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So we were talking of the conditions necessary for Harper-Dorn creep and this is what I had listed. We said that the material should have a very large grain size. For example in the case of aluminum that Harper and Dorn tested, they had grain size of 3.3 millimeter.

We also said the material should be of high purity which means 99.95 is what Harper and Dorn tested. So that kind of a purity is what is generally expected for Harper-Dorn creep to be manifested.

And then also the material should have a low initial dislocation density, that is the starting material which you are going to test during creep should have a low dislocation density. That means if you are using a cold work material, probably you may not see Harper-Dorn creep occurring within the material for test conditions that you choose.

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Newtonian viscous creep mechanism: Harper-Dorn Creep

- A few exceptions to the observations of Harper and Dorn are as follows
- Creep studies in alpha titanium, beta cobalt, alpha iron and alpha zirconium have shown the presence of a H-D regime at homologous temperatures of around 0.35 to 0.6 for applied stresses around $9 \times 10^{-5}G$ (G is shear modulus) and grain sizes of around 500 µm.

Ref: G. Malakondaiah and P. Rama Rao, "Grain size independent viscous creep in alpha titanium and beta cobalt," *Scripta Metall.*, 13 (1979) 1187-1190.

Ref: G. Malakondaiah and P. Rama Rao, "Viscous creep of beta-Co," *Mater. Sci. Eng.*, 52 (1982) 207-221.

Ref: J. Fiala, J. Novotny and J. Cadek, "Coble and Harper-Dorn creep in iron at homologous temperatures of 0.4-0.54," *Mater. Sci. Engg.*, 60 (1983) 195-206.

Ref: J. Novotny, J. Fiala and J. Cadek, "Harper-Dorn creep in alpha-zirconium," *Acta Metall.*, 33 (1985) 905-911.

Now having said there are a few exceptions to the observations of Harper and Dorn. So although this slide, previous slide discussed the conditions under which Harper and Dorn creep will be observed, there are a few cases where people have seen Harper and Dorn creep manifestation at temperatures lower than what Harper and Dorn had used.

For example creep studies in alpha titanium, beta cobalt, alpha iron and alpha zirconium have shown the presence of Harper-Dorn creep regime at homologous temperatures of around 0.35 to 0.6. Please note this is very low compared to what Harper and Dorn had used which was around 0.99 T m.

So which means Harper and Dorn is possible, it is possible

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to observe Harper and Dorn creep at very low, homologous temperatures also. And also the grain sizes that they used in these studies were

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in the range of around 500 micrometer. This is again smaller than what was observed with aluminum

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Newtonian viscous creep mechanism: Harper-Dorn Creep

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$0.99 T_m$

$3.3 \ mm \ Al.$

with a grain size of 3.3 millimeter.

So these are some of the studies listed here where people have seen Harper-Dorn creep being manifested at, under conditions different than what we discussed in the previous slide.

So that was about the Newtonian viscous creep mechanisms. We spoke about Nabarro-Herring creep, Coble creep and then Harper-Dorn creep.

And these all were known as Newtonian viscous creep mechanisms because the strain rate of deformation in steady state was found to be directly proportional to the applied stress. So we have a epsilon dot is equal to sigma to the power 1

Newtonian viscous creep mechanism: Harper-Dorn Creep

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$0.99 \, T_m$

$3.3 \, mm \, Al$

$\dot{\varepsilon} \propto \sigma^1$

kind of a relationship.
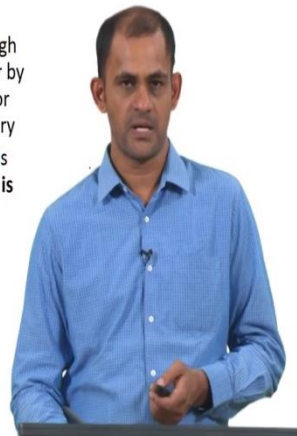
Now let us move on to another mechanism of

Grain boundary sliding creep mechanism, n = 2

- Grain boundary sliding (GBS) as a mechanism of creep bears a stress exponent value of 2.
- The grain size exponent (p) is also equal to 2
- GBS is typically a response of grain boundaries (usually high angle) to an applied shear stress and is supposed to occur by the relative motion of grains along a common boundary or along a narrow zone immediately adjacent to the boundary
- The relative motion of grains along a common boundary is known as Lifshitz GBS when the **accommodating process is diffusional creep**.

Ref: I. M. Lifshitz, "On the theory of diffusion viscous flow of polycrystalline bodies," *Soviet Phys. JETP* **17**, 1963, 909-920.

creep which is known as Grain boundary sliding. And grain boundary sliding as a mechanism of creep bears a stress exponent value of 2, which means your strain rate of deformation in the steady state creep regime is proportional to

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$\dot{\varepsilon} \propto \sigma^2$

the square of the applied stress.

So please note that the epsilon dot relationships, so the different equations that have been shown so far, that is the relationship between epsilon dot and sigma and temperature, diffusivity etc all correspond to the

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$\dot{\varepsilon} \propto \sigma^2$

steady state creep regime.

So we are talking about strain rate in this regime. So this is what epsilon dot is.

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It is a slope of the straight line in the steady state creep regime. So grain boundary sliding requires a stress exponent value of 2 or basically it is known to be controlling the deformation behavior if the stress exponent value is 2.
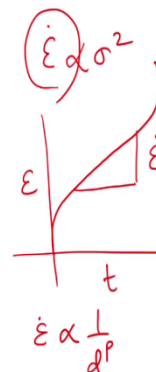
Now in addition to this stress exponent value what people have also observed is that the grain size exponent is also equal to 2.

So when I say the grain size exponent you may recall that epsilon dot is proportional to 1 over d to the power p,

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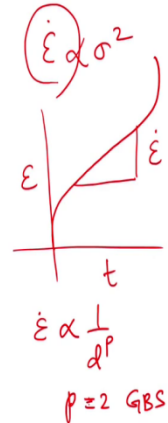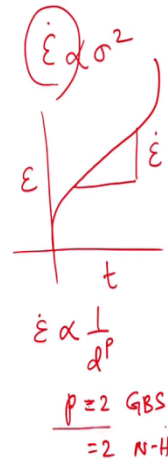
and when p is equal to 1, p is equal to 2, we say it is grain boundary sliding,

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Grain boundary sliding creep mechanism, n = 2

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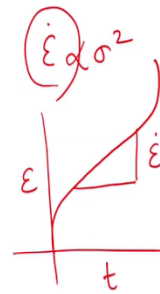
$$\dot{\varepsilon} \propto \sigma^2$$

$$\dot{\varepsilon} \propto \frac{1}{d^p}$$

$$p = 2 \quad GBS$$

of course p is equal to 2 also corresponds to Nabarro-Herring creep but p

(Refer Slide Time: 04:07)

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$$\dot{\varepsilon} \propto \sigma^2$$

$$\dot{\varepsilon} \propto \frac{1}{d^p}$$

$$p = 2 \quad GBS$$
$$= 2 \quad N\text{-}H$$

is equal to 2 along with, p is equal to 2 along with n is equal to 2

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$\dot{\varepsilon} \propto \sigma^2$

$\dot{\varepsilon} \propto \dfrac{1}{d^p}$

$n = 2, \quad p = 2 \quad \text{GBS}$

$= 2 \quad N\text{-}H$

is indicative of grain boundary sliding.

Now what exactly is grain boundary sliding? Of course as the term signifies we are talking about some kind of sliding and it is basically the response of the grain boundaries, generally high angle grain boundaries to an applied shear stress and is supposed to occur by the relative motion of grains along a common boundary or along a narrow zone immediately adjacent to the boundary.

So it is basically imagined to grains which are sliding against each other in the presence of a applied shear stress. So if you have, let us take this simple example of two squares

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$\dot{\varepsilon} \propto \sigma^2$

$\dot{\varepsilon} \propto \dfrac{1}{d^p}$

$n = 2, \quad p = 2 \quad \text{GBS}$

$= 2 \quad N\text{-}H$

as grains and then in the presence of the shear stress so one grain slides against the other.

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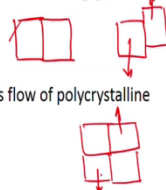- The relative motion of grains along a common boundary is known as Lifshitz GBS when the **accommodating process is diffusional creep.**

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$$\dot{\varepsilon} \propto \sigma^2$$

$$\dot{\varepsilon} \propto \frac{1}{d^p}$$

$$n=2, \quad p=2 \quad GBS$$
$$=2 \quad N\text{-}H$$

So that is what, in a very generic way is what grain boundary sliding is. It is the response of the grain boundaries and so not all types of grain boundaries are going to experience grain boundary sliding. So the grain boundaries have to be typically high angle grain boundaries.

So high angle grain boundaries means the grain boundary misorientation should be greater than 15 degrees. Typically your low angle grain boundaries with the misorientation angle lower than 15 and all the other boundaries which are misorientation greater than 15 degrees are known as high angle.

And so the relative motion of grains, so this is the relative motion of grains

Grain boundary sliding creep mechanism, n = 2

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along a common boundary, it could happen along a common boundary or along a narrow zone immediately adjacent to the boundary. So if it is along a common boundary then it is known as Lifshitz grain boundary sliding.

