Defects in Materials
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Lecture - 34
Interfaces-02

Welcome you all to this course on defects in material. In the last class we had a brief introduction to interfaces in material which is also a type of defect especially when deformation has to be propagated across r, when we look at optical properties of the material, gather few properties of the material, the interfaces have a lot of role to play in how it behaves. For example, if you take a crystalline material which is perfectly a single crystal you will find that like silicon, if you take it if it is a single crystal silicon, we will find that it light passes through when it is thing, but even it is the fine grained one we will find that it becomes translucent the opaque to light.

Similarly, when dislocations have to pass through that is if the if a material is a fine grained material we always say that it has a better strength at room temperature compared to the ones which a single crystal or essentially a crystal with a large grain size. Not only the grains, but what is the nature of this grain? What is the nature of this boundary, not only between the grains between the second phase particles all these things have a role to play on role to play and decide the strength of the material as far as mechanical properties are concerned?

So, having had the brief in a last class we had looked at the surface energy and then also what are the different types of classification of interfaces which we have looked at, when we consider a polycrystalline material essentially we know that it kinds of different types of grains, grain that is even in a single phase material, the grain is nothing, but a single crystal when we say that it is a polycrystalline material different types of single crystals oriented in different way, with an interface between them.

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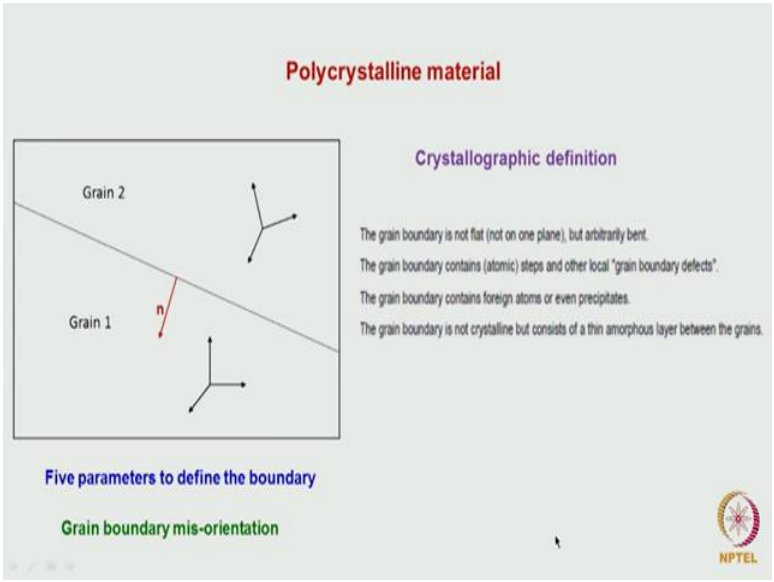

The simplest case what we are considering it here, with some interface and this is a 2 dimensional arm lag which we are showing it, but the axis which I am showing it is essentially a 3 dimensional one. If you look at the orientation of the axis here and the orientation of the axis here, because since the crystal structure remains that same the coordinate system if you see only thing it is differently oriented correct. So, suppose we have to represent the grain boundary.

Crystallographically: how will we represent this grain boundary? The way in which we can do it is that from this orientation to reach this orientation around different axis we may have to give some rotations correct. So, essentially 3 parameters are required to rotate this axis. So, that we can reach here that is how from this grain orientation we can reach this grain orientation. Then the boundary between them is with respect either this grain or the other grain. If you take the plane normal the coordinates of that itself can be it is like an h k l it can represent.

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Ok, but when we wanted to represent with respect to this plane normal, what we required these 2 parameters which are required that, because any vector if we consider and represent with respect to this coordinate system we normally require you will see the 3 angles have to be given correct.

Like for example, this is x if this is a vector r. We will find out the angle to gamma gives make, but essentially what is going to happen is that correct equals one; that means, that only 2 are necessary the other one is already defined by this relationship. So, essentially we require 5 parameters to define a grain boundary. This is as far as the crystallographic definition of a boundary is concerned, but then comes what is the structure of the boundary what is the energy of the boundary all these things. Here we assume that this boundary is a flat boundary perfectly a 2 dimensional surface, but that need not be the case quite often the boundary could be curved that is one and another is the boundary can have some steps on the boundary, that is one, that could be lot of defects which are present on the boundary. It is possible that the boundary contains other elements which are also sitting at different sites on the boundary.

Second phase particles could form at the boundary .in some cases one could assume that the boundary itself is an amorphous layer. So, if you think of that the grain boundary are defect is a very complicated one. Anyway, let us first consider the simplest case not the curved ones, but a simple one which is a flat 2 dimensional boundaries. Essentially what

we wanted to look at what is the structure of the boundary, what are the ways in which we can consider the boundary? One way in which we can look at it this is that yes, I mentioned just earlier we can represent it with respect to 5 parameters which are necessary correct. Using gbsd in an SEM, we can find out the misorientation or the orientation how with respect to one other grains are oriented.

In that process when we do that only 3 parameters which are necessary. How from one grain one orientation in a grain to an another grain, how by rotation we reach that orientation that is what we look at it. So, that only gives a misorientation angle correct, but there any be a see we never look at what is the grain boundary structure, we never bother about it. You only look at mis orientations that is how the work is run that should be very clear.

There are many other techniques like in TEM. We can look at using the fraction exactly get some information about that type of grain boundary structure.

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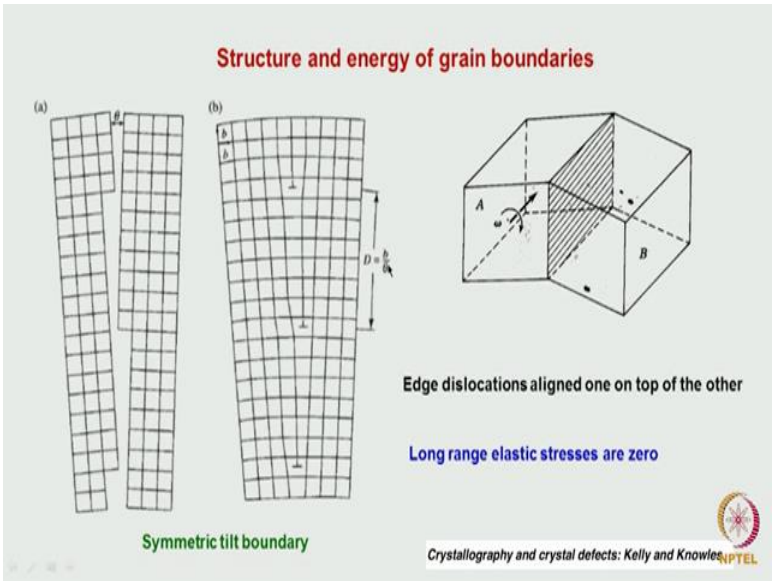

So, essentially suppose we take a crystal. They are single crystal and another identical one, we take it with the same orientation it is placed and if they are joined together. This can give rise to another one single crystal.

Like in the case of an ordered alloy, there is nothing like a stuff which is being created. What I mean by stuff is that, suppose there is assume that the planes are there. Here also

the planes are going to be exactly that same way. There are 2 ways in which it can happen. These itself of capi shifted by some value, which is smaller than that inter planar distance, the planes have to touch each then that is the way it happens in the case of a single phase. Whereas, when it is an ordered phase that I teach that, but different type of atoms could be there. That is how the anti-phase boundary or the translational boundary comes, which we have discussed earlier. So, if that is the case this is the way they will be touching each other.

So, it becomes as a one single crystal joining them together. Suppose we have this sort of a crystal one crystal and if I make a small cut like this making an angle and join them together. Then what will happen is that this is an exaggerated the huge. These is has if from here to here there is a tilt which has taken place that is what is being shown in this diagram. That is the around an axis normal to the plane grain boundary, that is around an axis which is in the grain boundary plane we are now rotating it by an angle.

