Electron Diffraction and Imaging
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Lecture - 21
Contrast from Strain Fields

Welcome you all to this course on Diffraction and Imaging. In the last class, we looked at analysis of defects, especially defects which are planar defects, what all various types of defects, the type of defect vectors associated with them, what sort of contrast they will give in the microscope.

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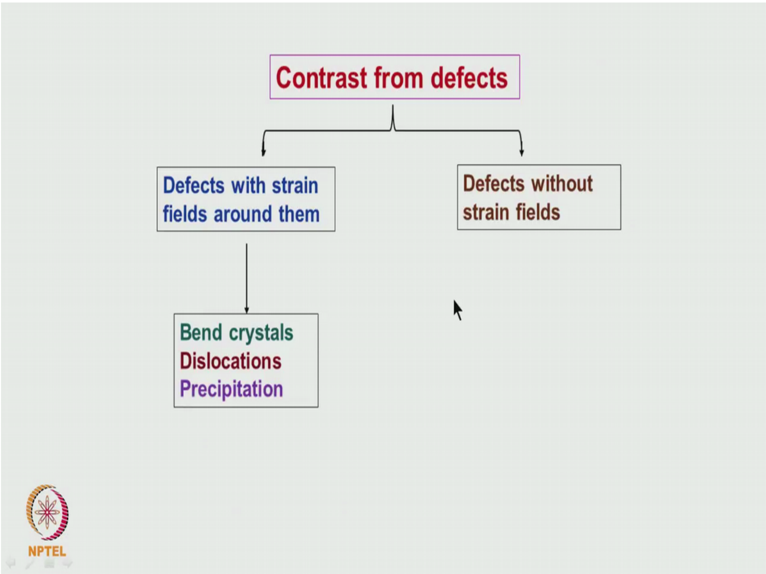

As I mentioned earlier there are two types of defects. One is defects with strain field around them and the defect just brings about only displacement without introducing any strain field. Today we will talk about defects which have got strain field associated with them, what is the sort of contrast which they give, how to analyze and interpret the results, right.

So, the type of defects which we mentioned are one is Ben Crystal itself. Suppose it is bent, that is one and another is some defects are there which causes either bending or shearing of planes. That is one that is what we call dislocation and there are coherent precipitates which are there in many materials. When this coherent precipitates are there to maintain the coherency between the matrix and the precipitate, some bending of the planes have to take place close to the interface. That also gives raise to some strain in the contrast. So, these are all the three cases which we will try to consider and then, these defects could be especially dislocation, they are imaged using a technique called as weak beam technique. That is also I will mention.

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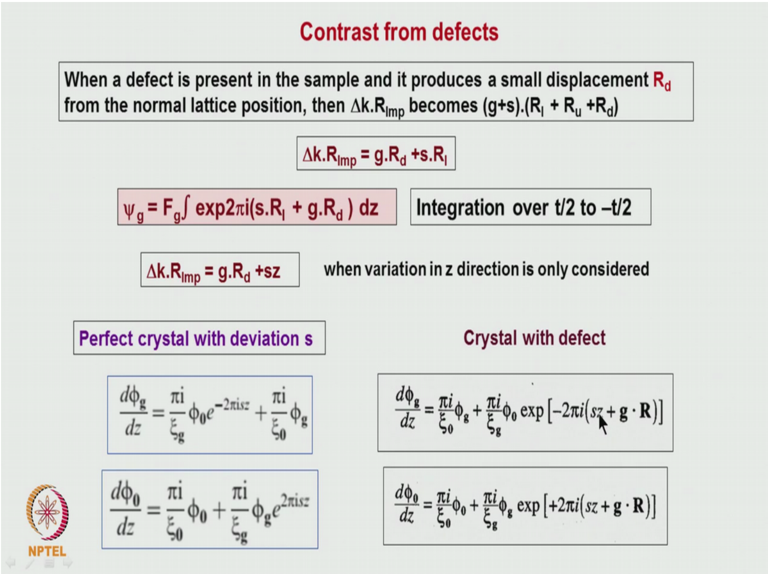

Essentially, what is considered that when a crystal is there in the phase factor, it adds one more term that g dot R correct and R turns out to be in the case of stacking fault which only a planar defect vector whether which we have to see it, but what happens in the case of a sample, ok which is bend where that distortion is taking place.

