Creep Deformation of Materials
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Mechanisms of Creep Part IV

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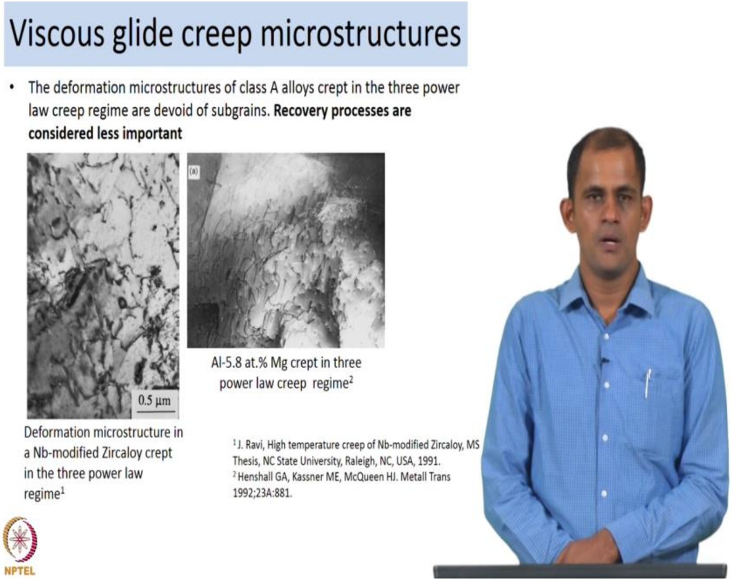

Ok so we were talking of the kind of microstructures that one could expect in a material that has crept in the viscous glide creep regime. So we were talking about how the microstructure would consist of lot of random dislocations, no particular configurations of dislocations as you would expect in a dislocation climb control regime.

So talking of dislocation climb control regime

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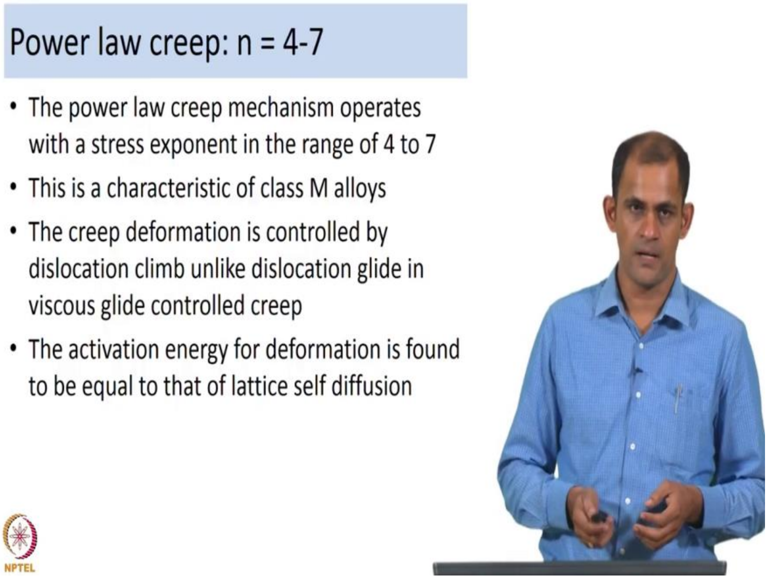

this is the kind of behavior that one would expect in a creep regime known as the power law creep regime. So the power law creep regime is, is signified by a stress exponent which varies in the range of 4 to 7. So materials which usually experience a power law creep, they tend to provide stress exponent values in the range of 4 to 7.

Now we were talking about two types of alloys. Earlier I was talking about class A alloys and then class M alloys. Now class A alloys is what we dealt with in the viscous glide, viscous glide creep regime so where the dislocation glide is actually the rate controlling step.

And in class M alloys which is what we are talking now, here the dislocation climb is the rate controlling step. And the stress exponent values of 4 to 7 is the characteristic of class M alloys. So I already mentioned it is dislocation control by dislocation climb.

Now the activation energy for deformation in materials which creep under power law creep regime is this deformation proceeds by activation energy equal to that of lattice self diffusion. And another interesting aspect of this regime is this is a grain size, so when you have

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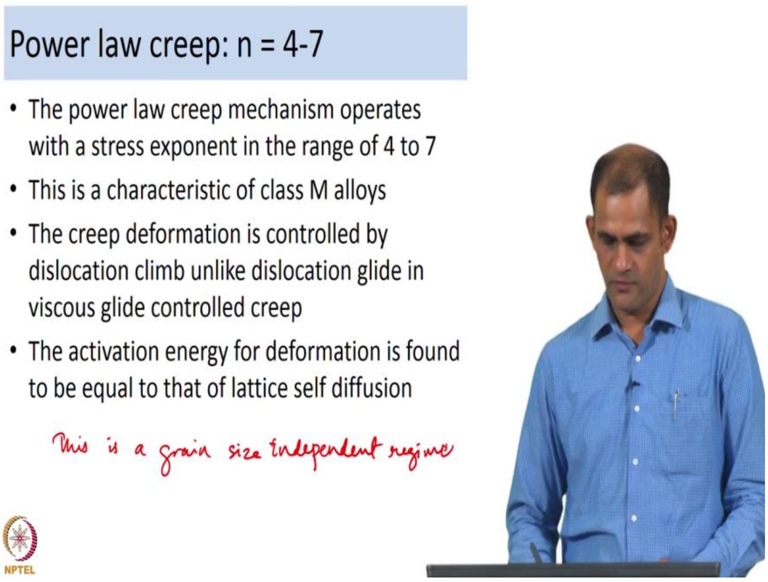

materials creeping in the power law creep regime you do not see any grain size dependence that means p is equal to 0.

So if you contrast this with what we learnt about Nabarro-Herring creep and Coble creep, Nabarro-Herring you saw p is equal to 2, Coble creep you saw p is equal to 3, grain boundary sliding we saw p is equal to 2, and in power law creep there is no grain size dependence, so p is equal to 0.

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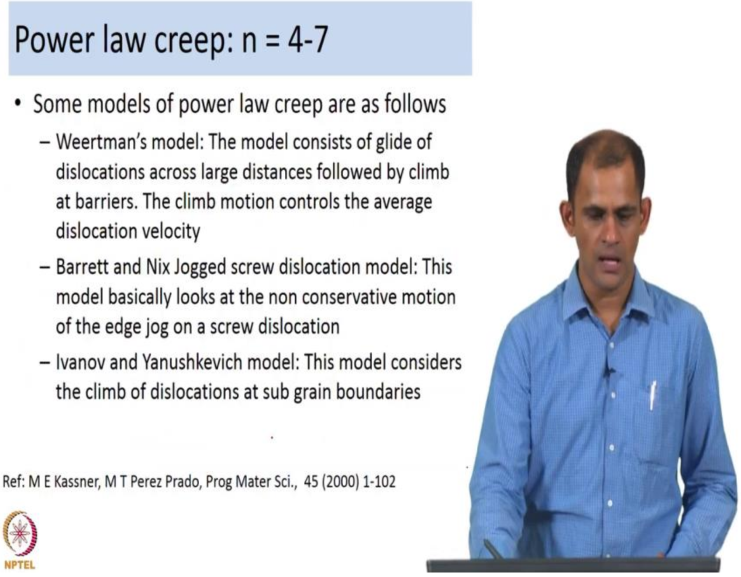

Now power law creep, we have n is equal to 4 to 7 and there have been different models which people have proposed to explain the material behavior in power law creep regime. These models also provide equations for correlation for the strain rate of deformation and its dependence on the applied stress and temperature and things like that.

So some of the different models are listed here. One of the first models to be

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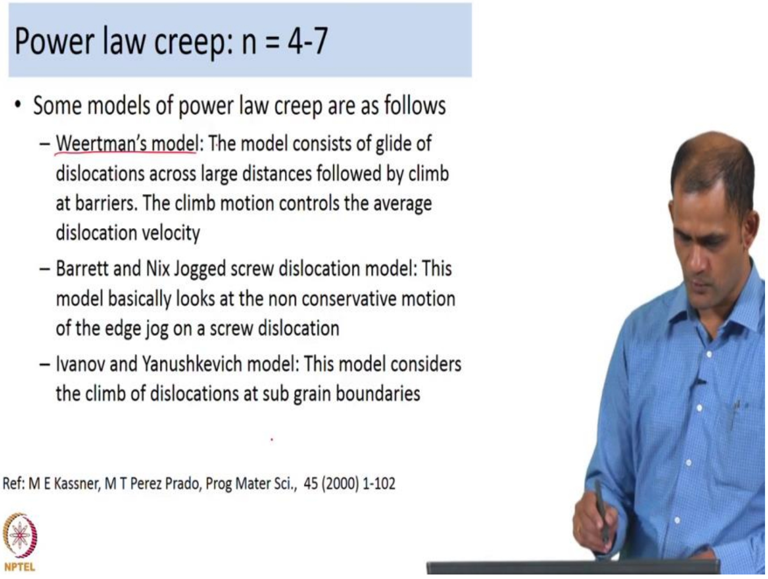

proposed in, to explain power law creep is the Weertman's model.

So the Weertman's model basically consists of, this model basically, the whole concept is about, you have source of dislocations, the source emits dislocations, these dislocations glide along the slip plane. They travel large distances and during the motion of the dislocations, plastic strain is generated.

Now whenever the dislocations come across any barriers as an example is like, long range stresses acting on the dislocations, long ranges stresses coming from other dislocations or networks of dislocations, so in those cases dislocations find it difficult to move further.

And at that point of time, the dislocations would start climbing in order to move up the barrier, probably meet the dislocations on the other slip plane and highlight each other or continue with their motion.

So in this model, the Weertman's model the climb motion controls the average dislocation velocity. So you have two steps. You have the glide step as well as climb step and what people have noticed or Weertman's proposal is that the climb step is what is the rate controlling and it actually influences the average dislocation velocity involved during creep.

Another model is the Barrett and Nix Jogged screw dislocation model. This model is also used to explain power law creep. So in the Jogged screw model, the model basically looks at non-conservative motion of the edge jog on a screw dislocation.

So if you recall we were talking about edge jogs and kinks in dislocations and one of the breaks in dislocation which we considered very important, very important from a plastic deformation point of view is edge jogs on screw dislocations.

Now edge jogs on screw dislocations are important because the edge jogs will not glide along with the screw dislocations. They actually, because the Burgers vector is lying out of the slip plane so the Burgers vector for the edge dislocation is lying out of the slip plane, so the only way the screw dislocation with an edge jog would move is by non-conservative motion.

So the screw dislocation has to drag the edge dislocation along with it and so dislocation climb is basically involved in the process. So Bartett and Nix suggested that when you have screw dislocation which has jogs then the motion of these at high temperatures can also lead to a stress exponent in the range of 4 to 7.

Another model that has been proposed is by Ivanov and Yanushkevich. And this model basically considers the climb of dislocations at sub grain boundaries.

So Ivanov and Yanushkevich said that the presence of sub grain boundaries acts as barriers to dislocation motion and in their model they said the climb of dislocations at the sub grain boundaries is what would determine the stress exponent and this is what would generally give you a stress exponent in the range of 4 to 7.

So a review of these different models has been provided by Kassner and Perez Prado in their review article in Progressive Material Science and I would encourage you to go through this article to get an understanding, a more detailed understanding of these different models.

Having said that, in this particular lecture I am going to take you a little bit into the detail of Weertman's derivation for the stress exponent using model and then I am also going to talk to some detail about the Barrett and Nix Jogged screw model.

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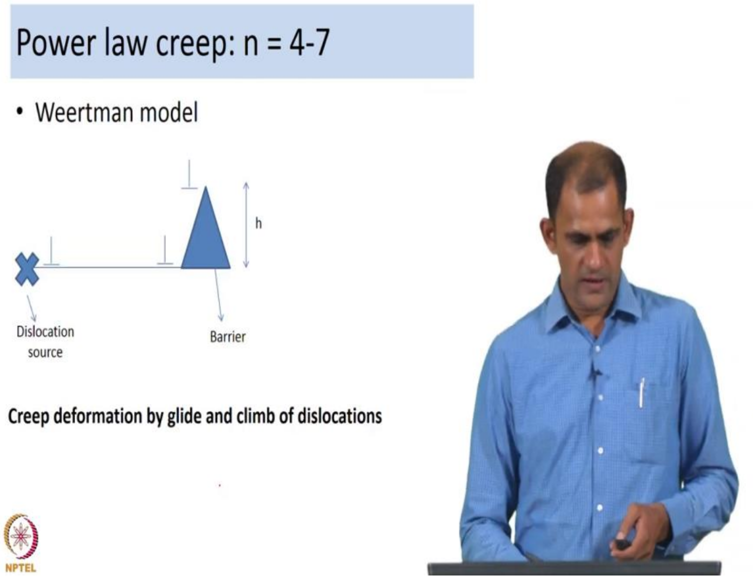

So first the Weertman's model, so here Weertman's model was the first model that was developed to explain the observation of a stress exponent in the range of 4 to 7. So

Weertman's model is as follows. So you have a dislocation source. So this cross actually indicates a dislocation source. And it is emitting some edge dislocations.

So dislocation source, a dislocation source is a Frank Read source basically, you can say a Frank Read source

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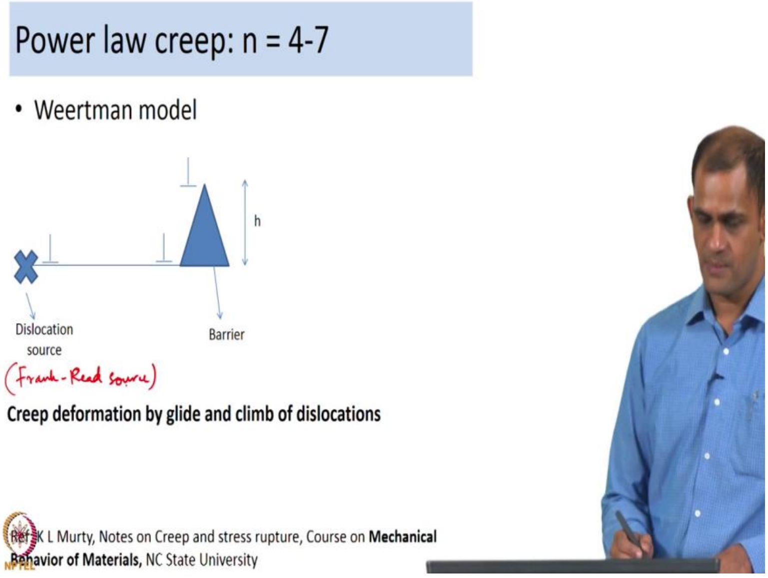

emitting dislocations. So these dislocations glide along there, slip plane

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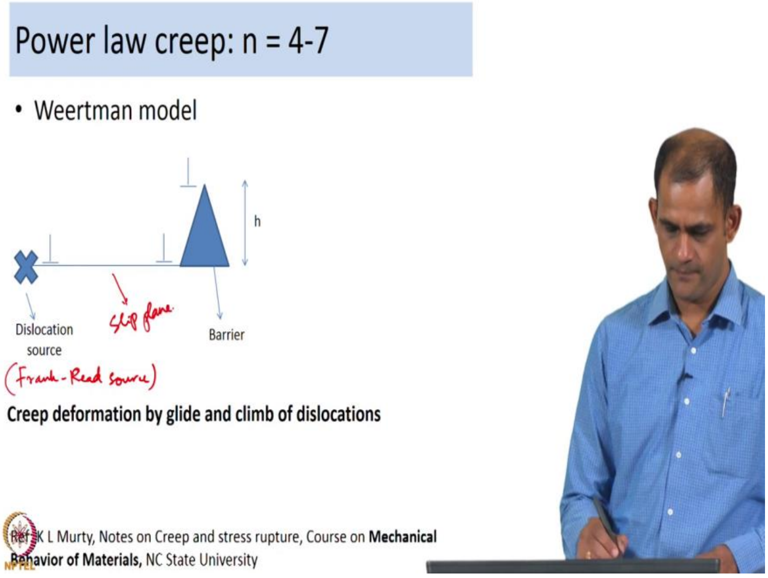

till they encounter a barrier. An example of a barrier is a Lomer Cotrell barrier. So this is an example of a

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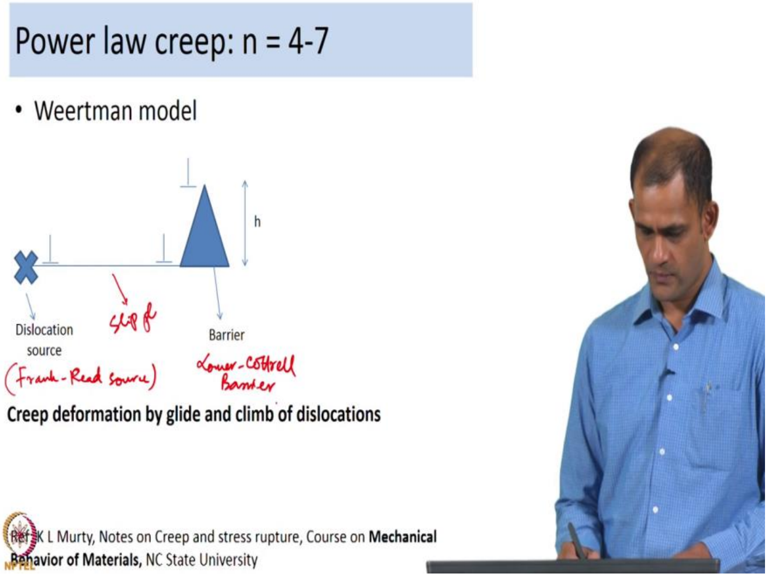

barrier to dislocation motion.

So now further emission of dislocation, so the moment this is, as you can recall from our previous portions, when a dislocation encounters a barrier and the dislocations continue to pile up at the barrier, there is going to be a back stress acting and this back stress will prevent further emissions of dislocation. So for creep deformation to continue this barrier has to be overcome in some form.

So the overcoming of this barrier can happen by dislocation climb.

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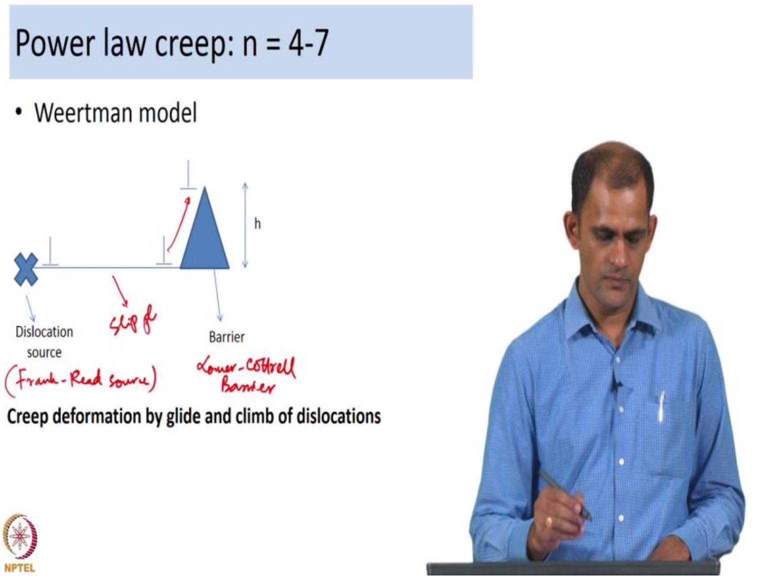

So he, Weertman basically used this concept and he derived equations and arrived at a correlation between the strain rate and stress

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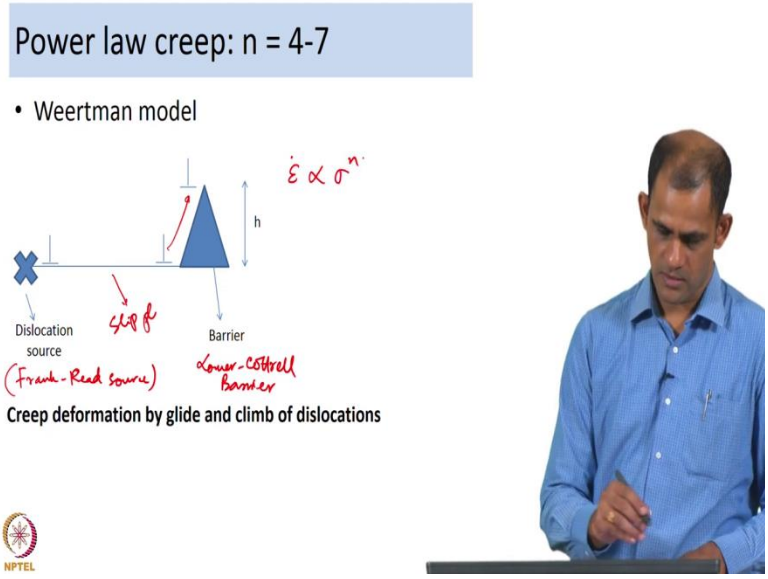

using that. So from here on I will be talking about the Weertman model and how he has derived the equations.

So we are assuming a source of dislocations. Assume a dislocation source, so here it is a Frank Read source emitting dislocations and we are talking of edge dislocations. The barriers to the edge dislocations is Lomer Cotrell barrier. So the height of the barrier is equal to h

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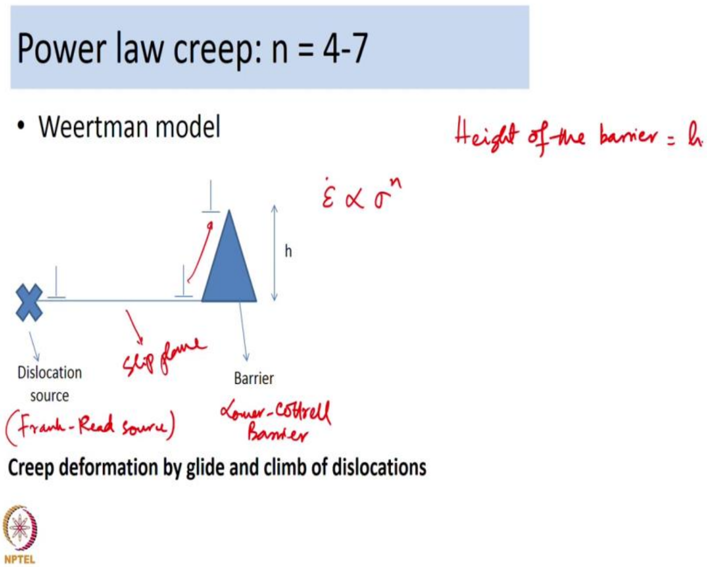

and the distance L, and the dislocations have to climb to overcome this barrier.

The distance the dislocations move before they encounter the barrier is L. So

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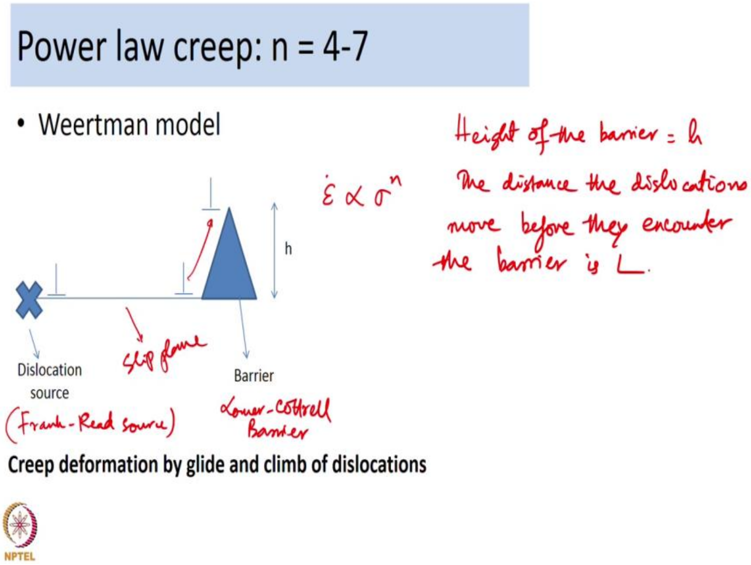

it is basically the distance along the glide on the slip plane. So that is the distance the dislocations move before they encounter the barrier. Now the total creep strain is a result of the glide plus climb event.

So you have two events

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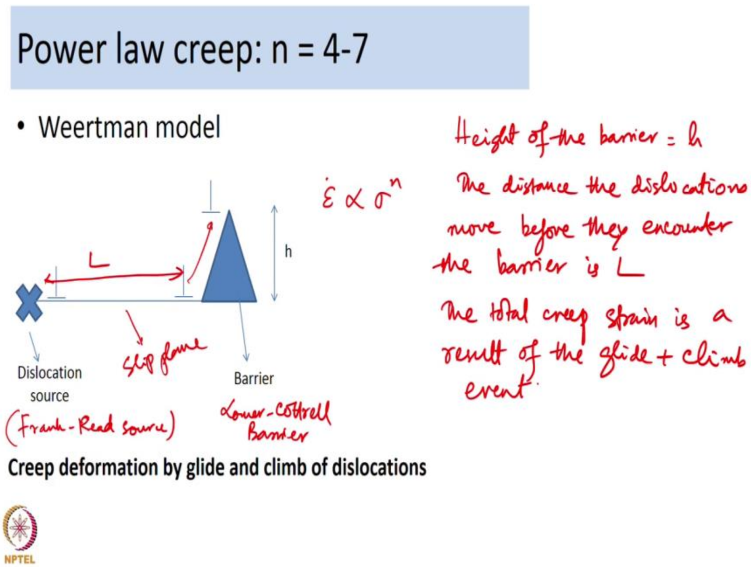

here, one is the glide event and one is the climb event. So the total creep strain or creep deformation is a result of this. So here what you learn during the process derivation is that the glide event actually determines the extent of the plastic deformation and the climb event determines the strain rate of the plastic deformation.

So now let us say that the total shear strain coming out of this, which is a result, total shear strain which is a result of this glide plus climb event will have, let us say, let us call it the total creep shear strain delta gamma. So delta gamma will be strain during glide plus strain during climb.

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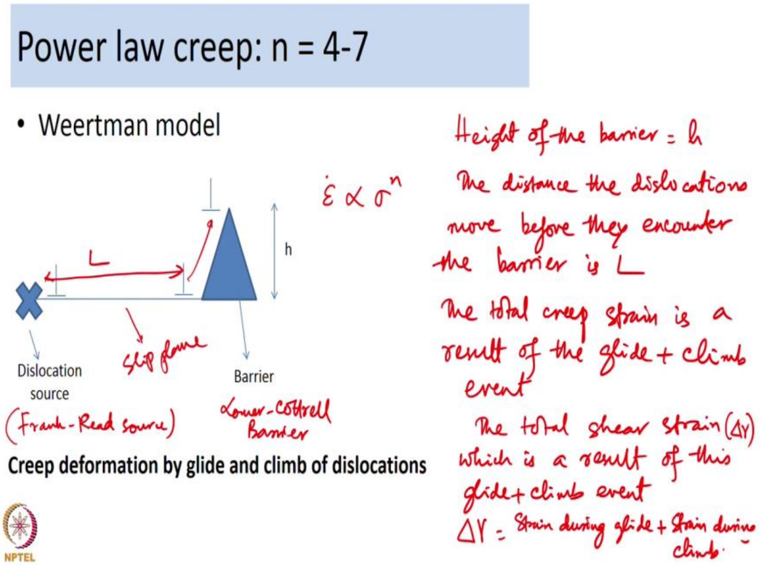

So these are the two terms that we will come about, so we will call delta gamma g as strain during glide and delta gamma c is equal to strain during climb.

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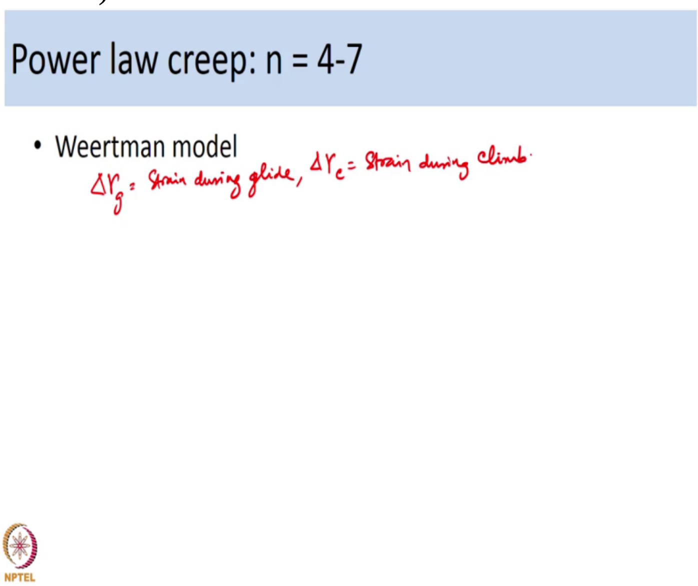

So the strain generated, the strain due to climb is lower than the, significantly lower than the strain due to glide.

So therefore

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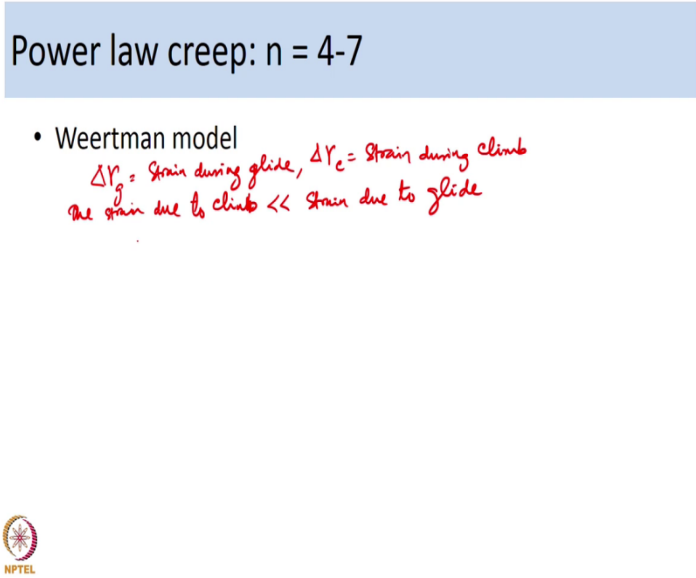

delta gamma which was, which we originally started by saying delta gamma g plus delta gamma c

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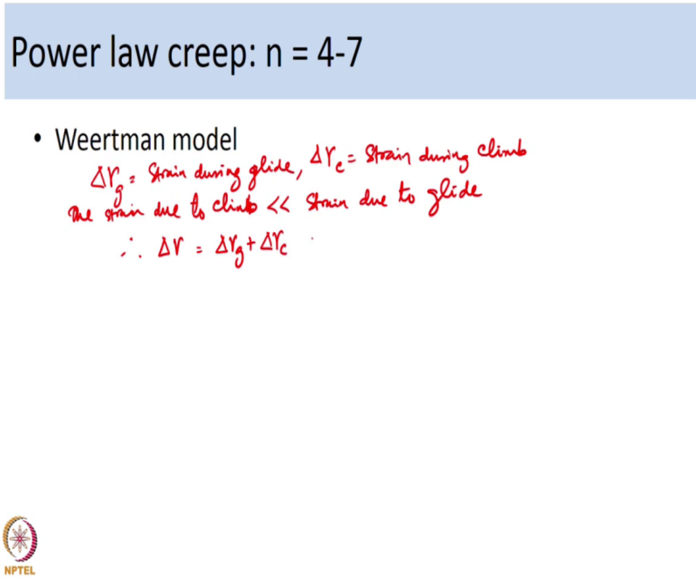

can be approximated as delta gamma g.

