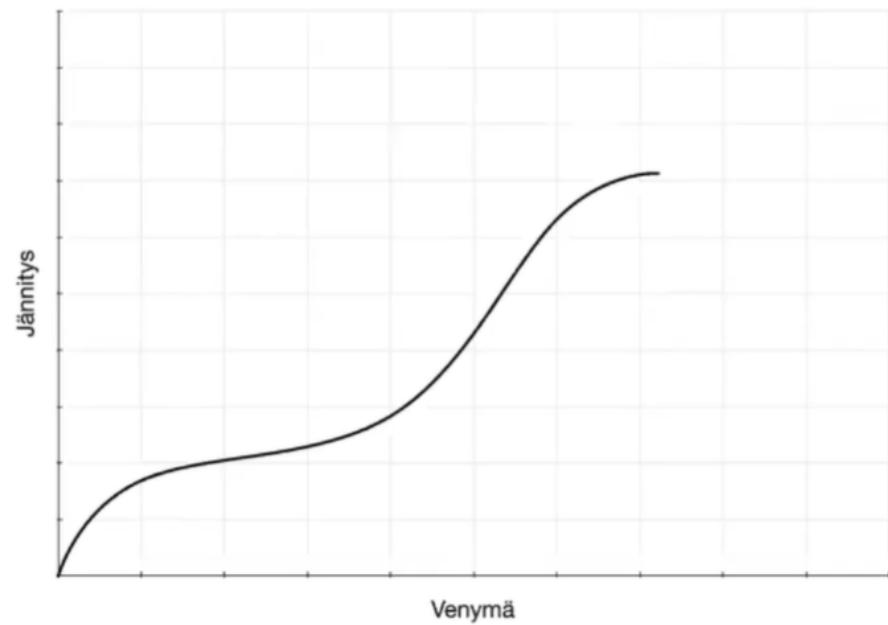
Dynamic straining ageing



FCC

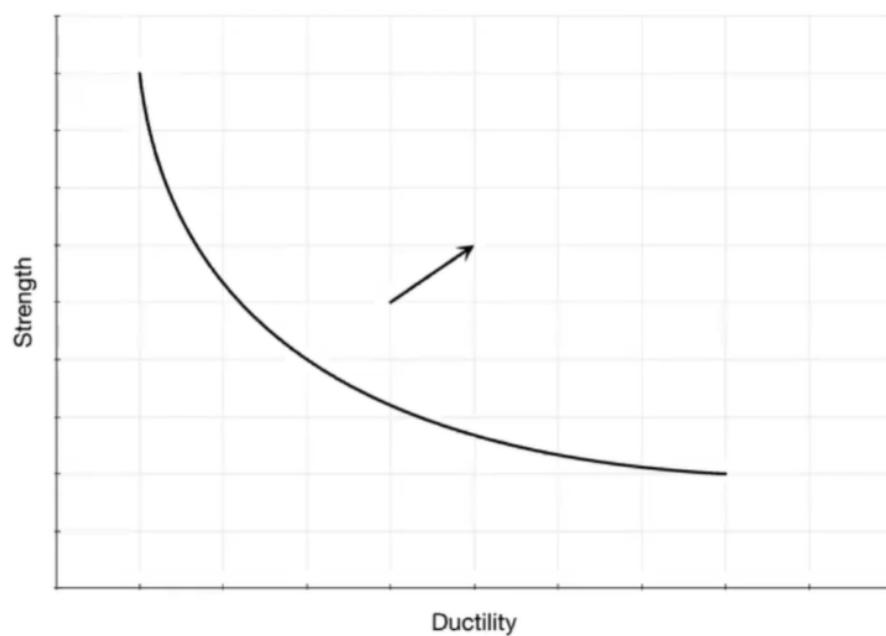


Single crystal

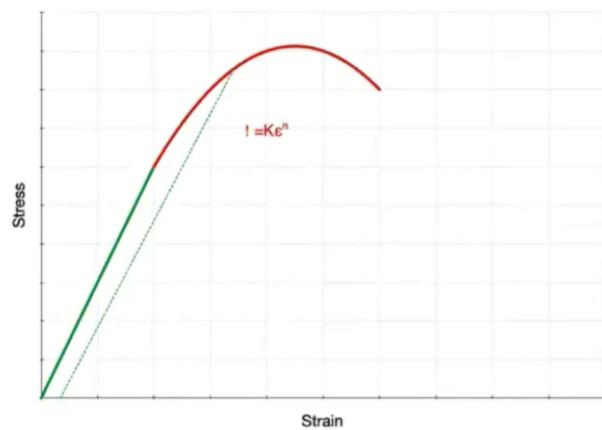


Strengthening mechanisms

What we win in strength...