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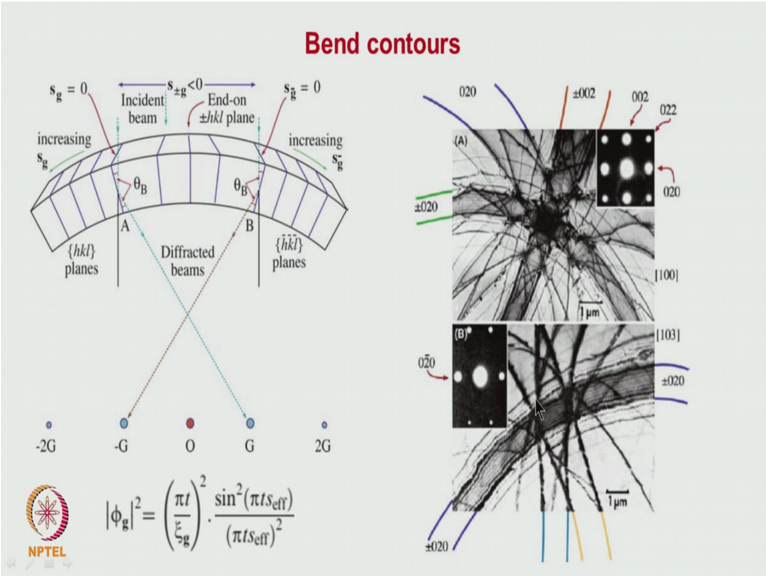

Everywhere the simplest case which we can consider that is a bend contour, you look at here, this is a sample which has been bend. When the sample gets bend, you assume that the sample is a straight sample planes are like this.

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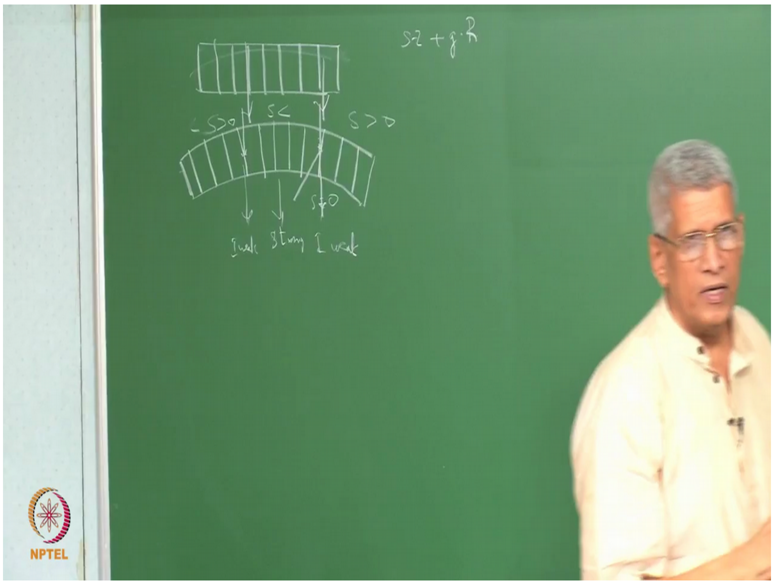

If we bend this sample that lattice planes itself gets bend, right. If a parallel beam is entering, then what will happen the regions which are very close to it, they need not satisfy the Bragg condition since the planes are bend here at this region as well as in this region it may satisfy exact Bragg condition. If an exact Bragg condition is satisfied from those regions, the strong diffraction or scattering is occurring. What will be the effect of it on the incident beam? In these directions, the intensity is going to be weak, right. That is a strong scattering.

Similarly, from here also intensity will be weak in this region. The intensity of the transmitted beam is strong, then we will find that when we go further away also, the intensity will be increasing it. This itself can be considered in two ways, that is mathematically if we have to use it in the expression, what is the way in which we can consider either the atoms are displaced. That means, the term here it comes s into z plus g into R. This is the phase time comes, right that is R is displaced, g remains that same. That way one can consider or other way in which we can consider when the planes are bend, the separation between the planes has not changed. So, this can be considered as bending of the planes is equivalent to using an additional shift from the exact Bragg condition, right. So, that is as adding s. So, in this region s equals 0. Here s may be positive, s may be positive here, here also, in this region s could be negative. This sort of way also we can consider because as the sample has been bend, we see that at some condition when the exact Bragg condition occurs at that condition, the deviation from the Bragg condition is 0, correct. On either side, it increases. That way also we can consider the strain itself.