So this is the mechanism proposed by this gentleman Lifshitz. He said when the relative motion of grains happens along a common boundary, it is known as grain boundary, Lifshitz grain boundary sliding and of course there has to be some form of accommodating process and in the case of this mechanism the accommodating process is diffusional creep.

So why do you need a accommodating process? Well the reason is when, so let us assume you have 4 grains like this and

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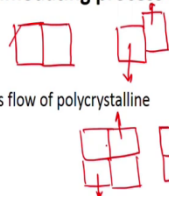the grains are tending to slide

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against each other. Then what happens is when this set of grains, let us say these two grains move with respect to the other two grains then you are going to have some form of stress concentration that
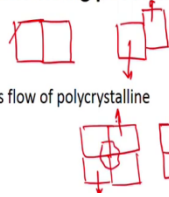
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Grain boundary sliding creep mechanism, n = 2

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$\dot{\varepsilon} \propto \sigma^2$

$\dot{\varepsilon} \propto \frac{1}{d^p}$

$n=2, \quad p=2 \quad GBS$

$=2 \quad N\text{-}H$

is generated within this region.

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Grain boundary sliding creep mechanism, n = 2

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$\dot{\varepsilon} \propto \sigma^2$

$\dot{\varepsilon} \propto \frac{1}{d^p}$

$n=2, \quad p=2 \quad GBS$
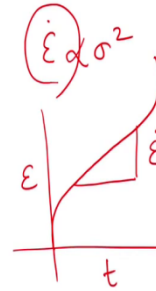
$=2 \quad N\text{-}H$

And if there is no accommodating process then what would happen is you will tend to develop probably voids in that region. So you may see voids like that during the sliding process.
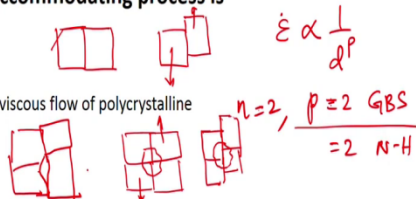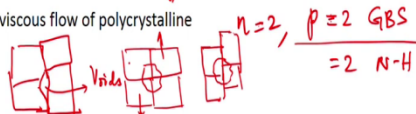
So in order to

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Grain boundary sliding creep mechanism, n = 2

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$$\dot{\varepsilon} \propto \sigma^2$$

$$\dot{\varepsilon} \propto \frac{1}{d^p}$$

$$n = 2, \quad p = 2 \quad GBS$$
$$= 2 \quad N\text{-}H$$

prevent the formation of the voids, so this is just a simple example of how the voids might form, so

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Grain boundary sliding creep mechanism, n = 2

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$$\dot{\varepsilon} \propto \sigma^2$$

$$\dot{\varepsilon} \propto \frac{1}{d^p}$$

$$n = 2, \quad p = 2 \quad GBS$$
$$= 2 \quad N\text{-}H$$

because you are trying to force a set of grains against each other and when the grain starts sliding against each other, there is a scope of loss of contact between the grains.

And the loss of contact will happen if there is no accommodation of the stress concentration that is generated during the process of the slide. So Lifshitz said that this accommodation process can happen by diffusional creep.

When we say diffusional creep it means grain boundaries
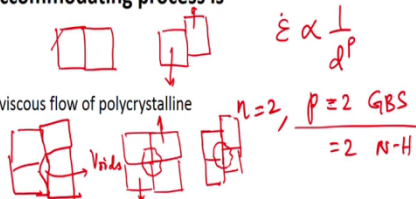
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Grain boundary sliding creep mechanism, n = 2

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which are perpendicular to the applied stress, there will be diffusion of vacancies from there to the grain boundaries parallel or the reverse of it so you need some amount of mass flow to happen to this region where voids are going to form.

So this mass flow which is basically material flow into that region so that the void does not happen, form, this material flow is going to happen by diffusional creep. So that is what is this mechanism of creep.

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Grain boundary sliding creep mechanism, n = 2

- On the other hand, when the process of **accommodation occurs by glide and climb of dislocations**, the GBS process is termed as Rachinger sliding. GBS with the deformation limited to a zone around the boundary comes under the category of Rachinger sliding.
- Accommodation is necessary to avoid the formation of voids at the grain boundary.

Ref: W. A. Rachinger, "Relative grain translations in the plastic flow of aluminum," *J. Inst. Metals* **81**, 1952–1953, 33-41.
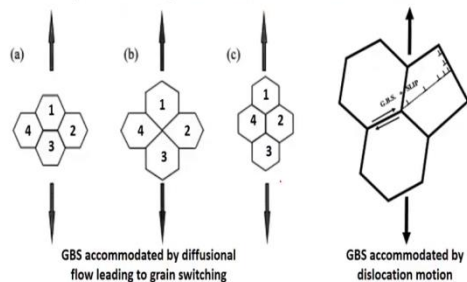
The other mechanism of creep is proposed by Rachinger. So what

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Grain boundary sliding creep mechanism, n = 2

- On the other hand, when the process of **accommodation occurs by glide and climb of dislocations**, the GBS process is termed as Rachinger sliding. GBS with the deformation limited to a zone around the boundary comes under the category of Rachinger sliding.
- Accommodation is necessary to avoid the formation of voids at the grain boundary.

Ref: W. A. Rachinger, "Relative grain translations in the plastic flow of aluminum," *J. Inst. Metals* **81,** 1952–1953, 33-41.

Rachinger said is if the process of accommodation occurs by glide and climb of dislocation then that process of creep is known as the Rachinger

(Refer Slide Time: 08:35)

Grain boundary sliding creep mechanism, n = 2

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- Accommodation is necessary to avoid the formation of voids at the grain boundary.

Ref: W. A. Rachinger, "Relative grain translations in the plastic flow of aluminum," *J. Inst. Metals* **81,** 1952–1953, 33-41.

sliding. So here instead of diffusional creep, you have the motion of dislocations.

So I am going to talk a little more about this in the coming slide but basically the point is, the accommodation if it occurs by glide and climb of dislocations then the grain boundary sliding process is Rachinger sliding.

And here the grain boundary sliding is, the deformation associated with the grain boundary sliding is limited to a zone around the boundary and so this is the category of sliding. And like I mentioned before the accommodation is necessary to prevent the formation

(Refer Slide Time: 09:10)

Grain boundary sliding creep mechanism, n = 2

- On the other hand, when the process of **accommodation occurs by glide and climb of dislocations**, the GBS process is termed as Rachinger sliding. GBS with the deformation limited to a zone around the boundary comes under the category of Rachinger sliding.
- Accommodation is necessary to avoid the formation of voids at the grain boundary.

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of voids, the grain boundaries in triple junction region.

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Grain boundary sliding creep mechanism, n = 2
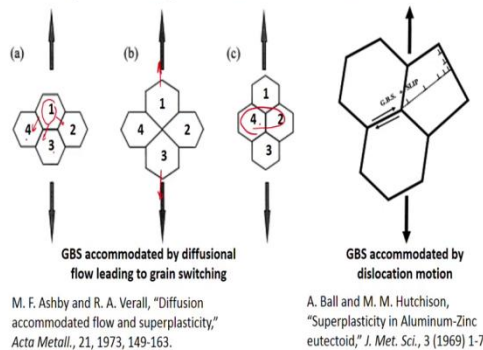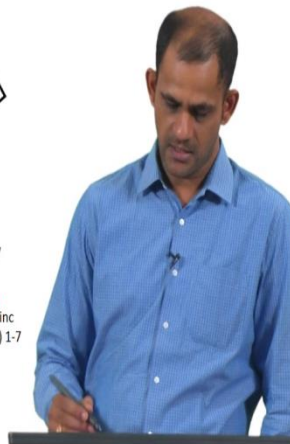- Illustration of the accommodation of GBS induced strain incompatibilities by diffusional flow and dislocation motion

GBS accommodated by diffusional flow leading to grain switching

M. F. Ashby and R. A. Verall, "Diffusion accommodated flow and superplasticity," Acta Metall., 21, 1973, 149-163.

GBS accommodated by dislocation motion

A. Ball and M. M. Hutchison, "Superplasticity in Aluminum-Zinc eutectoid," J. Met. Sci., 3 (1969) 1-7

So, so here is an illustration of the two mechanisms, so Lifshitz as well as Rachinger. So the first case is a model proposed by Ashby and Verall, they were trying to explain the switching of neighbors that happens during grain boundary sliding and they invoked the concept of diffusional flow.

So the point is experimentally it was observed under, in certain materials that when your grain boundary sliding happening there is switching of the neighbors. So for example here are four grains, 1, 2, 3, 4 and let us take the case of 1.

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So 1 is in contact with 2, it is in contact with 3

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and it is in contact with 4.

Now when grain boundary sliding is happening, Ashby and Verall proposed that there is extension of these grains 1 and 3 in that direction

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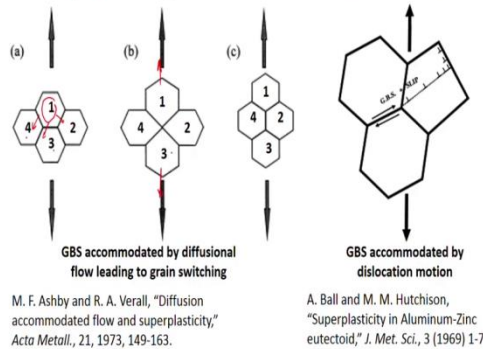
and slowly what happens is 1 loses its contact entirely with 3 and it ends up as having only 4 and 2 as its neighbors.