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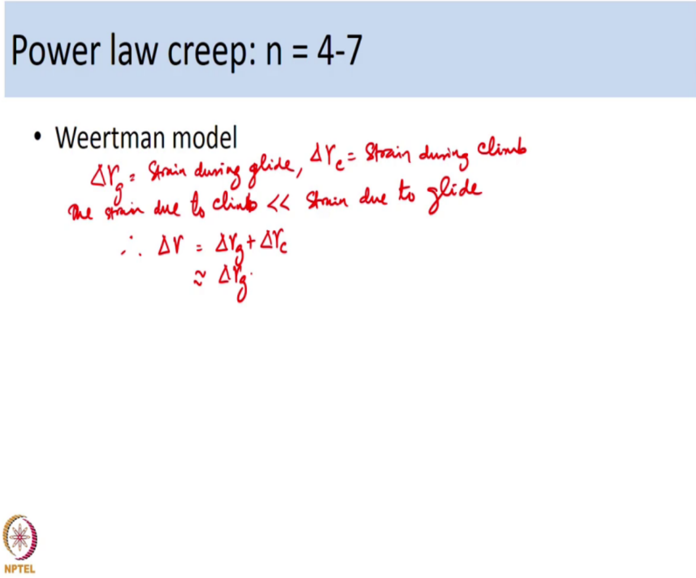

Now delta gamma g is a result of the dislocations moving along the glide plane so delta gamma g can be written as rho b L.

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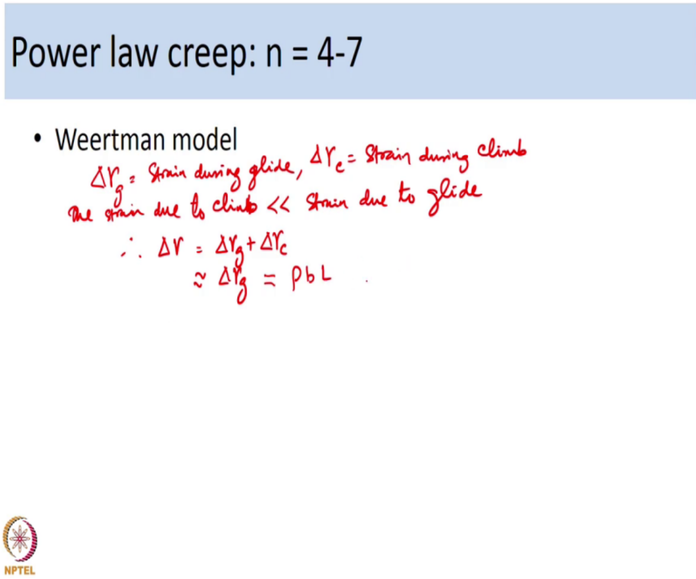

So rho is dislocation density,

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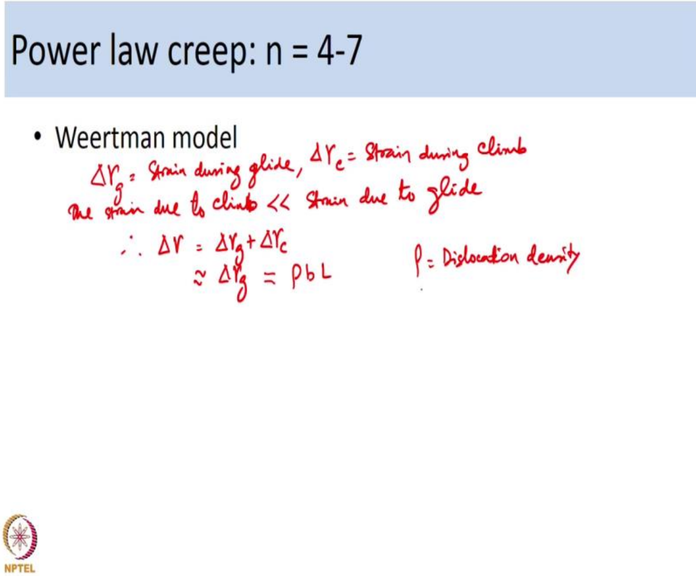

b is Burgers vector

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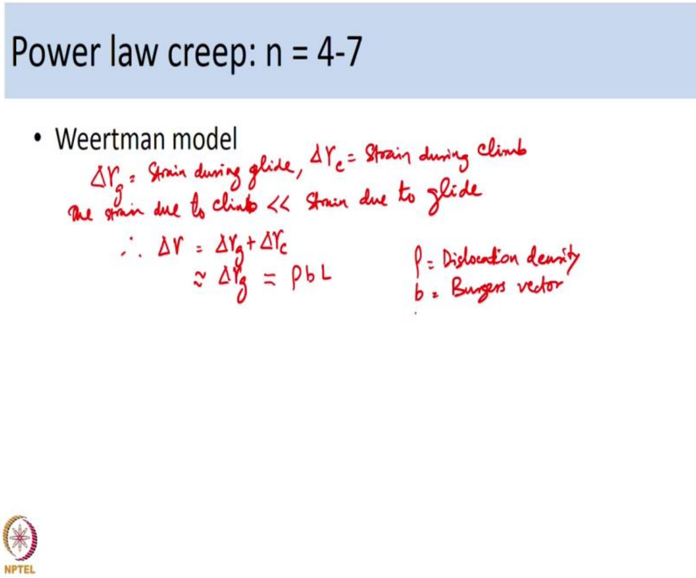

and L is the distance the dislocations have, which I already mentioned earlier L is the distance from the source to the barrier so it is basically the distance the dislocations have glided.

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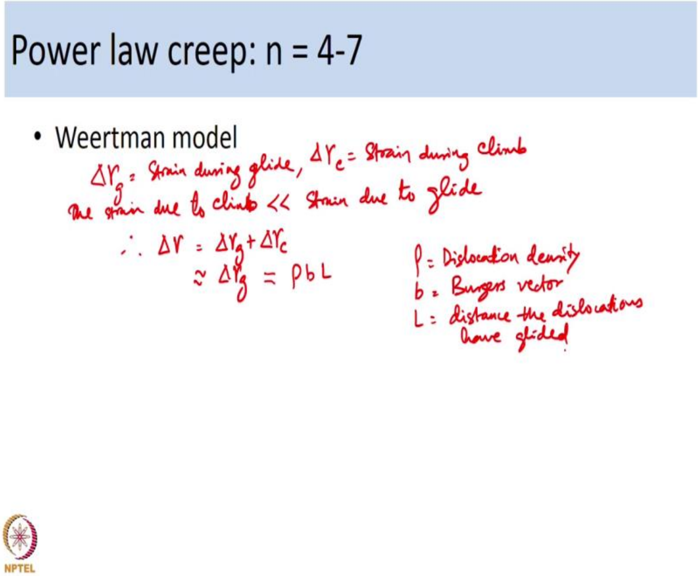

Now the total time of the glide plus climb event can be written as t is equal to t g plus

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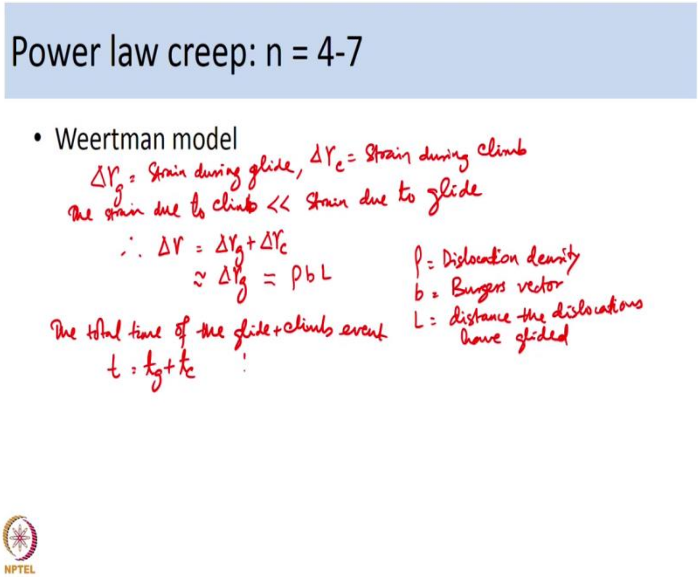

tc. t g is the time for glide

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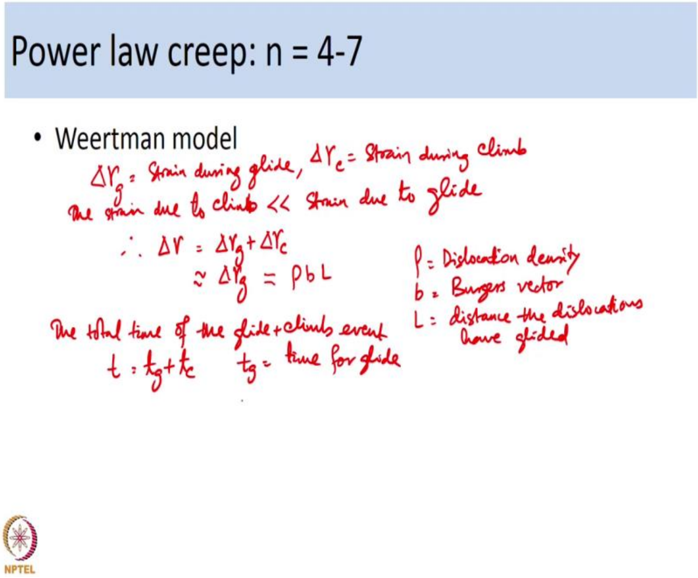

and tc is the time for climb.

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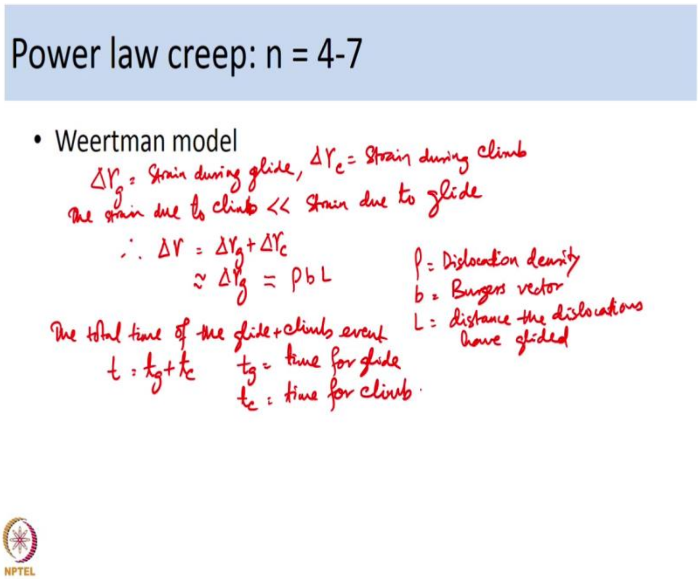

Now what has been observed or what Weertman proposed is the time for climb is going to be significantly larger than the time for glide,

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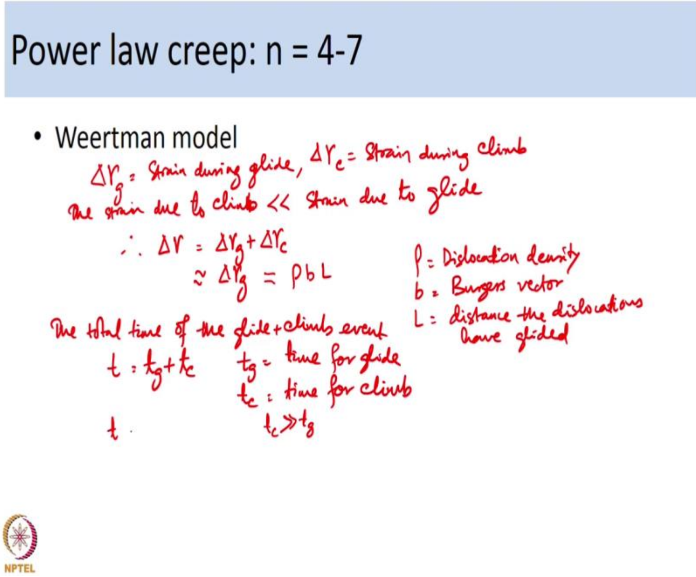

so t can be approximated as t is equal to

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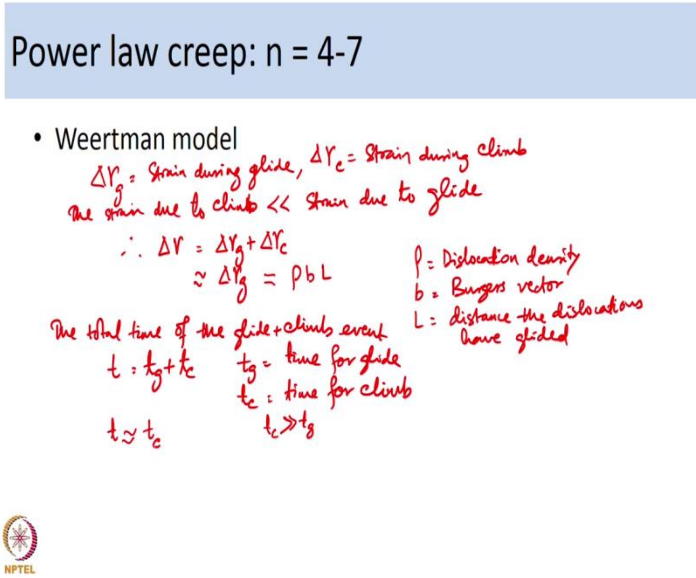

t c. And t c can be written, so t is approximately equal to t c which can be written as the height of the barrier divided by the velocity

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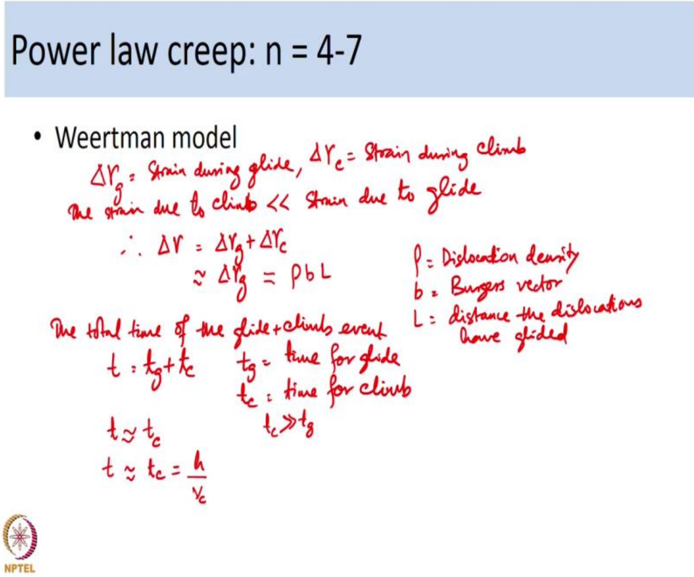

of the dislocation during dislocation climb.

So v c is equal to the climb velocity.

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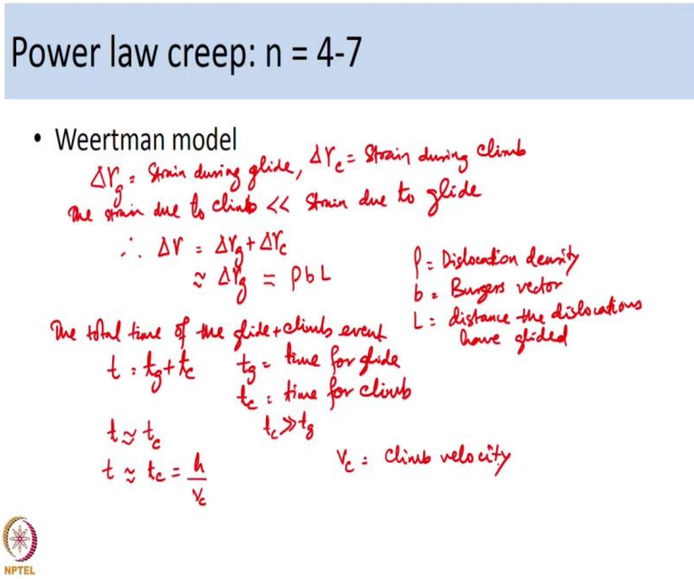

So which means the shear strain rate is going to be delta gamma over t so that will be equal to delta gamma g over t c,

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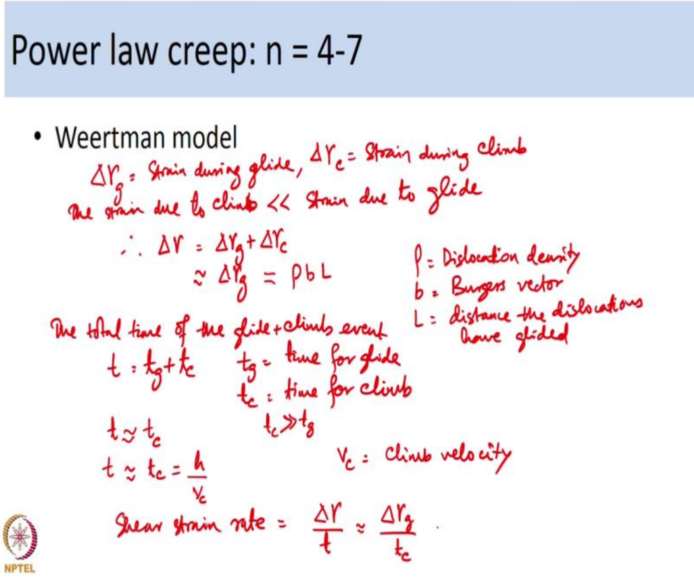

so that is equal to rho b L divided by h over v c.

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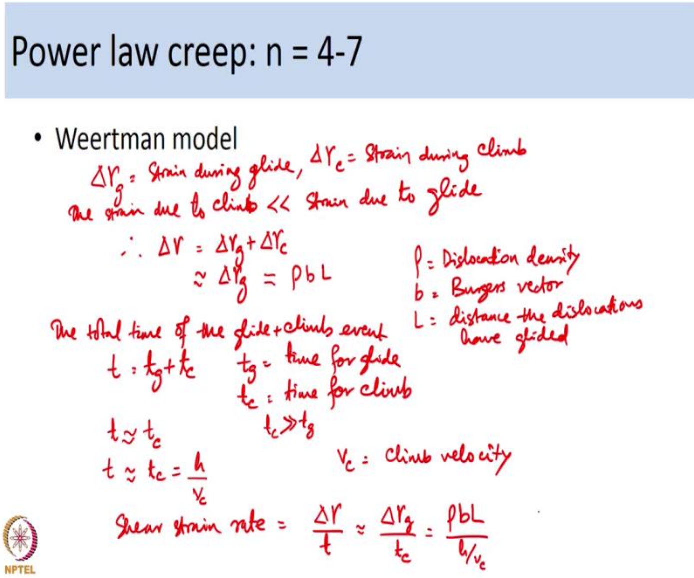

So shear strain rate gamma dot is equal to rho L, L over h into

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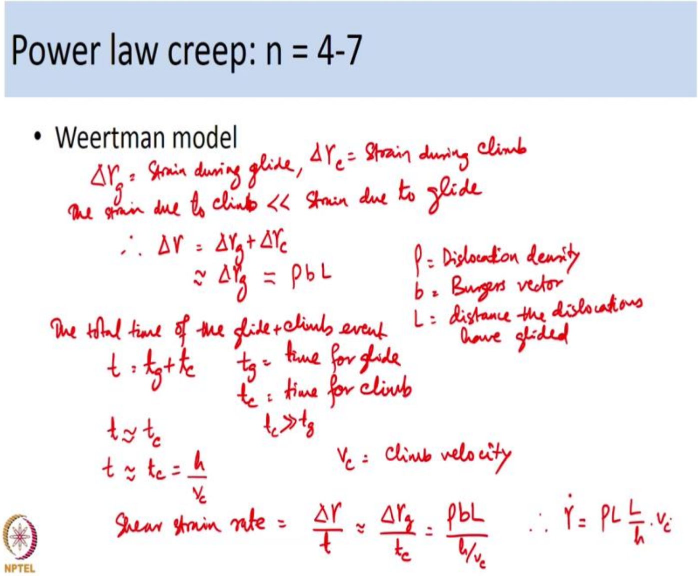

v c. So that is the relation he got for the shear strain rate as a function of the climb velocity. Now Weertman further said that the climb velocity v c is proportional to the vacancy concentration, the gradient in, the climb velocity is proportional to the vacancy concentration

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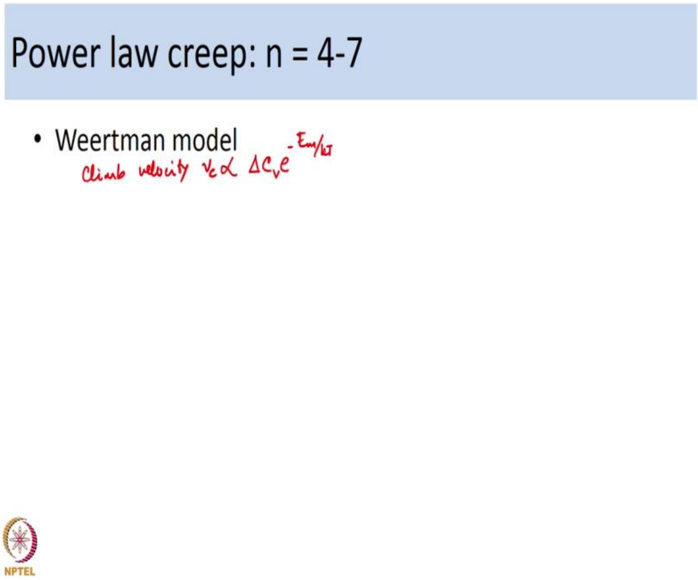

times the effect of the activation energy.

So the, since the climb, so dislocation climb requires flux of vacancies. So dislocation is climbing up so you need vacancies to go to the dislocation if you, if the dislocation is climbing down, a positive dislocation I mean then you have vacancies moving away from the dislocation core.

So in either case there is a flux of vacancies and this flux of vacancies depends on the concentration gradient which is basically the equilibrium vacancy concentration minus the vacancy concentration available at the dislocation region. So, so there is, so the climb velocity is dependent on this gradient of concentration of vacancies.

So here E m is the activation energy for

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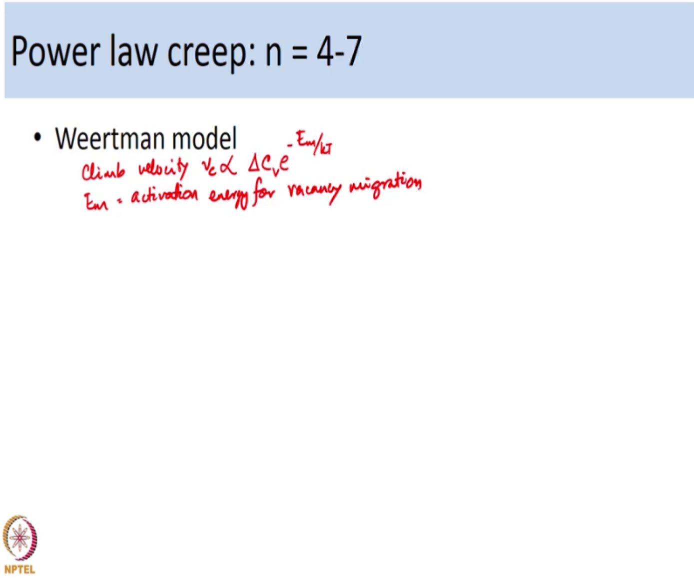

vacancy migration, delta C v is the concentration gradient of vacancies helping the dislocation climb process.

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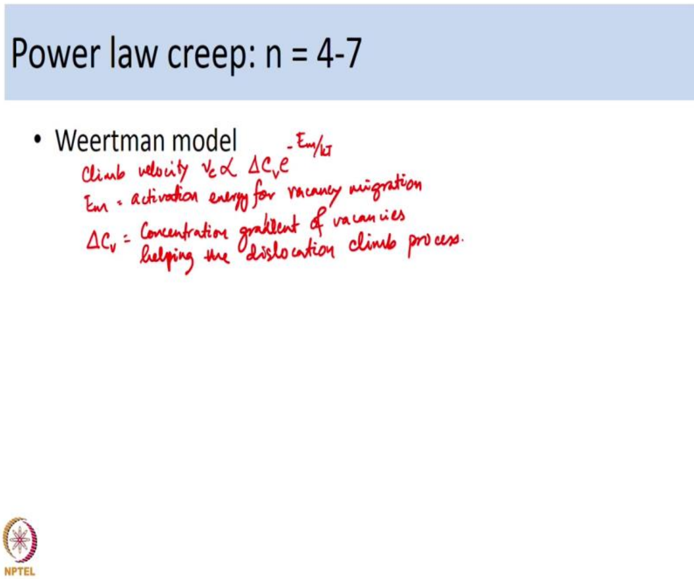

So delta C v was, can be written as C v plus minus C v minus,

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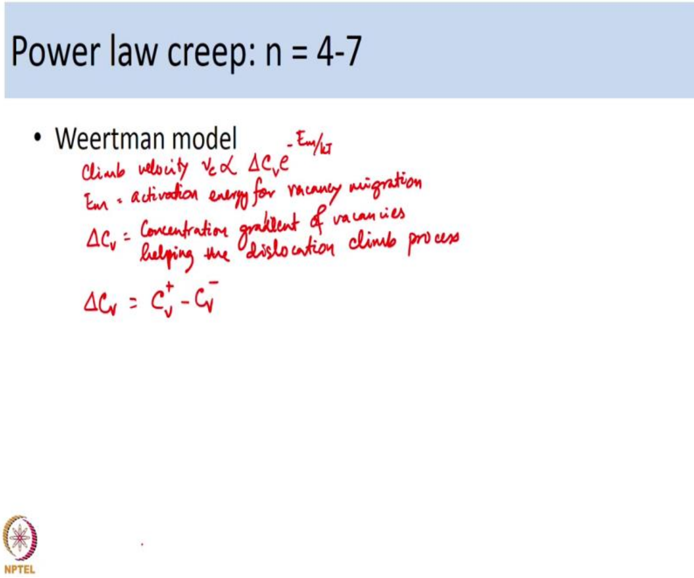

so if you recall it is a similar concept that we used in derivation of the Nabarro-Herring creep equation.

So in the presence of a stress, in the presence of a tensile stress you will have excess of vacancies, in the presence of a compressive stress you will have a reduction in the concentration of vacancies.

So this is a similar concept was used by Nabarro, so, by Weertman so he said C v plus can be written as C v e to the power sigma v over k T and C v minus will be written as C v naught e to the power minus sigma v over k T.

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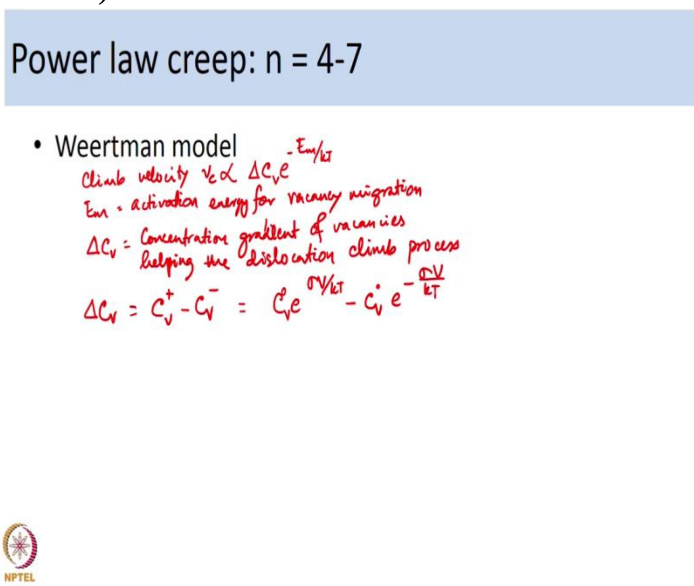

So delta C v will be then C v naught into 2 sin h sigma v over k T. So this is the influence of the stress, applied stress and

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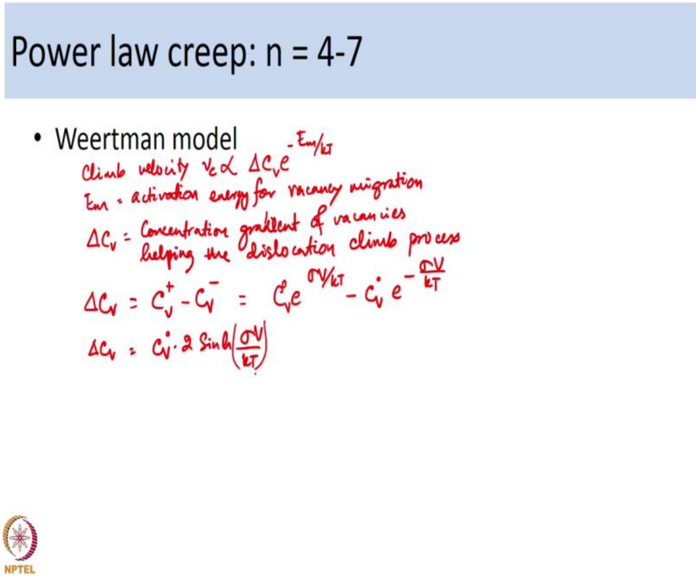

the, what it basically says is that in the presence of a stress, the climb velocity is going to be influenced.

So now if we include this into the shear strain rate equation so then gamma dot will eventually turn out as rho b L by h C v naught e to the power minus E m over k T into sin h, 2 sin h into sigma v over k T. So

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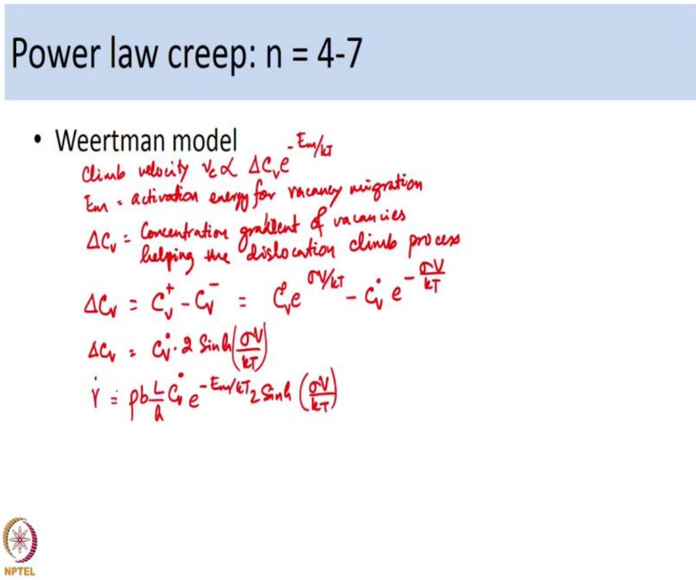

we can use a constant, say A so it is a, gamma dot will become basically a constant A into all this.