So this angle could be very small angles. Generally, what happens is that even if we take a single crystal of a material, which has been drawn by various processes, you will always find that this sort of boundaries do exist, very small angle boundaries are there. In this particular case what we are showing it is essentially is that it is a simple cubic lattice 2 dimensional lattice which we are showing it and here, it is a close packed plane with lot of ledges which has being created each of this ledge, if we join both of them together this ledge will become like an extra plane which will be equivalent to introducing.

And one extra plane, so it is dislocation, if you join them together now we will have one dislocation and another on top of another on top of it like this will come. And with respect to this plane look at this boundary plane, the crystal on either side is symmetrically tilted. So, this is called as a symmetric tilt boundary. We should understand that essentially one type of dislocation is nil there, only an edge type of a dislocation kept one on top of the other, and this boundary is generated. What is the other important characteristic? Whenever we considered as a grain boundary another important aspect which we say that if you assume that these boundaries are made with dislocation, generally we know that when dislocations are present the long range stresses like if a dislocation is present here, the displacement field will be there all over the long range elastic stresses is there large distances.

When these sort of boundaries are there even if they contain dislocations. The long range stresses should not be there within the grain; that means, that the dislocation should be oriented in such a way are the nature of the dislocation should be such that long range stresses are not present. Is it clear?

That is long range elastic stresses should be 0. Then only we can say that this is a boundary otherwise it is something like a crystal which has been deformed correct. Which long range stresses are present? When you defence a grain that is we assume that this grain is essentially does not contain any deformation this dislocation are present at the boundary such that the rest of the region is stress free. How can it be done? That is what we are trying to look at it if you look at the stress field around the single dislocation. I am just showing the stress sigma ez, ez over an edge dislocation.

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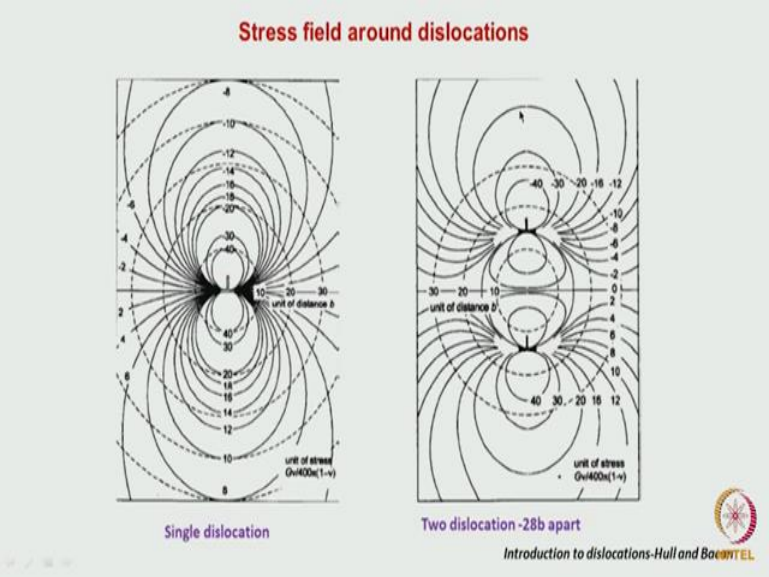

If a single dislocation is there this is the way, the stresses it is iso controls of the stresses are being or shown both tensile as well as the compressive stresses.

Both are being shown if we have one dislocation and then another dislocation on top of it, just above one another, then what we will see that at this bottom there should be a tensile stress. Corresponding to this one qualitatively looking could there should be a compressive stress should come. So, these stresses will cancel. When this happens on either side you find that when we have an array of dislocation, finally, what will be left

with these long range stresses will not be present. So, though we have an array of dislocations are present in to the grains, they are stress free.

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It is essentially a dislocation one on top of the other which it comes. So, from this figure we can understand that it is qualitatively all that is how why it is necessary to have dislocation to one on top of the other. That is precisely what it happens when we take a material. And during recovery process when the polygonization takes place. The dislocations with opposite sinks they cancel each other.

Other dislocations come one on top of the other to minimize that energy and so far the considered as a symmetric boundary.

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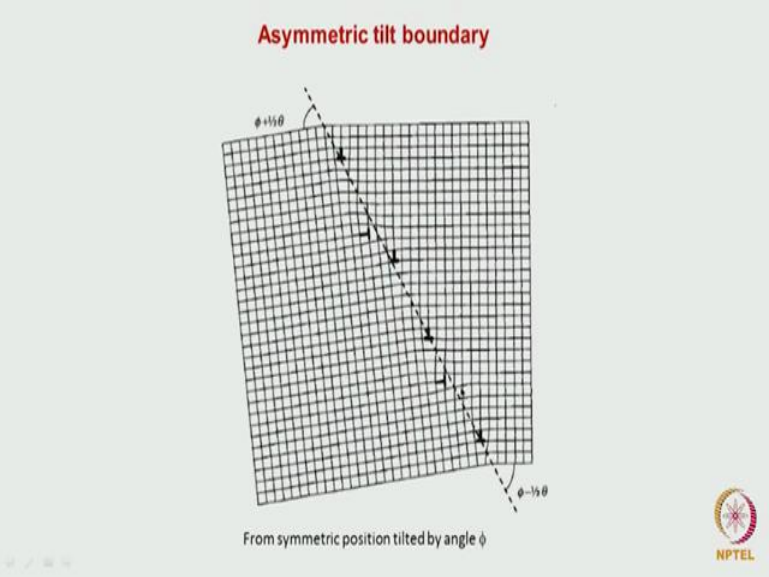

Now, let us look at here if you have a dislocation of an another that is all the dislocations have the same type of a burgers vector, the symmetrically boundary. Before dislocation with a different type of a burgers vector oriented with respecting this if it is being if it is present. Again this is a small angle boundary, tilt boundary, but this is called as an asymmetric tilt boundary because from this region to this region if you try to come, with respect to this boundary plane, it is not symmetrically tilted. The tilt angle is here it is 5 plus half theta here is 5 minus half theta.

So, essentially these are called isometric tilt boundaries. I will show you some examples where it has been this type of boundaries have been imaged. So, far we looked at just the structure of the simple tilt boundary. And another aspect which we have to look at it, here is the angle that tilt itself is controlled by what is going to be the burgers vector. And the separation between the dislocations, that is angle theta after tilt will be burgers vector by d.

That means that if the angle of tilt becomes large, the separation between the dislocations have to take place. That is higher the dislocation density in the plane and only one type of a dislocation is required then the angle increases. Then how does the energy of the boundary changes as a function of misfit angles, the tilt angle. This has been calculated by Read and Shockley in 1950 itself in a physical review paper it has been published.

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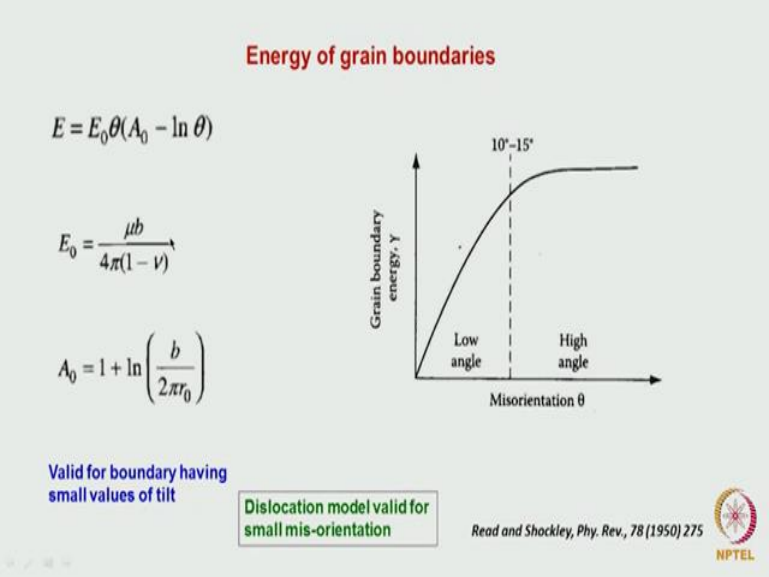

Essence of it, essentially is that this is e 0, which is essentially depends upon the material properties. We are it is assumed that it is a material having a modulus mu, an infinites material and Poisson's ratio nu. And r 0, is the core of the dislocation. This is the way the energy is varying that final formula which we get it; that means, that as theta increases the energy of the boundary also increases.