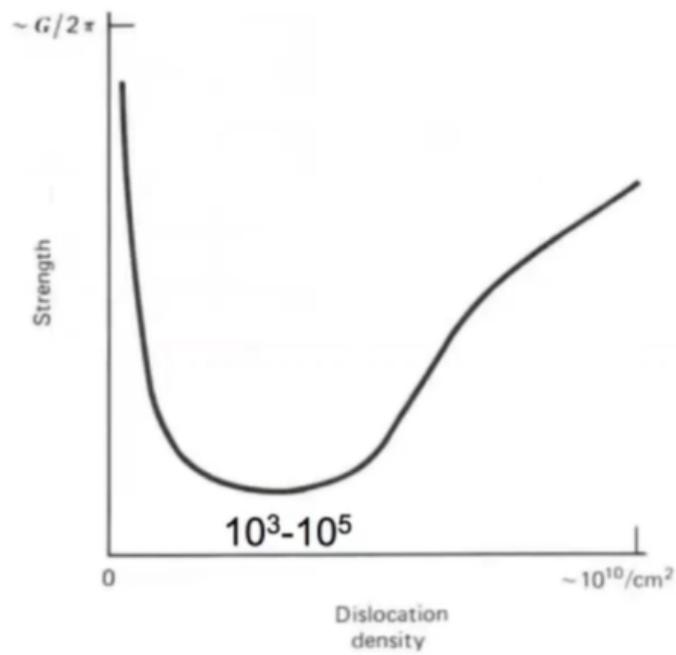


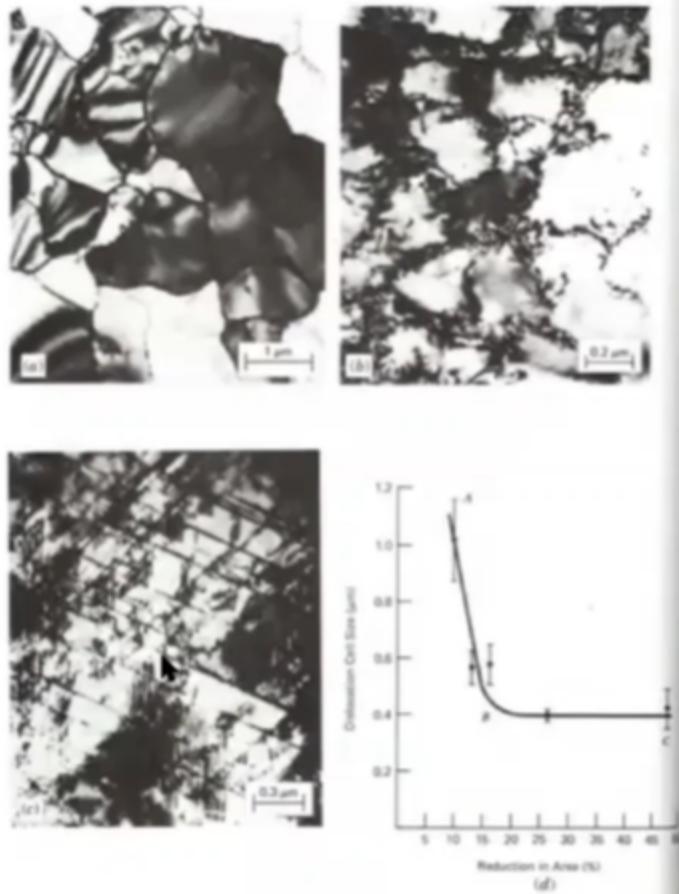
Strain hardening



Increase dislocation density by forging, rolling, etc.