So your creep strain rate is another constant K into the shear strain rate, so basically epsilon dot will become, say another constant A 1 into rho b L over h into C v naught e to the power E m over k T into sigma v over k T. So

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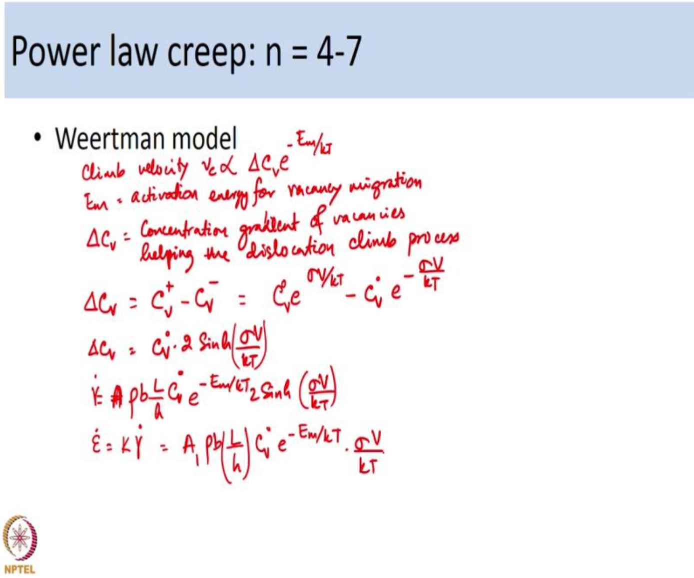

sigma v over k T comes from, at lower stresses sin h sigma v over k T becomes sigma v. So, over k T.

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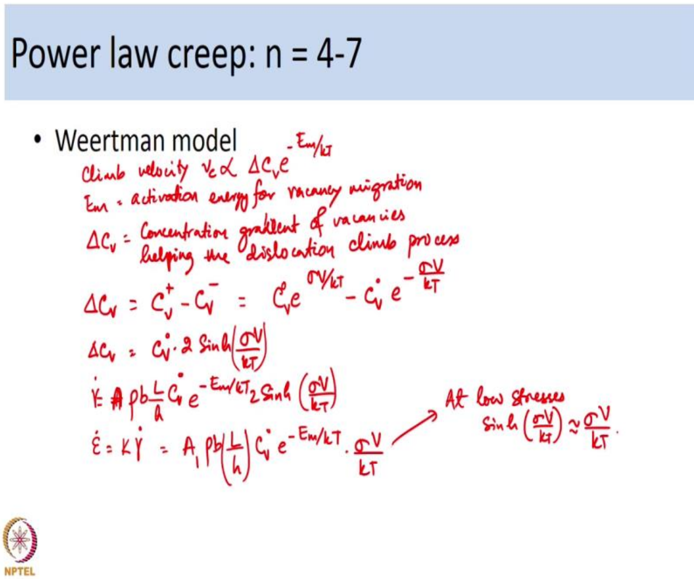

So that is the assumption he made. So then that becomes, so strain rate becomes, so A 1 into rho b L by h, so this term he takes as D L,

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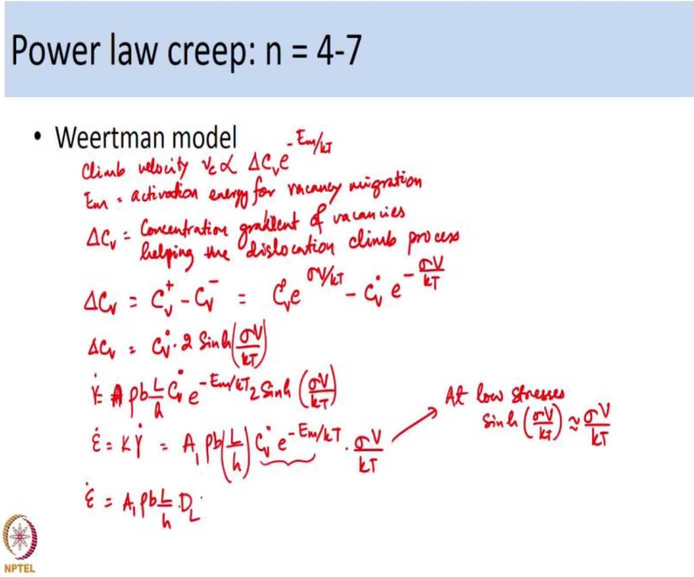

so C v naught is basically the equilibrium concentration of vacancies. C v naught is e to the power E v over k T, so

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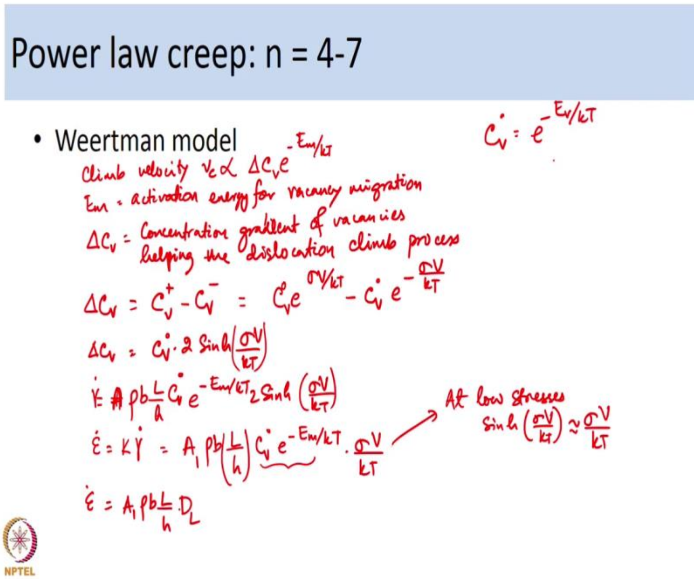

E v is the activation energy for formation of vacancies.

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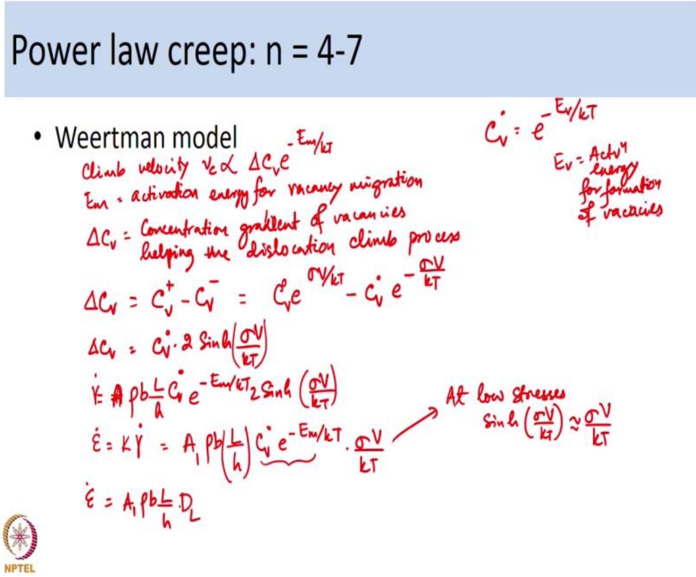

So this term becomes D L into sigma v over k T.

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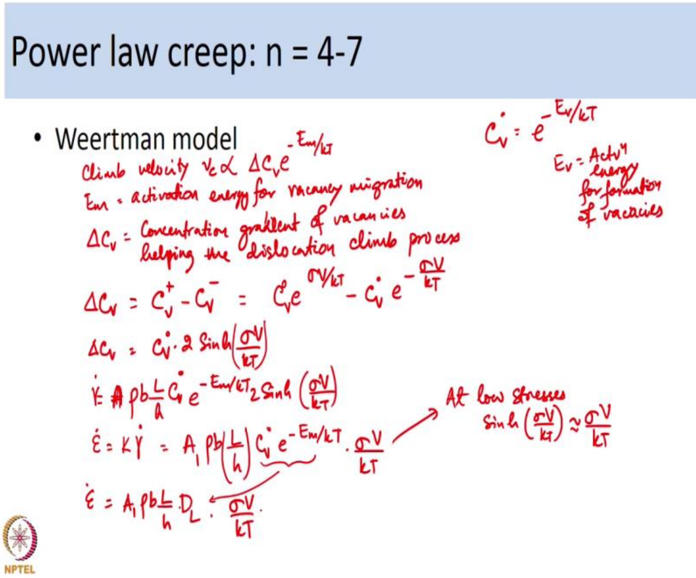

So that is the strain rate.

And then from Taylor's relation, so if you recall in one of our earlier portions I was talking about the relationship between dislocation density and applied stress, so what people have noticed is the dislocation density is some constant times the applied, square of the applied stress.

So he used

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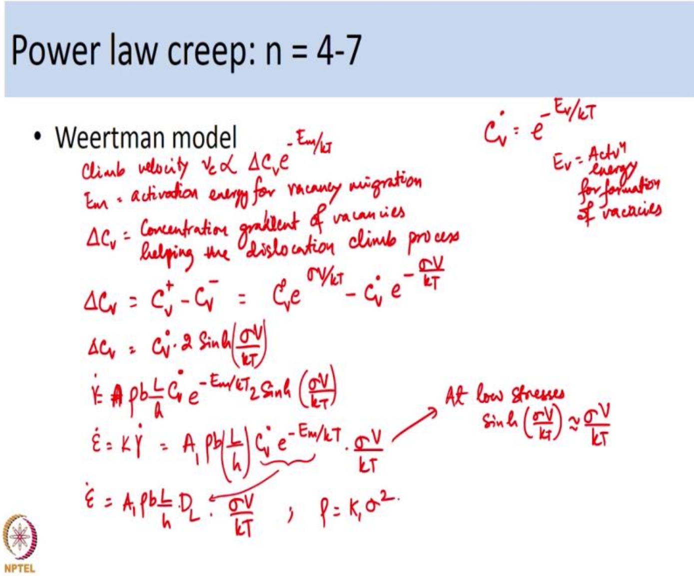

the relation here and so this equation then becomes, epsilon dot is some constant A 2 into L by h into D L into sigma cube.

So this is how

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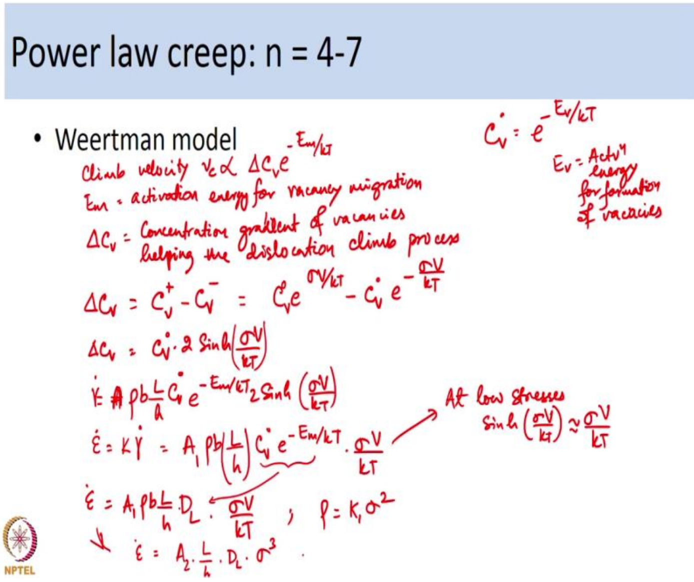

the equation turns out to be and if you see there is a dependence of sigma cube, epsilon dot proportional to sigma cube.

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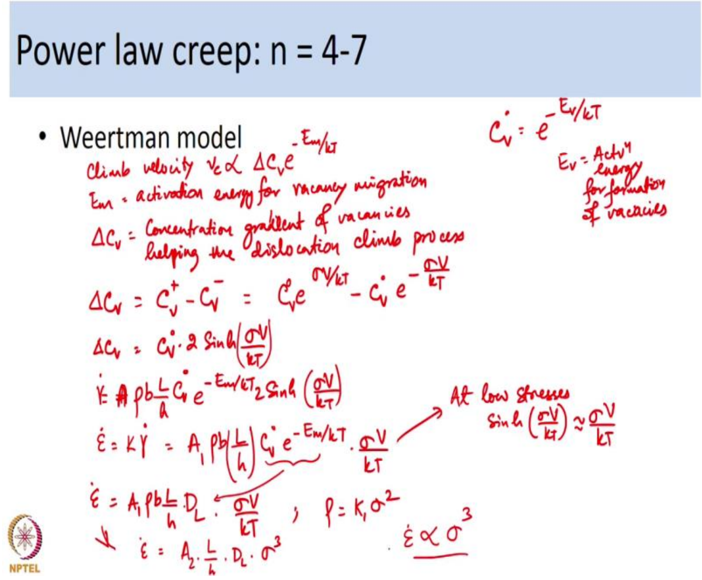

So this is called the natural power law creep equation.

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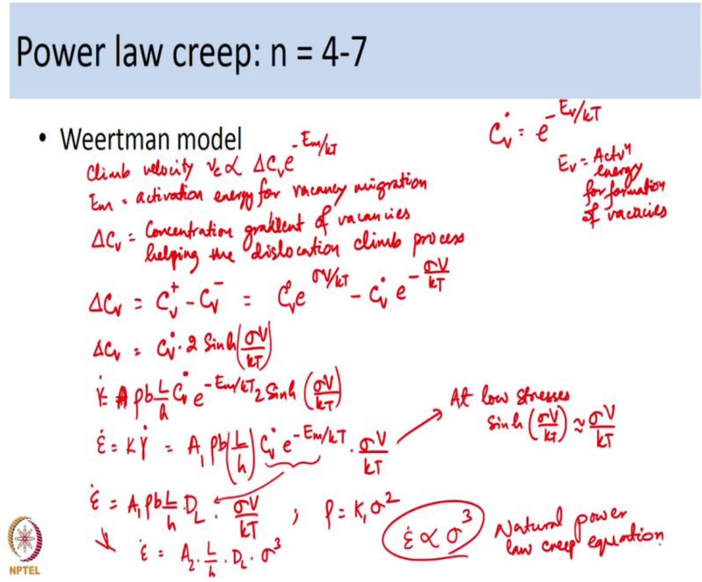

So this is what he derived from his relation. So epsilon dot is equal to A into L by h D L into sigma cube.

Now Weertman also further showed, so showed that L by h can be approximately given as sigma to the power

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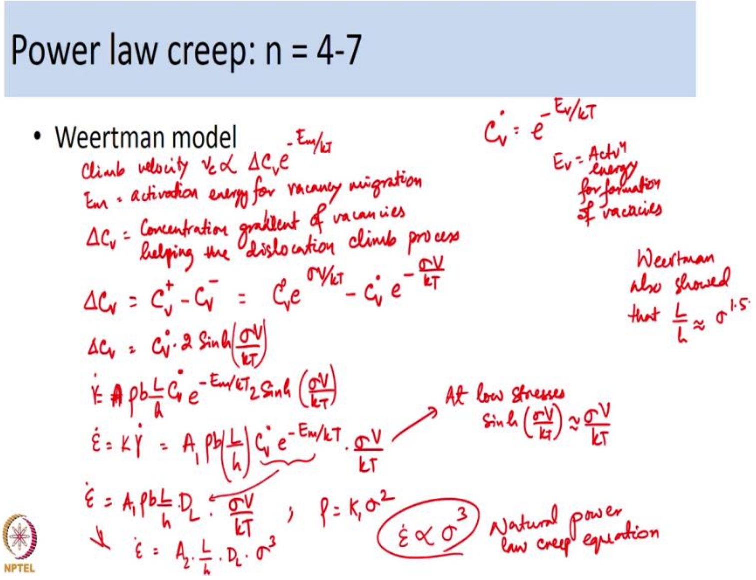

1.5 and that makes epsilon dot is equal to A some constant A 3 sigma to the power 4 point 5 into D L.

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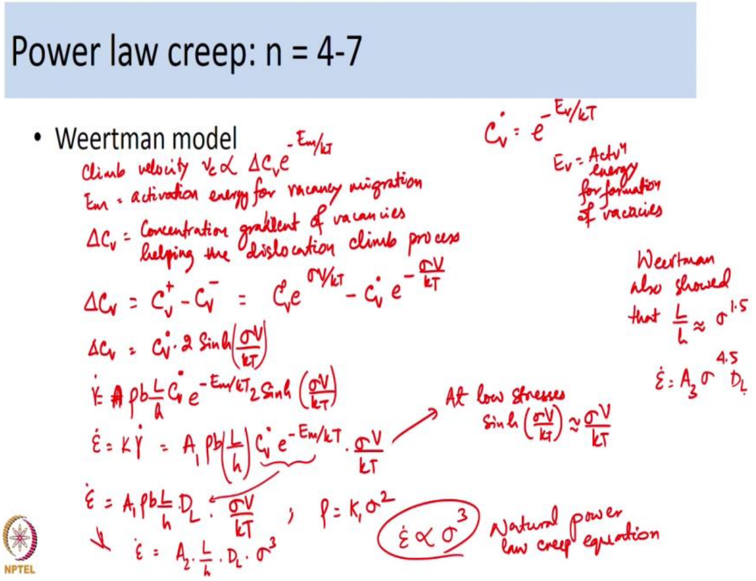

So sigma to the power 4 point 5 so this now is known as the power law creep equation and broadly or in general this is known as the Five power law creep equation.

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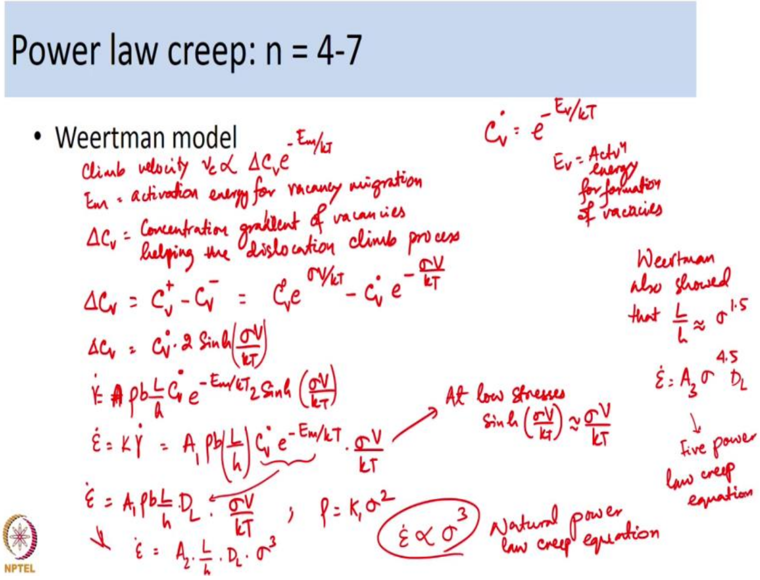

So this is the derivation of Weertman. So this is how the Weertman model works and how it eventually leads us to an equation where the strain rate is now dependent on sigma to the power 4.5 also, and it turns out that in general materials in the power law creep regime, they can have values between 4 to 7.

So that was the Weertman's equation.

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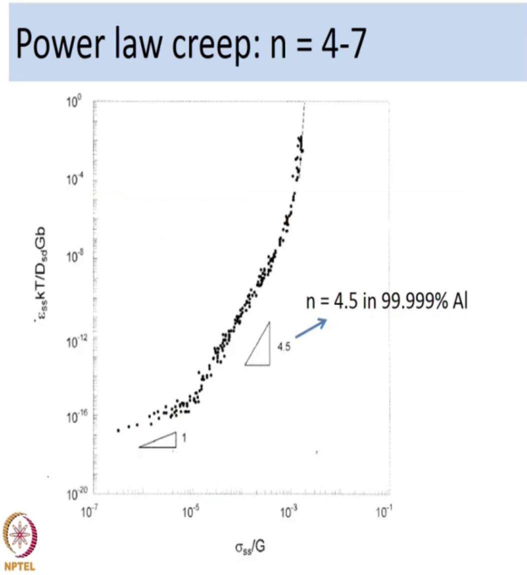

So this relationship between strain rate and applied stress has been observed in a variety of materials and this is here I am showing the relationship between normalized strain rates.

So this is normalized strain rate

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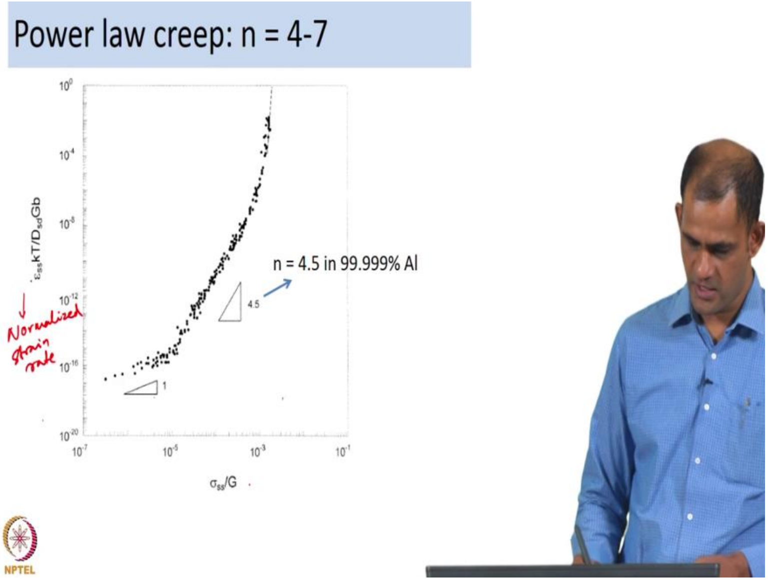

and this is

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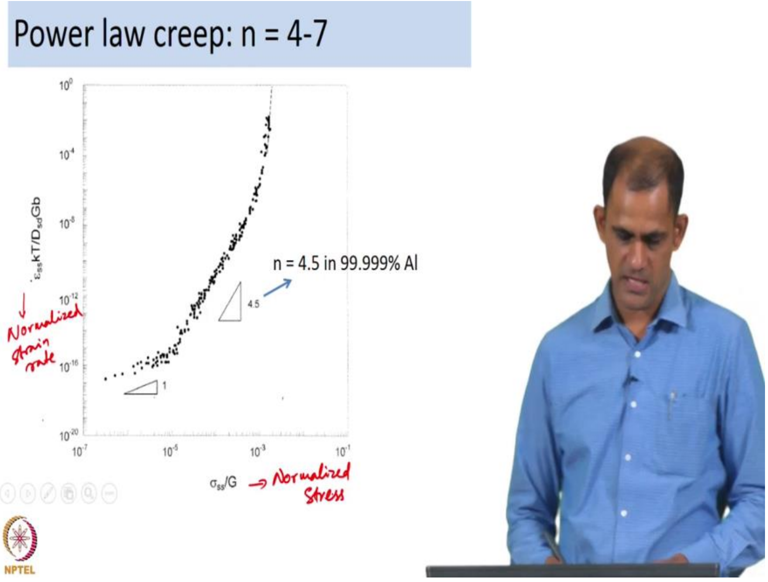

normalized stress. So a relationship between normalized strain rate and normalized stress, here you see n is equal to 4.5,

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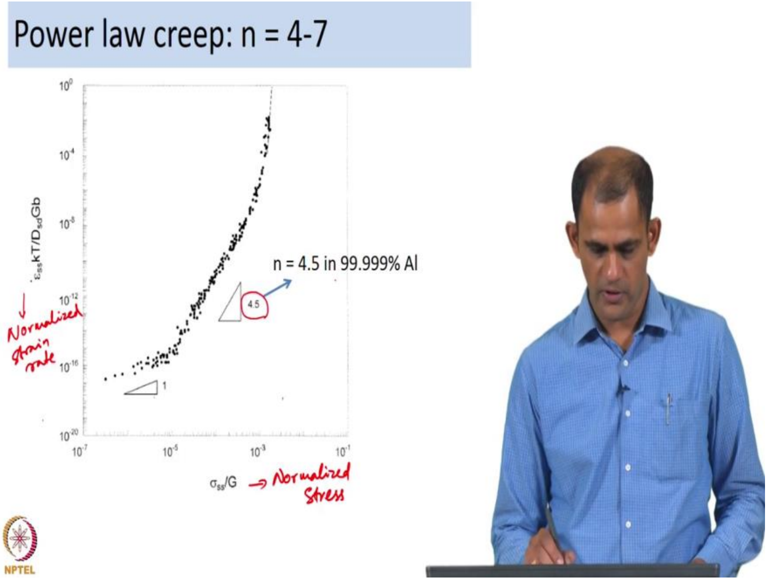

this has been observed in

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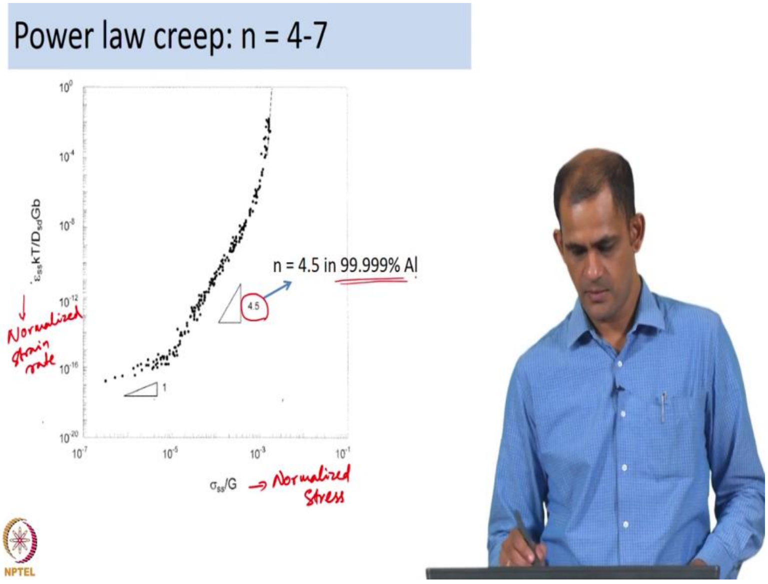

aluminum of very high purity. So this is a validation of the derivation that Weertman carried out to come to, to arrive at the power law creep equation.

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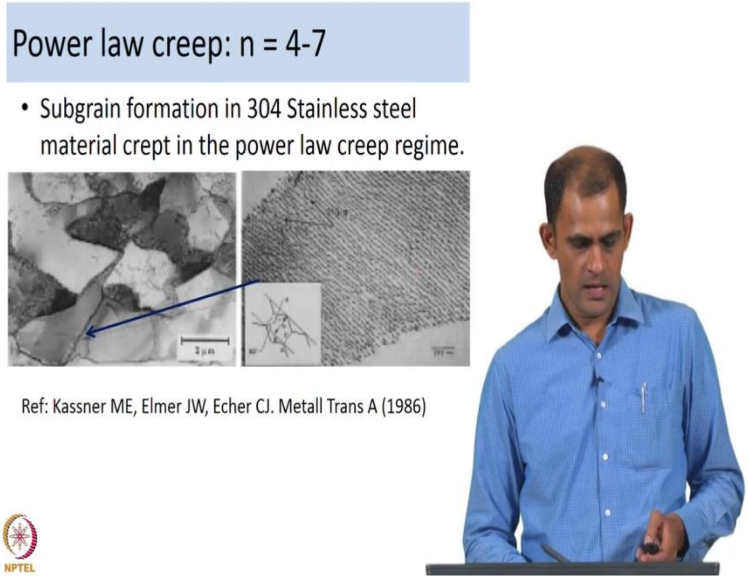

Now what kind of microstructures would you expect in the power law creep regime? Well, since this includes, involves dislocation climb so the climb is what we are describing as the rate controlling step and a little bit about why it is known as the rate controlling step?

Well the reason is creep strain is happening because of dislocation glide but the creep strain to continue to happen; you need climb to happen so that the back stresses are relieved. And if the back stresses are not relieved then creep strain is not going to happen. That is the first aspect.

So the efficiency of the deformation, in a way, if we call it as the efficiency of deformation, that is the efficiency of process is dependent on how fast the dislocation climb process happens. So because it involves dislocation climb, so the kind of microstructures that people have observed is what I am showing here.

So generally for a material that has undergone creep by the power law creep regime so the microstructure consists of well-defined sub grains. So you have sub grains forming within the material and the boundaries are basically arrangement of dislocation. So here is a T E M micrograph of dislocation. So these are all dislocations that have arranged themselves along this,

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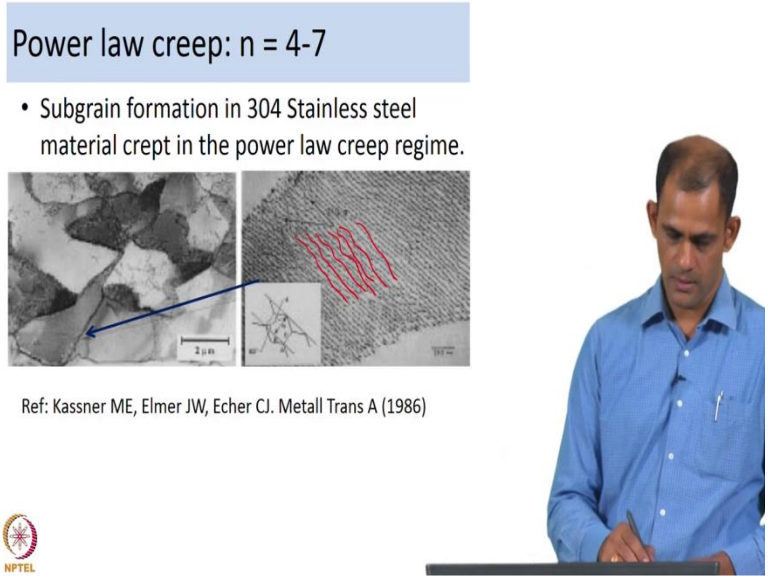

so boundaries.

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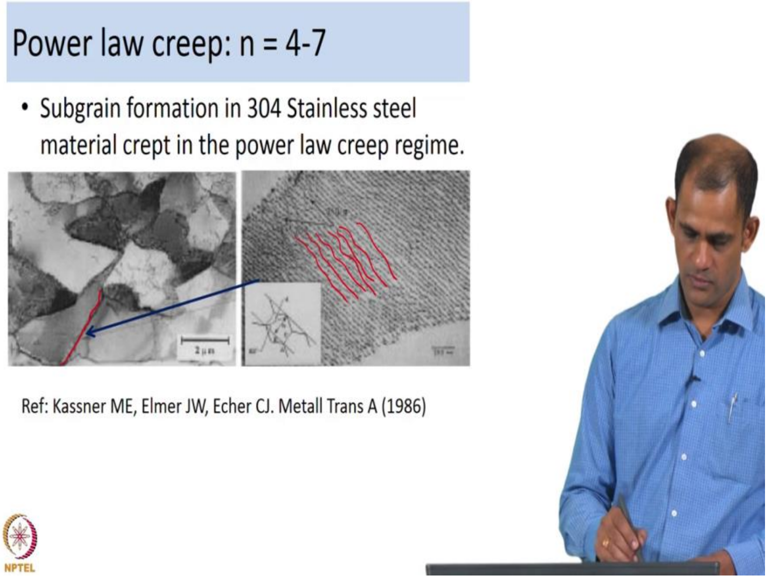

So this boundary if you take, magnify it further so that is kind of configurations that the dislocations develop. So sub grains, if you see sub grains in your material with n is equal to 4 then you can very easily say that it is a dislocation climb controlled creep.