The basic way in which they have done this calculation is assuming, that assume that, that the end of each of the boundary, like we have some dislocations which is presented a plane, which comes from here there are some dislocations are there, like this from here,

also; that means, that there are 2 grains oriented differently with dislocations; that means, the energy of the grain is high, if they join together the energy has to add a body is going to be the energy, that is how this entire calculation has been worked out.

That one can look into details in the paper. Essentially from this sort of an expression we can immediately understand that, that has the misorientation increases the grain boundary energy gamma increases, but we need go on like this go on increasing, it cannot happen the reason essentially is that this is based on the assumption that each dislocation from minus individual dislocation. Beyond a particular angle theta, we can calculate what is going to be the separation between the dislocations. If the separation between the dislocations become so small, that the core of the dislocation starts merging. Then the concept of telling that the dislocation is responsible for this boundary itself loses meaning.

Correct. So, this sort of a equation is applicable only for low angle theta. What is the grain boundary energy which has been seen beyond this value, it is being shown as it is nearly a straight line, but let us see whether it is so, or whether it can have some variations?

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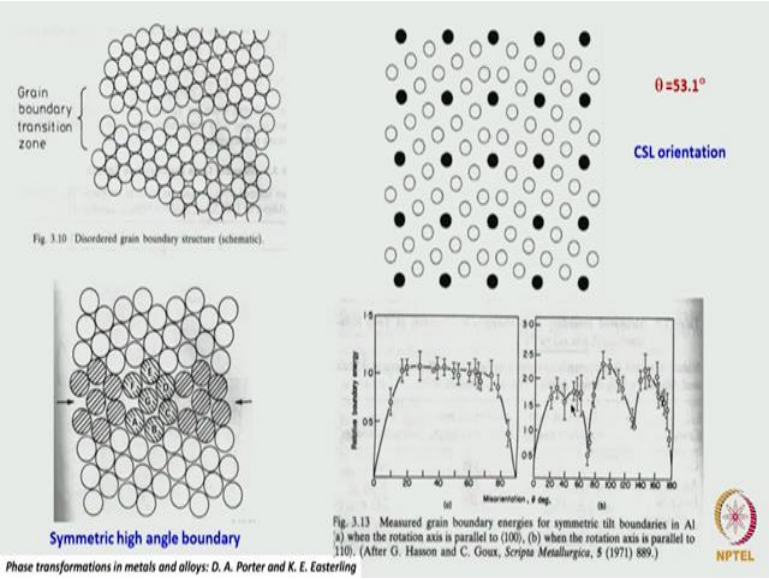

That will become clear. Here what we are showing it is there atomic planes you look at it close packed plane how they are arranged you look at this direction here. And in this one

the close packed direction is in this way. These 2 if you look at it the misorientation between them is very large correct.

When they come and meet at the boundary in large angle boundary. Essentially there are a lot of regency atomic planes do not exactly match each other at the interface. In the small angle boundary when the tilt angle is very small the atomics planes are matching perfectly.

Here the atomic plane matching is not there. There are some cases where what it can happen is that this matching may not be there at the boundary, but if you look at these planes they appear to be symmetrically dispersed with respect with respect to a boundary correct symmetrically tilted with respect to a boundary; that means, that if we consider these 2 that is the interface where the 2 orientations, and just continuously changed the angle. That is take one grain and another grain what we are considering, it is that the misorientation between them is if we continuously increase. Is it that some orientations there will be a perfect matching which can happen up to as interface, this is what exactly we can be here you see that this is one orientation in which these atoms are there. This is an another orientation in which it is there this black dots essentially represent the atoms which are common to both of them.

So, now this sort of an orientation if you try to look at it, here if you try to find out how many atoms are exact orientation in this super cell compared to the other. We can tell them this is what we call it as a coincidence site lattice. This sort of orientation this is for a simple cubic lattice which has been considered at an angle 53.1 degree centigrade. One degree we can get a perfect matching can be this is like a something like a twin orientation with a coherent twin boundary comes in between them what will be the consequence of it is that when we try to find out the energy of the boundary. At some of these angles that is where this sort of orientation exists. There will be an energy minimization of the boundary will take place. Here we have only just looked at that interpenetration of it.

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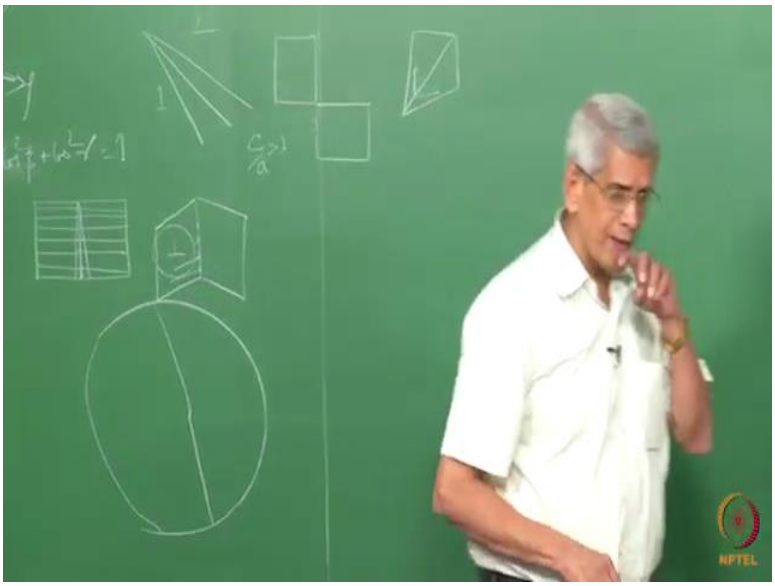

We do not know what that boundary could be that is, what we can see is that, if we have 2 crystals that is grain 1 and grain 2 we can have a grain boundary here or boundary here or boundary there, with different types of misorientation between them locally when we choose them depending upon the mis orientation the boundary energy can change. What boundary it will choose is the one which generally which will have the lowest energy correct. For various angles of mis orientation we can have curves like this in the energy. These are also possible these are all more favoured, compared to other boundaries with very high energy.

These are all what we consider it is equilibrium calculation, but generally when they grow they have form from the melt there are some restrictions and how they come and join because of that over a course of time, when we do the aging treatment we find that initially there will be boundaries with all types of orientations gradually during the grain growth many boundaries will vanish finally, it will reach and oriented with boundaries which are having minimum energy is possible, that sort of an orientation will be achieved in material.

So, essentially what is the important significant fact which one can notice is that, there are boundaries with minimum energy configuration high angle boundaries are there with minimum energy configuration is possible. The simplest example is itself is a twin

boundary that in FCC if we consider, 1 1 1 is the close packed plane 1 1 1 twins can form the coherent twin boundary has got the lowest energy.

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Table 13.2 Some coincidence lattices for c.c.p. and b.c.c. crystals [21]

Fraction of lattice points on coincidence lattice	Rotations producing coincidence lattice		Closest-packed plane of coincidence lattice written as plane of parent lattice	
	Axis	Angle (°)	c.c.p.	b.c.c.
1 : 3	110	70.5	111	112
	111	60		
	210	131.8		
	211	180		
	311	146.4		
1 : 5	100	36.9		
	210	180	210	310
	211	101.6		
	221	143.1		
	310	180		
	311	154.2		
	331	95.7		
1 : 7	111	38.2		
	210	73.4	123	123
	211	135.6		
	310	115.4		
	320	149		
	321	180		
	331	110.9		

Here what I am showing it is essentially is that when we considered this sort of a coincidence site type of a lattice model.

This one is to 3 that is sigma equal 3, or sigma equals 5 sigma equal 7. For different axis over which this is the angle by which a rotation should be given so that the coincidence site lattice of this type could be generated, that is essentially what this shows.