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And interestingly 4 and 2 which originally were not neighbors now have become neighbors to each other. So this is a mechanism of grains switching and in order to explain this mechanism, Ashby and Verall said that there is some amount of diffusional flow that is happening and this diffusional flow allows this grain switching to happen.

Now the other mechanism, we are talking about grain boundary sliding accommodated by dislocation motion, so this was the model proposed by Ball and Hutchison.

And what they said was when you have this grain boundary sliding happening, you are going to have the stress concentrations in

this triple junctions as well as in grain boundaries and in order to relieve this stress concentration what happens is the grain boundaries start emitting dislocations.

So we know that whenever there is a stress acting and if the stress is higher than a critical value then you are going to have emission of dislocations. And so in this particular case when the applied stress creates a stress concentration and if the stress concentration is higher than some critical value you have dislocations that get emitted from there.

Grain boundary sliding creep mechanism, n = 2
- Illustration of the accommodation of GBS induced strain incompatibilities by diffusional flow and dislocation motion

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And what happens is these dislocations after they are emitted; they travel to the

Grain boundary sliding creep mechanism, n = 2
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GBS accommodated by dislocation motion

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grain boundary opposite, present opposite to them. Now when they go to the grain boundary there is going to be some amount of pile up, as we know when dislocations encounter a barrier they start piling up and, so there is some barrier, there is a pile up that happens.

Now what it means is once there is a pile up there is a back stress that is going to act on the dislocations, which means further emissions of dislocations is going to be slowed down and if the emission of dislocations is slowed down then that means you will have cracks generated at region.
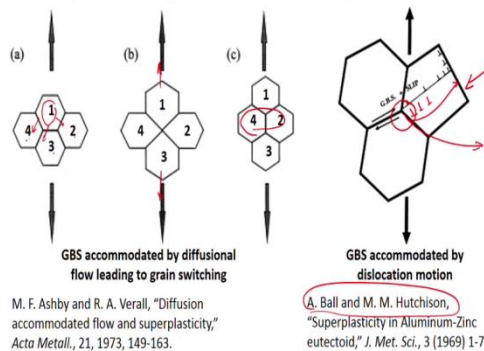
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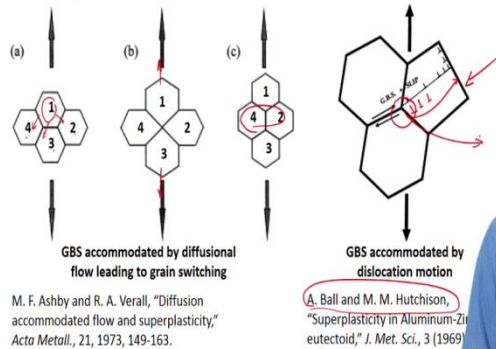
The fact that we do not see any cracks forming means this pile up is somehow relieved and what happens is this pile up is relieved by dislocation climb. We discussed about dislocation climb in

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previous portions. So climb is a way of relieving the pile up stresses that are developed due to the dislocations getting stopped by the grain boundary.

So for the grain boundary sliding to continue inhibited or unimpeded what is necessary is the dislocations that are produced from the stress concentration regions, they go and get start climbing at the opposite grain boundary.

So this is the grain boundary sliding mechanism accommodated by dislocation motion.

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Grain boundary sliding creep mechanism, n = 2

- Ashby and Verall's model fails to capture correctly the stress dependence of the strain rate during grain boundary sliding although it is able to explain the switching of equiaxed grains during deformation.
- Gifkins presented a core and mantle model to explain GBS. All deformation was assumed to occur only in the mantle region of the grain.
- The strain rate of deformation for Grain boundary sliding is approximately given by

$$\dot{\varepsilon} \approx K \left(\frac{b}{d}\right)^2 D \left(\frac{\sigma}{G}\right)^2$$

Here D is equal to grain boundary diffusivity

Now between the two models, what people have noticed is that the Ashby and Verall's model of grain switching and being carried out, being accommodated by the diffusional flow apparently fails to capture correctly the stress-dependence of the strain rate.

So since the strain rate of, should be dependent as a square of the

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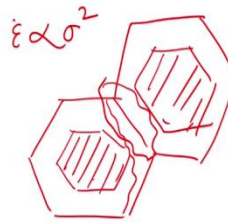**Grain boundary sliding creep mechanism, n = 2**

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$\dot{\varepsilon} \propto \sigma^2$

applied stress, this model that they proposed, Ashby and Verall, was unable to capture that stress dependence and that is why this model has not been widely applied or used for explaining grain boundary sliding.

Now in addition to the Ashby Verall's as well as Ball Hutchison's model, Gifkins also presented a core and mantle model, what were, what Gifkin said is, so if you have a grain, so he said you have what you know, as a core and then

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the surrounding region is all mantle.

And he said during grain boundary sliding, all deformation was supposed to happen in the

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mantle region only. So that is the model proposed by Gifkin. So the deformation is concentrated in this mantle
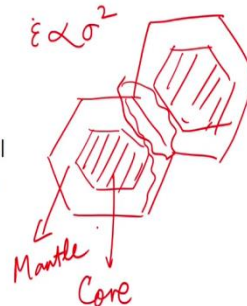
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Grain boundary sliding creep mechanism, n = 2

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region.

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Grain boundary sliding creep mechanism, n = 2

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Here D is equal to grain boundary diffusivity

$\dot{\varepsilon} \propto \sigma^2$

Mantle Core

So this is another viewpoint, this is another way of looking at the grain boundary sliding process. Now by and large the strain rate of deformation for grain boundary sliding can be approximately given or described by this equation.

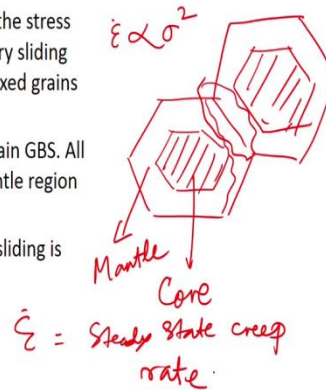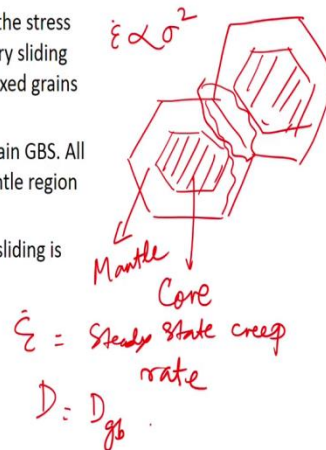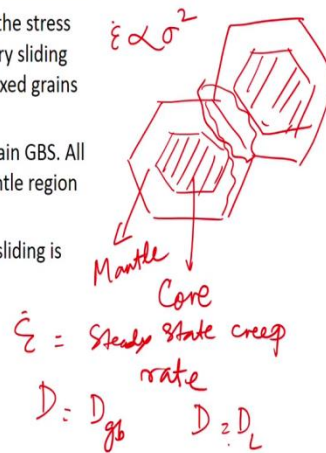So again epsilon dot is the steady state creep rate and what this

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Grain boundary sliding creep mechanism, n = 2

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- The strain rate of deformation for Grain boundary sliding is approximately given by

$$\dot{\varepsilon} \approx K \left(\frac{b}{d}\right)^2 D \left(\frac{\sigma}{G}\right)^2$$

Here D is equal to grain boundary diffusivity

$\dot{\varepsilon} \propto \sigma^2$

Mantle

Core

$\dot{\varepsilon}$ = Steady State creep rate.

equation tells us is that the steady state creep rate is dependent on the inverse of the square of grain size and it is also directly proportional to the applied stress.

And there is a diffusivity term D and this diffusivity term D is equal to

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Grain boundary sliding creep mechanism, n = 2

- Ashby and Verall's model fails to capture correctly the stress dependence of the strain rate during grain boundary sliding although it is able to explain the switching of equiaxed grains during deformation.
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Here D is equal to grain boundary diffusivity

$\dot{\varepsilon} \propto \sigma^2$

Mantle

Core

$\dot{\varepsilon}$ = Steady State creep rate

$D = D_{gb}$

grain boundary diffusivity although there have been cases where people observed that D is equal to D L

Grain boundary sliding creep mechanism, n = 2

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$\dot{\varepsilon} \propto \sigma^2$

Mantle

Core

$\dot{\varepsilon}$ = Steady state creep rate

$D = D_{gb}$ $D = D_L$

that is lattice diffusivity but by and large grain boundary sliding because the deformation is associated in the regions around the grain boundary. So the grain boundary diffusivity is what is applicable to grain boundary sliding deformation mechanism.

Microstructural changes during grain boundary sliding creep

- Fiducial markers are generally employed to study the contribution of grain boundary sliding to the total creep strain[1].
- Grain boundary sliding leads to shearing of the fiducial markers and the shear offset provides a measure of the strain contribution.