That is a kind of microstructure that one could see for a material crept in a n is equal to 4 to 7 regime.

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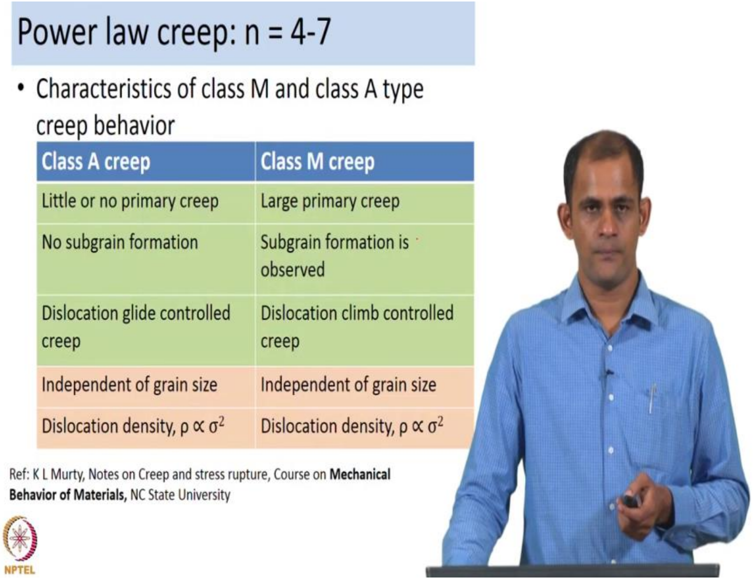

Now, so this is a behavior of a class M alloy.

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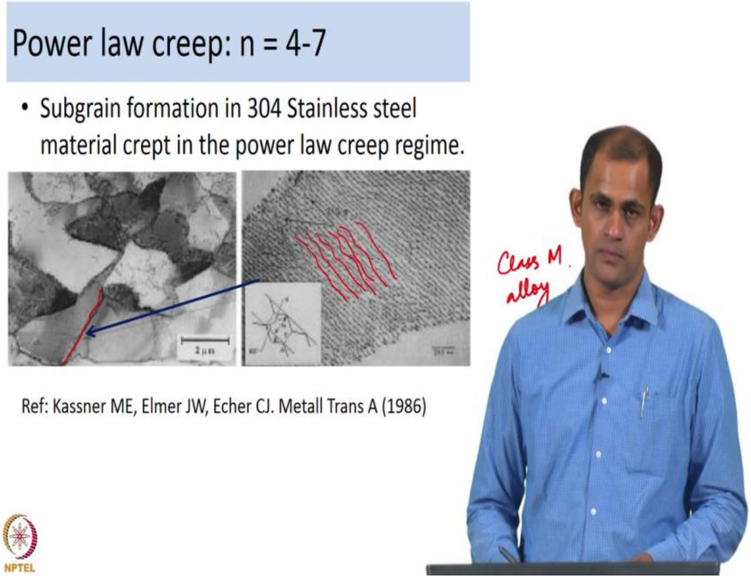

So class M alloy, so that is the kind of behavior that you would expect.

Now since, so now we have so far, we have talked about the two types of alloys so we have talked about class A alloy and we have also talked about class M alloy.

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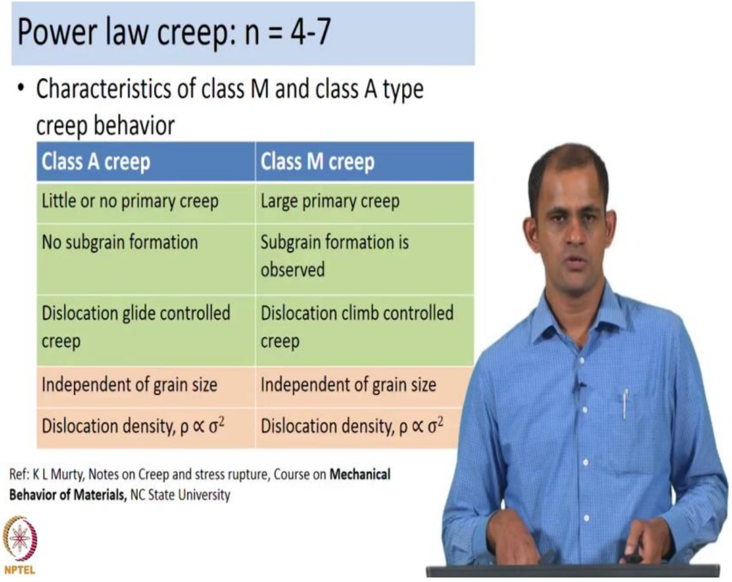

So just to give you a summary of the characteristics of a class A and class type, M type of alloy, so class A there is very little or no primary creep.

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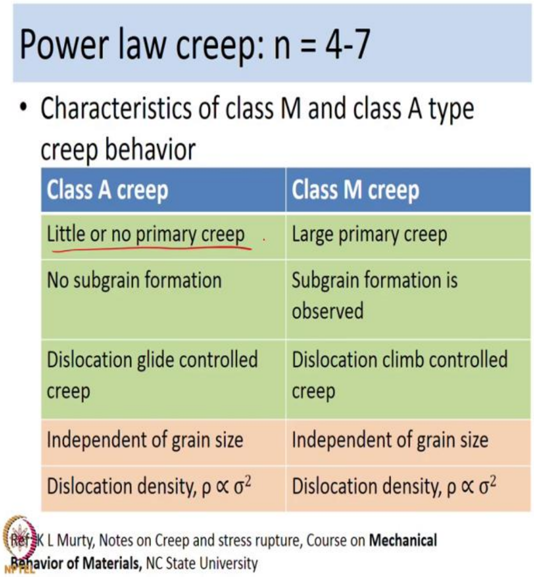

So this I had talked about in some detail. I had shown you some creep curves of material which was deforming in the n is equal to 3 regime and of material it was deforming it n is equal to 4 to 7. So this is epsilon versus time. So this is kind of behavior.

So you have very little primary creep in class A, so materials of class A

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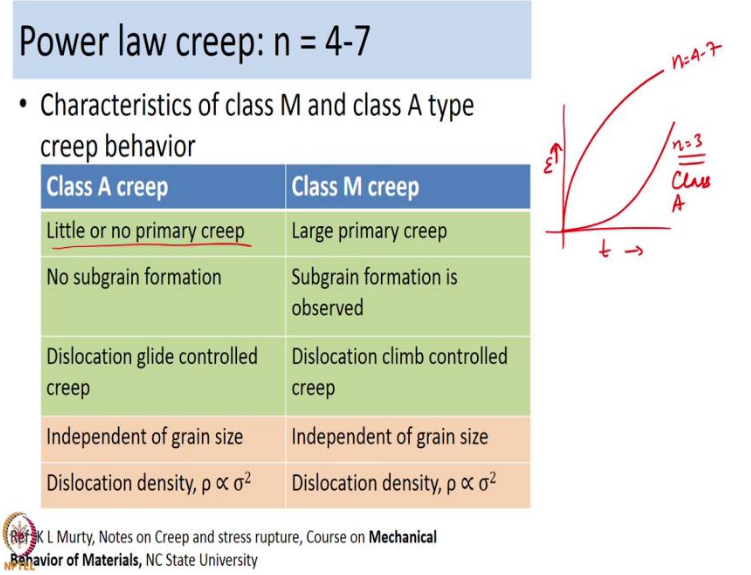

type, they show very little primary creep whereas materials of class M, they show a prominent primary creep. So you see a

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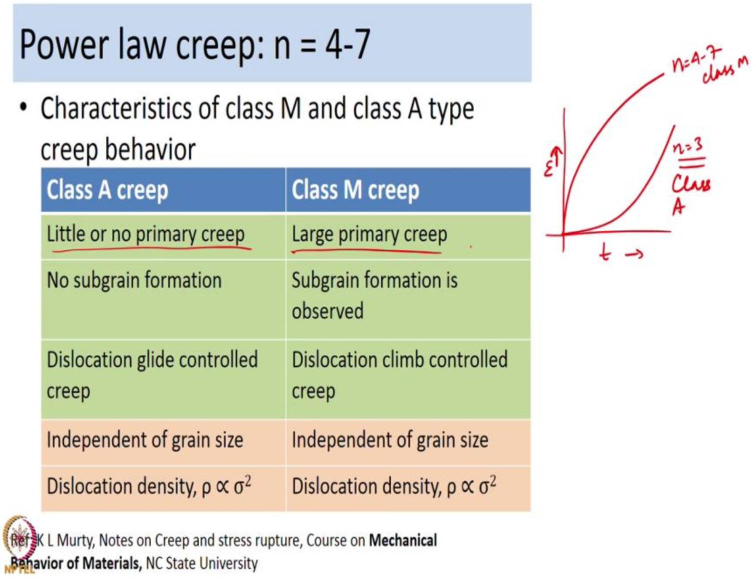

prominent or large primary creep.

Secondly you do not see any sub grain

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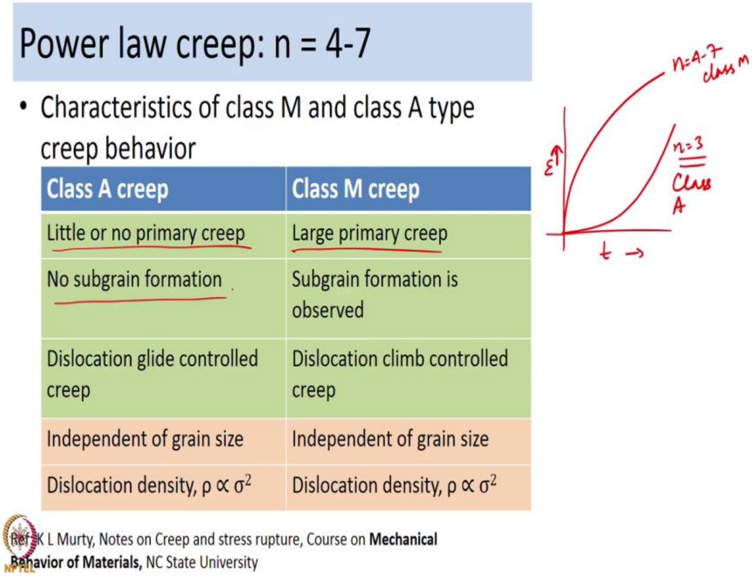

formation in class A creep. So I had shown you few micrographs towards to the end of the portion on viscous glide creep. So I had said that you generally have random dislocations. You do not have any particular configuration of dislocation.

So you do not see any sub grain formation in class A creep whereas in class M creep you see distinct

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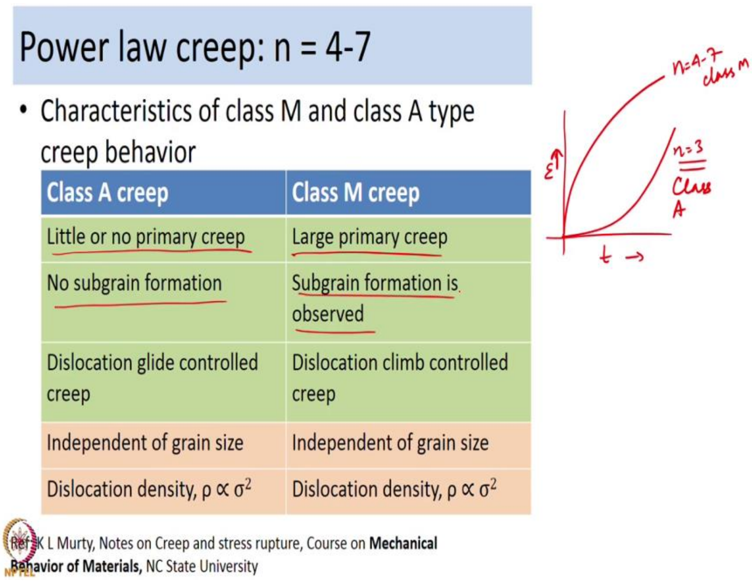

sub grains forming within the material, so sub grains it is basically something like that. So the grain has broken down into smaller grains.

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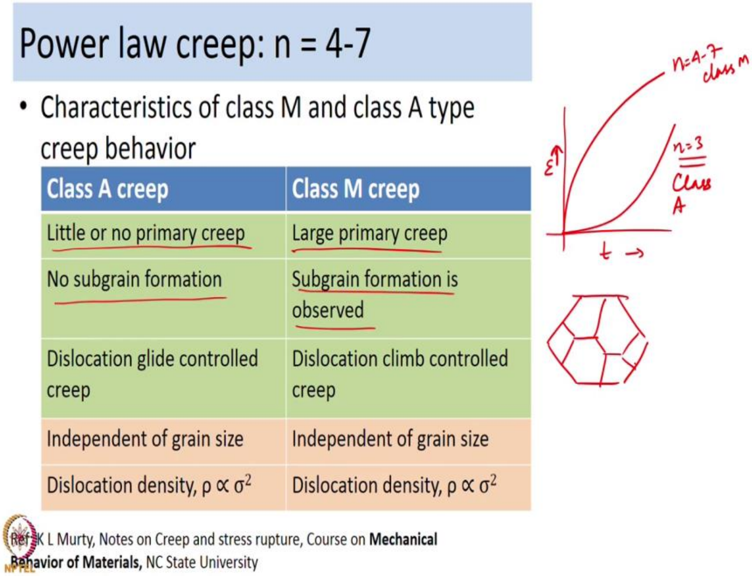

So class A creep we understood that it is dislocation glide controlled creep. It is dislocation glide controlled creep because you have solute atmospheres around the dislocation which prevent easy motion of the dislocation. So they prevent the dislocation from gliding easily along the slip plane so that is why it is dislocation glide.

Whereas in class M alloys we are talking about dislocation

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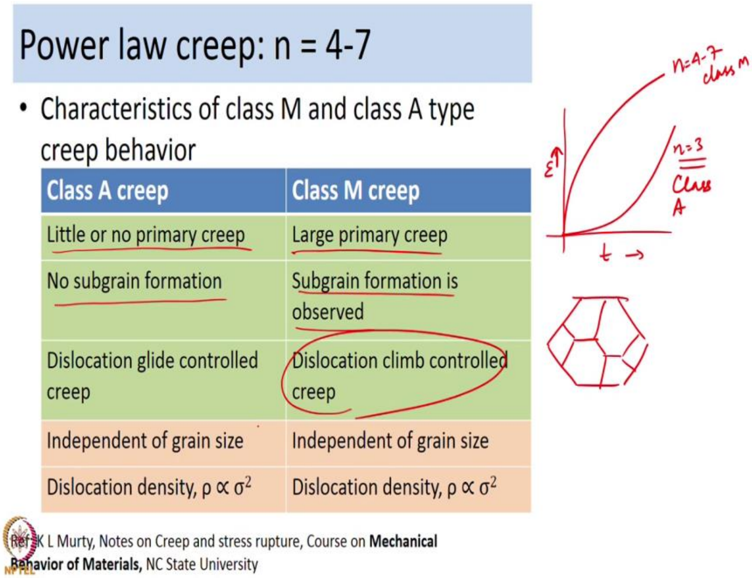

climb controlled creep, so the creep deformation happens only if the climb is happening properly. So these are the differences, these three are the differences between

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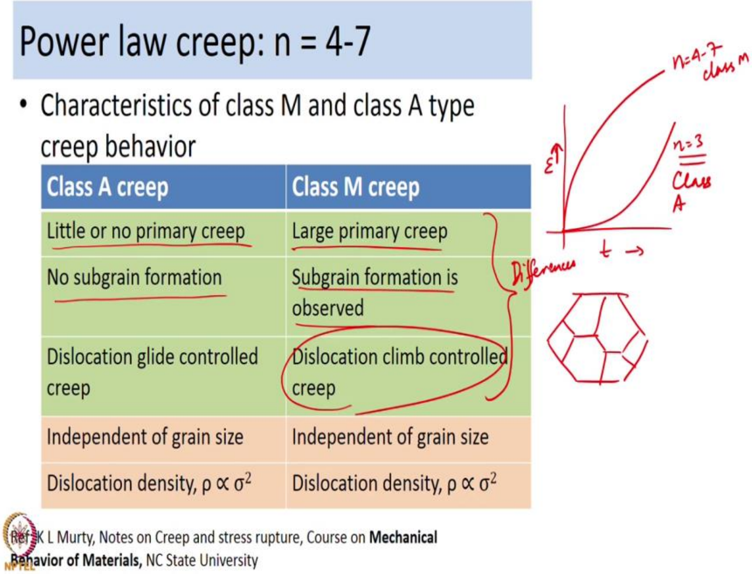

the class A and class M alloys and their creep behavior.

Whereas the similarities

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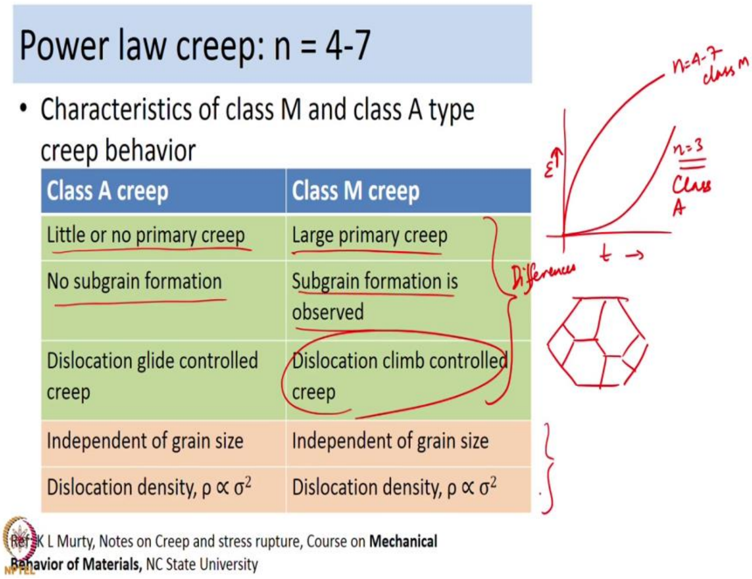

between the two types of alloys is both class A alloys they are independent, the strain rate of deformation during creep is independent of grain size and it is the same in class M alloys also. So it is independent of grain size.

And another similarity is that the dislocation density

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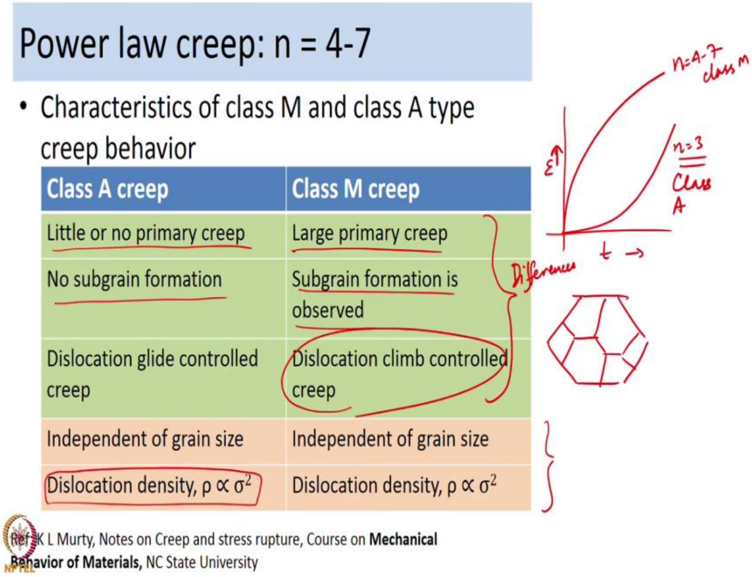

follows the Taylor law. So rho is proportional to sigma square and a similar thing

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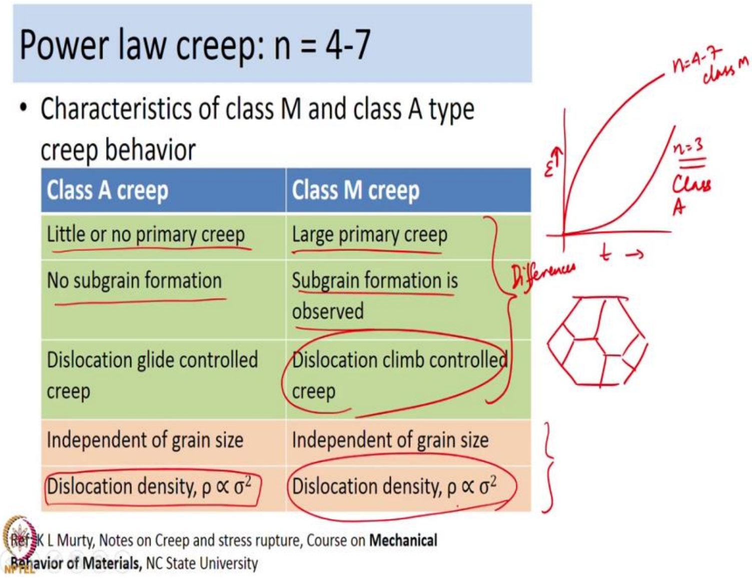

is observed in class M alloys as well. So you see rho is proportional to sigma square. So these are the

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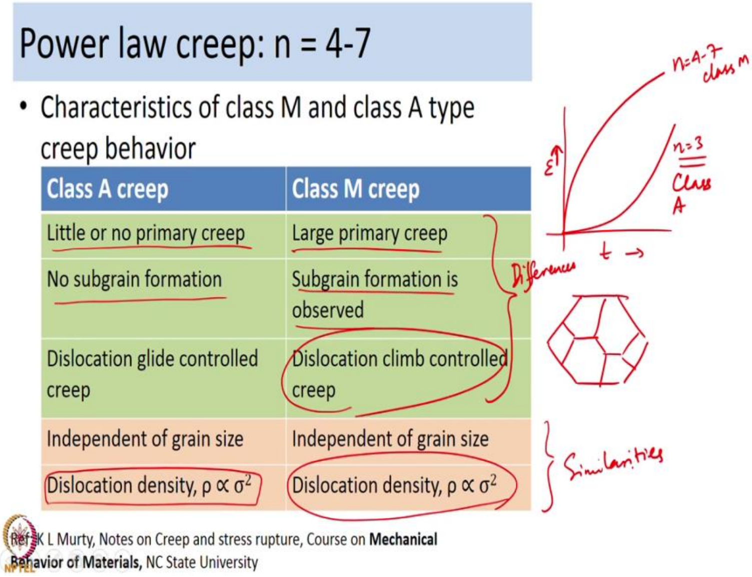

the similarities between the class A creep and class M creep.

Ok,

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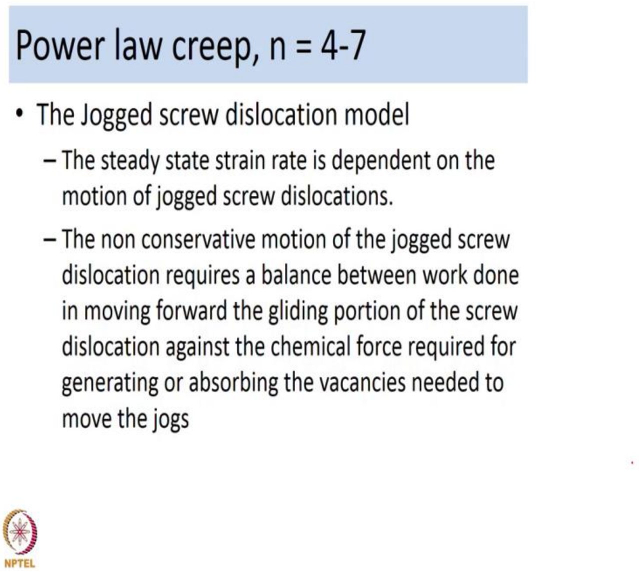

so now we are going to talk about the second model. So we spoke about the Weertman's model in detail. So we said sub grain formation and we derived the equations. So now we will talk about another model which was proposed by Barrett and Nix and this is known as the Jogged screw dislocation model, a model used for explaining n is equal to 4 to 7 behavior.

So here in the Jogged screw dislocation motion model, steady state strain rate is dependent on the motion of the Jogged screw dislocation. So you recall dislocation like that and if the Burgers vector,

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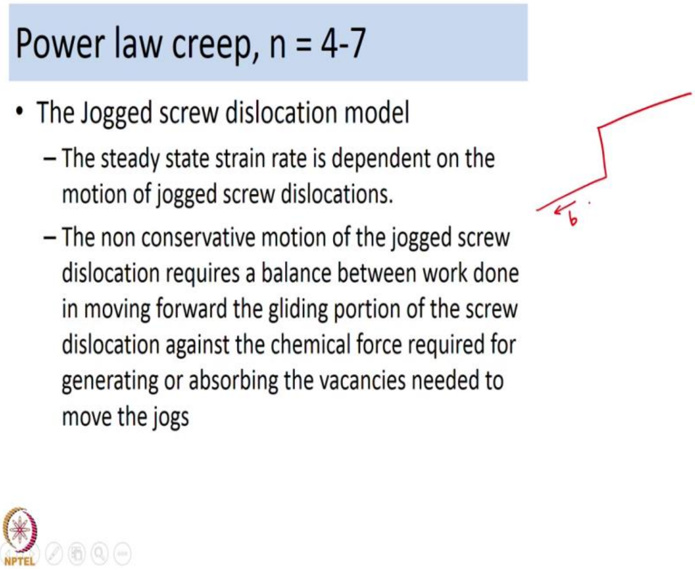

screw dislocation and so it is perpendicular to the jog and

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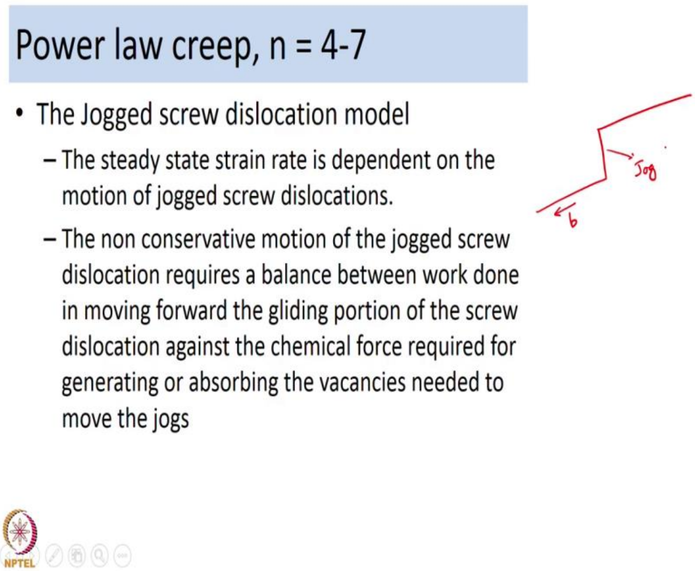

the jog prevents easy motion of the screw dislocation.

So the only way the screw, jogged screw dislocation moves is by non-conservative motion and during, so Barrett and Nix they used this aspect of the screw dislocation.

They said the non-conservative motion would require the balance between the work done in moving forward the gliding portion of the screw dislocation against the chemical force required for generation or absorption of vacancies at the jogs.

So in dislocation climb, so when extra half plane of atoms

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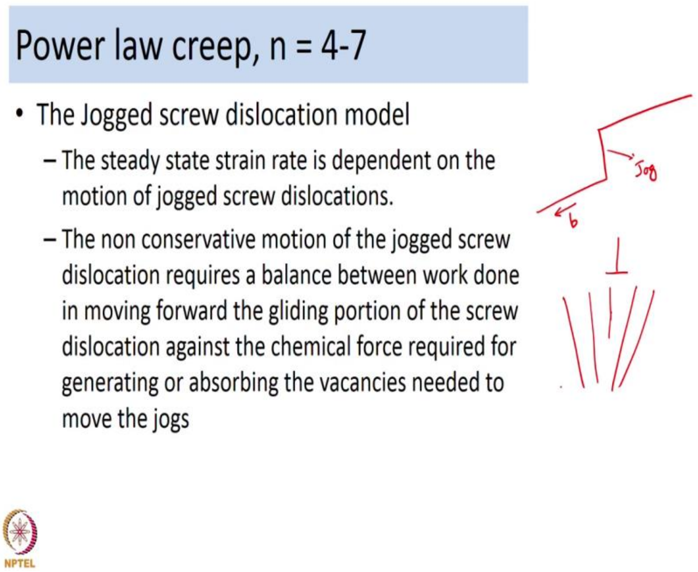

is to climb, so if the positive dislocation is moving up

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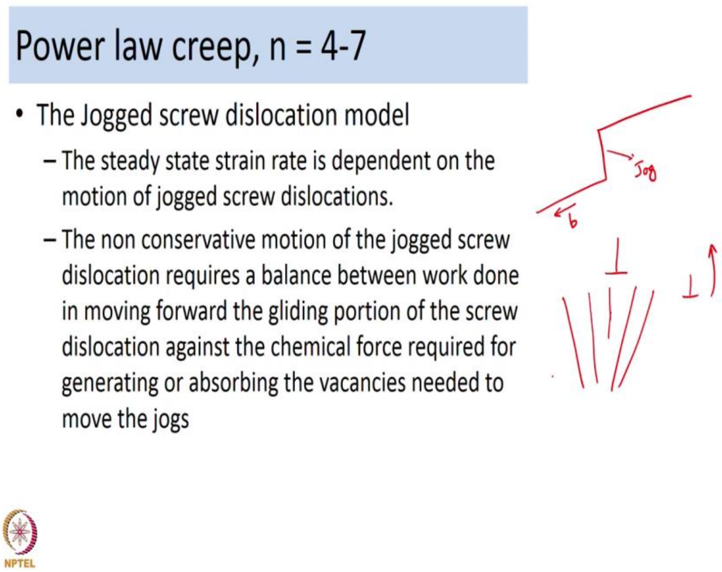

so you basically have vacancies going and getting absorbed there.