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Table 3.2 Measured Grain Boundary Free Energies

Values selected from compilation given in *Interfacial Phenomena in Metals and Alloys*, by L.E. Murr, Addison-Wesley, London, 1975.

Crystal	γ_b/mJ m^{-2}	T/°C	γ_b/γ_{sv}
Sn	164	223	0.24
Al	324	450	0.30
Ag	375	950	0.33
Au	378	1000	0.27
Cu	625	925	0.36
γ-Fe	756	1350	0.40
δ-Fe	468	1450	0.23
Pt	660	1300	0.29
W	1080	2000	0.41

This is one plot which I had, this is one table which I have included because this table gives the relationship between the grain boundary energy and the surface energy. That is between the solid and the vapour surface. Generally, it is seen that this ratio of grain boundary energy to a surface energy transfer to be close to around 0.3.

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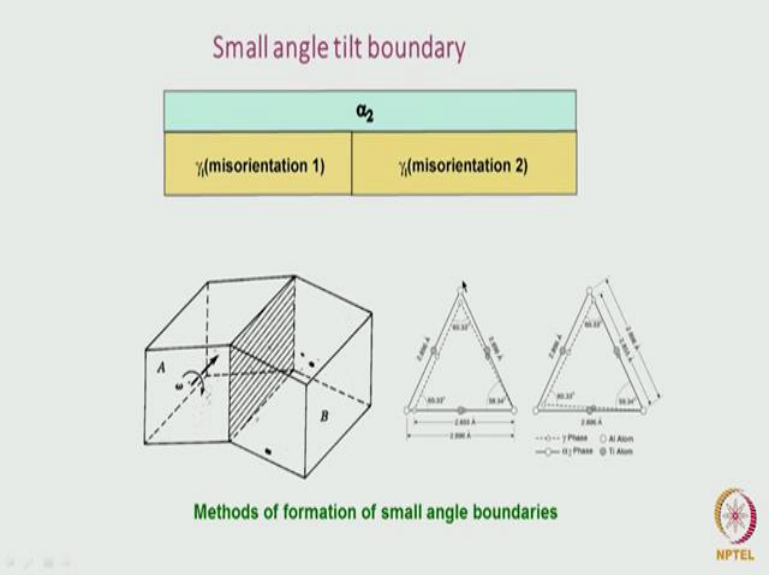

So far we have considered the tilt boundary the structure of the tilt boundary we have looked at it because it essentially made up of this location range one on top of the other. I am taking a very specific case where, your transformation which is taking place in a material where, it goes from hexagonal to a face and the hexagonal to a tetragonal system; that means, that there is some change in the c b a ratio. This can happen in FCC also. Because of that with respect to the matrix there are 2 orientations of my crystal is possible. This is exactly like this like, suppose I take that there is a tetragonal lattice is here this an another variant which has nucleated. This is the c axis. If these to come and join together since this c by a ratio assume that is slightly greater than 1 less than 1. What will essentially happen is there, the interface continuity has to be maintained when that continuity is being maintained across the interface.

Essentially the c axis will be slightly tilted not exactly 90 degrees some angle (Refer Time: 27:53), that angle will be nothing, but the misorientation; that means, from here to here, there is a slight misorientation has been introduced. That is what essentially has happened in this crystal structure.

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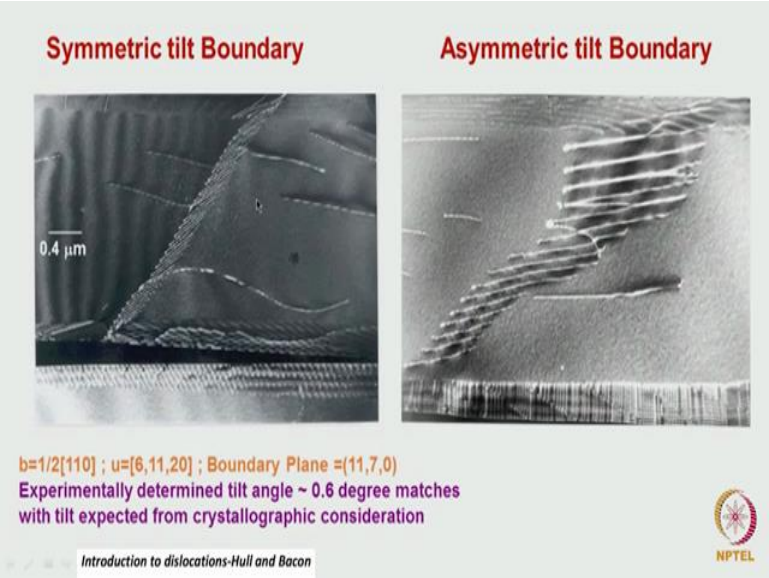

This gives rise to a symmetric tilt boundary. The dislocations of the boundary has been imaged using a weak beam technique. We have found out that the all the dislocations are the same type of burgers vector which is one characteristic which the tilt boundary will have. And the line direction is also almost that is same then we could find out the boundary plane. And from which we could determine what is the tilt angle. And this tilt angle corresponds to what is expected from the severe ratio. And this is in the same alloys one at some other region. It is not only one type of a dislocation. Now you can see that there is an another type of a dislocation is also being proceed right oriented in a different way. This gives rise to you can see that there is a tilt which there this boundary is essentially an isometric boundary, by doing an analysis we could find out what is the asymmetric angle all these details summing there.

So, essentially what we have looked at is that, from low angle to high angle, just we varying the tilt, itself we have looked at the tilt boundary. When the boundary angle is very small essentially by dislocations we could the try to model the boundary. In other cases, like using and some angle of tilt. Essentially there is a perfect matching of the lattice planes occur which are the like coincidence site lattice, model that could be for those type of boundaries essentially the energy is going to be very small.

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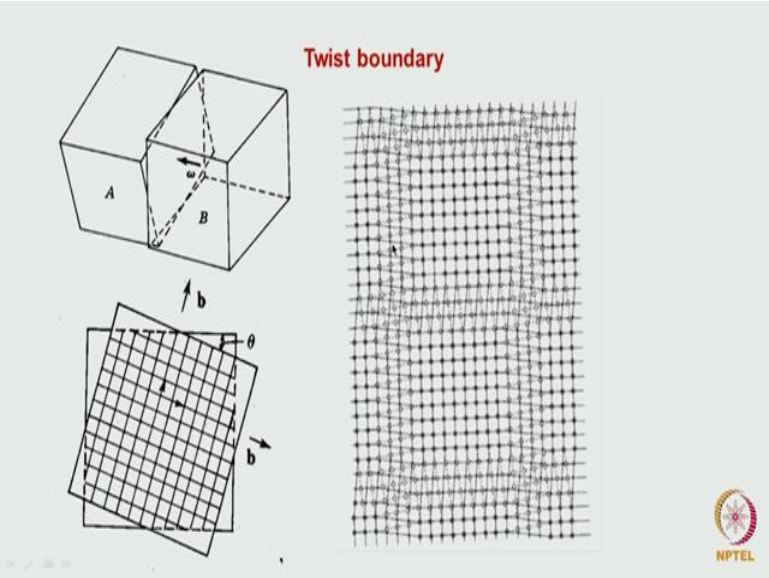

This is one type of a boundary. What is an another type of a boundary, which we can have instead of a tilt we can have a twist like for example, what has been done here is that this is a one cube and this an identical cube, but with respect to one to the other. It is just given a rotation by a small angle. This is essentially what you can see that the planes look at the lattice planes are there. If you just try to give a twist, but what is essentially important is that here that twist is very small. This is something which is similar to if I take a grid and put another grid on top of it.

Generally, we will be getting a moyero fringe pattern correct. Essentially the moyero fringe pattern will have which will correspond to similar to the dislocation network. That is essentially what is being shown here. The fringe pattern if you try to look at it, if some dislocations are generated at this region, perfect matching could be there in the region in between and by having this dislocation we could stabilize this boundary, but still maintaining the small rotate twist.