SFE affects dislocation structures





Grain boundary strengthening

Dislocations pile at grain boundaries
Small grain size increases ductility as well
Hall-Petch: $\sigma_{ys} = \sigma + k/\sqrt{d}$

Solid solution strengthening

Solute atoms distort lattice=> interaction forces with dislocations

Symmetric force field=> small effect on screw dislocations;

Asymmetric field => strong effect to both screw and edge dislocations

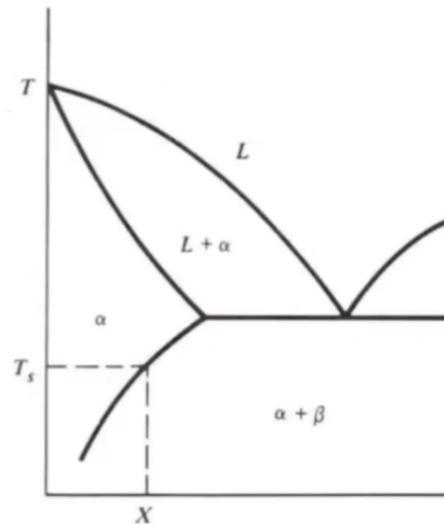
Misfit $\epsilon \sim da/dc$

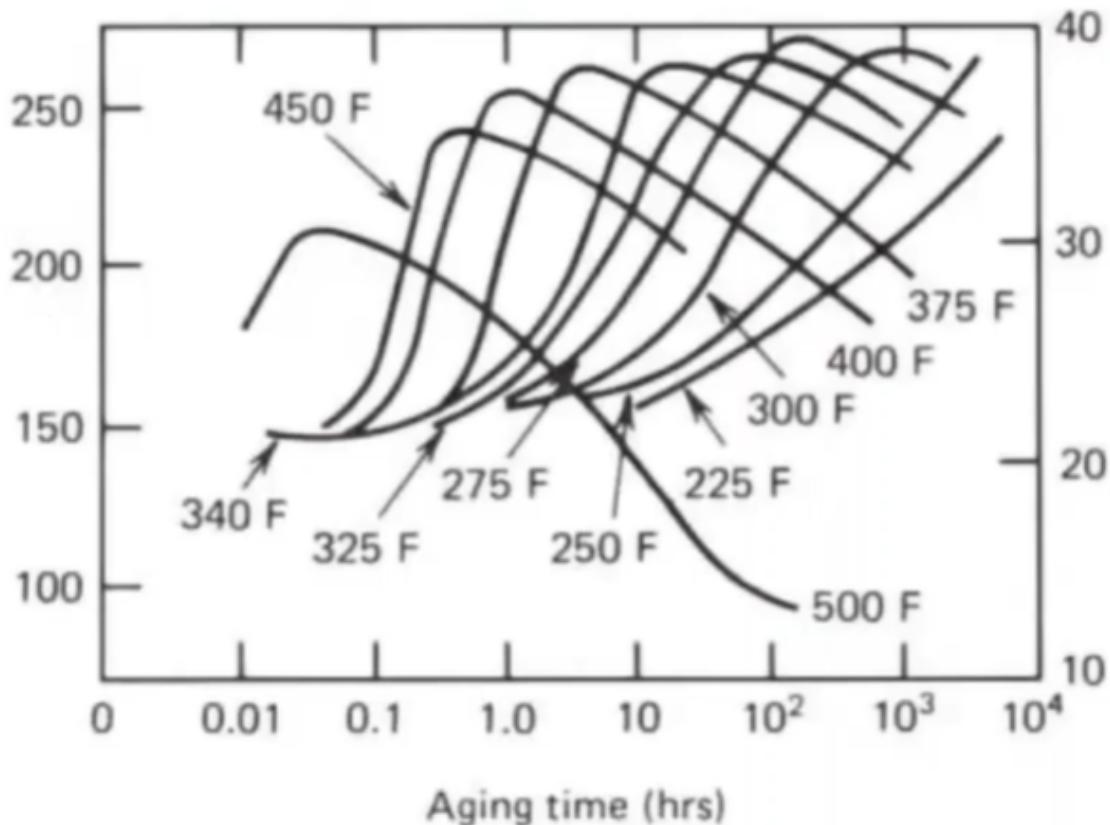
Solubility limits (large misfit => small solubility)