The other

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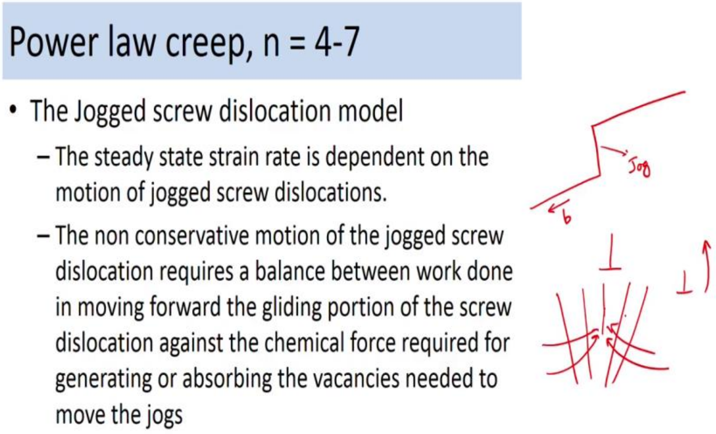

way, if the dislocation is climbing down,

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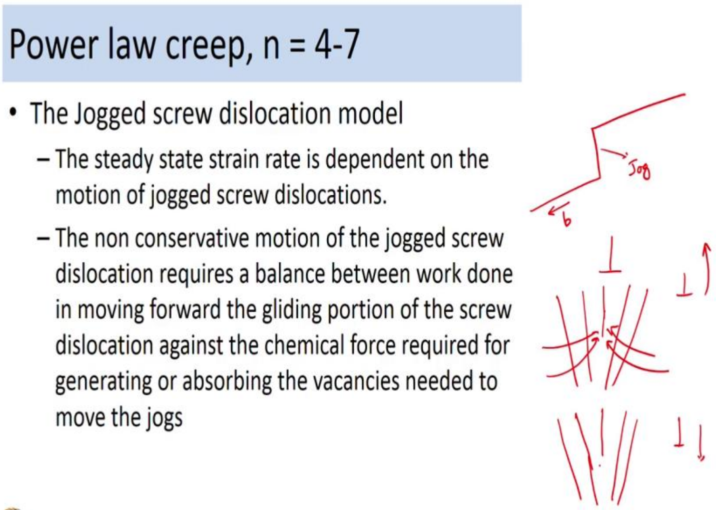

so if the dislocation is climbing down, you need vacancies to move away from them. So case 1, vacancies moving

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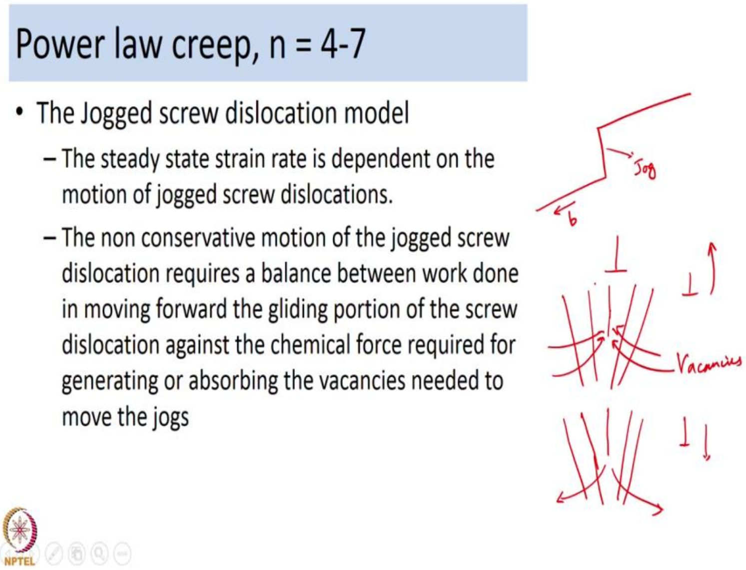

towards the dislocation core and case 2, vacancies moving away.

So either you have

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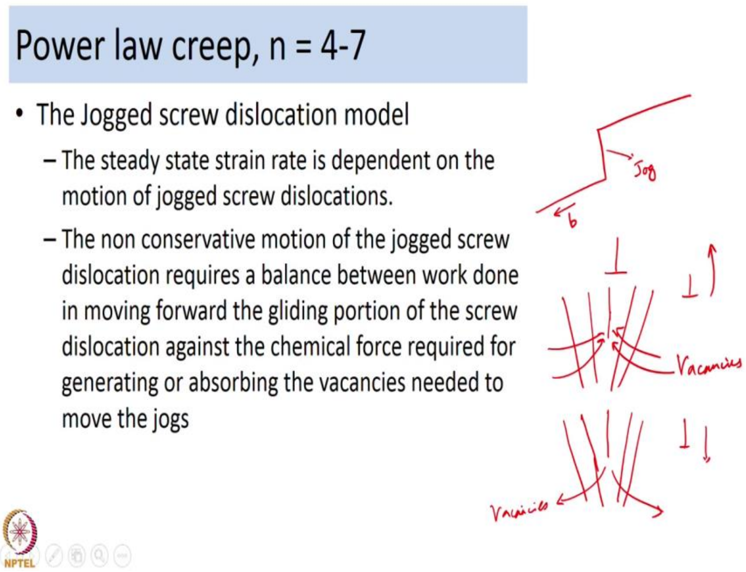

absorption of vacancies at the core or you have generation of vacancies at the core. And both are necessary for the motion of the, climb of the dislocation.

So this is a concept they used so they said you could have either absorption of the vacancies at the jogs, edge jogs so the jogs are edge, of edge character.

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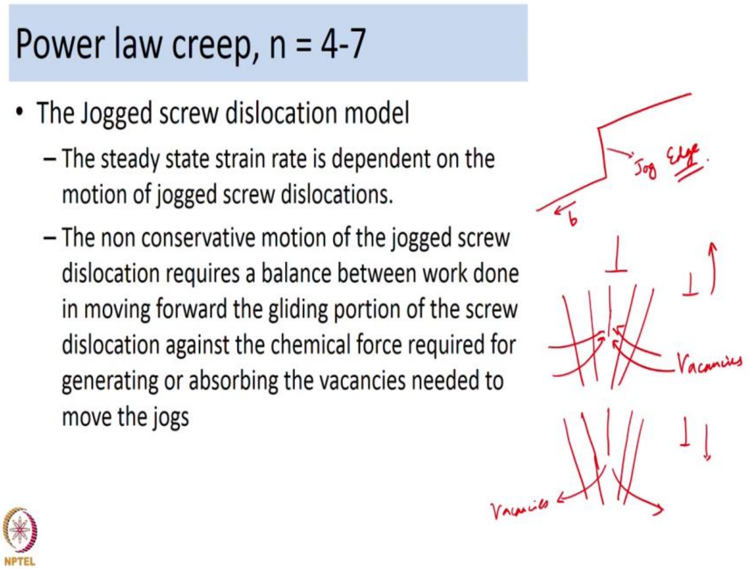

So you could have absorption of vacancies at the edge jogs or you could have generation of vacancies or emission of vacancies from the edge jobs.

So they used this and they said there is a balance between the work done in moving the dislocation, a gliding portion of the dislocation which is, this is the gliding portion of the dislocation to move it forward and this work done will be balanced by the chemical

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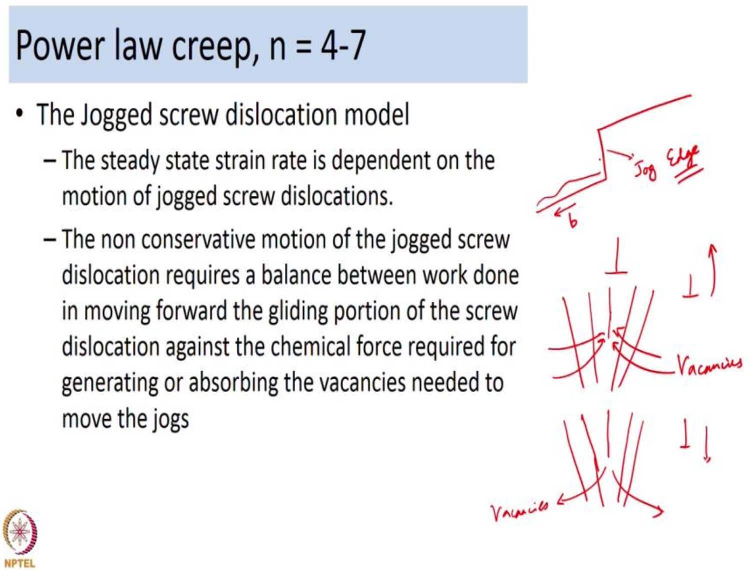

force required for the absorption of generation of

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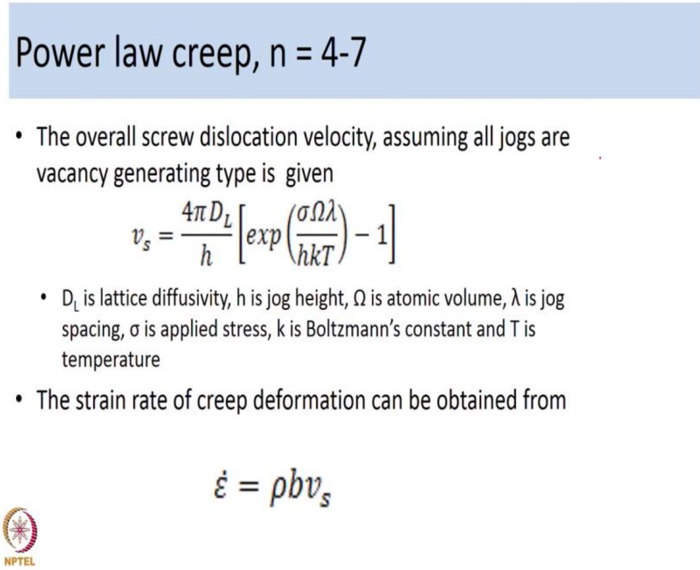

vacancies.

Now they made a small assumption. They said let us assume all jogs are the vacancy generating type of jogs, so which is vacancy generating type of jogs then the velocity of the dislocation will be given by the following equation.

So v s is the overall screw dislocation velocity, so it is the overall jogged screw, dislocation, the velocity of

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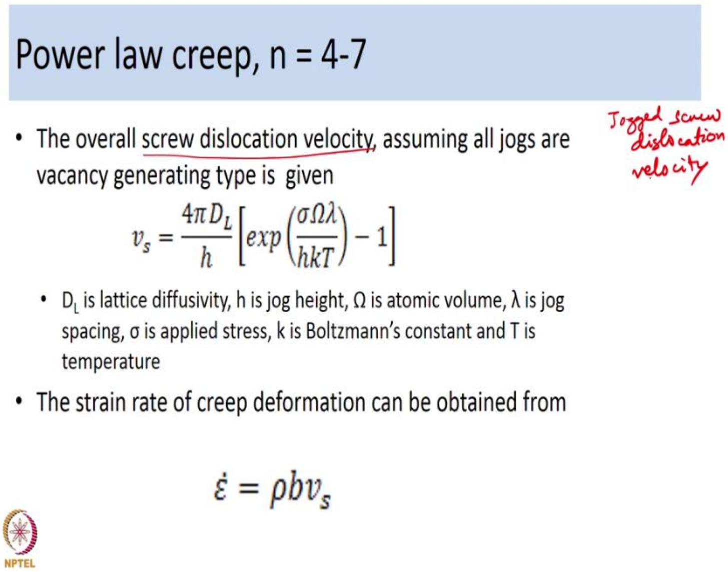

the jogged screw and it is given by the following equation so here D L is the lattice diffusivity,

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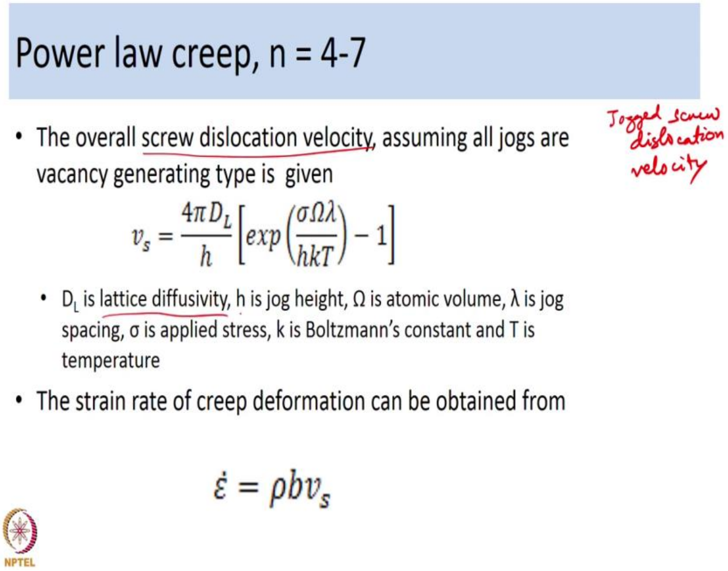

h is

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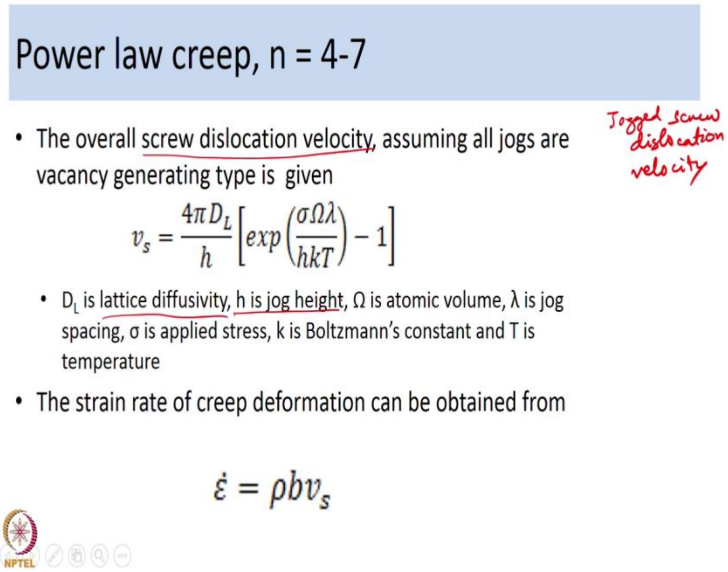

the jog height, omega

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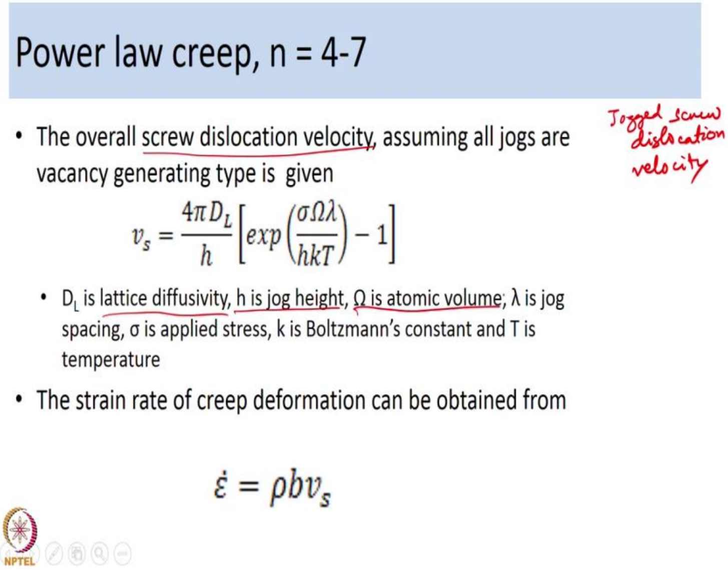

is the atomic volume, lambda is the jog spacing,

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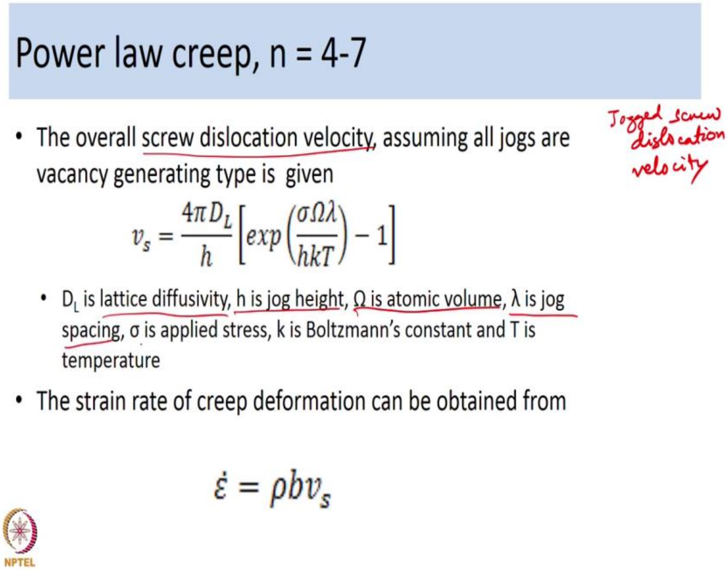

and sigma is the applied stress

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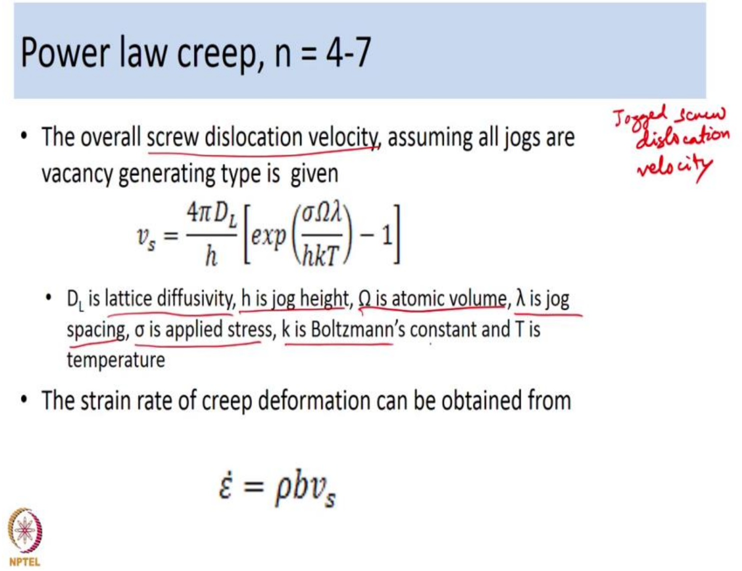

and k is Boltzmann's constant, T is temperature.

So if you have, say jog, so if you have edge jogs like that so this will be the spacing between the jogs

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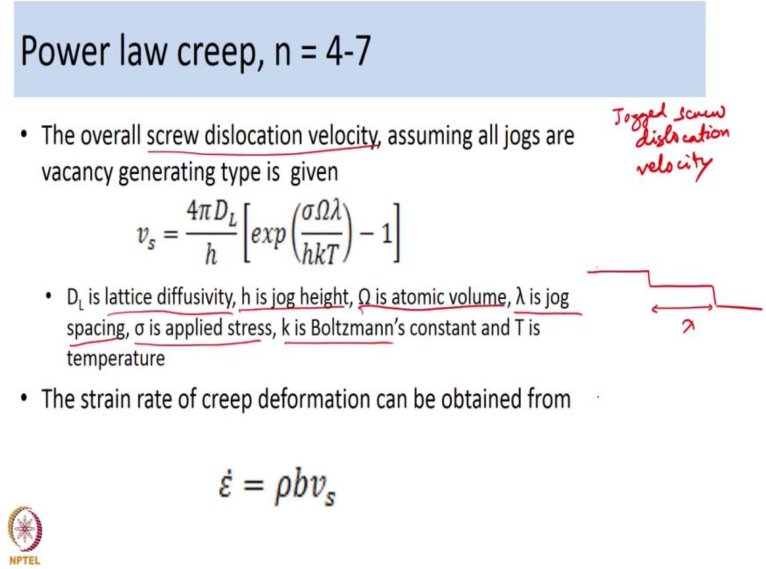

so another way of looking at it is say, jog is impeding a dislocation from moving, screw dislocation so this will be the

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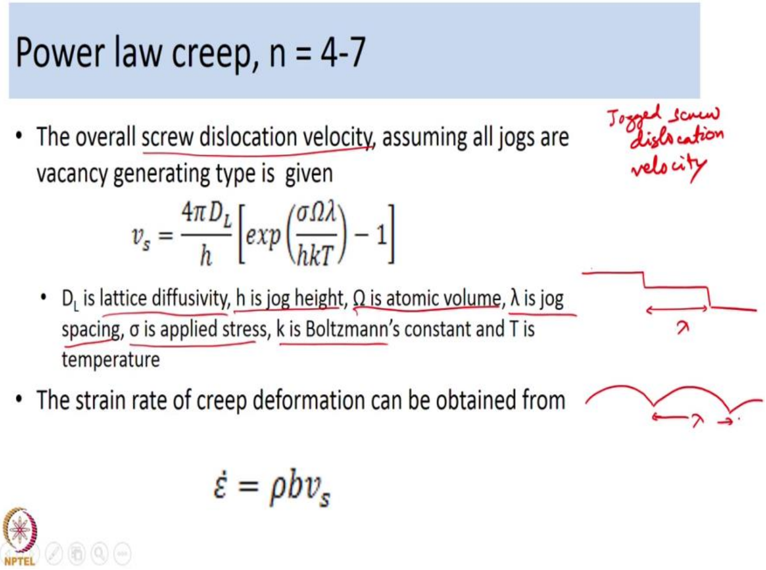

lambda, so that is the jog spacing. lambda is the jog spacing.

Now this is the equation that they came up with, for the overall screw dislocation velocity and then knowing that the

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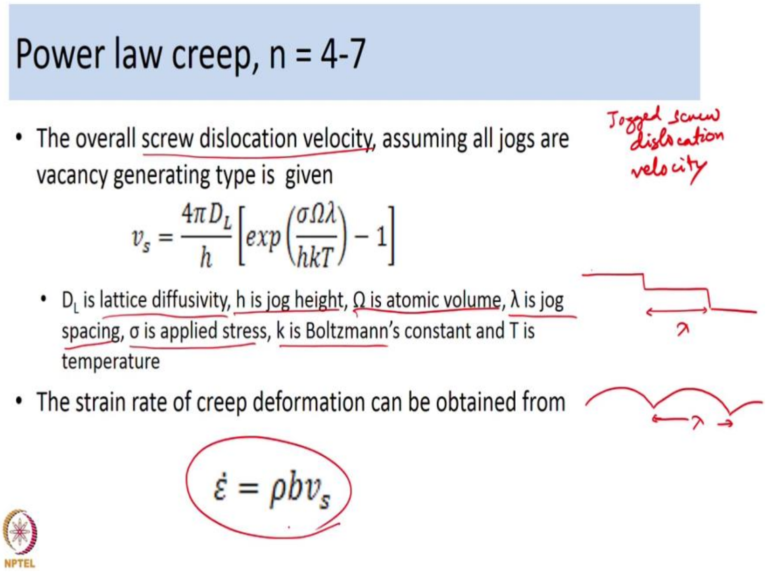

strain rate of deformation is dependent on the velocity of the dislocation by the following equation, so epsilon dot is equal to rho b v s.

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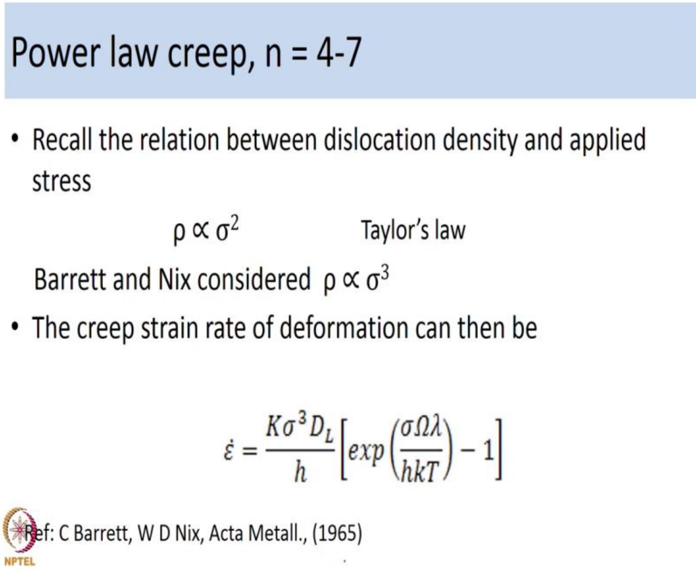

And then the Taylor's law basically says the rho is proportional to sigma square. So that is the dislocation in, relation between dislocation density and applied stress.

So rho is proportional to sigma square,

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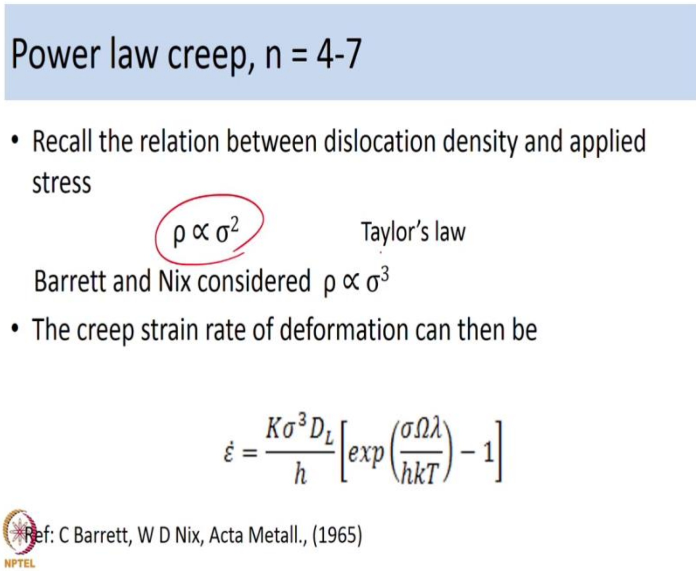

that is the Taylor's law.

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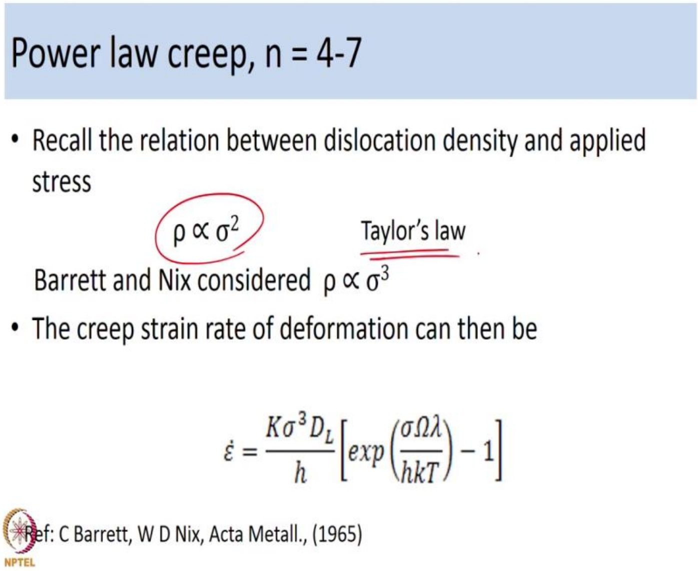

So Barrett and Nix instead of sigma square, they considered that rho is proportional to

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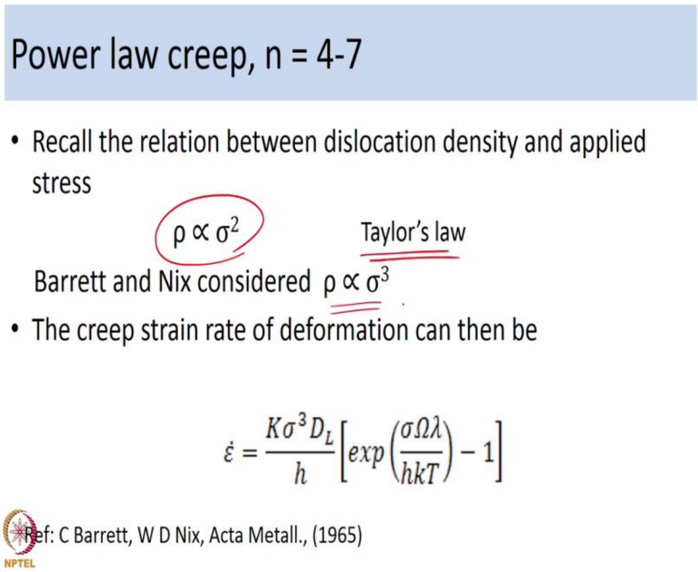

sigma cube, so they employed that and used it in the equation rho b v s.

So what

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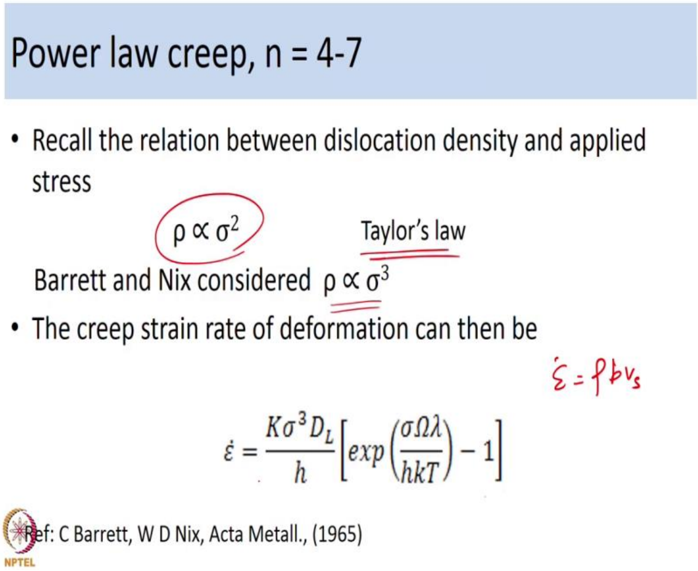

they ended up with is the creep strain rate of deformation can then be given by the

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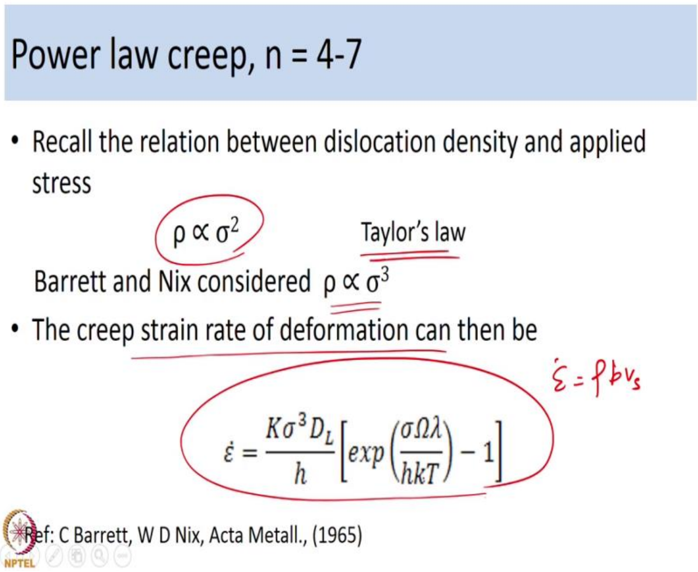

following equation, so epsilon dot is equal to k sigma cube D L over h exponential of sigma omega lambda h k T minus 1.