Generally, these dislocations are essentially a screw type of a dislocation, which always gives rise to a twist. Important difference between a twist boundary and the tilt boundary is that in the tilt boundary one type of a dislocation is sufficient to stabilize the boundary whereas; here one type of a dislocation is not sufficient. There should be another perpendicular type of a dislocation which is necessary that is for 2 reasons, one the long range elastic stresses has to be reduced, that is one and the other one also is that the when

you give a twist it is under torque. You should have an opposite force somewhere else to stop it that is what the second dislocation essentially is doing that part of it. So, generally these dislocations will be perpendicular to each other.

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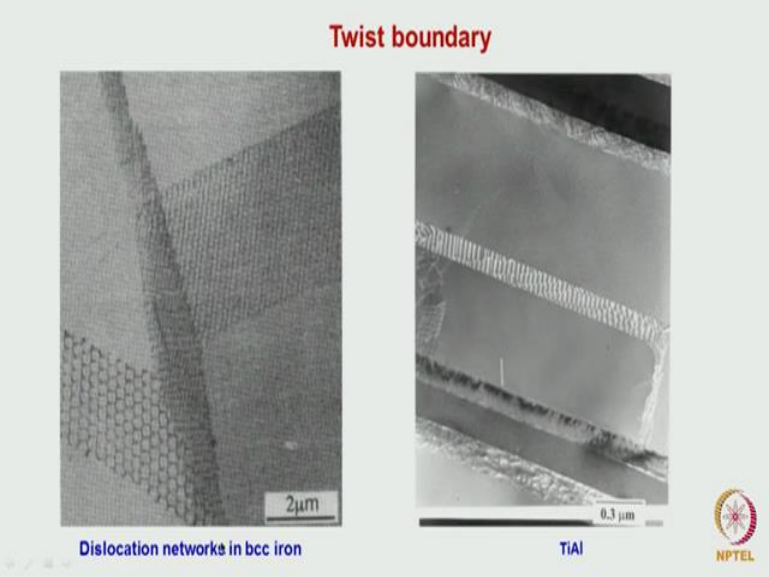

These are some of the pictures which has been taken in an electron microscope, where you can see that these are dislocation networks in BCC iron their hexagonal network of dislocations are there. This is in gamma titanium aluminium where we have observed this is between one variant and an another variant. Both of them the c axis are perpendicular to each other. And the small c by a ratio is not equal to 1 because light miss match is there. Because of that at the interface we have a twist boundary is created here

Essentially you can see here also one type of a dislocation and an another type of a dislocation could be observed correct.

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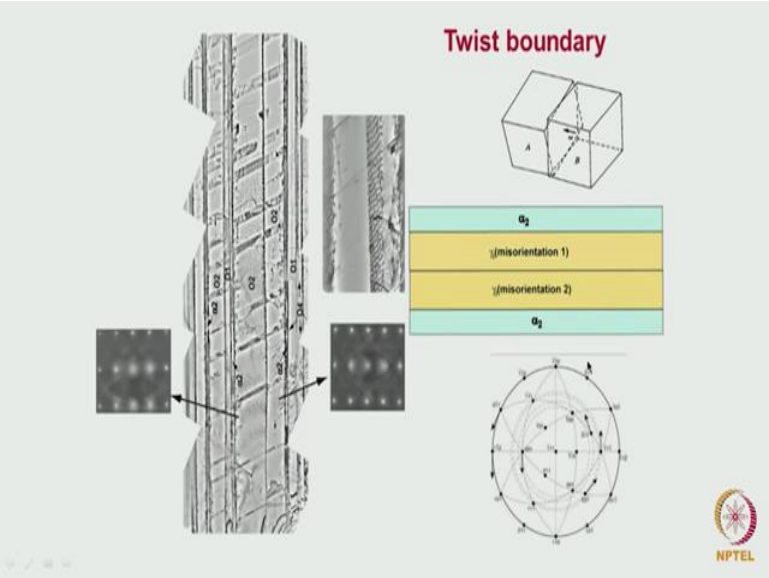

This is an another case of a twist boundary. Here what has happened is that it is the same material, but this side and this side we have the same variant.

I will not going to a detail how it happens, but finally, what is happening is that a light twist is generated. A twist boundary is generated. So, these 2 misorientation using deflection pattern are essentially using the kikuchi patterns which one can see it that this kikuchi pattern is slightly shifted because otherwise if you look at the normal diffraction pattern they appear the same, but if you look at the kikuchi pattern here, this ellipse like one it is slightly shifted from here to here. Using this we can try to find doing it a different orientation, we can find out what is going to be there tilt which is associated with this boundary. And here also what is essentially important is that in a twist boundary here you can see that 2 types of dislocations you could see one, and on this direction also another type of a dislocation is there. 2 dislocations which are almost perpendicular to each other set up dislocations are necessary to stabilize this boundary.

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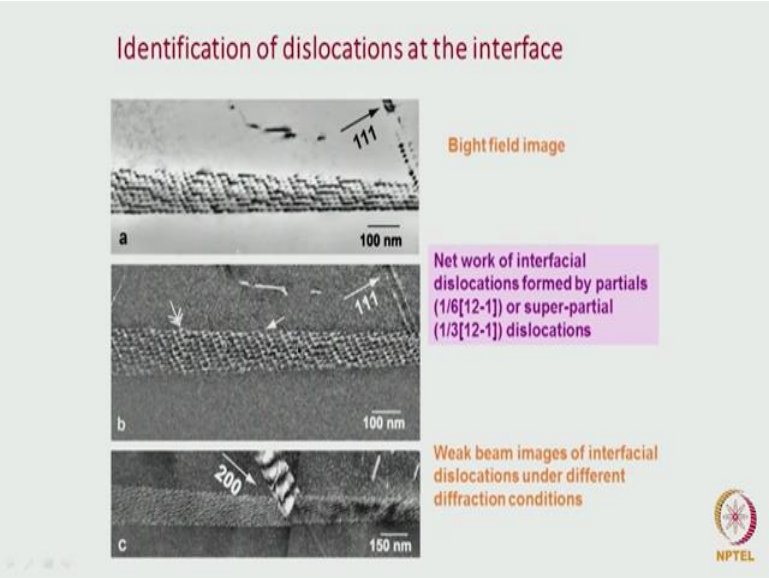

This is an another example. Here when you look at the weak beam you can see that this is one type of a dislocation. This is an another type of a dislocation get these 2 together or the ones which is form a stabilizing this twist boundary. So, far we had considered this twist boundary, which is essentially with respect to a small twist angle, but we can have a large to if we go on increasing the twist then what will happen, the same question which arises like what has for the case of a tilt boundary the same thing will happen because that dislocation networks has to become smaller and smaller whether it will be able to accommodate.

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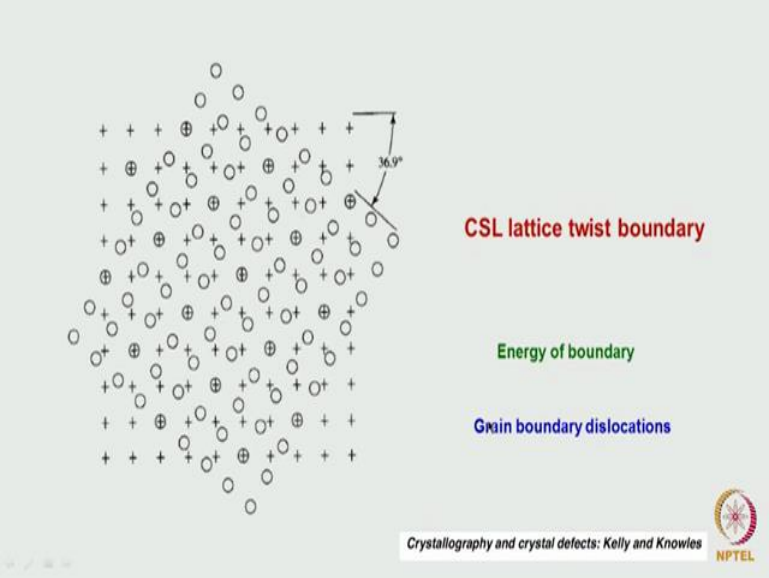

There are some conditions under which, atom positions some of the atom positions can perfectly match and these that like the coincidence site lattice the CSL lattice. We can have corresponding to this orientations, the boundary energy will be very small. Though the twist is very large that is essentially this high angle boundary could be considered either as a tilt it could form, increasing the angle of tilt or we can consider increasing the angle of twist that is what we are doing. Quite often it can happen that we can have a combination of both could be present; we are considering it as 2 different cases.