If we do that, then we know that the intensity is depended strongly on s as a function position. Then, you can see that you see here, there is one line which is dark and another intense line which is dark on either side. We get some fringes also, correct. Why do this fringe comes? Tell me, what is the cause for these fringes to come? It is because the sample thickness has not changed here. What is it which is changing then?

Student: Lattice.

That is also not changing, but as a function from distance s is changing, now when if s changes t into s can oscillate because t into s depending upon how it become the value becomes integer or a half integer depending on the intensity can fluctuate in a value of psi g can be 0 and maximum it can fluctuate. So, either thickness changes or if s changes, also it will be effectively the same. So, that is why we get some fringes also. Here you understand that this sort of contours, these are called as bend contours which is very common in metallic samples which are prepared using jetting technique, especially if it is a solution treated sample, where the sample is not stiff.

These sort of contours, the problem of this contour is that imaging becomes you can see that so many lines are coming, it is very difficult to see a very clean image, but dislocation can also give raise to many defects, can give rise to this sort of a contrast. How we can differentiate it? Suppose I tilt the sample, what will happen is some other region of the sample will exactly satisfy the Bragg condition. That means, this band itself can move within the sample by tilting the sample whereas, if a defect is present, the defect fixed in a particular region. In the sample even if you tilt it, the defect will not move which is fixed in the position. This is the way one can differentiate between.

Student: We are focusing the beam of dislocation the beam can.

No no no, beam is focused on the sample.

Student: But dislocations can get activated by the beam.

Which one?

Student: Dislocations.

True, generally the beam current which we choose is such that they need not, but normally that is not. It happens essentially in a sample that if a sample is deformed and you have dislocations which are there on the sample, there are some stress fields. If I focus the beam onto a very small region, what will happen in that region? It will raise the temperature. When the temperature is being raised, it is equivalent to relaxing that stress and then, the dislocations will start moving. That is different from that because that is the beam is being focused. Dislocation moves here and the beam is kept constant. We are tilting the sample. When we tilt the sample, this image shifts. You understand this is the difference between these two.

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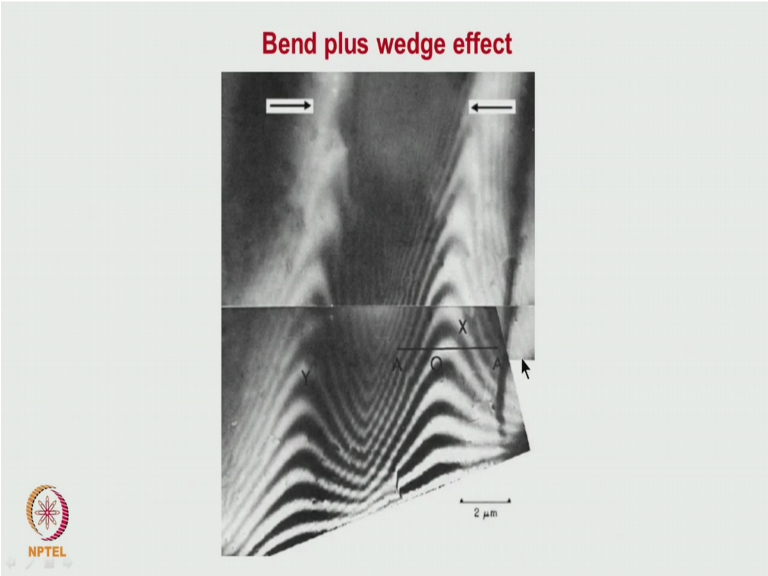

So, here it is another case where you are just showing an example taken from the literature. You have it from the edge of the file. It is a wedge shaped of a file. So, that when it is a wedge shape is there, that is going to be a fringe which will come which I had explained in addition to it. That is a bending is also there. Both effect are simultaneously present. Then, this sort of being which it will appear, so when you look at a sample unless that sample is perfectly flat and strained free, otherwise you will not get an ideal uniform intensity. There will be a lot of variations. There are new methods by which this sort of defects could be avoided. We will talk about in the last class when I come to.

So, bend contour when you see the bending causes a strain in the sample right, but the strain is spread over the full volume. It is not confined to any particular region. The same thing happens with the dislocation, correct when it is there even though dislocation core of the dislocation is there surrounding it.