Precipitation hardening

Large number of small precipitates

Hardening with certain thermal treatment and certain alloy systems

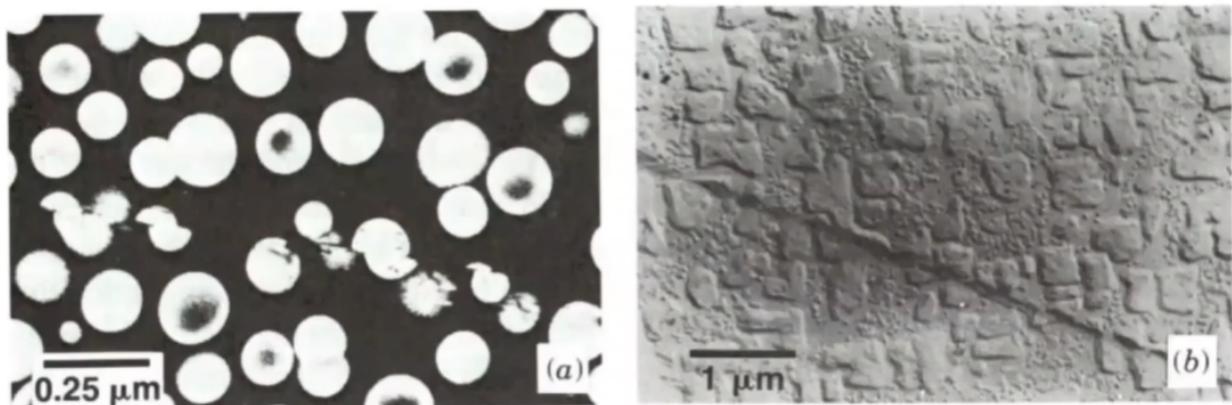




Hardening is affected by

precipitate

- size,
- distribution,
- coherence
- misfit



Dispersion strengthening

incoherent particles

E.g. oxide-dispersion strengthened alloys

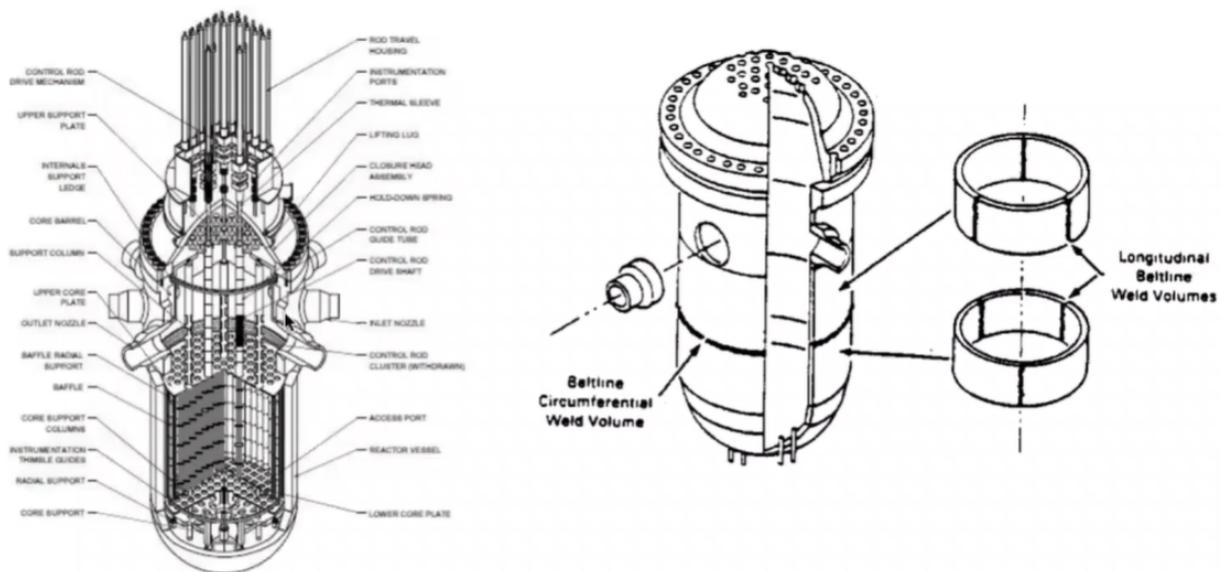
TABLE 4.2 Chemical Composition of Mechanically Alloyed ODS Superalloys