Offset following grain boundary sliding [2]
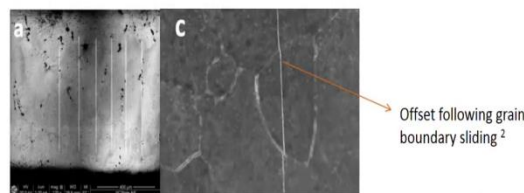
[1]F. Ochoa, X. Deng and N. Chawla, "Effect of cooling rate on creep behavior of a Sn-3.5Ag alloy," *J. Electron Mater*, 33 (2004) 1596-1607.
[2]P. Roodposhti, A Sarkar, K L Murty et al., MSE A, 2016.

So how do you know that the material has experienced grain boundary sliding? Of course one way is by determining the p value and determining the n value but in addition to this, just like giving evidence in the form of microstructural evidence, something that we discussed in Newtonian viscous creep mechanism, so similar concept can be applied here.

You can find out if grain boundary sliding has happened in the material or not by looking at the post-creep microstructures. So after you have carried out your creep tests, if you take the

microstructures and you look at them then probably you can get some evidence of grain boundary sliding.

So one of the way of doing this is you create some fiducial markers on the sample before carrying out the creep test.

So you can use a diamond scribe to create some gentle scratches along the surface. So here are some scratches along the surface, and what is noticed is if the material has experienced grain boundary sliding then you start seeing, so what was originally a straight line starts getting offseted

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Microstructural changes during grain boundary sliding creep

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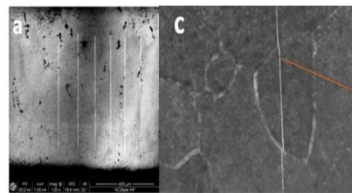
Offset following grain boundary sliding [2]

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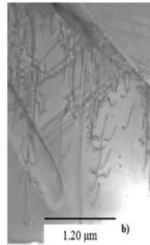[2]P. Roodposhti, A Sarkar, K L Murty et al., MSE A, 2016.

so the diamond scribe line gets offseted during the process of grain boundary sliding.

So here is an example of an offset created in a Magnesium alloy. So if you see in this line there is a small offset but it is there nevertheless and,

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Microstructural changes during grain boundary sliding creep

- Fiducial markers are generally employed to study the contribution of grain boundary sliding to the total creep strain[1].
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Offset following grain boundary sliding [2]
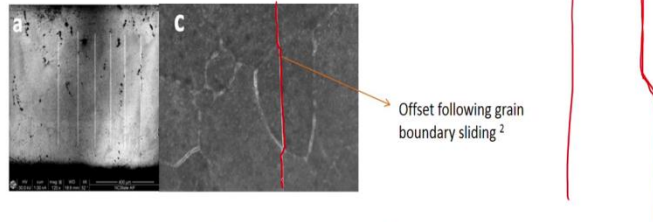
[1]F. Ochoa, X. Deng and N. Chawla, "Effect of cooling rate on creep behavior of a Sn-3.5Ag alloy," *J. Electron Mater*, 33 (2004) 1596-1607.
[2]P. Roodposhti, A Sarkar, K L Murty et al., MSE A, 2016.

so that offset is indicative of grain boundary sliding and it is, very interestingly you can determine this strain involved or the contribution of the grain boundary sliding to the total creep strain by using this offset.

So one article from where you can pick up the related equations is this. So

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Microstructural changes during grain boundary sliding creep

- Fiducial markers are generally employed to study the contribution of grain boundary sliding to the total creep strain[1].
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Offset following grain boundary sliding [2]
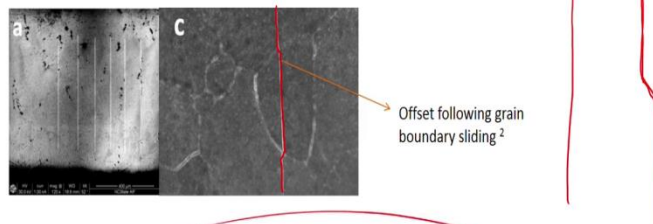
[1]F. Ochoa, X. Deng and N. Chawla, "Effect of cooling rate on creep behavior of a Sn-3.5Ag alloy," *J. Electron Mater*, 33 (2004) 1596-1607.
[2]P. Roodposhti, A Sarkar, K L Murty et al., MSE A, 2016.

you can find out how to determine the strain involved in grain boundary sliding, the contribution of grain boundary sliding to the total strain by, from the offset method. And so that is one set of microstructures that you can see.

Another microstructure

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Microstructural changes during grain boundary sliding creep

- Since stress concentration gets relieved by emission of dislocations, dislocation activity close to the grain boundary is expected.

Dislocation activity close to a grain boundary in a Ti alloy crept in the grain boundary sliding regime

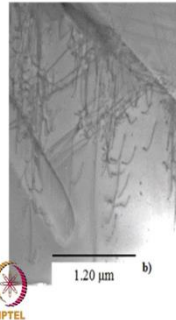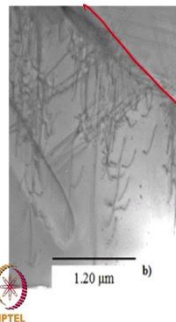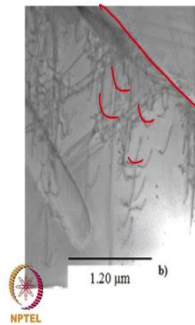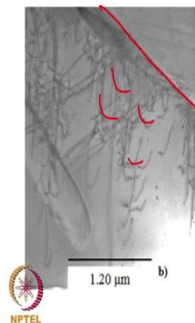S. Gollapudi, I. Charit, K. L. Murty, Acta Mat (2008)

1.20 µm

that could also be observed is you may see a significant amount of dislocation activity close to the grain boundary. The reason is, we said the stress concentration can be relieved by dislocations getting emitted from one grain boundary and these dislocations go and get piled up at another grain boundary where they eventually start climbing.

So which means, if your material to start with had very low dislocation density before creep

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Microstructural changes during grain boundary sliding creep

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Dislocation activity close to a grain boundary in a Ti alloy crept in the grain boundary sliding regime

S. Gollapudi, I. Charit, K. L. Murty, Acta Mat (2008)

1.20 µm

Before creep
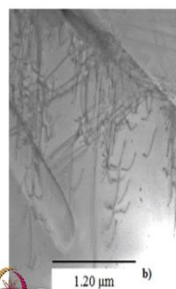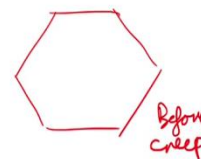
there is a possibility that after creep you may see a lot of dislocations activity next to the grain boundaries. And that would imply that the grain boundaries were involved in the deformation process.

So here is an example.

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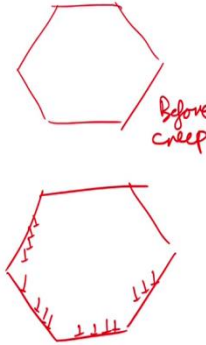
So this is the grain boundary and

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you have lot of dislocation activity close to the grain boundary and observed in

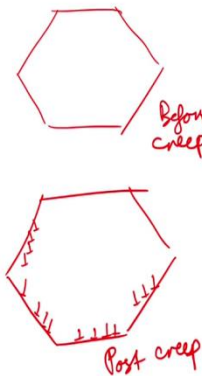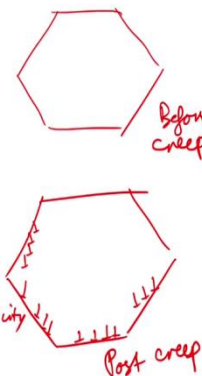Microstructural changes during grain boundary sliding creep

- Since stress concentration gets relieved by emission of dislocations, dislocation activity close to the grain boundary is expected.

Dislocation activity close to a grain boundary in a Ti alloy crept in the grain boundary sliding regime

S. Gollapudi, I. Charit, K. L. Murty, Acta Mat (2008)

a titanium alloy. So this is also a type of microstructure that you could expect in a sample that has experienced grain boundary sliding.

Microstructural changes during grain boundary sliding creep

- Since stress concentration gets relieved by emission of dislocations, dislocation activity close to the grain boundary is expected.

Dislocation activity close to a grain boundary in a Ti alloy crept in the grain boundary sliding regime

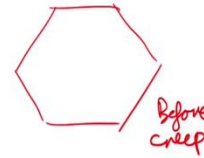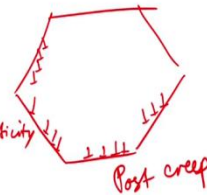S. Gollapudi, I. Charit, K. L. Murty, Acta Mat (2008)
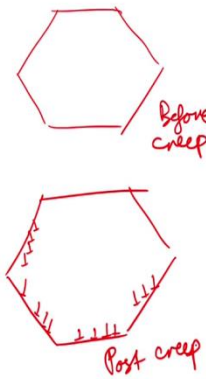
So grain boundary sliding as a creep mechanism is very important for understanding superplasticity.

So superplasticity is the ability of a material to experience or provide high strains before failure and typically the elongations that you get from a superplastic material can be hundreds of strain. So it could be more than 100 percent strain that is something that you can get in superplasticity.

So people have observed that one of the key deformation mechanisms and in fact the key deformation mechanism in superplasticity is grain boundary sliding and the strain rate sensitivity high, so the strain rate sensitivity m is required to be around 0.5.