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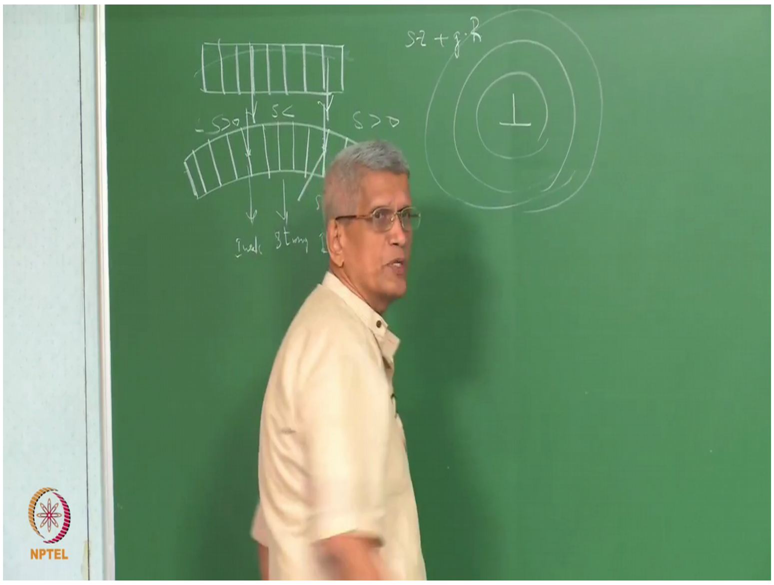

If a dislocation is being present surrounding this, if you see there is going to be a displacement which is going to take place over the large area, how it is going to effect in that. So, in a planar defect, it is only a defect which is only in a particular direction. Here that is not the case. How this has to be incorporated in finding out the intensity that is what we will discuss for the next few minutes.

Before that what are we looking for in dislocation, suppose we have deformed a sample? There are many types of dislocations. We wanted to identify what all the types of dislocations which are there if you take FCC itself, there are perfect dislocations. There can be six types. What is a perfect dislocation first tell me? No, the metallurgists will know. Others what do you define as perfect dislocation, mechanical engineers that is if the burgers vector of the dislocation is equal to your translation vector in the lattice, we called that dislocation as a perfect dislocation, right. If the burgers vector of the dislocation is not a translation vector in the lattice, then we define them as partial dislocation. Is this clear? That means that is a simple example.

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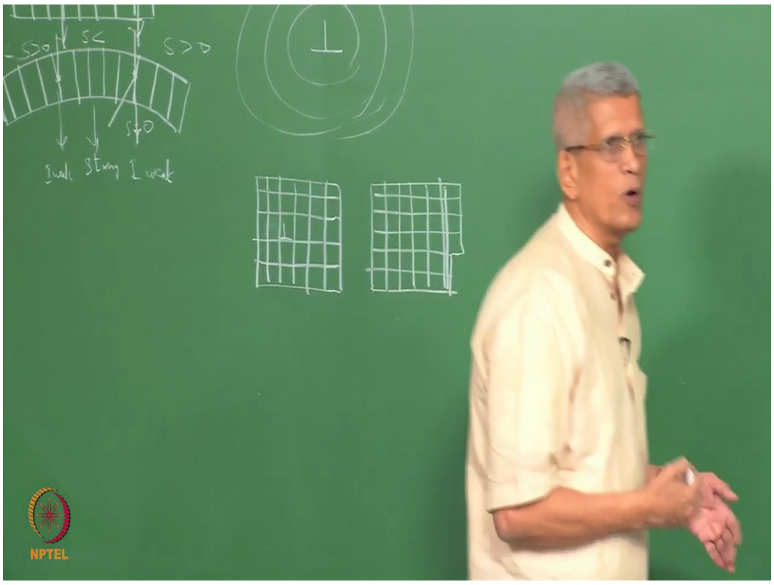

Suppose I have a sample. I introduce these are all I will draw it like this. This is a unit cell which I am showing it. Suppose some dislocation has been introduced. If this dislocation comes out of the sample, what it will do? I may not have drawn the planes exactly. Now, you see that after the dislocation has come out, only one step has come outside, but otherwise the crystal structure looks identical, right. All the lattice planes are there. So, that means that what is the distance by which the dislocation has moved that atom in the slip line corresponds to shifting atoms from one lattice point to another lattice point. So, this sort of translation vectors are called as perfect dislocations. The partial dislocations are the one which give stacking fault that will come later. So, we wanted to know all the characteristic of a dislocation. What all characteristics of a dislocation we should know? What is the burgers vector? What is the line direction of the dislocation? These are all the information which we require about a dislocation, correct burgers vector direction and its magnitude and the line direction.