Alloy	Ni	Cr	Al	Ti	Y_2O_3	W	Mo	Ta
Inconel MA 754	bal	20	0.3	0.5	0.6	—	—	—
Inconel MA 6000	bal	15	4.5	2.5	1.1	4	2	2
Alloy 51	bal	9.5	8.5	—	1.1	6.6	3.4	—

Case: Loviisa pressure vessel

A'' Aalto University
School of Engineering

10.9.2019
135

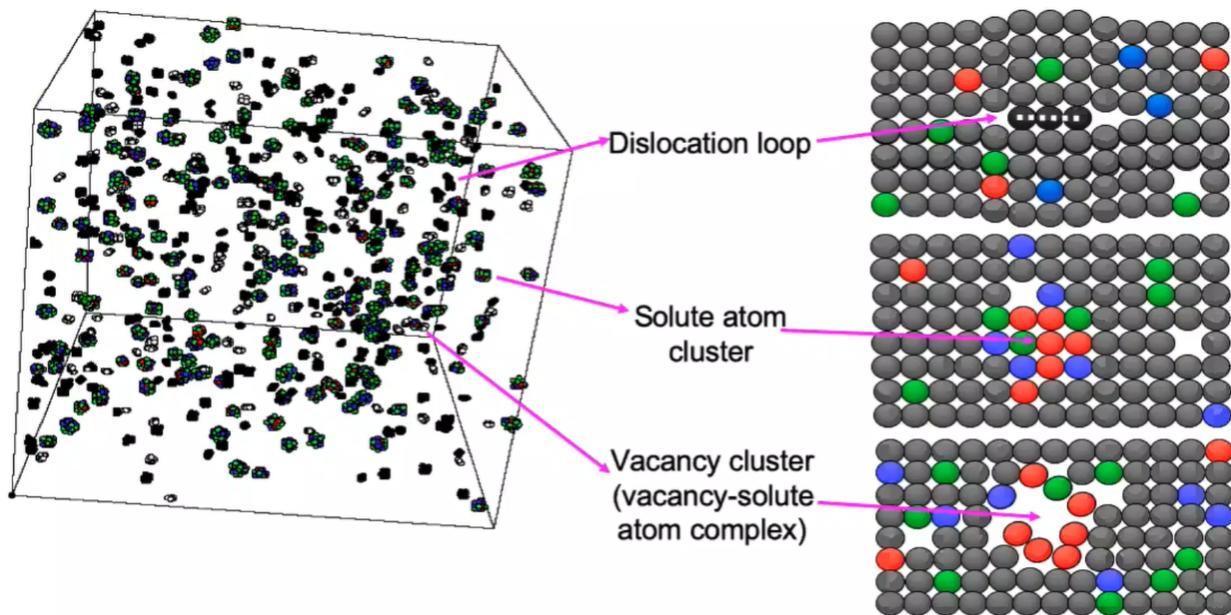


Neutron flux strengthens

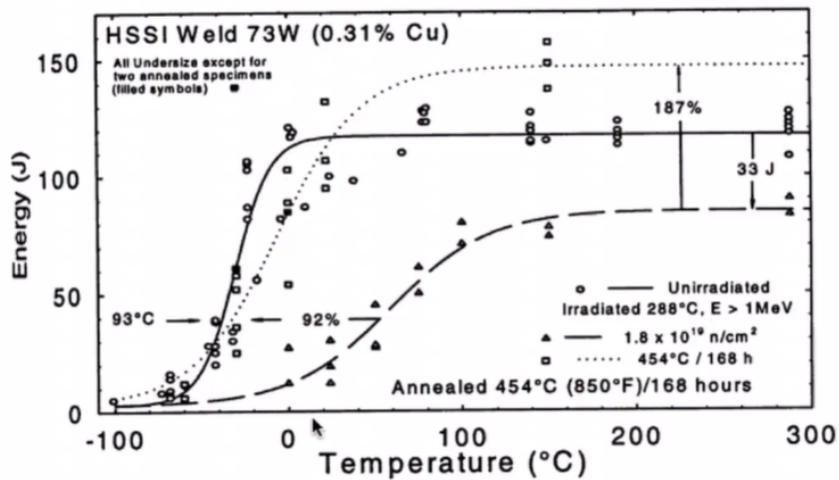
**High-energy neutron diffraction causes (self-interstitials)
These gather into lower-energy formations (loops, clusters,
nanovoids)**