So if you have m is

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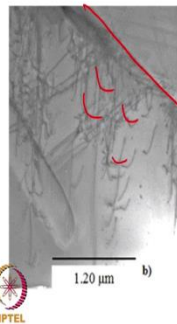
equal to 0.5 then you get high plasticity

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and m is actually equal to 1 over n and since n is equal to 2

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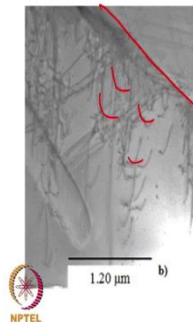
for grain boundary sliding you get m is equal to point 5.

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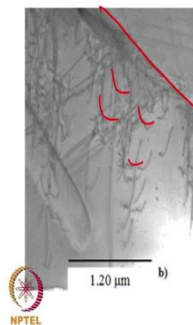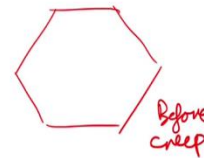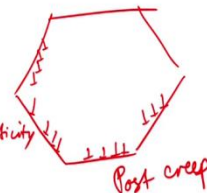
So that is why if you have grain boundary sliding playing an important role then you can expect high superplasticity. Superplasticity on its own is a vast topic to be covered, so we are not going to discuss that in this set of lectures.

Now after grain boundary sliding which was n is equal to 2, let us move on to another creep mechanism which is known as the viscous glide creep mechanism.

Here the stress exponent value n is equal to 3.

So the n is equal to 3 regime, though in principle corresponds to the Power law control so creep mechanism where we generally say n is equal to 4 to 7 but it differs from the same at a mechanistic level.

So this n is equal to 3 is very close to n is equal to 4 to 7 which is known as the Power law creep regime which we are going to discuss next in the coming portions.

So

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there is a lot of similarity between these two but in terms of the values that you are getting, but at the mechanistic level these two are very different from each other. The viscous glide creep regime is very different from the Power law creep regime.

Now the power law controlled creep mechanism is mostly dislocation climb controlled. So this is something that we will learn, like I mentioned, in the coming portions we are going to talk about power law creep in detail. So the power law controlled creep mechanism is mostly dislocation climb controlled and it is noted in pure metals also known as, pure metals or class II or metal

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Viscous glide creep mechanism, n = 3

- The *n* = 3 regime though in principle, corresponds to the power law controlled (*n* = 4-7) creep mechanism, it differs from the same at a mechanistic level.
- The power law controlled creep mechanism is mostly dislocation climb controlled commonly noted in pure metals and class-II or metal-class alloys .
- The viscous glide creep mechanism on the other hand is dislocation glide controlled and is usually exhibited by alloys known as class-I or class-A, and hence the n = 3 regime is sometimes referred to as alloy type creep behavior
- In class-A alloys, the rate of glide is controlled by the diffusion of the solute atoms, thereby leading to relatively slower rate of glide compared to that of climb thereby the viscous glide of dislocations becomes the rate controlling process with *n*=3

class alloys.

So you can observe power law creep in pure metals, or you can observe it in some alloys which are known as class II alloys or class M alloys. Now

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Viscous glide creep mechanism, n = 3

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- In class-A alloys, the rate of glide is controlled by the diffusion of the solute atoms, thereby leading to relatively slower rate of glide compared to that of climb thereby the viscous glide of dislocations becomes the rate controlling process with *n*=3

on the other hand, the topic of interest which is viscous glide creep, so the viscous glide creep mechanism on the other hand is dislocation glide controlled and is usually exhibited by alloys known as class I or class A.

So please note, one is dislocation climb controlled

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Viscous glide creep mechanism, n = 3

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- The power law controlled creep mechanism is mostly dislocation climb controlled commonly noted in pure metals and class-II or metal-class alloys .
- The viscous glide creep mechanism on the other hand is dislocation glide controlled and is usually exhibited by alloys known as class-I or class-A, and hence the n = 3 regime is sometimes referred to as alloy type creep behavior
- In class-A alloys, the rate of glide is controlled by the diffusion of the solute atoms, thereby leading to relatively slower rate of glide compared to that of climb thereby the viscous glide of dislocations becomes the rate controlling process with *n*=3

n = 3
is n = 4-7
power law
regime

Pure metals
class II alloys
or class M
alloys

and the other one is dislocation

(Refer Slide Time: 21:17)

Viscous glide creep mechanism, n = 3

- The *n* = 3 regime though in principle, corresponds to the power law controlled (*n* = 4-7) creep mechanism, it differs from the same at a mechanistic level.
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n = 3
is n = 4-7
power law
regime

Pure metals
class II alloys
or class M
alloys

glide controlled. And the alloys which exhibit viscous glide creep are known as class I alloys or class A alloys. And hence the n is

(Refer Slide Time: 21:31)

Viscous glide creep mechanism, n = 3

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equal to 3 creep regime is also referred to the alloy type creep

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Viscous glide creep mechanism, n = 3

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behavior. Now we are talking about dislocation glide controlled.

So how is the glide of the dislocation controlled? Well, the guide of the dislocation is controlled by the diffusion of the solute atoms.

Viscous glide creep mechanism, n = 3

- The $n = 3$ regime though in principle, corresponds to the power law controlled ($n = 4$-7) creep mechanism, it differs from the same at a mechanistic level.
- The power law controlled creep mechanism is mostly dislocation climb controlled commonly noted in pure metals and class-II or metal-class alloys .
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- In class-A alloys, the rate of glide is controlled by the diffusion of the solute atoms, thereby leading to relatively slower rate of glide compared to that of climb thereby the viscous glide of dislocations becomes the rate controlling process with $n=3$

Handwritten annotations:
n = 3
is n = 4-7
power law
regime

Pure metals
class II alloys
or class M
alloys

class I or
class A alloys

So if you remember dislocations that is edge dislocations can interact with solute atoms because the solute atoms can occupy either the compression sites or the tensile sites present in the dislocation region in the core, near the core of the dislocation. So the solute atoms can interact with edge dislocations.

So if you have solute atoms interacting with the edge dislocations then what would happen is you will have dislocations getting controlled because for the dislocations to move ahead now it has to either carry these atoms along with it or it has to separate itself from these atoms.

So because in either case it needs higher stress, so what happens is you have to provide a certain amount of stress before the dislocations can move completely away, separate away from the solute atoms.

So as long as that does not happen, so if your stress level is such that the dislocation level is not able to separate itself from the solute atoms then what will happen is the glide is going to be controlled by the solute atoms and the diffusion of the solute atoms.

So here you achieve a, what happens is the rate of glide is slower compared to that of climb and therefore the glide of the dislocations or the viscous glide of the dislocations becomes the rate controlling process with n is equal to 3.

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Viscous glide creep mechanism, n = 3

- The three power law creep region is known to occur in solid solutions with a large atom size mismatch.
  - Alloys with higher concentration of the solute atoms seem to prefer the three power law creep regime as the mode of creep
- Class A alloys usually exhibit either no or little primary creep or a region characterized by an increasing slope (increasing strain rate)
- Examples of alloys that creep by this mechanism are Al-Zn, Al-Ag and Ni-Fe alloys. Even intermetallics such as Ni_3Al, Ti_3Al etc have been known to exhibit viscous glide creep behavior

So the three power law creep that is viscous glide creep is generally known to occur in solid solutions where the solute atom has a large atom size mismatch with the matrix atom.

And so alloys with higher concentration of the solute atoms seem to prefer the three power law creep regime as the mode of creep. So interestingly class A alloys usually exhibit either no, or very little primary creep and the creep regime is characterized by an increasing slope. So this is in contrast to class M alloys or pure metals where primary

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Viscous glide creep mechanism, n = 3

- The three power law creep region is known to occur in solid solutions with a large atom size mismatch.
  - Alloys with higher concentration of the solute atoms seem to prefer the three power law creep regime as the mode of creep
- Class A alloys usually exhibit either no or little primary creep or a region characterized by an increasing slope (increasing strain rate)   → Class M alloys
- Examples of alloys that creep by this mechanism are Al-Zn, Al-Ag and Ni-Fe alloys. Even intermetallics such as Ni_3Al, Ti_3Al etc have been known to exhibit viscous glide creep behavior

creep is prominent.

And examples of alloys that creep by this mechanism are aluminum zinc, aluminum silver, nickel iron alloys, and interestingly even some intermetallics such as Ni3Al and Ti3Al, have also been found to exhibit viscous glide creep behavior.

Why I say it is interesting, the reason is we are talking about solute atoms interacting with the dislocations and preventing their motion so you are mainly talking about solid solutions but intermetallics have also been observed to display this creep behavior and I will explain why that might be the case in the coming portion.

Now let us pay, take our attention back to the statement that the class A alloys usually exhibit either no or very little primary creep regime.