So this is the equation that Barrett and Nix arrived at using concept of motion of Jogged screw dislocations. Now

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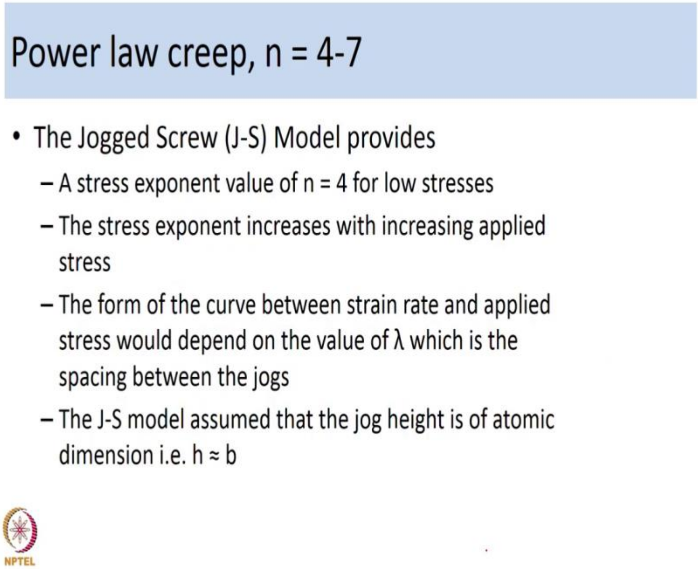

the Jogged screw dislocation, the characteristics of this model, the Jogged screw model is as follows.

So it allows a stress exponent value of n is equal to 4 at lower stresses.

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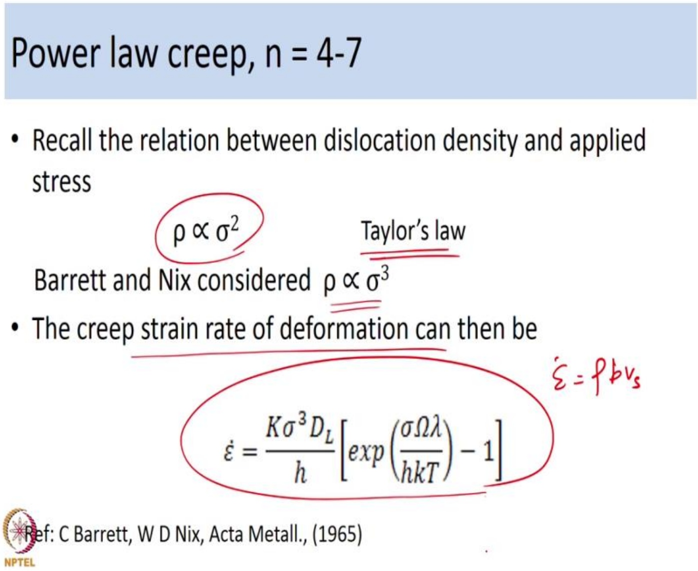

So if your stress values are low then e to the power sigma omega by h over h k T will be approximately equal to sigma omega lambda by h k T,

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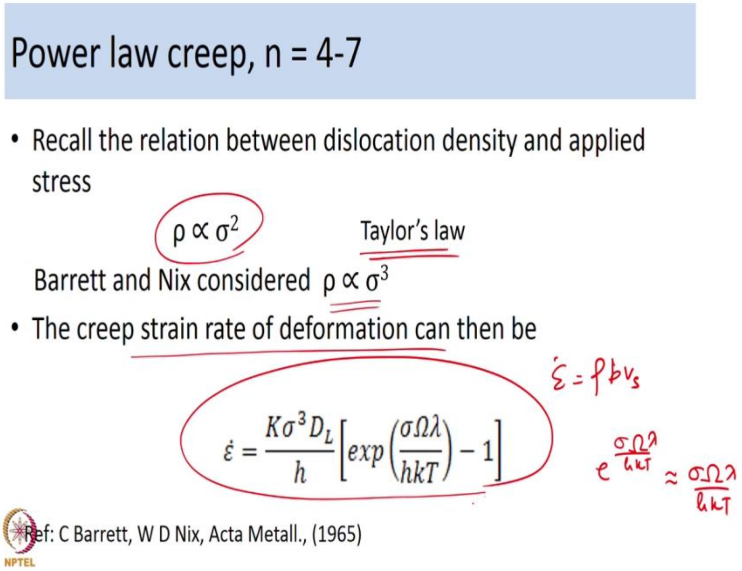

so then your strain rate is essentially sigma cube proportional to, into sigma some constant into D L.

So

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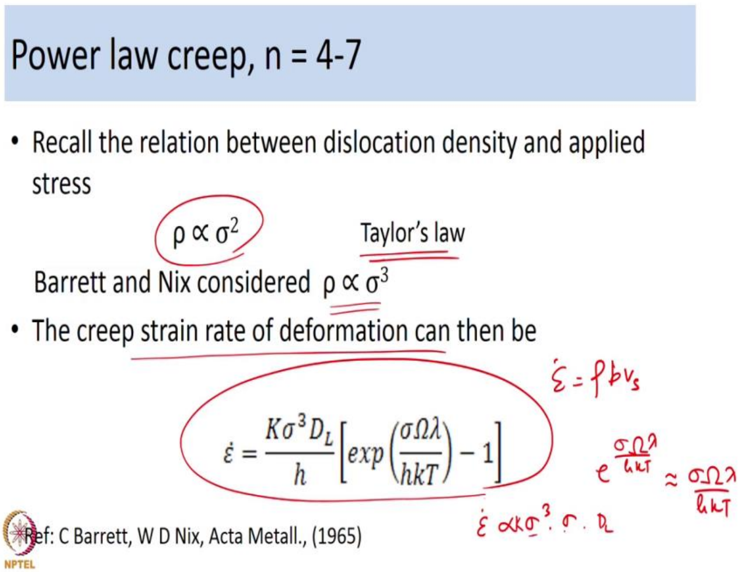

what you end up with is epsilon dot is proportional to

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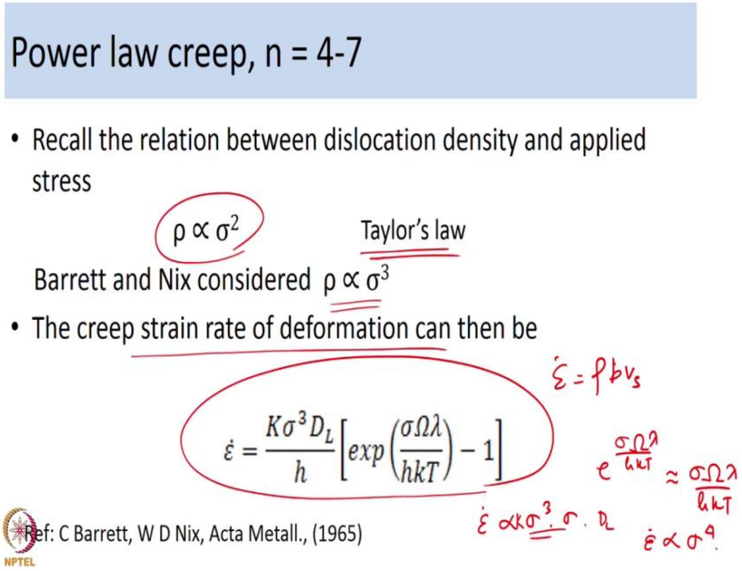

sigma to the power 4. So that is what is happening here.

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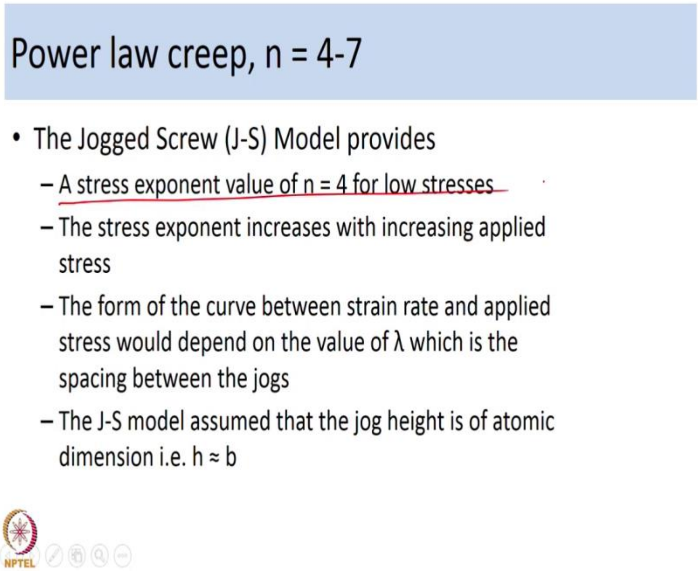

So you get a stress exponent value of 4 for lower stresses. Now as the stress increases, one of other observations of this model is the stress exponent will increase with increase in applied stresses.

So as the applied stress increases, you may not see e to the power, you may not see it equal to, so you may gradually

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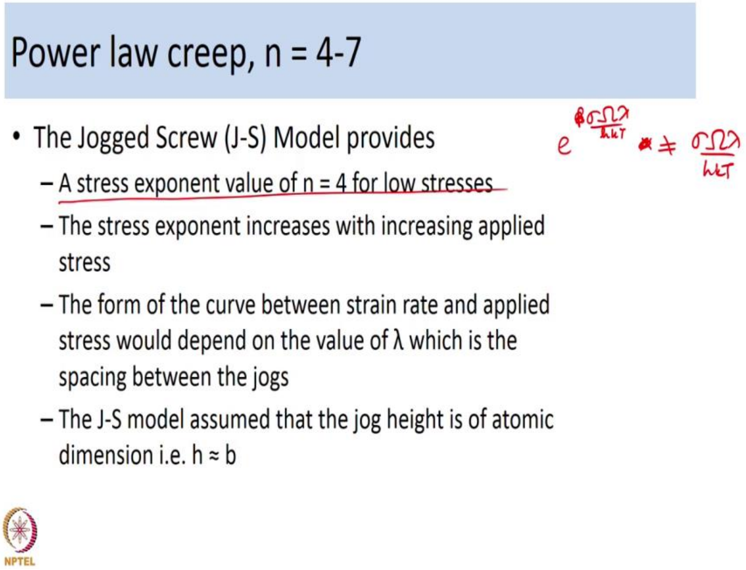

see a different relationship so gradually the stress exponent value would increase with increase in applied stress.

And another observation is that the form of the curve between epsilon dot and sigma, the form of the curve between the strain rate and the applied stress would depend on the value of lambda which is the spacing between the jogs.

So we may expect n is equal to 4

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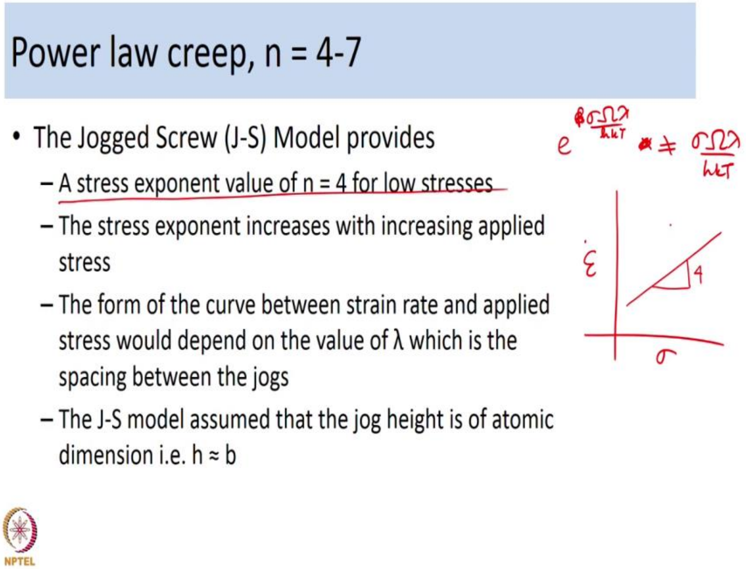

at certain stress ranges but as I said epsilon dot versus sigma, so the value may gradually change from n is equal to 4 to higher values

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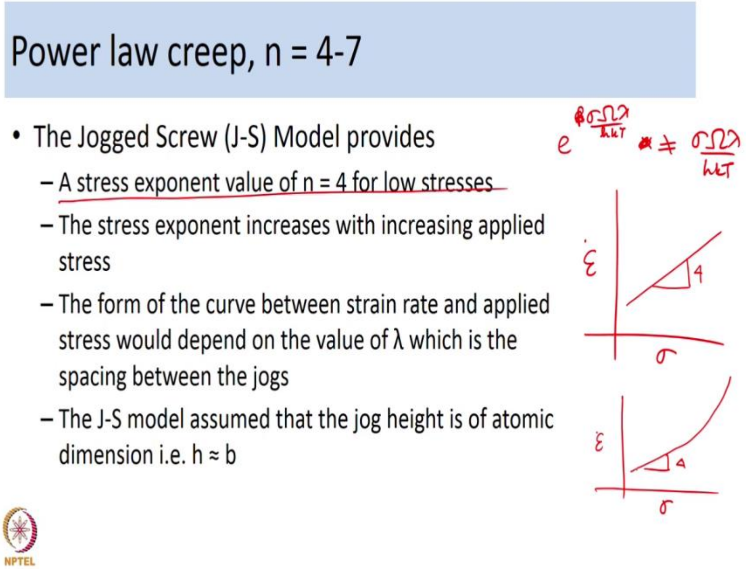

and the nature of the curve, the form of the curve will depend on this value of lambda.

And one of the significant assumptions of the model is that the jog height is of atomic dimension, so you are talking of a screw dislocation with a jog and Barrett and Nix said the height of the jog h is roughly equal to atomic height so they said, it can be taken as h is equal to b. So you can approximately take

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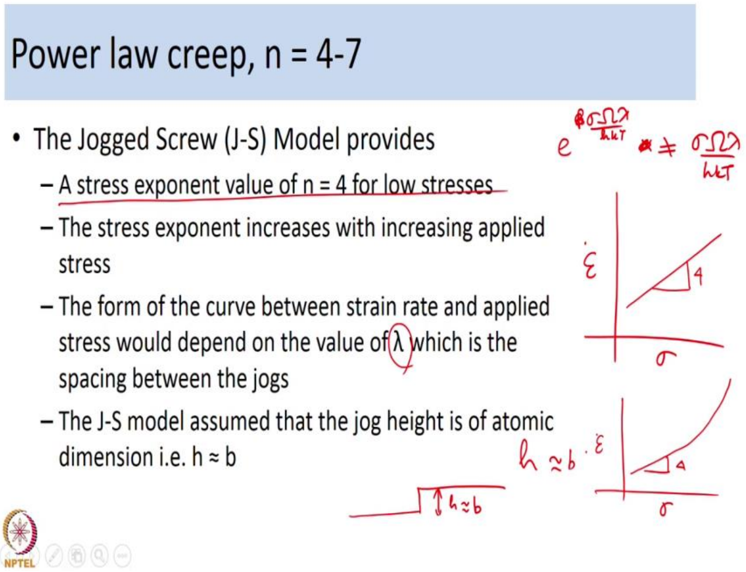

it h is equal to b, b is the Burgers vector.

So that is the model assumption.

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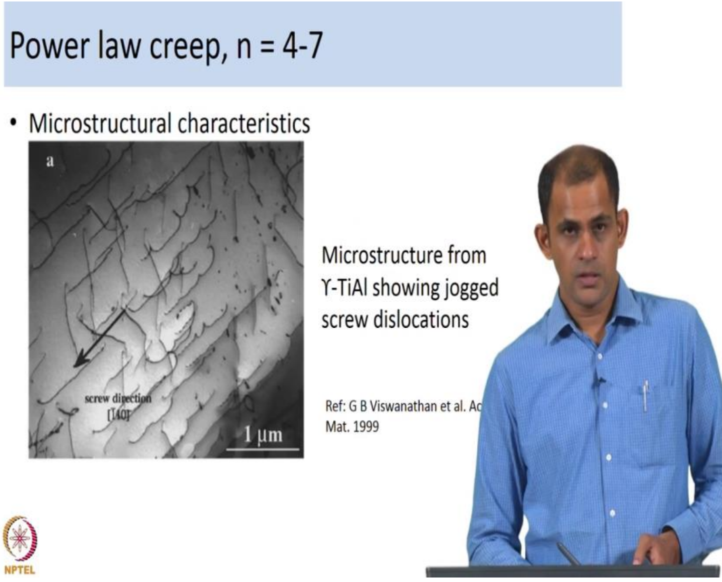

So how do we know if the material is creeping as per this model? So how do you know whether you should use the Jogged screw model for describing your creep data?

Well if you do T E M analysis of, of your crept sample so, say your material has crept already, you took some samples, did a T E M analysis.

So if you see Jogged screw dislocations like those shown here, so this is a screw segment with a jog,

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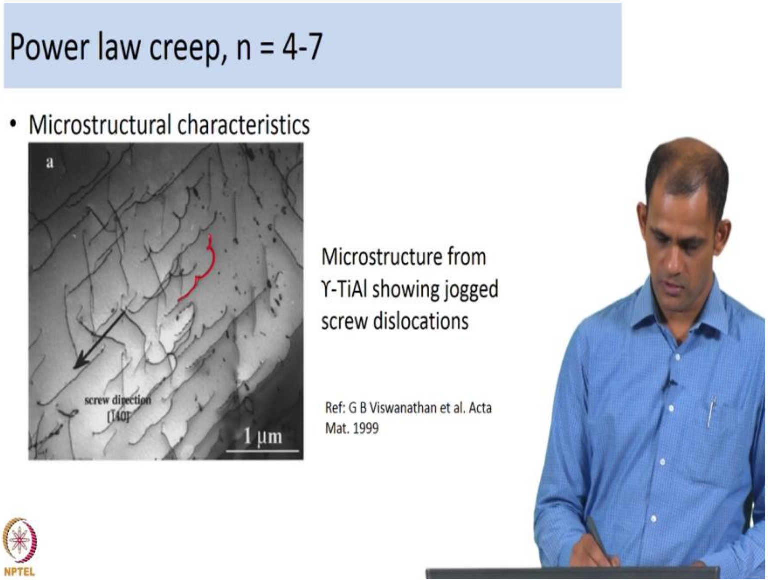

another... so if you see something like that then you

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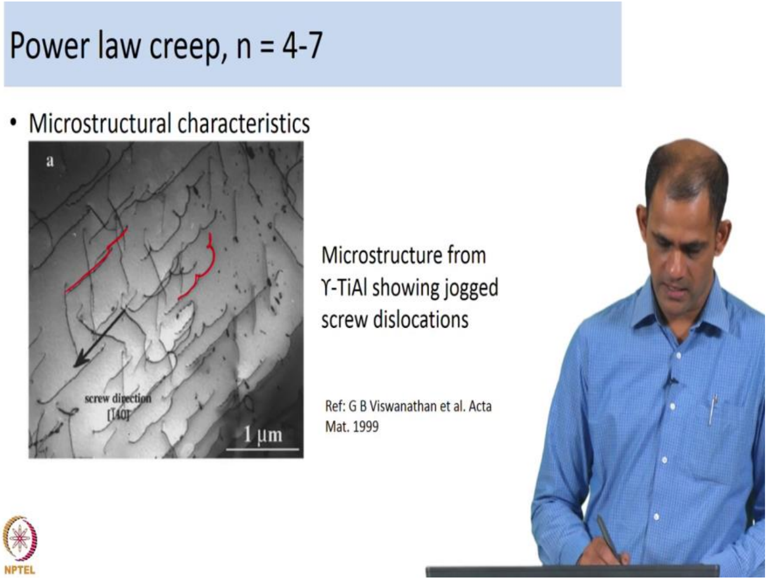

can safely assume that the material is,

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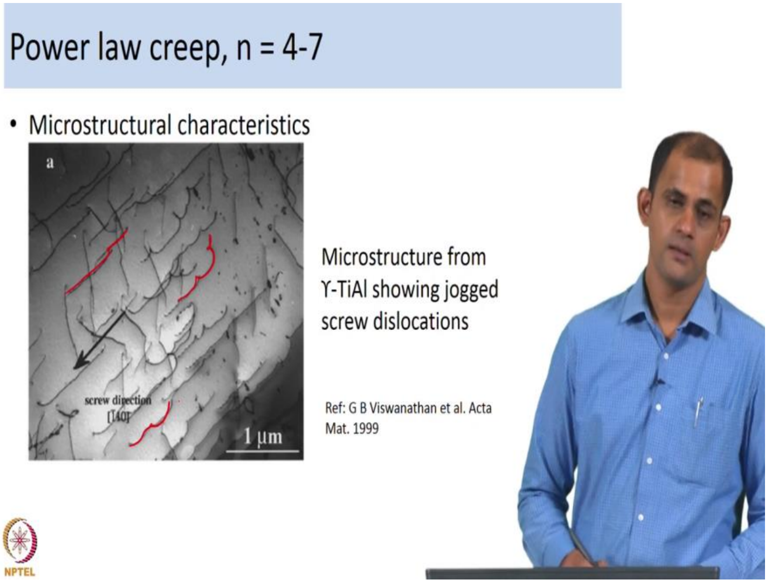

so that you need the Jogged screw model to describe your creep data.

So this is work, this is data from; this micrograph belongs to Titanium aluminide, microstructure of gamma titanium

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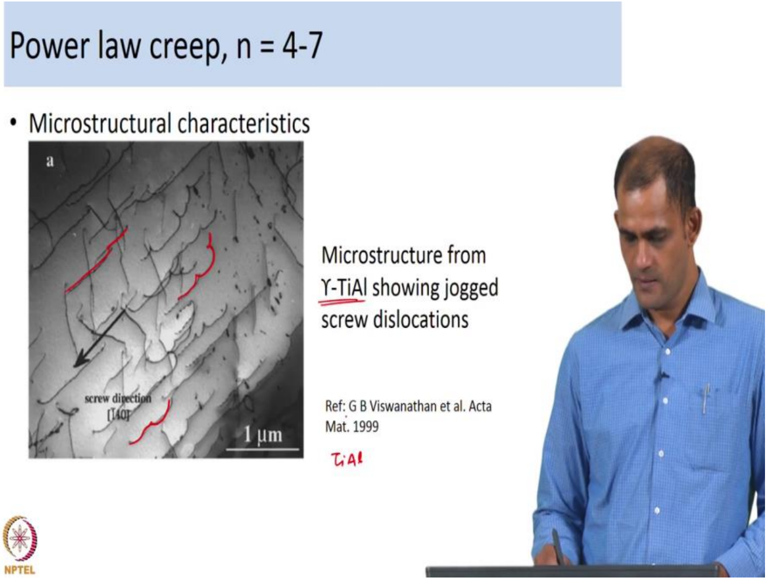

aluminide and work by Viswanathan, Vasudevan and Mike Mills. So, so the screw dislocation with the jogs and that is the microstructure they got so they wanted to use this Jogged screw model to explain their creep data.

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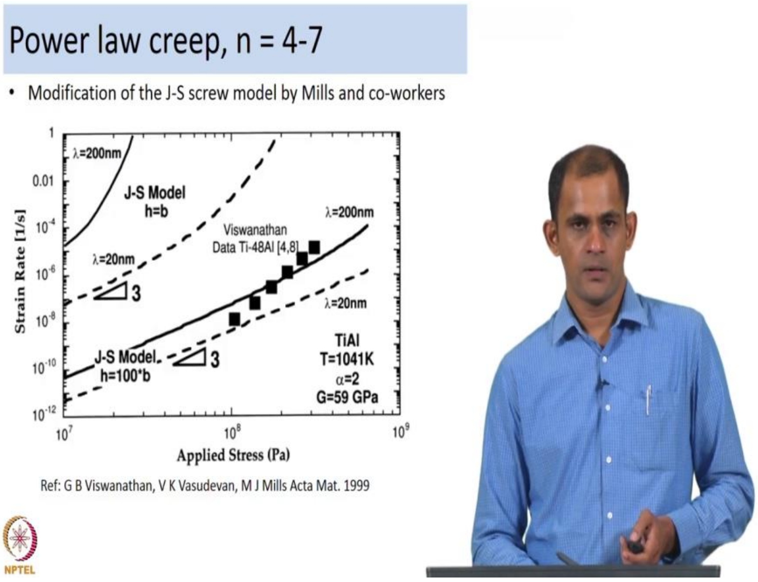

But very interestingly, so in that paper by Viswanathan and Vasudevan and Mills and subsequently there is a lot of work that Mills group carried out to investigate the Jogged screw model in detail and its application to the creep behavior of titanium aluminides and subsequently alpha titanium alloys as well as zirconium alloys.

So what they found out is the basic equation of the Jogged screw model was not suitable for explaining their or describing their experimental data.

So the squares here, so this is experimental creep data from gamma titanium aluminide.

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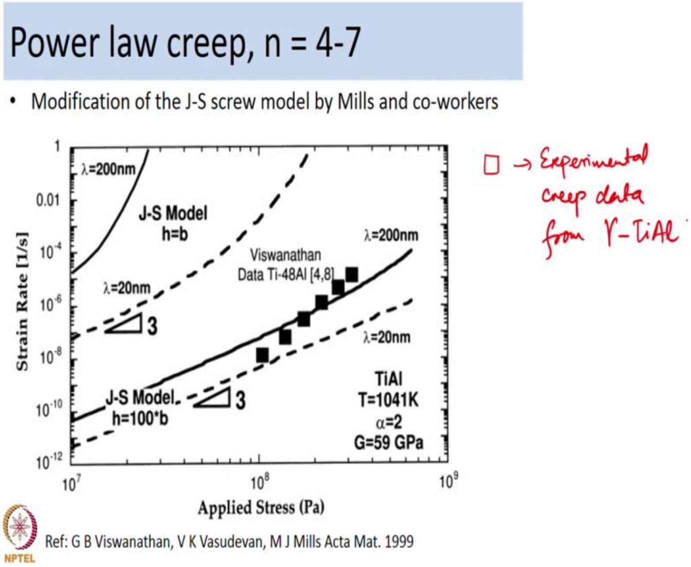

So they saw Jogged screw dislocations in the T E M micrograph so they thought they should use the Jogged screw model to explain their creep data.

So they tried that. So what happened was the Jogged screw model with

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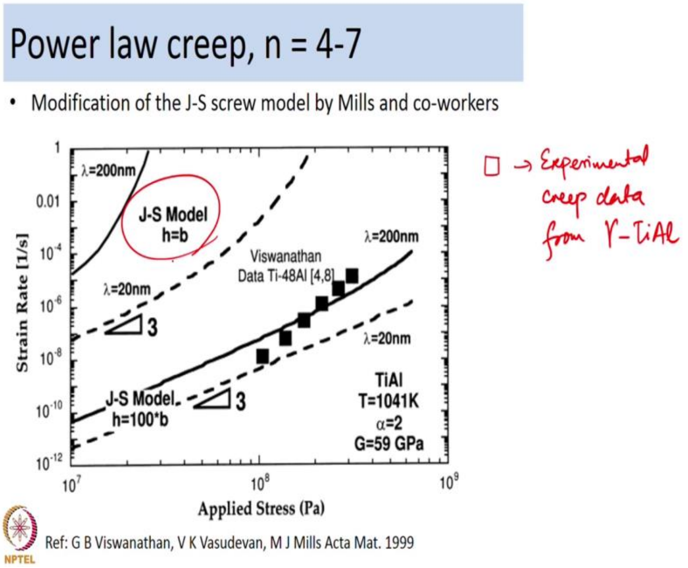

h is equal to b was not able to explain the data. So the h is equal to, Jogged screw model was predicting strain rates which were at least 3 to 4 orders of magnitude higher than that they observed experimentally.

So this is the experimental data and this is the prediction of Jogged

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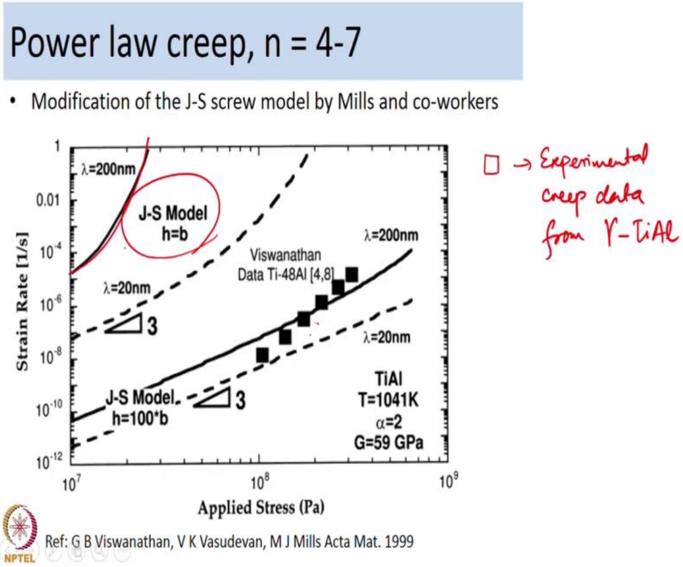

screw model with lambda is equal to 200

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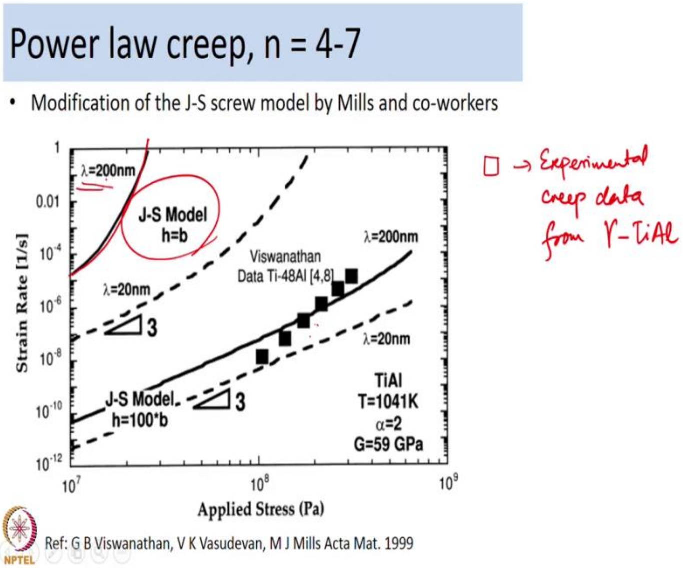

nanometer.