Interaction with solute atoms (form nanoprecipitates)

Especially copper-rich nano-precipitates form



Strengthening causes embrittlement



Known and followed

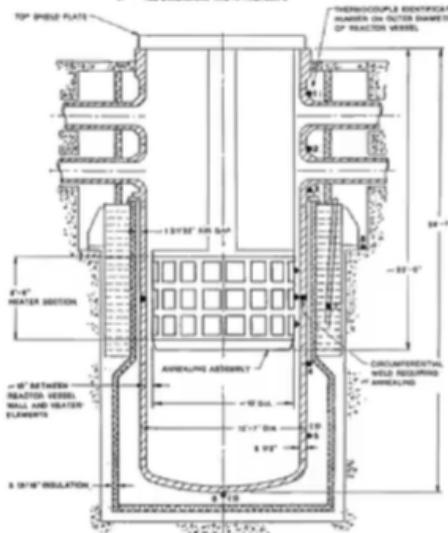
Noticed on other plants

Follow-up samples in the pressure vessel

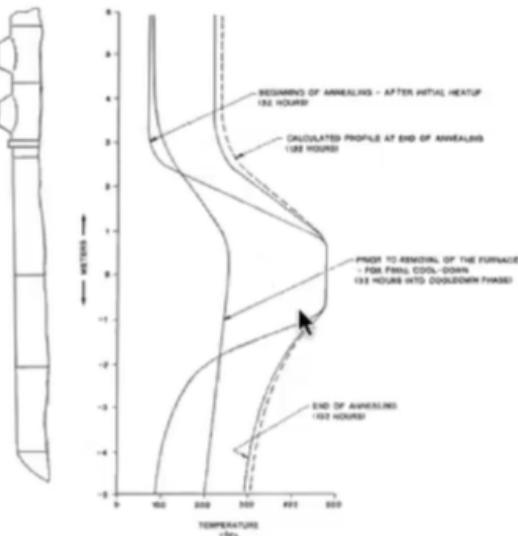
=> Thermal treatment

Reactor	Year	Temperature/time	SS clad
Novovoronezh-3	1987	430±20 °C/168h	No
Armeniya-1	1988	450±50 °C/168h	No
Greifswald-1	1988	475±15 °C/150h	No
Kola-1	1989	475±15 °C/150h	No
Kola-2	1989	475±15 °C/150h	No
Kozloduy-1	1989	475±15 °C/150h	No
Kozloduy-3	1989	475±15 °C/150h	Yes
Greifswald-2	1990	475±15 °C/150h	No
Greifswald-3	1990	475±15 °C/150h	Yes
Novovoronezh-3 (re-anneal)	1991	475±15 °C/150h	No
Novovoronezh-4	1991	475±15 °C/150h	No
Kozloduy-2	1992	475±15 °C/150h	No
J. Bohunice-2	1993	475-503 °C/160h	Yes
J. Bohunice-1	1993	475-496 °C/168h	Yes
Loviisa-1	1996	475±15 °C/100h	Yes
Rovno-1	2010	475±15 °C/150h	Yes