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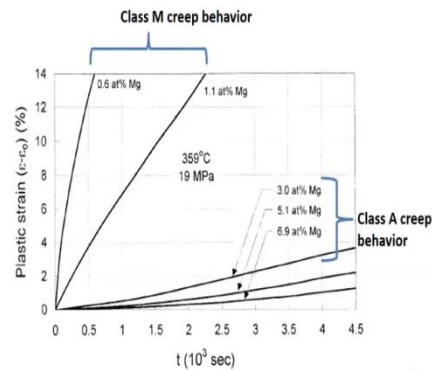
So let us look at this slide. So here we have two types of curves. This is aluminium magnesium alloy. So you have two samples where one with point 6 atomic percent
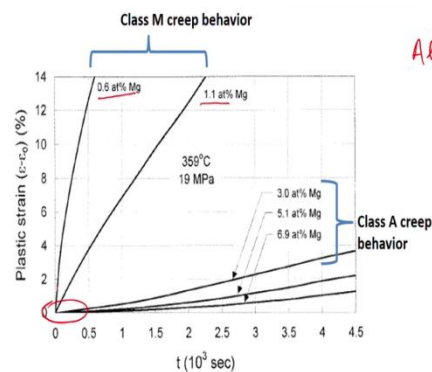
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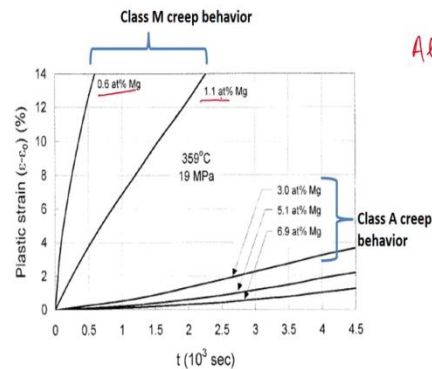
and one with 1.1 atomic percent magnesium and these two samples are showing a fairly large amount of primary creep regime before entering into the steady state.

On the other hand the alloys with 3, 5.1 and 6.9 atomic percent magnesium demonstrating very small
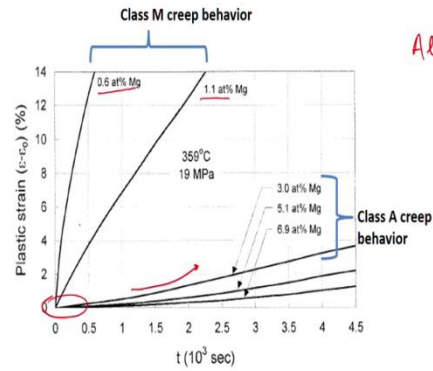
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primary creep regime and in fact the creep rate is increasing right from the beginning.

Whereas here your creep rate is decreasing towards the steady state regime, here the alloy is exhibiting an increasing
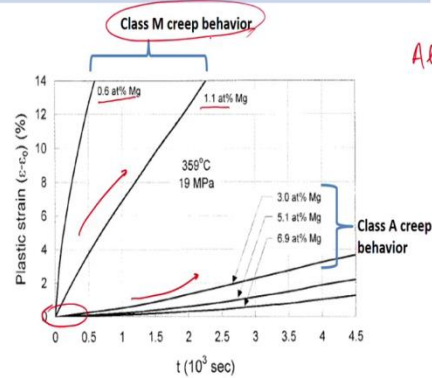
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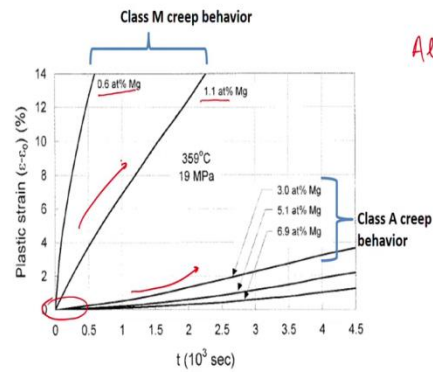
creep rate. So this is class M behavior.
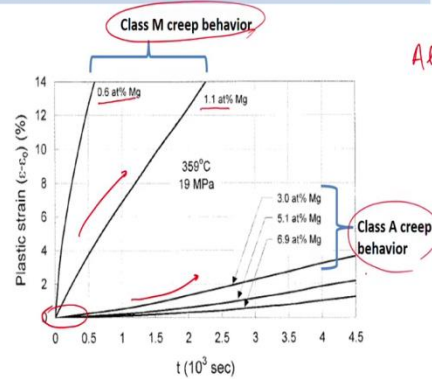
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So even though it is a solid solution, on account of the low quantity of magnesium present in the alloy, the behavior of the alloy is still like a metal. It shows a prominent primary creep and it shows a decreasing strain rate behavior.

On the other hand this is class A behavior,

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creep behavior but the primary creep is very small and the alloy content is high and so that is influencing this behavior.

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Now the following theories have been proposed to explain the glide controlled creep process. So according to Cottrell and Jaswon, the dragging force that limits the motion of, dislocation further limits the dislocation glide, the dragging force could be due to the segregation of solute atmosphere to the moving dislocation.

This is something that I already mentioned. And dislocation speed in this case is controlled by the rate of migration of the solute atoms.

So because the solute atoms have a tendency to segregate to the dislocation so the, as the dislocations move they will have to carry the solute atoms along with it and the solute atoms will move with the dislocation, migrate and the migration rate is what is controlling the dislocation speed.

On the other hand Fisher suggested that the viscous process had its origin in the destruction of the short range order in solid solution alloys so you could have short range order in some materials

And when you have short range order the dislocation, for the, and when the dislocation comes across these regions it is essentially going to destroy the order if it passes through that region, it is going to create disorder within that region.

And the disorder created by the dislocation motion would result in the formation of a new interface thereby the interfacial energy would become the rate controlling process. So that was the suggestion given by Fisher to explain the viscous glide process.

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Viscous glide creep mechanism, n = 3

– Suzuki suggested that the dragging force was an **outcome of solute atoms segregating to stacking faults**.
– Weertman suggested that dislocation glide is impeded in long range ordered alloys because of the energy penalty associated with creation of anti phase boundaries

Ref: H. Suzuki, *Sci. Rep. Res. Inst. Tohoku University*, 4A, 1957, 455.
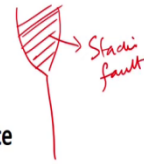Weertman J. Trans AIME 1960;218:207.
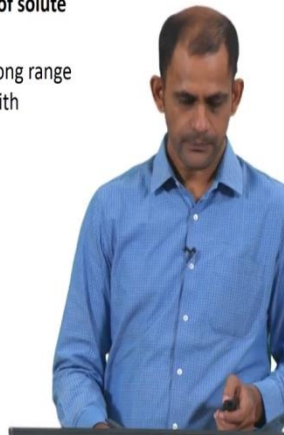
Elsewhere Suzuki came up with the suggestion that the dragging force was an outcome of solute atoms segregating to the stacking faults.

So solute atoms will have a tendency to segregate to the stacking faults and the stacking faults are basically partial dislocations, two partial dislocations separated and the enclosed region is the stacking fault.

And the stacking fault actually provides sites

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Viscous glide creep mechanism, n = 3

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– Weertman suggested that dislocation glide is impeded in long range ordered alloys because of the energy penalty associated with creation of anti phase boundaries
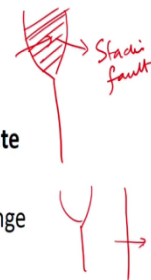
Ref: H. Suzuki, *Sci. Rep. Res. Inst. Tohoku University*, 4A, 1957, 455.
Weertman J. Trans AIME 1960;218:207.

for solute atoms to segregate. And once the solute atoms segregate to the stacking faults, the stacking fault energy is going to come down further which means the dislocations will find even more difficult to join back or the partials will find it difficult to join back to form a complete dislocation.

And when you compare a partial dislocation to a single dislocation, a single dislocation will find it

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Viscous glide creep mechanism, n = 3

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– Weertman suggested that dislocation glide is impeded in long range ordered alloys because of the energy penalty associated with creation of anti phase boundaries

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rather easy to move through the lattice compared to the partial dislocation. So that is why once you have partial dislocations and if you have the solute atoms segregating to the

stacking faults then there will be a dragging force on the dislocation motion. So that was the suggestion of Suzuki.

And Weertman suggested that dislocation glide is impeded in long range ordered alloys because of the energy penalty associated with creation of anti-phase boundaries.

So when you talk of ordered alloys like Ni3Al,

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Viscous glide creep mechanism, n = 3

– Suzuki suggested that the dragging force was an **outcome of solute atoms segregating to stacking faults**.
– Weertman suggested that dislocation glide is impeded in long range ordered alloys because of the energy penalty associated with creation of anti phase boundaries
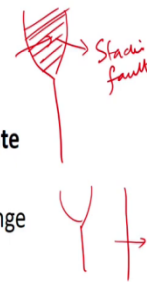
Ref: H. Suzuki, *Sci. Rep. Res. Inst. Tohoku University*, 4A, 1957, 455.
Weertman J. Trans AIME 1960;218:207.

so there is a certain structure, that is a certain amount of ordering of atoms in order to achieve that stoichiometry and once a dislocation moves, it may force atoms of the same type to come against each other or become neighbors to each other.

And that would lead to the creation of an anti phase boundary and the energy associated with the anti phase boundary is what, according to Weertman is the impediment, impediment to the dislocation glide. So these are some of the mechanisms proposed in favor of the viscous glide creep mechanism.