So here what is going to happen is that, in this particular case also because this is also a simple cubic lattice which is being considered, this is sigma equals 3 bound sigma equals 3 that type of a boundary here just showing a twin boundary.

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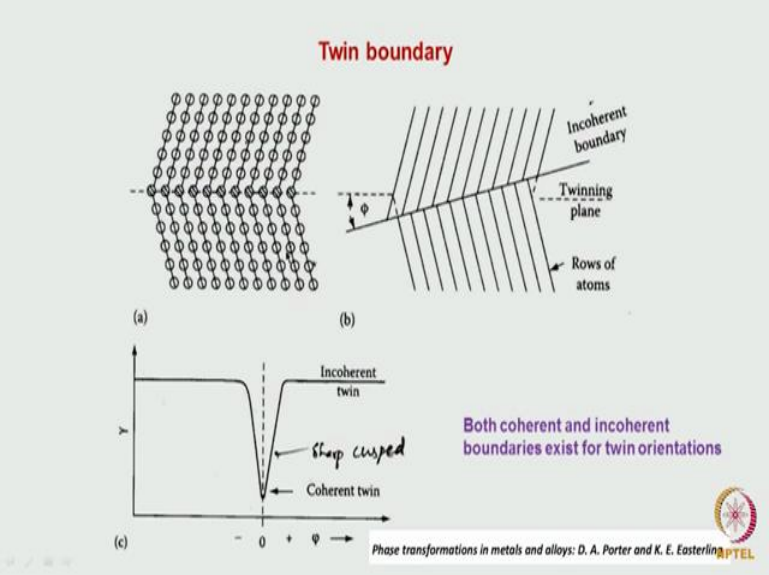

If you look at this particular type of orientation, the boundaries perfectly coherent, right. And if you look here this also orientation remains the same, but the boundary plane is a different one. This becomes an incoherent boundary. In fact, in many materials like copper, nickel, many stainless steel, where when we annealed. Them quite often one could observe within a grain. Sometimes this sort of some small stuff could be seen. This is a coherent boundary, this is also a coherent boundary, this is an incoherent one.

The energy of the incoherent boundary is very high. Energy of the coherent boundary maybe of the order of a 10 to 15 milli joule; the incoherent boundary could be of the order of 200 300 that; that means, that what is the misorientation between these 2

regions? They remain that same correct. So, how the boundary will move? Everything is going to be decided by the specific boundary the particular boundary energy, correct. So, many of the kinetics of grain growth and all are going to be controlled by the nature of the boundary and not just the misorientation. That is what is being shown here, also is that the boundary energy if you plotted with a function of a mis orientation on either side the small misorientation the energy goes up and here it is sharp decreases there.

n fact, one can see also that in a microscope if you try to see it, generally close to the incoherent boundary there will be lot of dislocations could be seen whereas, the coherent boundary will not contain in any dislocations because the lattice matching is perfect across the boundary. This if you remember about twins we have discussed, in few classes earlier. There we have considered that 1 1 1 plane is a close packed plane in a FCC that is a plane becomes that is the plane which becomes us the coherent in boundary plane.

Let us look at what will happen. So, far what we have considered the different type of boundaries. That is low angle boundary to high angle boundary correct. What is then and how we can construct these boundaries using dislocations, or for the special type of boundary where the coincidence sight model could be used. Then when this type of boundary is form, in a poly crystalline material different boundaries can have different type of energy depending upon what is going to be the orientation between the various grains, that is what is being shown here there is a grain 1 grain 2 and grain 3.

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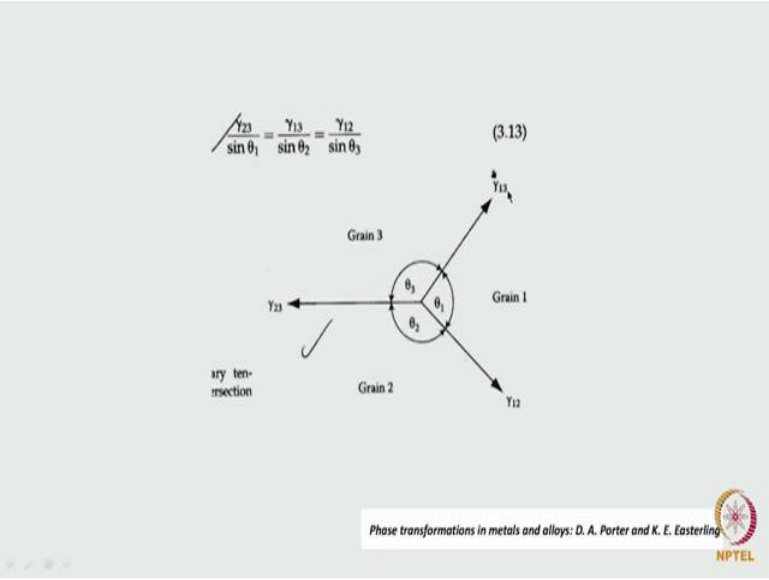

So, this is the between grain 1 and grain 3, this is the grain boundary energy between 2 and 3 this is the grain boundary energy gamma 2 3 here, gamma 1 2. If that is the case the angle between this boundary this easy to derive, and only just giving that formula, that is the gamma 2 3, divided by this sin theta 1, 3 divided by the opposite angle gamma one 2 divided by obtuse angle, which is sin theta 3. This have to be equal. That is because it is an energy balance which has to be done.

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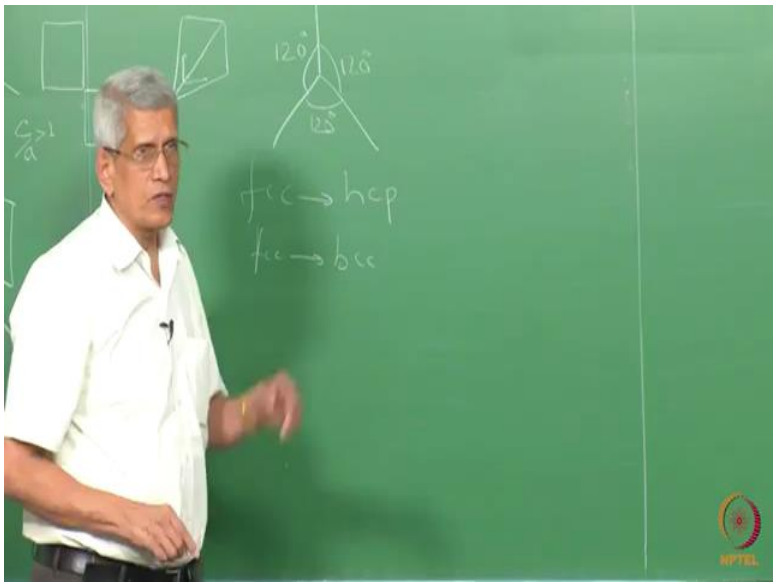

If the boundary turns out to be the same type of a boundary (Refer Time: 41:29) the equilibrium condition then, what will happen is that this angle theta will turn out to be 120 degrees, that is where we will have a boundary that equilibrium all the boundaries have to be the same type this angle r will be 120 degrees. And in fact, we can find out the grain boundary energy from the grooves which are created, when we etch the sampling that is thermal etching chemical etching you can do that here. So, an example which is being shown is that this is the solid vapour boundary which is there on either side and this boundary the gamma s v is the energy this value be for most of the (Refer Time: 42:24) materials we have this value which is available.