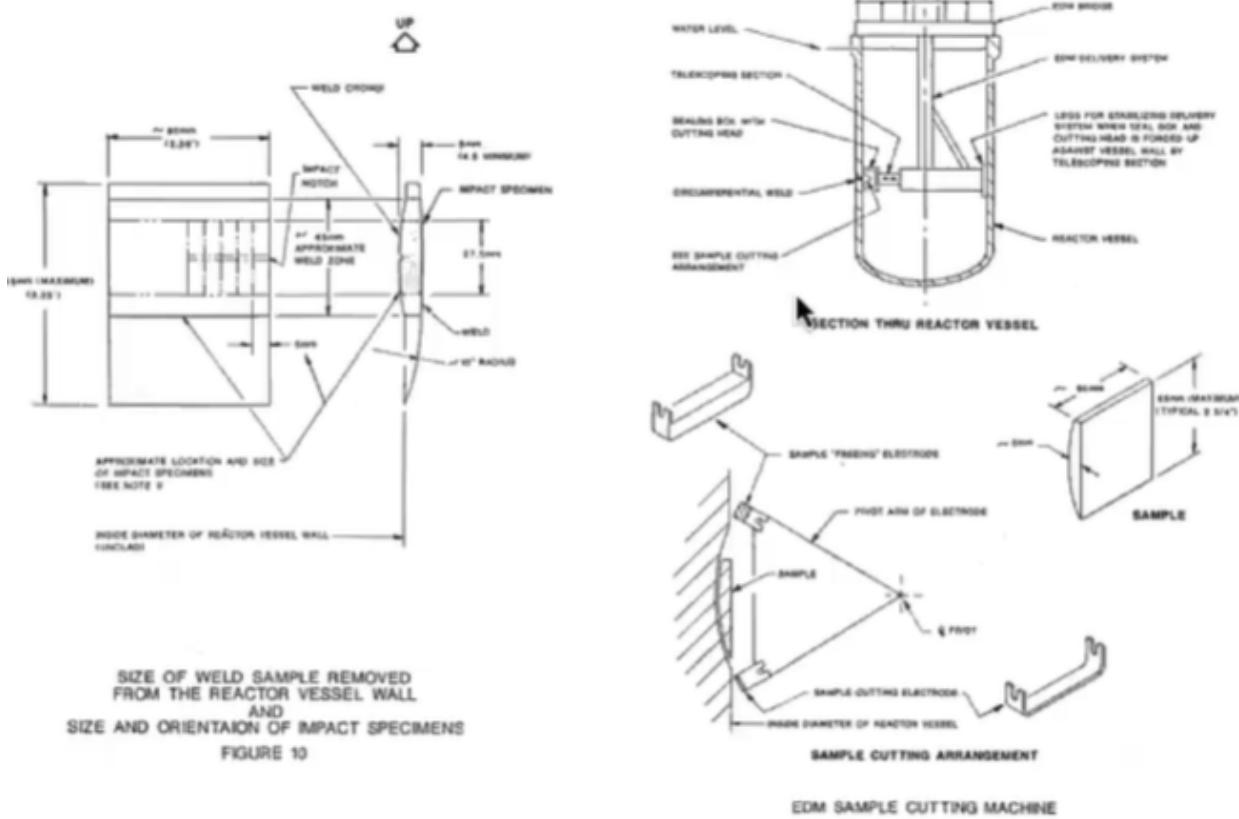
NOTE 1: ► THERMOCOUPLES ON INSIDE DIAMETER OF REACTOR VESSEL
- SIX AT EACH ELEVATION - TWO THERMOCOUPLES IN EACH 100 DEGREE SECTOR
(SEE FIGURE 10)
2. ► THERMOCOUPLES ON OUTSIDE DIAMETER OF REACTOR VESSEL
- THREE AT EACH ELEVATION, 20 DEGREES APART - EXCEPT AS SHOWN
3. ALL DIMENSIONS ARE APPROXIMATE



NOVOTORONEZH UNIT 3
THERMOCOUPLE LOCATIONS DURING
THE ANNEALING OF THE REACTOR VESSEL



NOVOTORONEZH UNIT 3
ACTUAL REACTOR VESSEL WALL
TEMPERATURE PROFILES ALONG VESSEL HEIGHT
FEBRUARY - MARCH 1991



In annealing

Structure heals with diffusion in high temperature

Not proper annealing (much lower temperature)

Long process time to compensate low temperature

Partial recovery

Loviisa I annealed successfully
Possible re-embrittlement followed
Neutron flux decreated (fuel loading changed)
Improved computational tools for cleavage monitoring

Loviisa II not annealed (and not planned)