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Viscous glide creep mechanism, n = 3

- The steady state strain rate of creep is given by [1]

$$\dot{\varepsilon} \approx \frac{0.35}{A} D_s \left(\frac{\sigma}{G}\right)^3$$

- Here A is the interaction factor that depends on the particular viscous creep drag mechanism controlling the dislocation glide, D_s is the solute diffusivity
- The activation energy of creep (Q) has been found to be equal to that of self diffusion activation energy [2]

[1] Weertman J. Trans AIME (1960);218:207
[2] K L Murty, F Mohamed, J E Dorn, Acta Metall., 20 (1972) 1009

Now the steady state strain rate of creep in this particular case, so if you have a material experiencing viscous glide creep mechanism with steady state strain of creep is given by

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Viscous glide creep mechanism, n = 3

- The steady state strain rate of creep is given by [1]

$$\dot{\varepsilon} \approx \frac{0.35}{A} D_s \left(\frac{\sigma}{G}\right)^3$$

- Here A is the interaction factor that depends on the particular viscous creep drag mechanism controlling the dislocation glide, D_s is the solute diffusivity
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this equation, and here A is the interaction factor that depends on the particular viscous creep drag mechanism controlling the dislocation glide.

So we spoke about four mechanisms by which dislocation motion could be dragged or dislocation glide could be impeded and so A is the factor. So depending on which mechanism is controlling the glide, you will have the value of A associated with that mechanism.

And Ds is the solute diffusivity and Ds will be proportional to e to the power minus Q over k T and

Viscous glide creep mechanism, n = 3

- The steady state strain rate of creep is given by [1]

$$\dot{\varepsilon} \approx \frac{0.35}{A} D_s \left(\frac{\sigma}{G}\right)^3$$

$$D_s \propto e^{-Q/kT}$$

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[2] K L Murty, F Mohamed, J E Dorn, Acta Metall., 20 (1972) 1009

what has been proposed by some researchers that Q is essentially equal to the activation energy for

Viscous glide creep mechanism, n = 3

- The steady state strain rate of creep is given by [1]

$$\dot{\varepsilon} \approx \frac{0.35}{A} D_s \left(\frac{\sigma}{G}\right)^3$$

$$D_s \propto e^{-Q/kT}$$

$$Q = Q_{sd}$$

- Here A is the interaction factor that depends on the particular viscous creep drag mechanism controlling the dislocation glide, D_s is the solute diffusivity
- The activation energy of creep (Q) has been found to be equal to that of self diffusion activation energy [2]

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[2] K L Murty, F Mohamed, J E Dorn, Acta Metall., 20 (1972) 1009

self diffusion. So that is what has been observed experimentally.
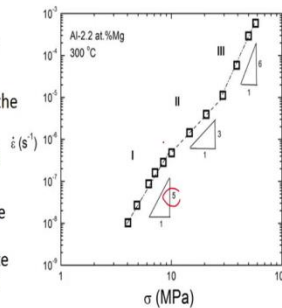
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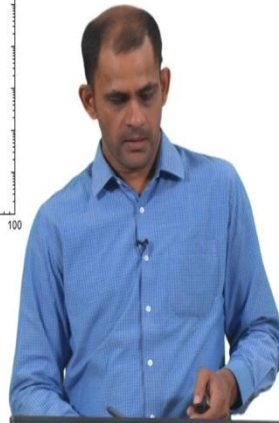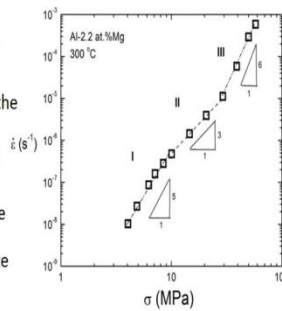
Now if you look at the strain rate versus stress behavior for a material which is capable of exhibiting viscous glide creep mechanism, so this is the kind of behavior you would notice. So you will initially at lower stress values you will get n is equal to 5.
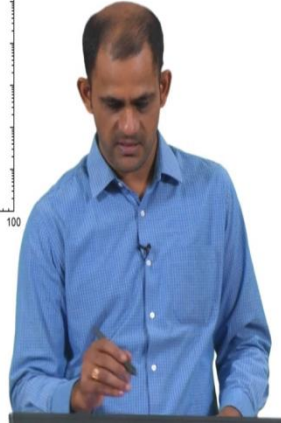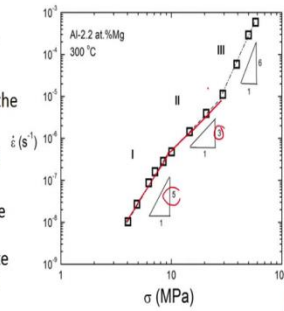
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So this is aluminum 2.2 atomic percent magnesium alloy and under certain conditions of stress and temperature, if you carry out a creep test, to begin with at lower stresses you will see a n is equal to 5 regime.

And then as you increase the stress further the material is going into the n is equal to 3 regime which is the viscous

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Viscous glide creep mechanism, n = 3

- At low stresses the alloy displays a class M type of behavior with n = 5
- At intermediate stresses, the alloy displays a class A type behavior with n = 3
- At high stresses where the dislocations are able to break away from the solute atmospheres, the class M behavior is again seen.
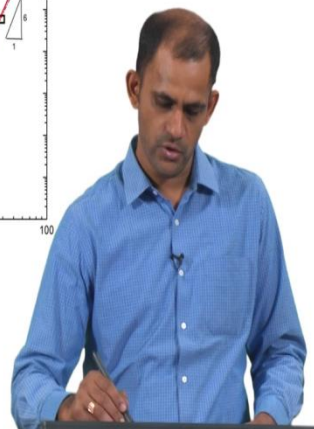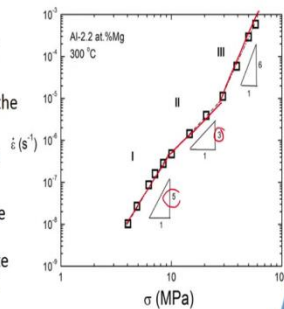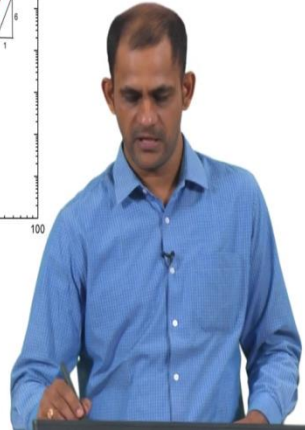
glide creep and then at higher stresses, further higher stresses again

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Viscous glide creep mechanism, n = 3

- At low stresses the alloy displays a class M type of behavior with n = 5
- At intermediate stresses, the alloy displays a class A type behavior with n = 3
- At high stresses where the dislocations are able to break away from the solute atmospheres, the class M behavior is again seen.
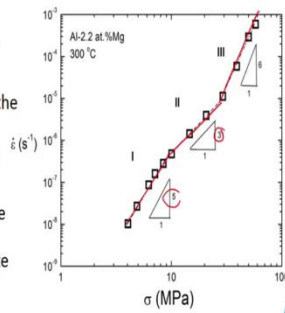
it is going to come back to the n is equal to 5

kind of behavior so which is power law kind of behavior.

So what is happening is at lower stresses you have n is equal to 5, at intermediate stresses A class type, class A type behavior, at high stresses when you apply very high stresses what is now happening is the dislocations are now actually able to break away from the solute atmosphere.

So the stress is high enough for dislocations to break away from the solute atmosphere and so now the material can again behave like a class M alloy. So
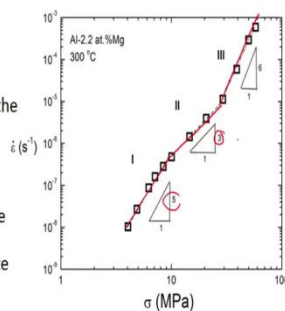
you have a class A alloy at intermediate

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Viscous glide creep mechanism, n = 3

- At low stresses the alloy displays a class M type of behavior with n = 5
- At intermediate stresses, the alloy displays a class A type behavior with n = 3
- At high stresses where the dislocations are able to break away from the solute atmospheres, the class M behavior is again seen.
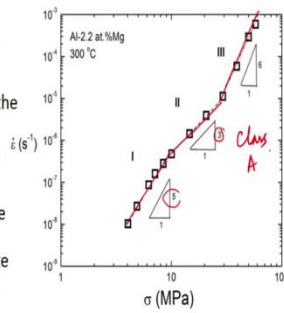
stresses that transits

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Viscous glide creep mechanism, n = 3

- At low stresses the alloy displays a class M type of behavior with n = 5
- At intermediate stresses, the alloy displays a class A type behavior with n = 3
- At high stresses where the dislocations are able to break away from the solute atmospheres, the class M behavior is again seen.
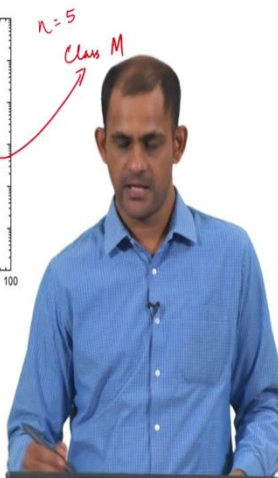
into a class M behavior at high stresses.