In the prediction of Jogged screw model

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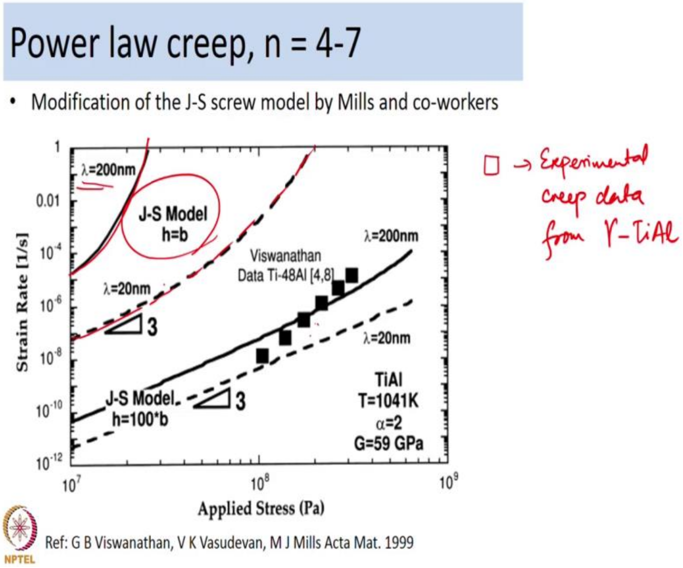

with lambda equal to 20 nanometer,

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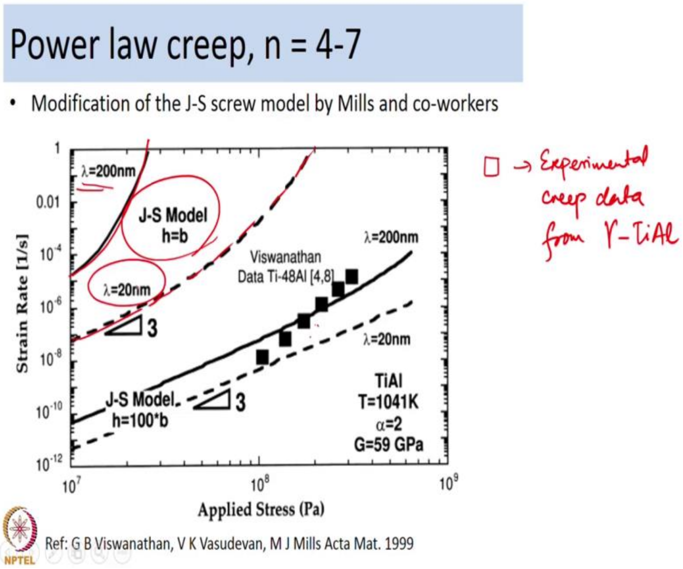

so you see at least, so

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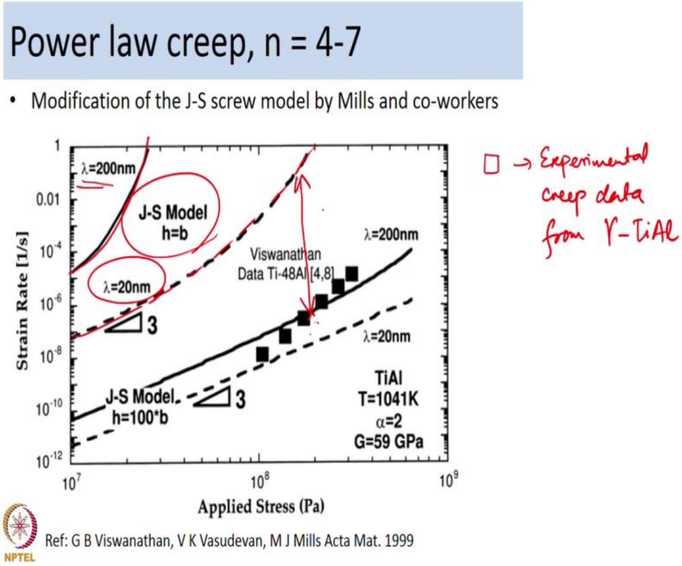

would say 10 to the power minus 2 is, say this is 10 to the power minus 2 approximately and data they have is 10 to the power minus 8, so the order of magnitude is, order of magnitude difference is 5 to 6.

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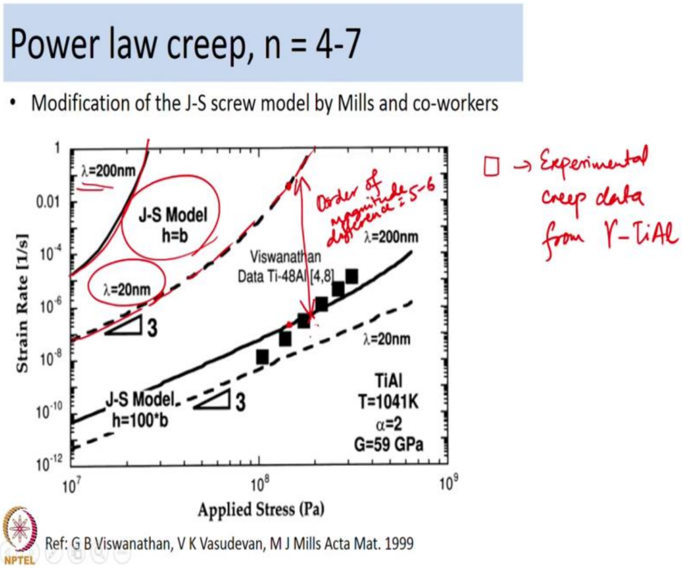

So you have 5 to 6 orders of magnitude difference between the original Jogged screw model which was, by assuming a Jog height of b and the experimental data. In order to understand this discrepancy, so they believed that, to use the Jogged screw model because microstructure was clearly showing lot of jogged dislocations, screw dislocations.

And what they found out was if they employed a jog height of

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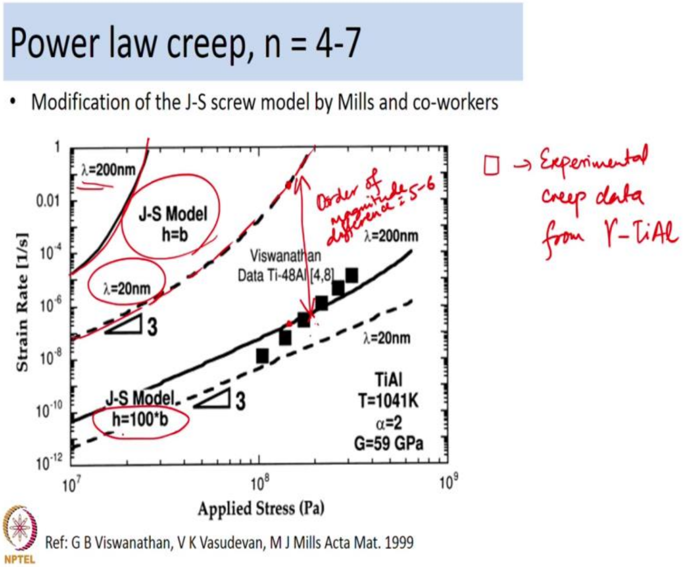

100 times the Burgers vector,

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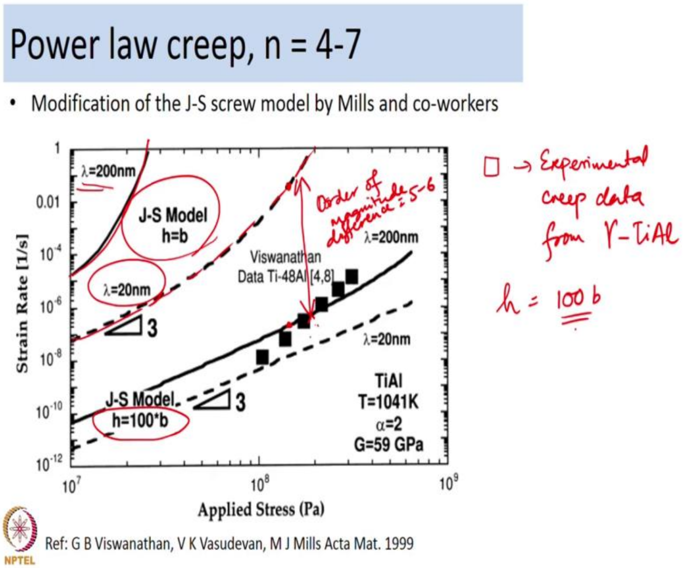

the jog height is going to 100 times the Burgers vector then for a jog spacing of approximately 200 nanometers and jog height of

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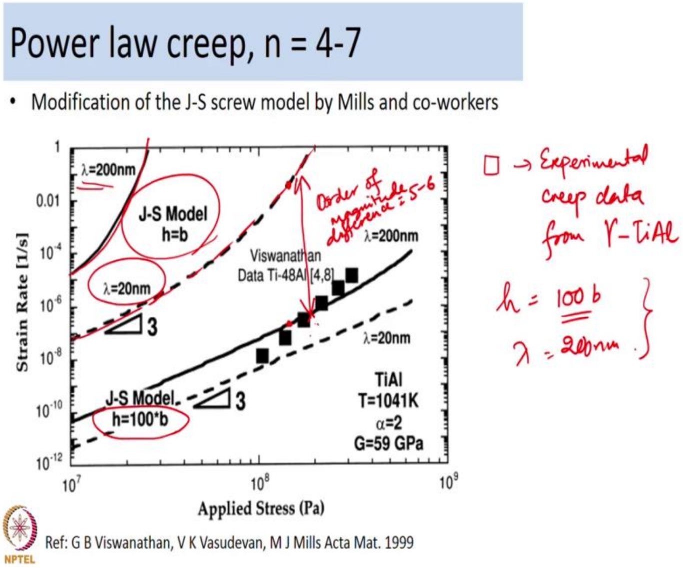

100 times the Burgers vector the model was able to describe the experimental data.

Actually what they found out was the entire data, creep data was working for a dislocation, jog spacing in the range of 20 to 200 nanometer. So

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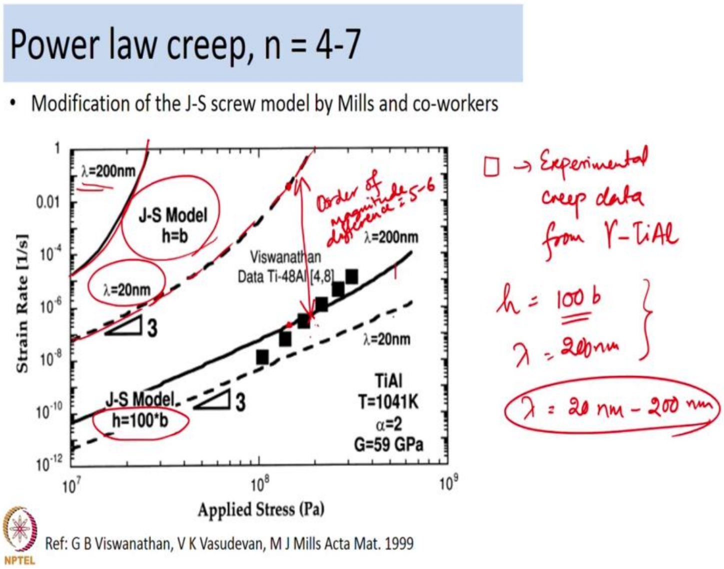

this is the upper and the lower bound. This is lower bound here and this is the upper bound.

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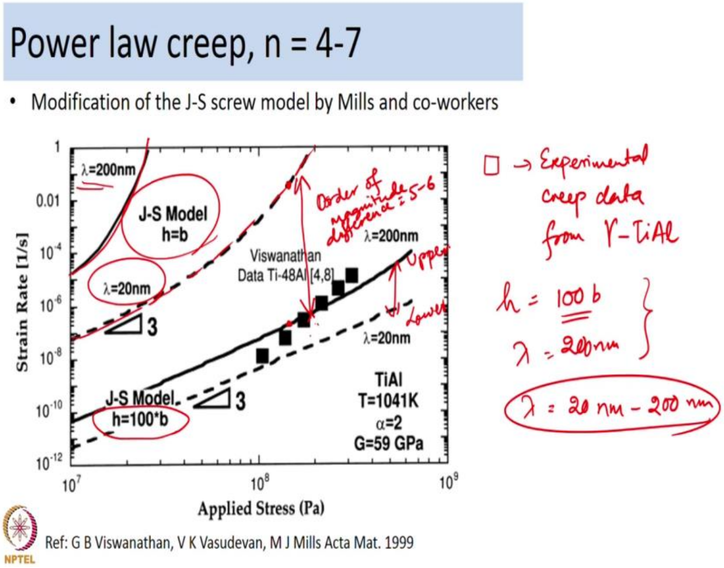

So the experimental data that they obtained was in that range.

So what they realized is the equation that Barrett and Nix came up with would not work unless it was modified and the modification they proposed was, in this particular paper was, the jog height can be several times higher than a single Burgers vector.

So for a jog height of around 100 times the Burgers vector the data was being explained well by the Jogged screw model.

So that was the modification they came up with and subsequently, now in support of this observation, that the jog height is 100 times the Burgers

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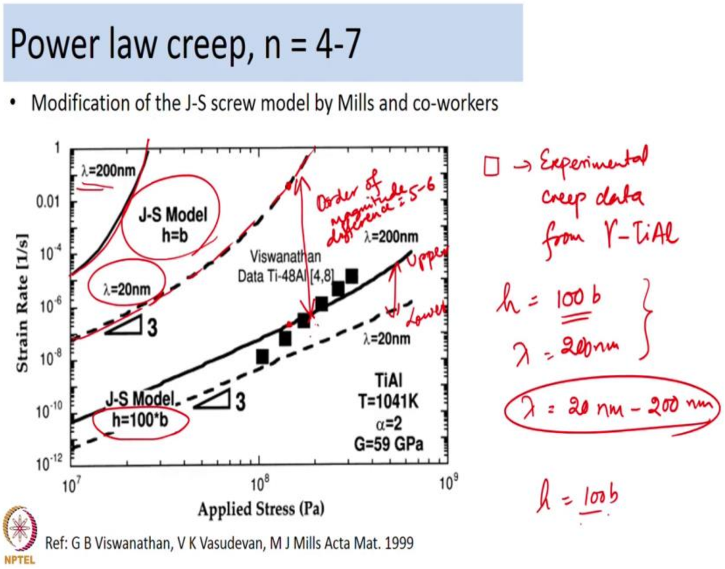

vector

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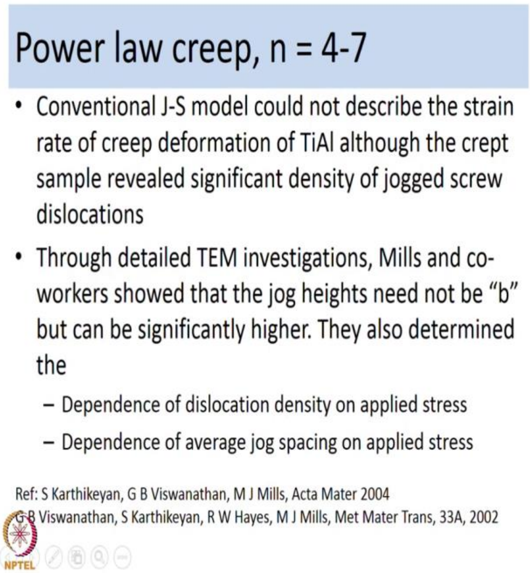

so the T E M data, they used the T E M to show that.

Through detailed T E M investigations, Mills and co-workers showed that the jog heights need not be **b** but can be significantly higher. They also, using T E M, they determined dependence of dislocation density on applied stress and the dependence of average jog spacing on the applied stress.

So in addition to the fact that the jog height has to be several times the Burgers vector they also decided to determine the dependence of dislocation density on the applied stress because Barrett and Nix said that the rho can be

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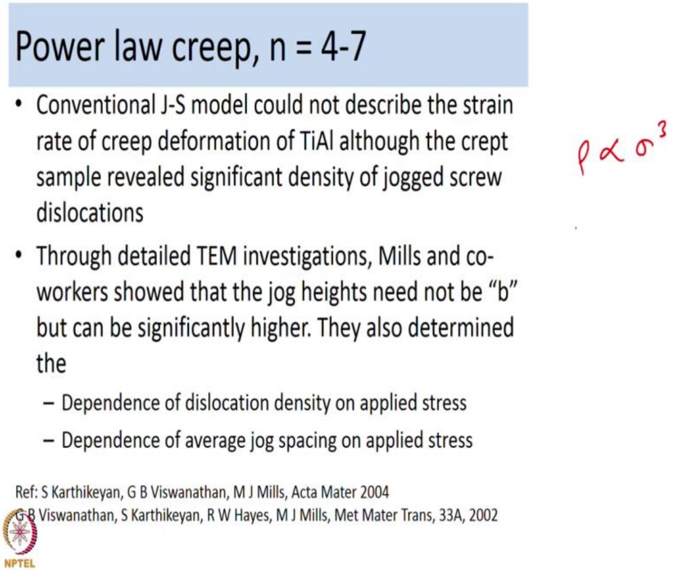

proportional to sigma cube.

Whereas traditionally people have been using sigma square

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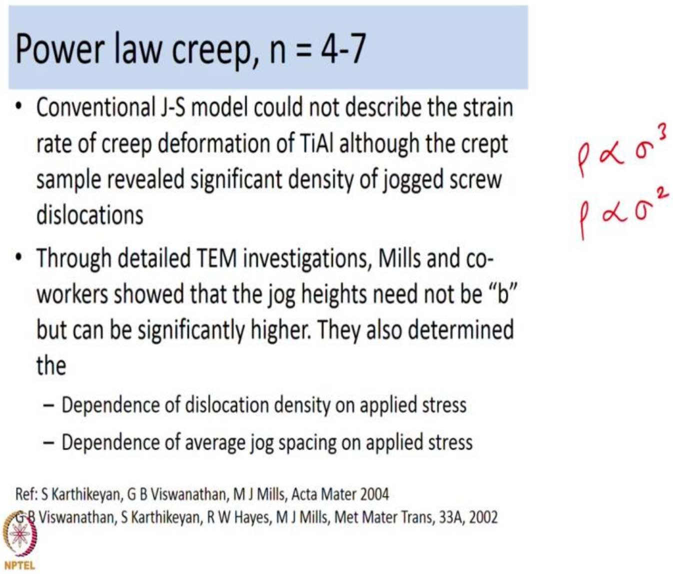

so they, and there is also a constant alpha, so essentially it is, so the constant involved on the dependence of dislocation density on applied stress, so to understand that they used a T E M. And also the lambda value, the dependence of lambda on sigma

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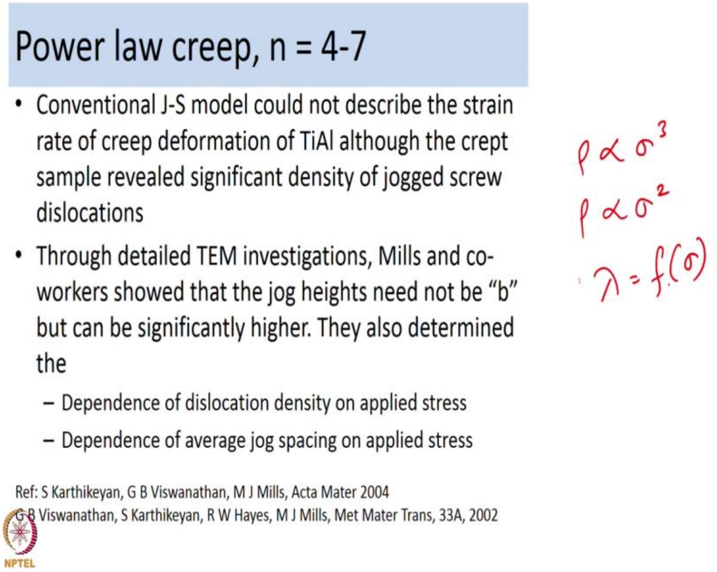

was also had to be determined and so they used the T E M for doing that.

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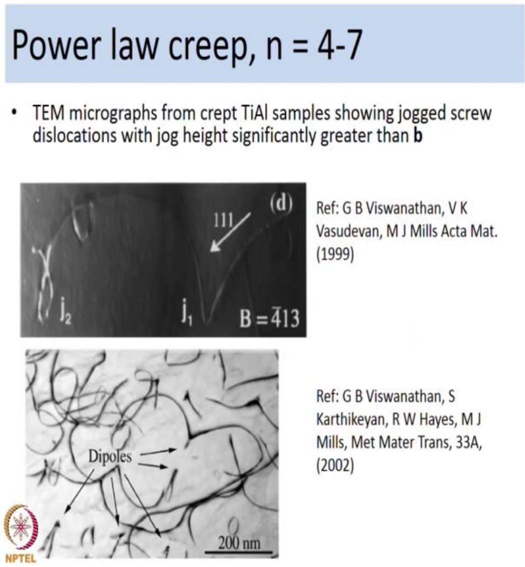

Now here is a T E M micrograph from a crept titanium aluminide sample and so this is a jogged dislocation and

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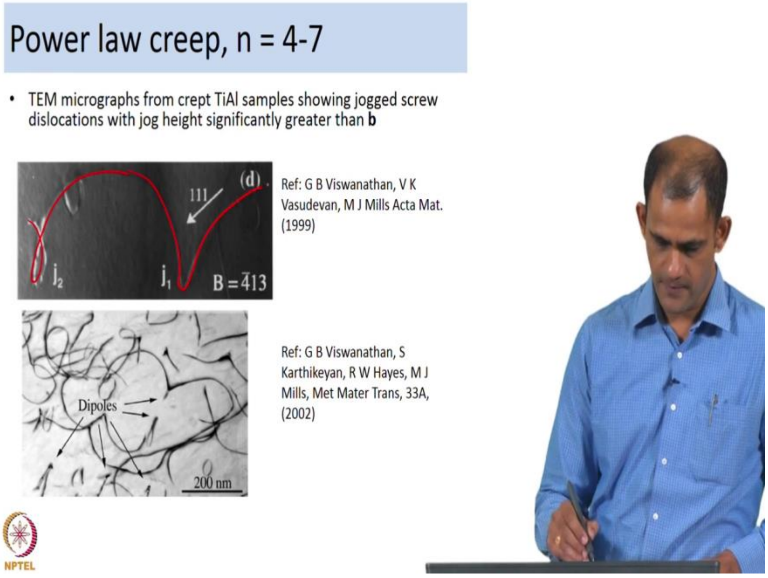

if you see, when you using some tilting experiments they found out the height of the jog or, so this is the height of the jog and if you can notice, definitely not a single Burgers vector, it is more than that. So they employed the T E M to understand that.

And another example is shown here. So you have another jogged dislocation

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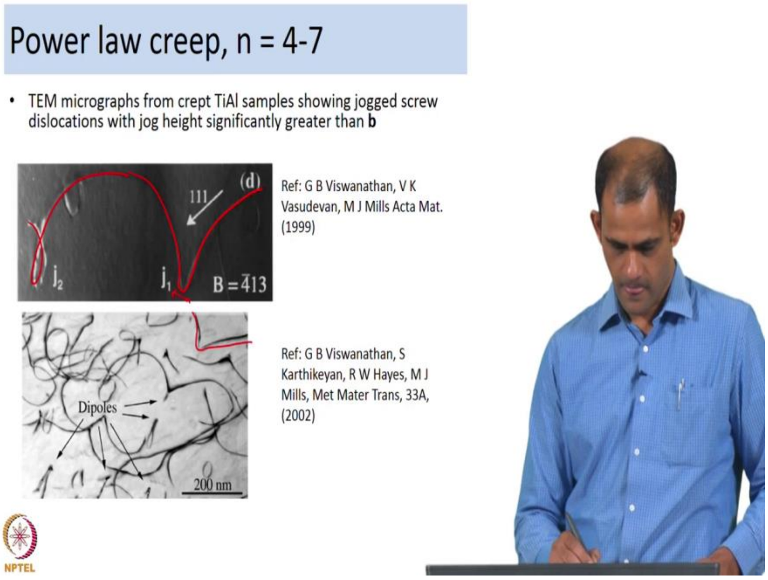

and clearly several times the Burgers vector. So that is how the group came up with a rationale to explain the difference between the experimental creep rates versus the original Jogged screw model.

So that was the two models, one was the Weertman's model and the second one was the Jogged screw model and its modification and its usefulness for explaining creep strain rate data as a function of stress and temperature.

Now, so power law creep we are

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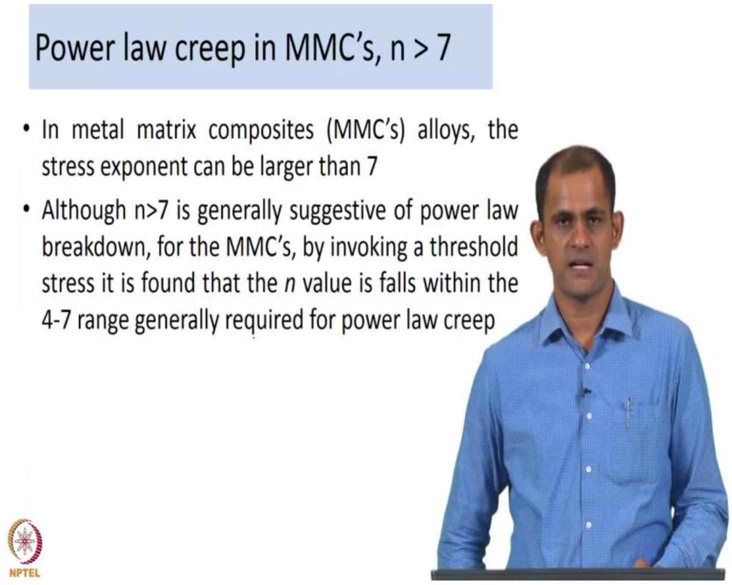

understanding as something where stress exponent is in the range of 4 to 7 but in certain cases you may get n greater than 7 but the

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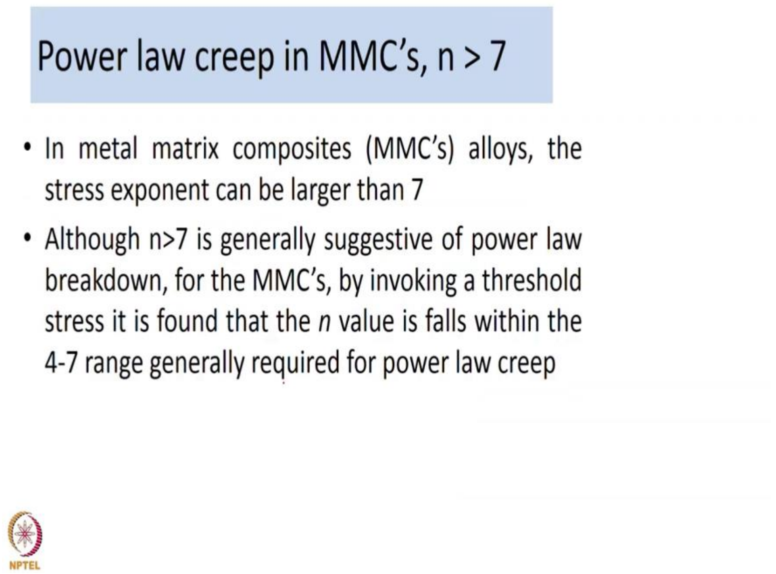

material would still be creeping in the Power law creep regime. Now having said that, when the stress exponent values are very large, say greater than 7 then people say Power law has essentially broken down.

So the Power law breakdown, there is another regime, the Power law creep regime, the Power law breakdown regime, P L B, so when you see stress exponent values greater than 7, then you typically say that is breakdown of the Power law. But in metal matrix composites you could see n greater than 7 but the matrix is still undergoing creep within the Power law creep regime.

So how is this happening is in metal matrix composites, because you have secondary phase, so say as an example aluminium and silicon carbide

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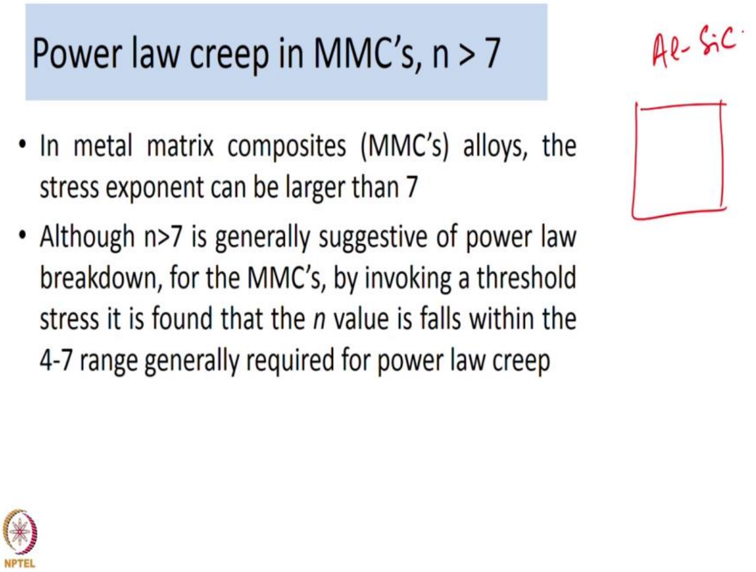

so you have a dispersion of silicon carbide precipitates, second phase silicon carbide within say aluminum matrix.

So the dislocations are now going to experience barriers to their motion from the dispersoids, you have the dispersoids

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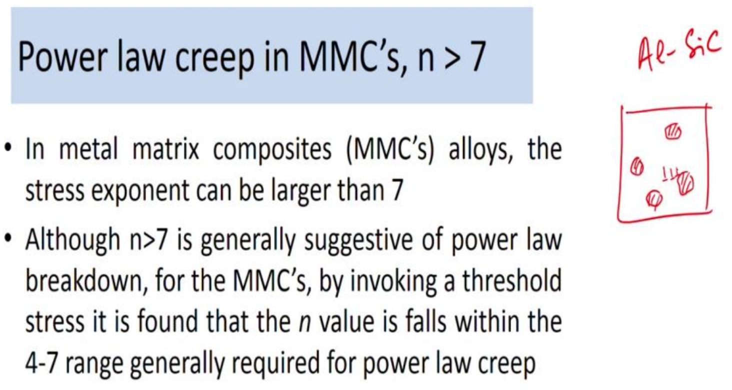

which are moving to, which are strengthening the material on account of the resistance they provide to the motion of the dislocations.

Now what researchers have found out is when you have a case like this, when you have a metal matrix composite, these dispersoids are going to, you need a minimum threshold stress. So when you have a situation like this

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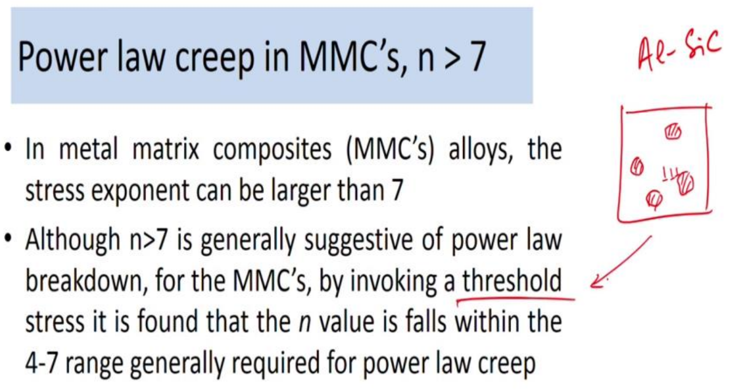

the stress that you are applying for creep to happen has to be higher than a threshold stress.

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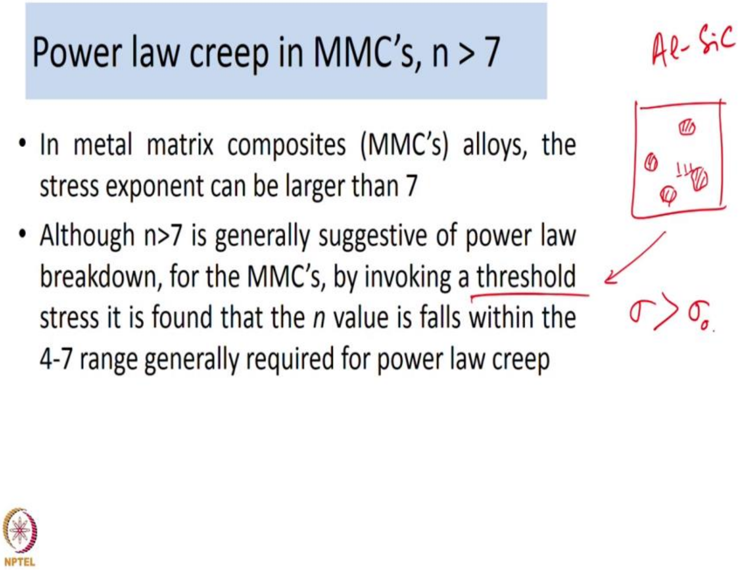

So let us call this threshold stress sigma0. So if the applied stress is greater than sigma0 then you can explain the creep behavior of these metal matrix composites.