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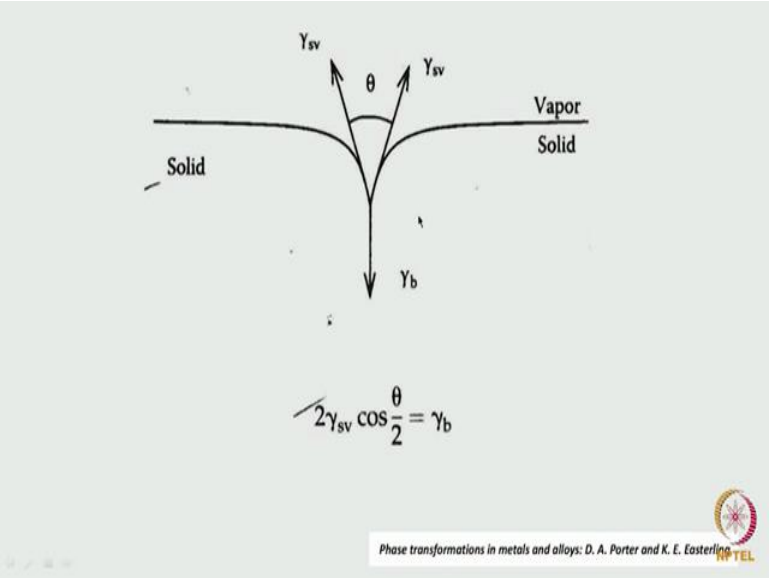

Then depending upon the angle which it makes with respect to if we drop this angle theta is you know then we can calculate. This expression could be derived. These essentially that balance of equation which we have written. 1 2 and 3 this boundary and then we can find out that grain boundary energy depends upon twice gamma s v 2 cos theta by 2. This way we can find out the energy of the various types of boundaries could build it. These are all experimental determination of grain boundary energy.

But generally calculations have grain boundary energy computationally, there are lots of difficulties are there. Because there is a discontinuity which comes at the boundary atoms are not at equilibrium position across the boundary because the add the separation between the atoms does not correspond to equilibrium separation, but it could be values which are different from that calculating what is going to be the energy is an involved process.

In fact, if one looks at the literature hardly much data is available, on computational way of looking at the grain boundary energy and minimization of the grain boundary and look at what is the type of a grain boundary structure which the material will have lot of experimental work is available.

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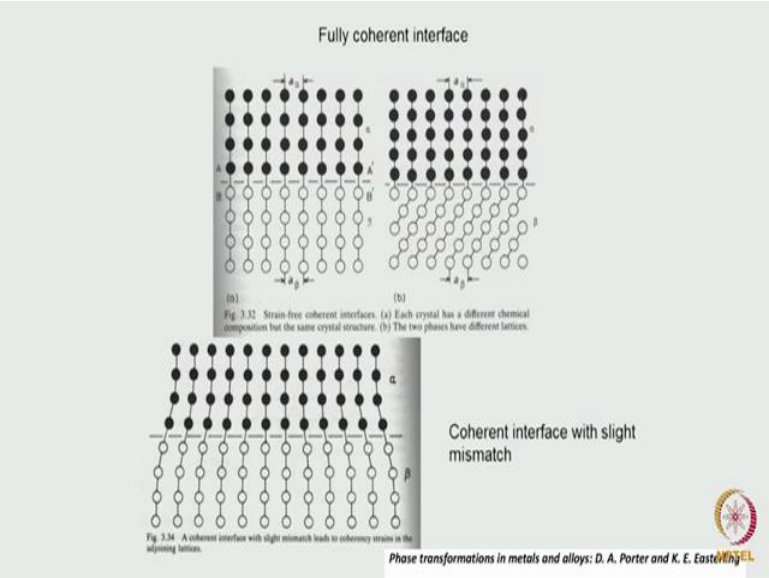

So far what we considered was essentially with respect to a single phase right, that small angle boundary and high angle boundary and special boundaries. These are all the type of boundary and depending upon the energy between the various types of boundaries what is the angle which will make between the boundaries. Other type of a boundary which we have to consider is essentially is when we have precipitates are there which has a different composition compared to that of the matrix.

Suppose the precipitates and the matrix has got the same crystal structure, the example which we can take it is copper and cobalt. Both phases being a FCC, or assume many condition many types of precipitation like this can be done well where the composition is of the matrix is different precipitate is different, but the crystal structure is the same. Then in such cases we will find that the lattice continuity is being maintained across the interface. If the lattice parameter is also turns out to be perfectly matching, then what is going to happen is that coherency in this particular case will be 0 correct because if the lattice parameter is the same it is perfectly coherent, right that is one case, but still the interfacial energy will be there because chemical composition here and the chemical composition is across this may be going to be different. So, interfacial energy will come.

This is an another case which we are showing, where the assumed this is the matrix. This matrix is inclined in other words, but across the interface if you look at it that is a matching is there perfect matching is being maintained correct. Suppose the lattice

parameter is different. In these particular cases because tomorrow when we talk about the interaction of dislocations. Suppose a dislocation to enter from here to here what will happen? Because there are some small difference the modulus will be different they will have an effect on the behaviour of the dislocations, especially on the moment of dislocations.

This will discuss later tomorrow. In this one if you consider it the lattice parameter is essentially slightly different here the lattice parameter is a, small and here there is a slight increase is there, but the parameter difference is. So, small that still it maintains the coherency; that means that how it maintains the coherency, by distorting the planes on either side of it. How much is the disruption which is going to be is determined by the modulus of the material on either side will decide to what extent the distortion is going to be accommodated whether within the precipitate or within the matrix correct.

And suppose this grows to a very large size; that means, that the total strain also increases at some particular point of strain when the strain becomes very large it may not be able to accommodate that total strain. Then it punches out some dislocations makes the interface incoherent.

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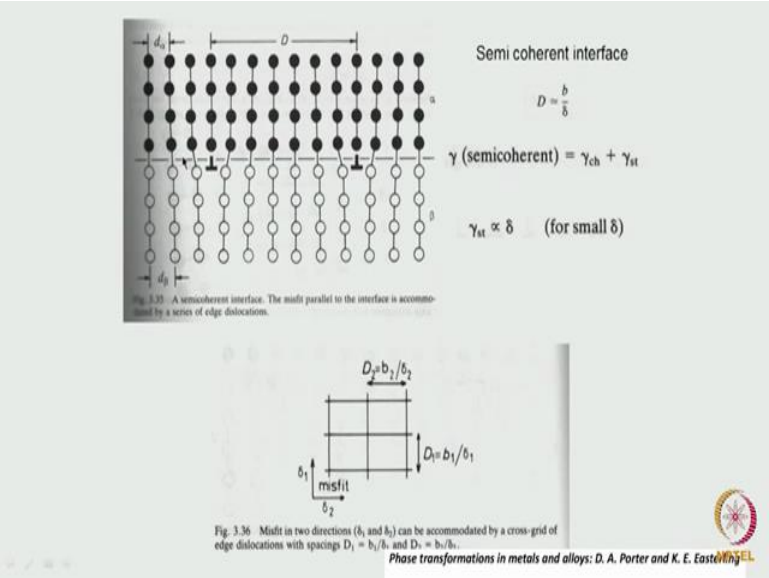

This can happen in a different way also. If the coherent is strain is large also even at small sizes of the second phase particles. When it punches out the dislocation like this at

the interface these are called as misfit dislocations. They all are the burgers vector which is actually and the interface plane.

Now, if you look at it earlier all the planes where distorted around the interface. Now most of the planes they maintain a perfect coherency only around that dislocation this is distortion is there correct. So, the coherence is the strain could be reduced. These type of interfaces are called as semi coherent interfaces. And if we know what is going to be the strain misfit strain, then we can find out what will be that b by (Refer Time: 49:18) what should be the spacing between the dislocations which is going to come at the interface that also we can calculate it.

So, for these boundaries if we look at it, the semi co this will have a chemical energy also is going to be there free energy chemicals. Plus, the strain energy is also going to be there. Both of them together will con decide what is going to be the total energy of the interface. And another thing also which one should understand is that as I mentioned.