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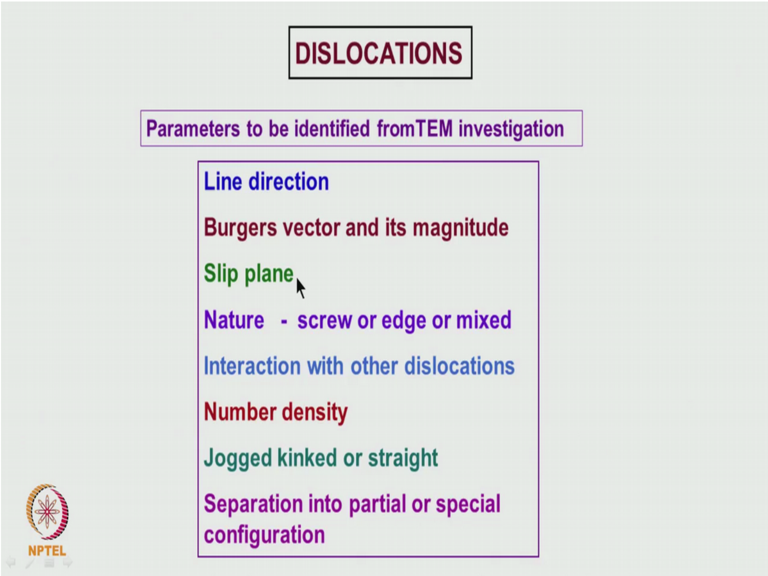

Then, which plane is this dislocation moving, these information is required and if we consider in FCC, the closed pack directions are essentially a by 2 1 1 0 type.

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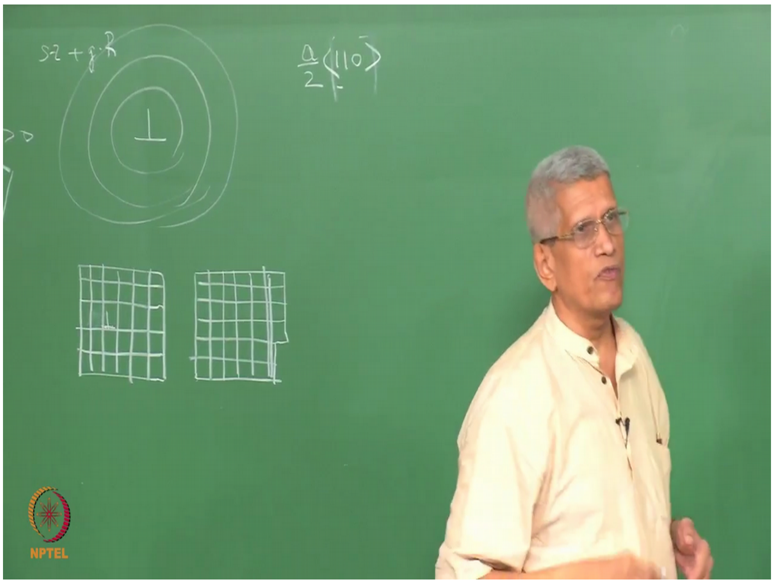

We can have six type of directions can be there. They can have different line directions. So, all these dislocations could be produced when we deform the sample. These dislocations will interact with each other and produce some dislocations which are called as sessile dislocations which does not move in the material. So, this sessile dislocations are the ones which are responsible for increasing the work hardening in the material. So, understanding the type of dislocations, how they are arranged, all this information we require. Sometimes the dislocations could be jagged or the dislocations could split into partial. All this information is required. These are all what we wanted, ok.

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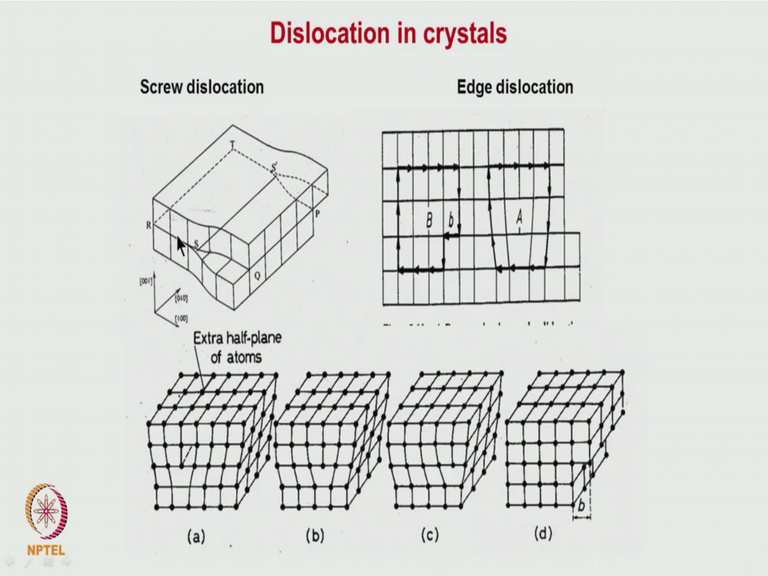