And once you use the threshold stress to rationalize your creep data, what you notice is the end value then comes within the 4 to 7

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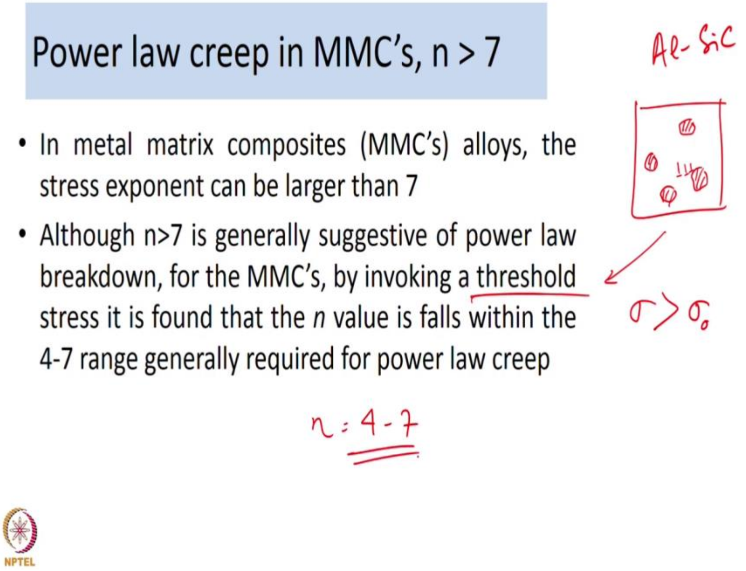

range that you understand

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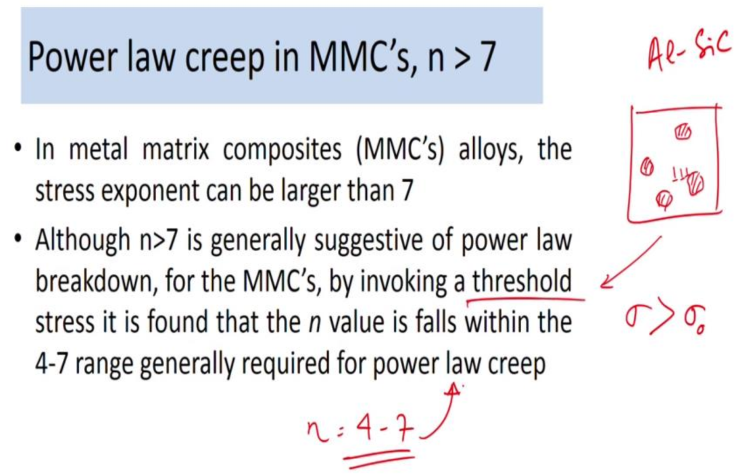

is implying, that you understand for Power law creep.

So basically you have to invoke a threshold stress to rationalize your data, creep data and once you do that then you will be able to understand, then you will be able to get stress exponent values in the range of 4 to 7.

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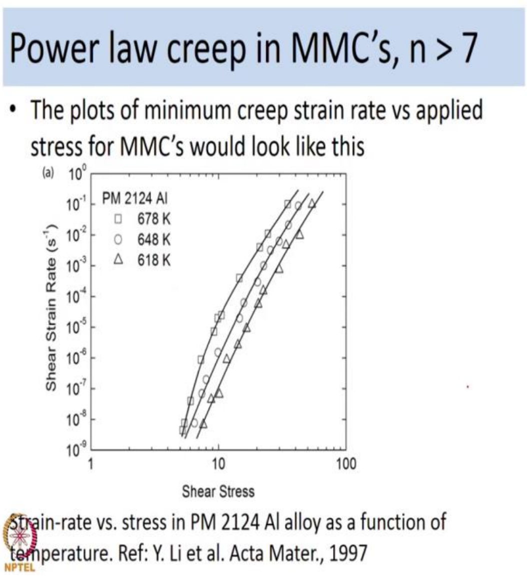

Now for metal matrix composites with n greater than 7 typically the strain rate of deformation, say epsilon dot versus stress,

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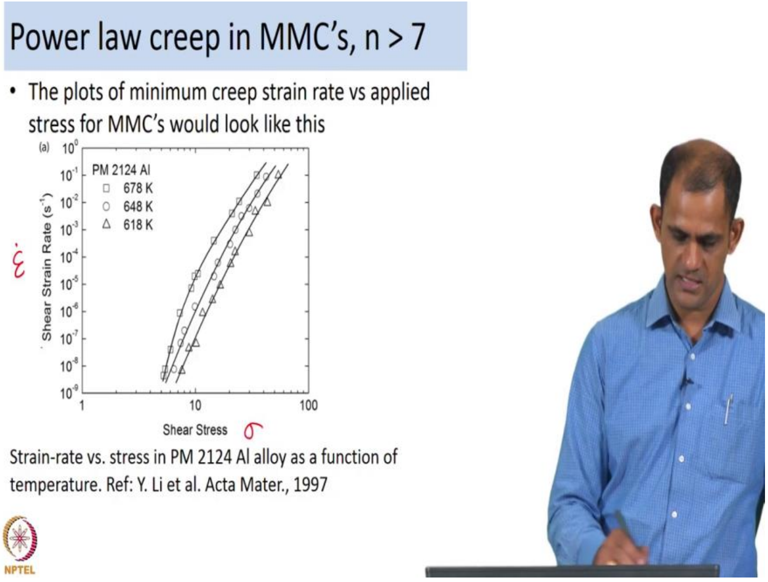

so this is shear strain rate which is gamma dot but in general if you have epsilon dot versus stress, so the kind of creep behavior, the creep behavior will be like that, so you will see a curvature, you will see a curvature in your plot.

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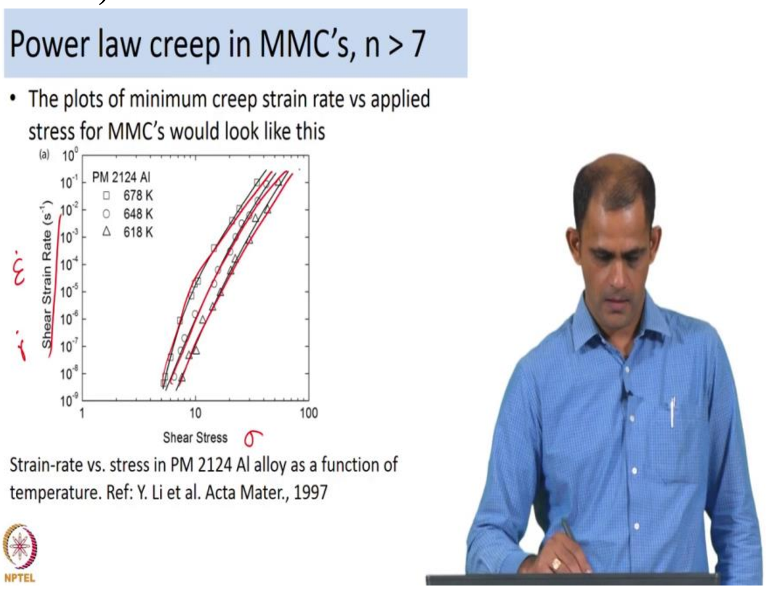

So it is not, typically you are, typically for a material behaving in the Power law creep regime, we are going to expect a linear

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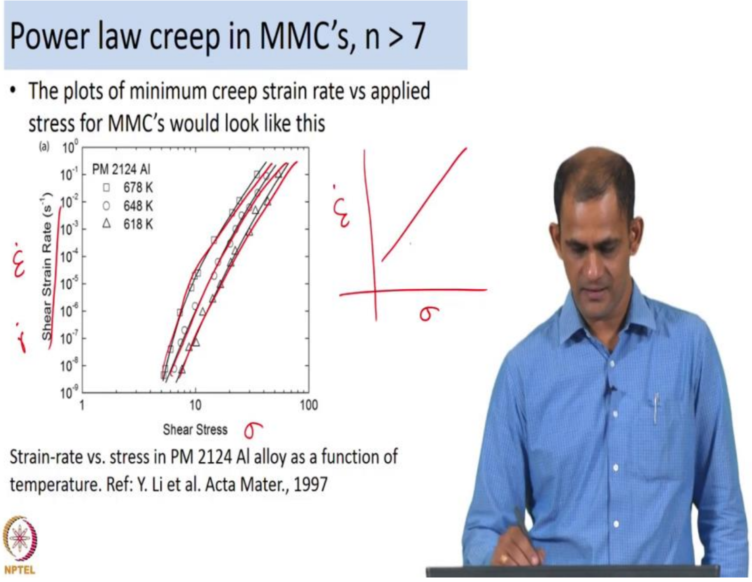

line like that and with n in the range of 4 to 7

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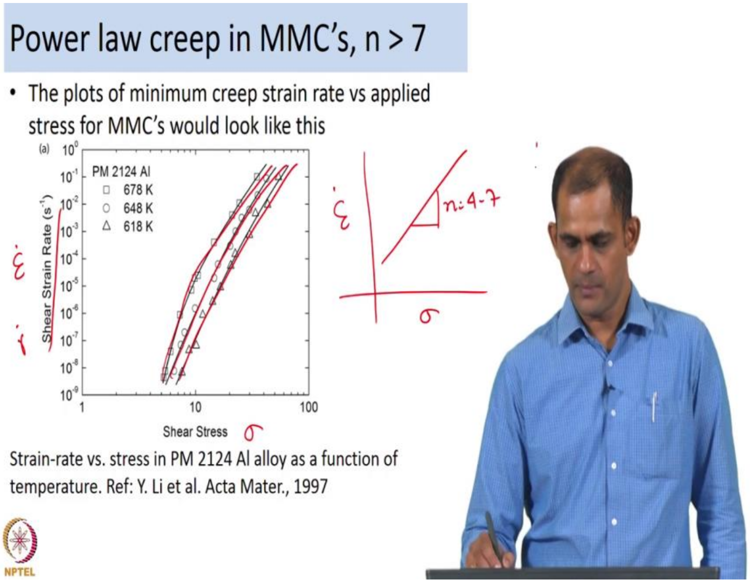

but for metal matrix composites, during creep you may not see a straight line but you may see something like that. So you may

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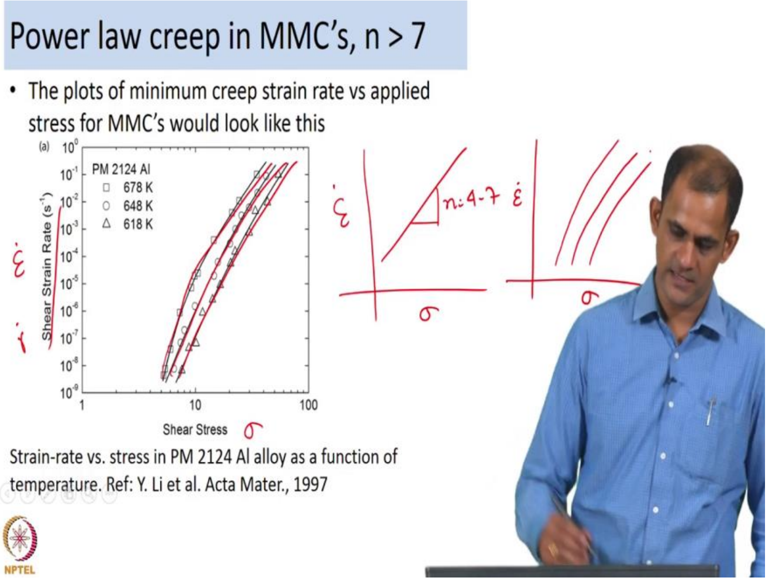

see a certain amount of curvature in the plot.

So

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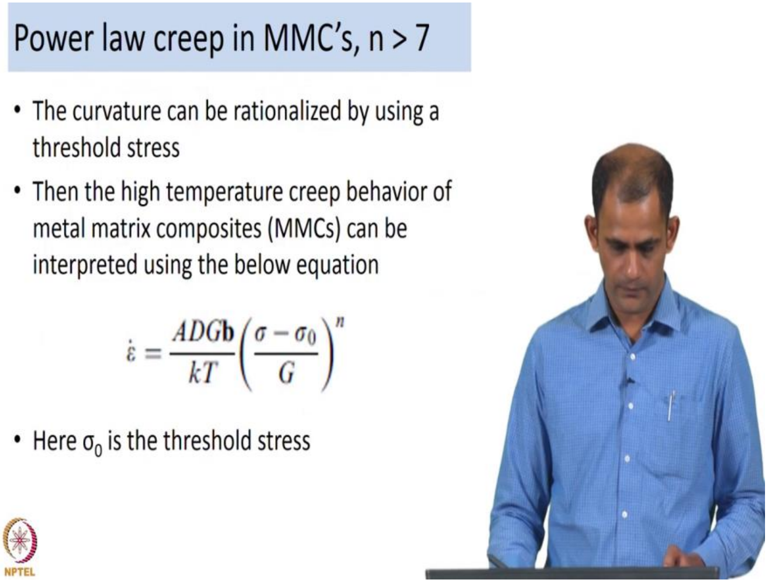

the equation that you should use, so the curvature like I mentioned can be rationalized by using a threshold stress and the equation that you

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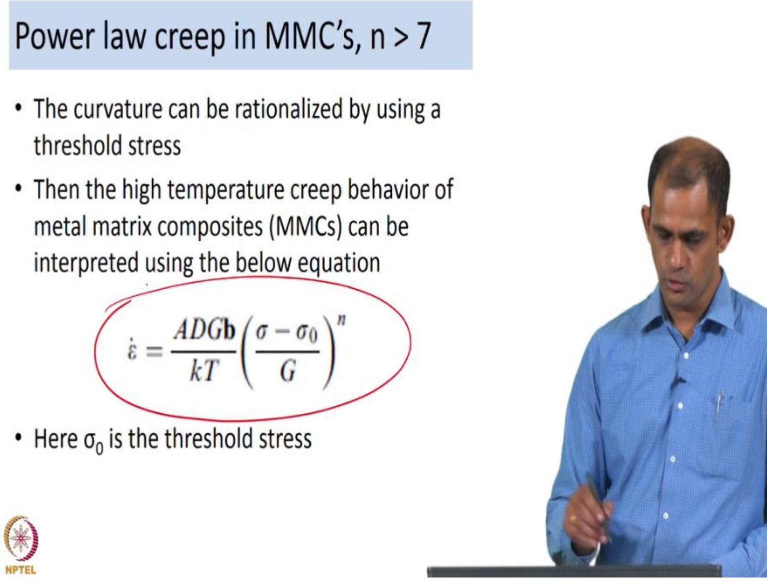

should use to describe the creep behavior of a metal matrix composite, he has given here so you have to have, the stress that is actually dragging the creep is an effective stress. It is not the applied stress but it is applied stress minus sigma0.

So the effective stress

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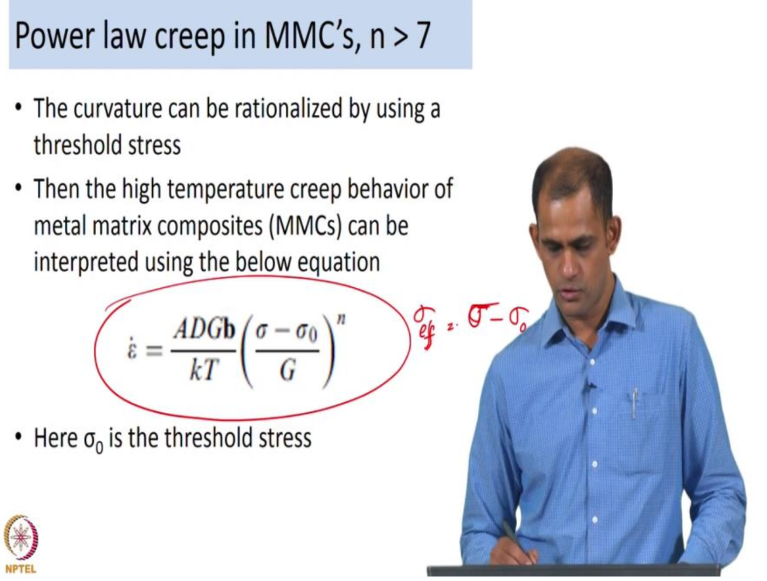

is equal to sigma minus sigma naught. Now you, one can find out these values of the sigma0 from the strain rate versus sigma plots. So from these plots one could find

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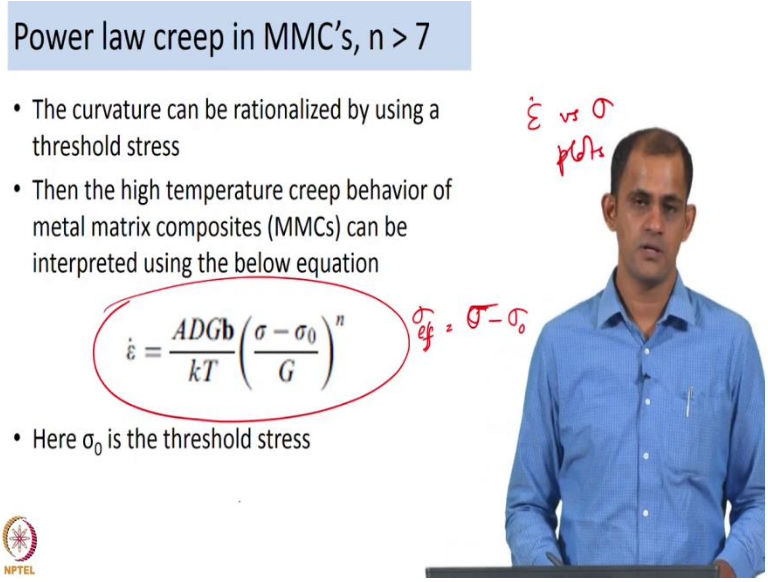

out the value of the sigma naught.

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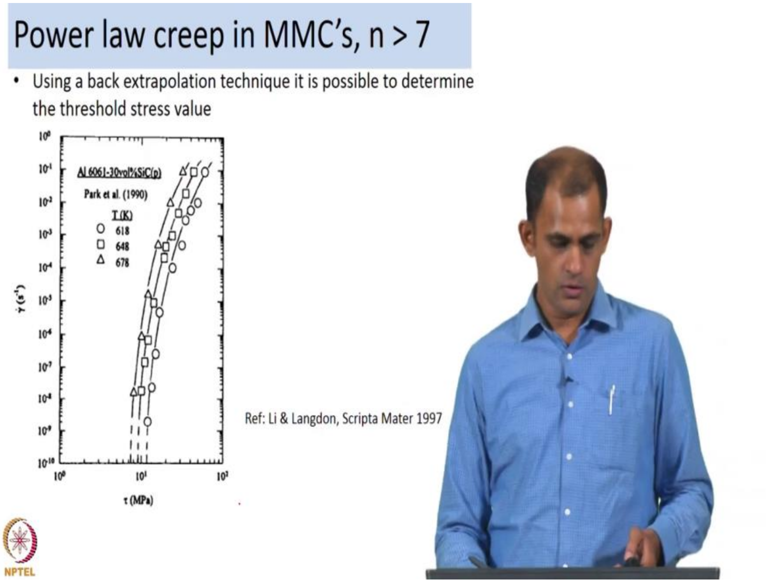

So one simple approach that has been proposed by Li and Langdon is

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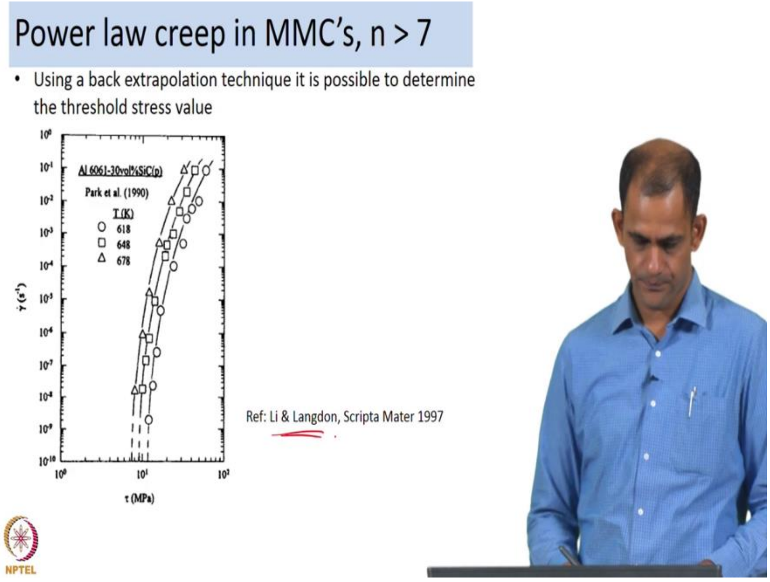

back extrapolation technique. So when, if sigma is very close to sigma naught,

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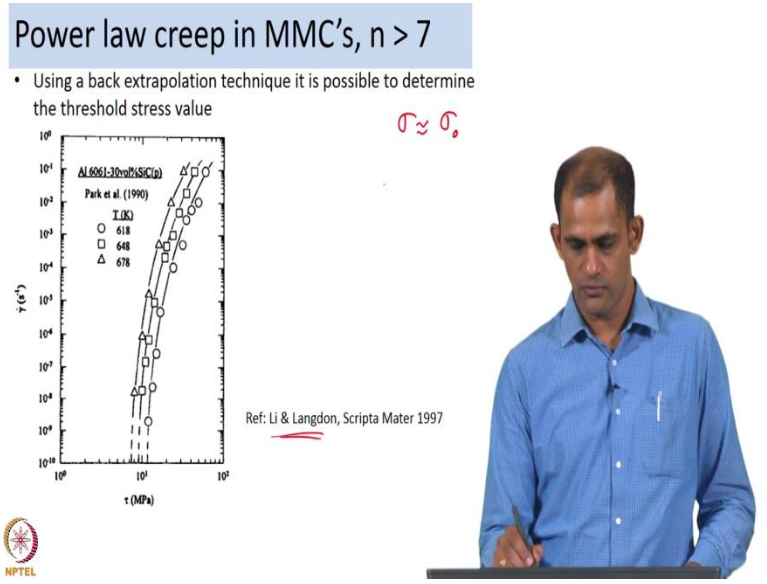

so we saw that epsilon dot is proportional to sigma to the, sigma minus sigma naught to the power n.

So if

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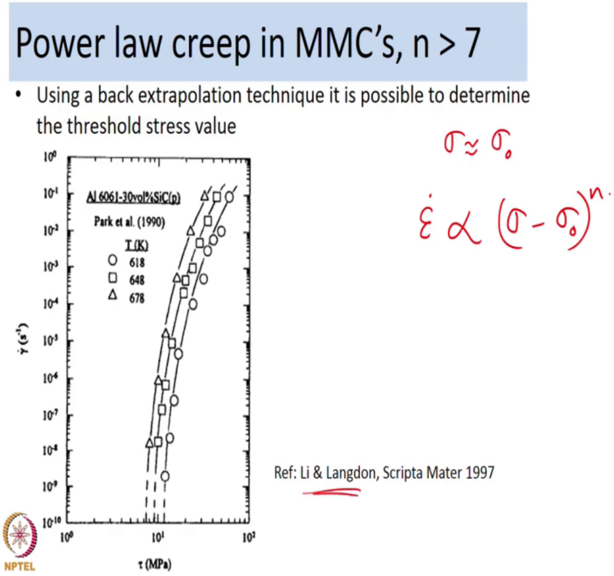

applied stress, if your applied stress is closer to

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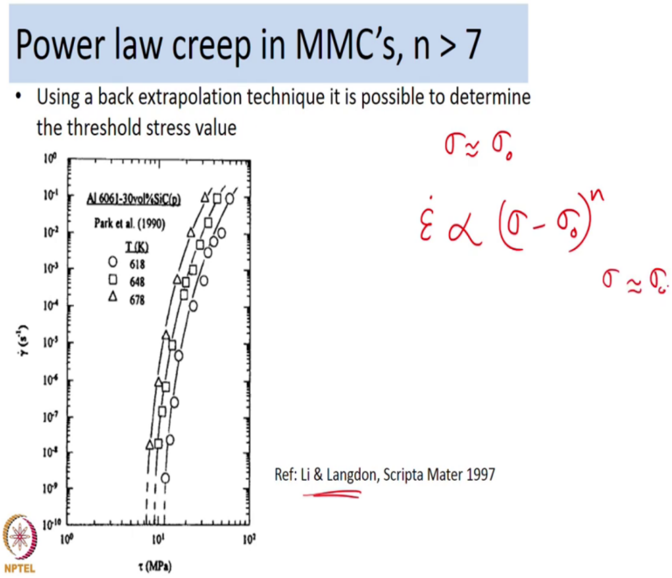

Sigma0 then your strain rate of deformation will tend towards

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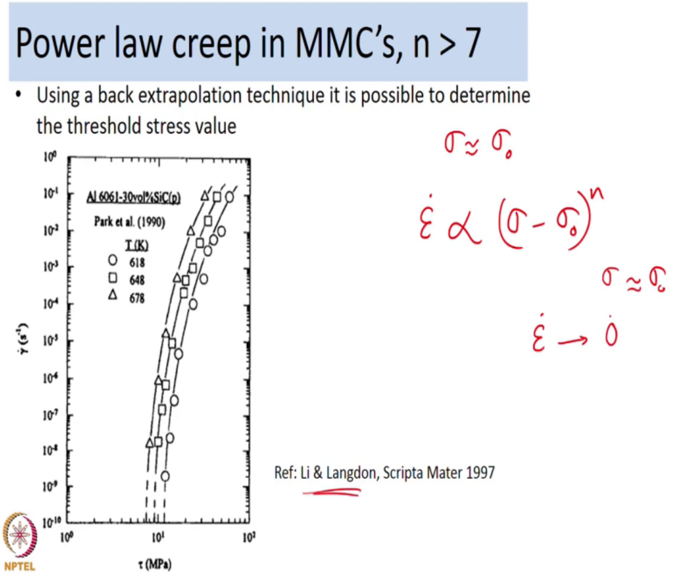

zero, right. So if sigma is close to sigma naught, then epsilon dot is close to 0.

So what Li and Langdon said is, if you extrapolate, back extrapolate your creep curve then it, sigma close to sigma0 you will get strain rates which are so slow, so small so that you can almost say that your strain rate of deformation

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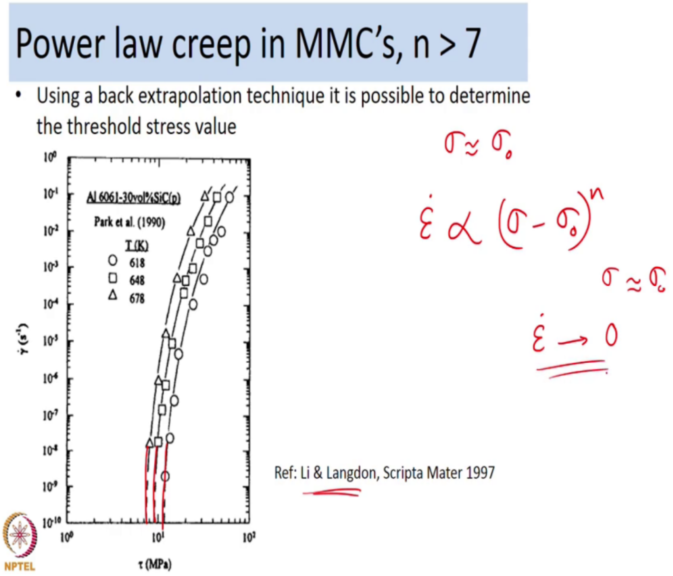

is close to 0.

So the idea of Li and Langdon was back-extrapolate and wherever the curve intersects the x axis, so here x axis is tau, we could also

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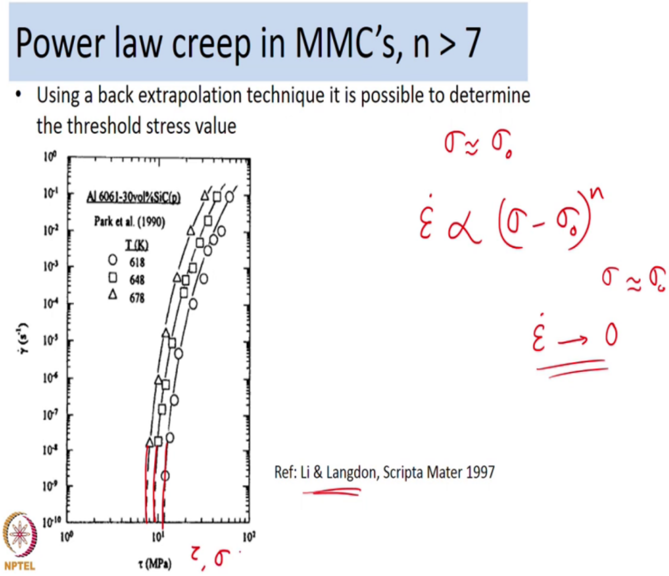

say it as sigma so wherever it intersects the x axis and assuming a strain rate of, considering strain rate of 10 to the power minus 10 as almost equal to

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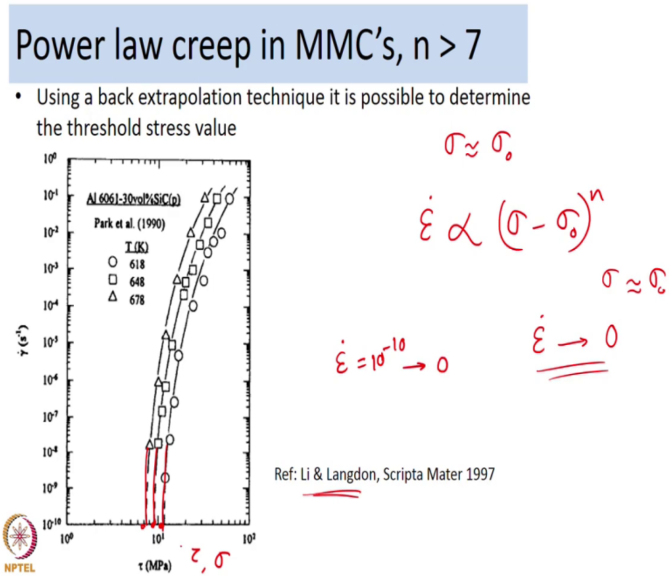

0, you can say that this stress corresponds to your threshold stress.

So that is the approach they suggested for determination of the threshold stress. There are other approaches as well, there is a mathematical method you would use these graphs also for determining the threshold stress and there will be one numerical that we will do on this concept to understand the other approach for determining the threshold stress for these materials.