The smaller that delta, what will happen? The dislocation the distance between the dislocation becomes very large. If the precipitate size is smaller than this, then what is going to happen no dislocations will be punched out? And essentially it will be maintaining a coherency. That is why when the precipitates are very fine, most of the time the coherent. And if they are coherent and they maintain coherency they have a spherical shape also associated with it right and as the precipitate grows one the morphological change takes place. Not only that add some particular sizes.

The dislocations will be punched out in at the interface which we call it as the misfit dislocations. This we can calculate also depending upon what is going to be the misfit in the 2 direction assuming this interface to be considered as a simple cubic type of a lattice. One can find out what is going to be the misfit in these 2 perpendicular directions which we can determined the misfit we can find out and once we know the misfit we can find out what will be should be the separation between the dislocation also.

But these dislocations which are generated one should understand that they are different from the normal met dislocations within the matrix. Because the matrix dislocation are this side, that they can move this dislocation need not be (Refer Time: 51:44). Those dislocations are generated just to reduce the strain. Suppose my interface itself that is the lattice parameter itself is quite different. Crystal structure could be quite different

between these 2 faces composition also could be different then the interface could be and incoherent the face.

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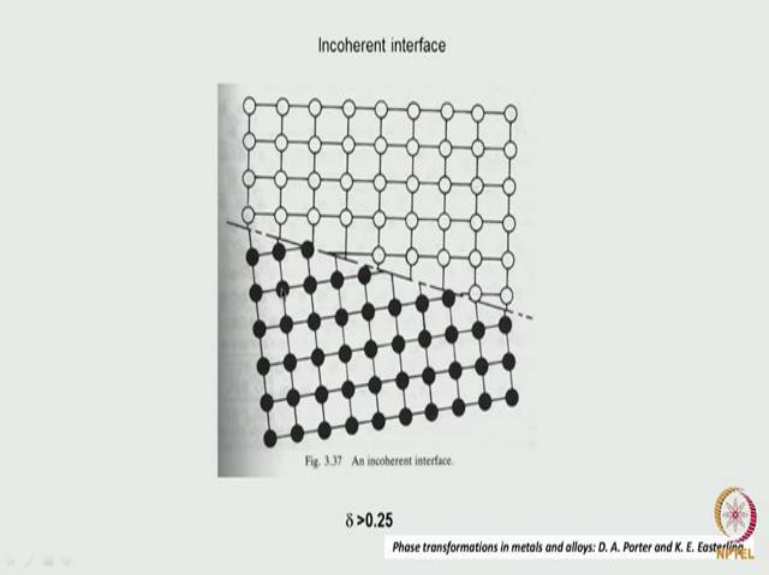

So, in an incoherent interface what happens is that no strain is being accommodated right. Up to the end up that interface the perfect lattice matching is being maintained on either side of the crystal, whatever the lattice parameter, which is there here the same thing we will be seeing right up to the interface. Because there is in the case of coherent is strain there is a lattice parameter is changing because they wanted to maintain the continuity across the interface correct.

Generally, when the misfit parameter is greater than 0.25, when we say that these are incoherent interfaces. All these type of move example what we can take it is many of the carpets which form a nickel based alloys they generally are incoherent. And quite often we can see that the perfect interface could be seen between the matrix and the precipitate. Sometimes to accommodate the strain what they do it is some dislocations are punched out. Because the volume strain there will be a difference between the 2 faces to accommodate the volume strain some dislocations are punched out into the matrix.

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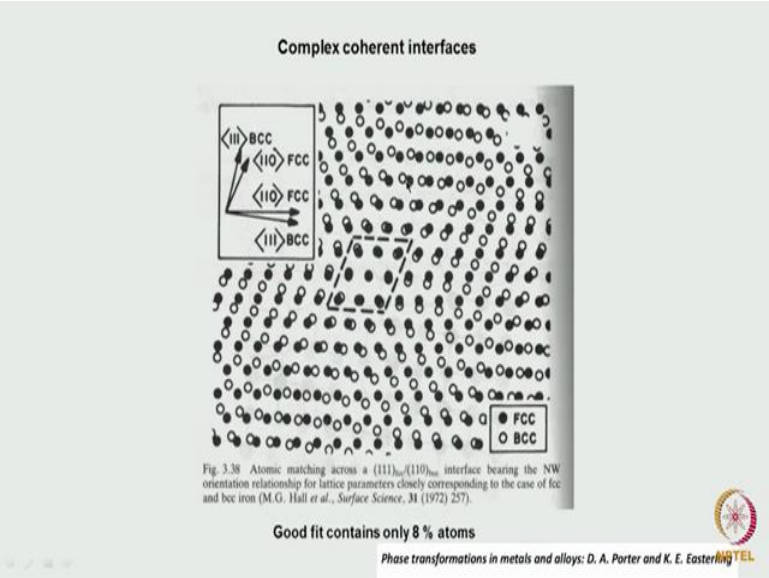

This is another type of a complex coherent interface which are seen in many materials. One of the examples which we can take it is let us take the case of FCC to HCF. The phase transformation if you take place. If the lattice parameter remains, the same FCC has got the hexagonal structure. And HCP also has got a hexagonal structure that matching can be perfect across the 1 1 1 plane correct. It can maintain the orientation relationship. The other example which we can take it is, in iron it is from BCC. What is the close packed plane in FCC? It is 1 1 1.

The close packed plane in BCC is 1 1 0, but if they form with these planes, it is not going to be this is the best plane where the interfacial strains could be reduced to a minimum, but it is not completely. Because of that in these sort of interfaces this called as a complex coherent interface here. If you look at the diffraction pattern there is a slight change in the orientation between the close packed direction in BCC, and the close packed direction in FCC correct that is what is being shown here when these 2 are forming with the 1 1 0 plane of a BCC parallel to 1 1 1 plane of FCC.

These sort of relationship this small relation, this could be seen in the diffraction pattern itself, from which we can identify looking at the diffraction pattern taken from this region we can get information about what is the direction in which the close packed plane in FCC is oriented and what is the direction in there, close packed plane in 1 1 1 is oriented close packed plane in BCC is going to which is 1 1 1 this information, one could

get it. So, far what we have considered essentially is that, we try to get a simple idea about the grain boundary stretching.

Consider single phase material, what are the types of boundaries which we can have both small angles as well as the high angle boundary. What is the type of structure which this boundary can have both for the small angle boundary as well as the high angle boundary? And what all a special type of boundaries like coincident site lattices which we considered it. And we looked at the special type of boundary which forms in FCC and many other structures like twin boundaries also we considered it which a special type of a boundary.

That is as far as single phase is concerned. If precipitation takes place the precipitates could be either fully coherent or incoherent or semi coherent. If it is fully coherent, the perfect matching across the boundary is there. Only way we can identify them then the boundary will be very sharp, when it is fully coherent these boundaries in electron microscopy transmission electron microscopy we if we look at it when it is perfectly coherent there will be a sharp interface will be seen between the matrix and the precipitates.

If it is semi coherent are highly coherent then what is going to happen is that coherency means that the planes are distorted; that means, some strain in has taken place there on the interface. Whenever straining is taking place that will appear as a contrast variation which will be gradually changing; that means, that we will in a coherent precipitate we will not be able to identify the interface sharply.

In such cases to identify the procedure to get the precipitates size we have to use dark field technique to image the precipitate separately. There are many things like the nature of this mismatch between various types of along various directions in the precipitate. They also do control the morphology of the precipitates. Those aspects we have not gone into the detail that becomes part of the phase transformation work. Essentially here what is essentially necessary to define the various types of interfaces?

So, we have a nil that is grain boundaries, small angle boundary, tilt boundary, twist boundary, coherent boundaries we have considered, special boundaries which we have considered. Interface which we have looked at in some few simple examples which we have considered, this is all which I wanted to tell about interfaces, but actually this whole

interface if it has to be covered both the energetics and all aspects of it. We are try to give a very over simplified picture of the boundaries right.

At the beginning I mentioned that it is are the boundary itself could be a curved one there could be some precipitation could be there a lot of complexities associated with the boundaries. At this point I had just deliberately omitted those who are interested can look into various books which I had given some of the references that will give some information about it. We will stop here now.