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Viscous glide creep mechanism, n = 3

- At low stresses the alloy displays a class M type of behavior with n = 5
- At intermediate stresses, the alloy displays a class A type behavior with n = 3
- At high stresses where the dislocations are able to break away from the solute atmospheres, the class M behavior is again seen.
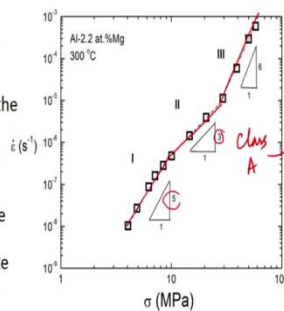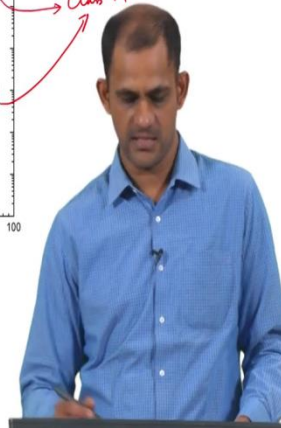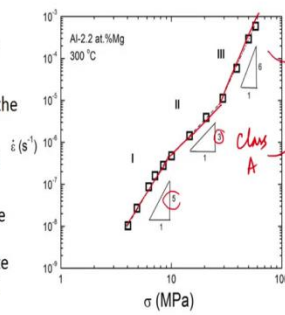
Now

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Viscous glide creep mechanism, n = 3

- The breakaway stress (τ_b) is given by

$$\tau_b = A_{11} \left(\frac{W_m^2}{kTb^3} \right) c$$

- Where A_{11} is a constant, W_m is the interaction energy between the edge dislocation and the solute atom and c is the solute concentration

Ref: J. Friedel, *Dislocations*, Pergamon Press (1964)

what should be the stress, the minimum stress that you need to apply so that the dislocations can break away from the solid atoms?

Well that stress is given by this

Viscous glide creep mechanism, n = 3

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$$\tau_b = A_{11}\left(\frac{W_m^2}{kTb^3}\right)c$$

- Where A_{11} is a constant, W_m is the interaction energy between the edge dislocation and the solute atom and c is the solute concentration

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following relation. So the breakaway stress tau b is a function of the constant A_{11}, and it is also a constant of, is also a function of

Viscous glide creep mechanism, n = 3

- The breakaway stress (τ_b) is given by

$$\tau_b = A_{11}\left(\frac{W_m^2}{kTb^3}\right)c$$

- Where A_{11} is a constant, W_m is the interaction energy between the edge dislocation and the solute atom and c is the solute concentration

Ref: J. Friedel, *Dislocations*, Pergamon Press (1964)

the interaction energy between the edge dislocation and the solute atom.

Well, because the solute atoms are going to stick to the grain, to the edge dislocation so your, the stress that you apply should be able to provide an energy higher than the kind of the binding energy between the solute atom and the edge dislocation.

So once your stress is able to account for that energy then, or higher, then your dislocation is going to separate from the solute atoms. And it is also, tau b is also dependent

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Viscous glide creep mechanism, n = 3

- The breakaway stress (τ_b) is given by

$$\tau_b = A_{11}\left(\frac{W_m^2}{kTb^3}\right)c$$

- Where A_{11} is a constant, W_m is the interaction energy between the edge dislocation and the solute atom and c is the solute concentration

Ref: J. Friedel, *Dislocations*, Pergamon Press (1964)

on the concentration of the solute atom. So that is, that is, this once you apply this amount of stress

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Viscous glide creep mechanism, n = 3

- At low stresses the alloy displays a class M type of behavior with n = 5
- At intermediate stresses, the alloy displays a class A type behavior with n = 3
- At high stresses where the dislocations are able to break away from the solute atmospheres, the class M behavior is again seen.
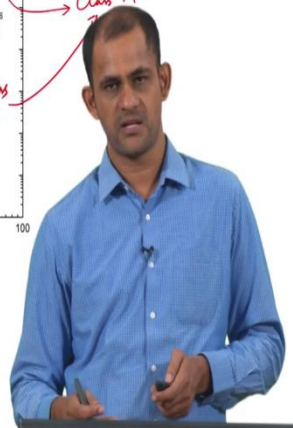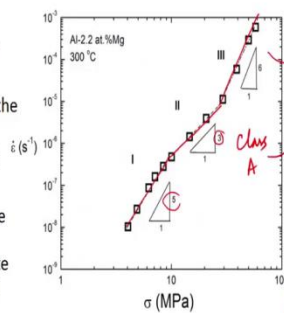
then you can move away from class A into class M type of creep behavior.

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Viscous glide creep mechanism, n = 3

- The breakaway stress (τ_b) is given by

$$\tau_b = A_{11}\left(\frac{W_m^2}{kTb^3}\right)c$$

- Where A_{11} is a constant, W_m is the interaction energy between the edge dislocation and the solute atom and c is the solute concentration

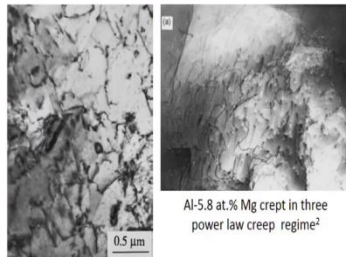Ref: J. Friedel, *Dislocations*, Pergamon Press (1964)

Now

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Viscous glide creep microstructures

- The deformation microstructures of class A alloys crept in the three power law creep regime are devoid of subgrains. **Recovery processes are considered less important**

Al-5.8 at.% Mg crept in three power law creep regime[2]

0.5 μm

Deformation microstructure in a Nb-modified Zircaloy crept in the three power law regime[1]

[1] J. Ravi, High temperature creep of Nb-modified Zircaloy, MS Thesis, NC State University, Raleigh, NC, USA, 1991.
[2] Henshall GA, Kassner ME, McQueen HJ. Metall Trans 1992;23A:881.

let us talk about some of the microstructures that are generally associated with a creep sample crept in this regime. Now the deformation microstructures of class A crept samples are generally devoid of sub grains.

So sub grain formation

Viscous glide creep microstructures

- The deformation microstructures of class A alloys crept in the three power law creep regime are <u>devoid of subgrains</u>. **Recovery processes are considered less important**

Deformation microstructure in a Nb-modified Zircaloy crept in the three power law regime[1]

Al-5.8 at.% Mg crept in three power law creep regime[2]

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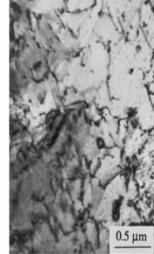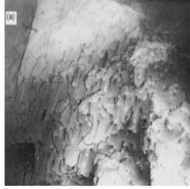
[1] J. Ravi, High temperature creep of Nb-modified Zircaloy, MS Thesis, NC State University, Raleigh, NC, USA, 1991.
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if you recall is generally, the sub grains form when the dislocations are arranging themselves in certain fashion, you need the dislocations to create some configurations. And this is generally associated with strain hardening and recovery, a balance between strain hardening and recovery. Now here, and that is controlled by dislocation climb.

And in this particular case, because it is glide controlled instead of climb controlled, so that is why, one we have small primary creep and secondly as a result of that you, a small primary creep also means you do not see sub grain formation. And so as if you can see here, the dislocations are by and large random.
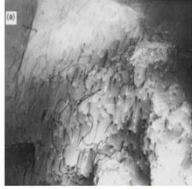
So you have two samples, one is

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Viscous glide creep microstructures

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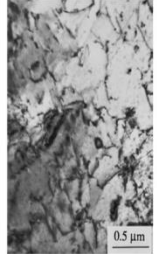
Deformation microstructure in a Nb-modified Zircaloy crept in the three power law regime[1]

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[1] J. Ravi, High temperature creep of Nb-modified Zircaloy, MS Thesis, NC State University, Raleigh, NC, USA, 1991.
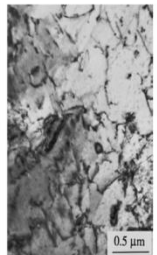[2] Henshall GA, Kassner ME, McQueen HJ. Metall Trans 1992;23A:881.
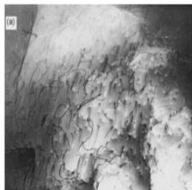
a crept Zircaloy sample and the other

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Viscous glide creep microstructures

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Deformation microstructure in a Nb-modified Zircaloy crept in the three power law regime[1]

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one is the crept Aluminum Magnesium sample. In both the samples what they observed was random set of dislocations, some
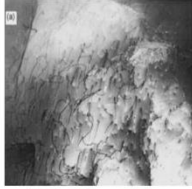
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Viscous glide creep microstructures

- The deformation microstructures of class A alloys crept in the three power law creep regime are devoid of subgrains. **Recovery processes are considered less important**

Al-5.8 at.% Mg crept in three power law creep regime[2]

0.5 µm
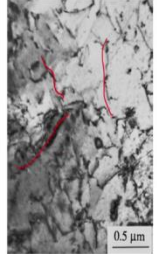
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amount of bowing 0:34:01.3 but by and large random distribution of dislocations.

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Viscous glide creep microstructures

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0.5 µm
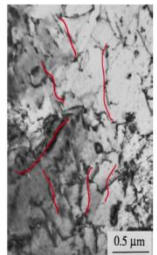
Deformation microstructure in a Nb-modified Zircaloy crept in the three power law regime[1]
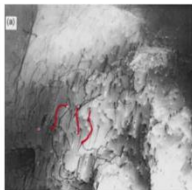
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[2] Henshall GA, Kassner ME, McQueen HJ. Metall Trans 1992;23A:881.

So this is, these are the kind of microstructures which you can expect to be associated with a viscous glide creep sample.