So, how do we define a dislocation? That is in the case of an edge dislocation. You see here that is one extra plane is there. If I take burgers circuit around it that is what is being shown here around a region which contains an extra plane and a region which does not contain an extra plane. You find that after the burgers circuit has met, that is here the burgers circuit is complete, here the burgers circuit is not complete, there is a vector gap is coming. This gap we call it as burgers vector of the dislocation, right. This is how we define it. The same way we can take burgers circuit around the region, where a screw dislocation is concerned. We will find that the burgers circuit does not close in a direction which is under line direction of the dislocation and the magnitude of that burgers vector turns out to be a by 2 1 1 0.

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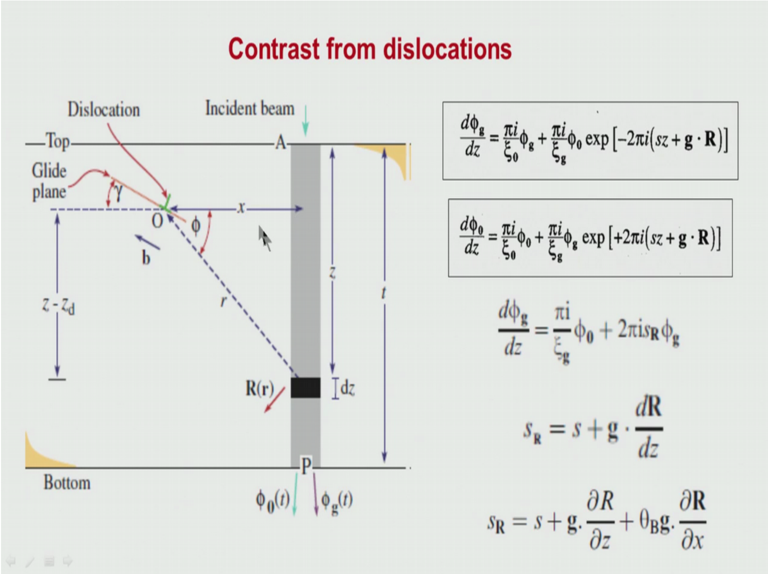

In the case of FCC, this one can find out for different crystal lattices, what the burgers vector of the dislocations will be, ok. This is how the dislocations are defined and the burgers vectors are defined. Is it clear?

Now, suppose we assume that this is the sample top and the bottom of the sample and a dislocation is present here. As I mentioned, the dislocation has got a strain field at different points, right. So, all along around this dislocation, it is going to generate a strain field around it, correct. So, when we calculate that intensity at every point on the sample surface x y, how you have calculated the intensity at this particular point x y, how the intensities varies as we move along and try to find out in the transmission direction, how the intensity has changed if we are trying to do it, we find that in these region. It will be some displacement will come lattice positions. This is i. If you remember I showed that this called as a column approximation which we use that size of the column is about 2 nanometer, where we try to calculate that intensity. So, at this position, there is some displacement. So, overall within this column because the dislocation is there, there is going to be displacement of atoms which is going to vary along this column like that it is going to happen everywhere. So, average of all these displacements which will cause the intensity which we have to take it to find out what is going to be the total intensity at the back of the sample.

This is how we have written that expression. Now, what happens is that this R, we could define it in a very simple way as a vector. Now, R could not be defined as a simple vector. That is what we will come to. Now, what is happening is that this s z plus g are written as an another deviation. If we write as this one, then this s plus this g dot what we are writing is d r by d z. What is d r by d z? It means that displacement at any point here, this is vector R and how it is going to be at this particular depth. So, that itself is going to change, right. So, this is the way now we have to find g dot. Instead of g dot R, we have defined g dot d R by d z, where R is the fault vector at that particular point along the column. In fact, strictly speaking what happens is that it is with respect with this column if I move in the same direction along this direction, if I move then also when we go further way from it, the displacement will change. That will give an additional term d r by d x like that it will be defined.

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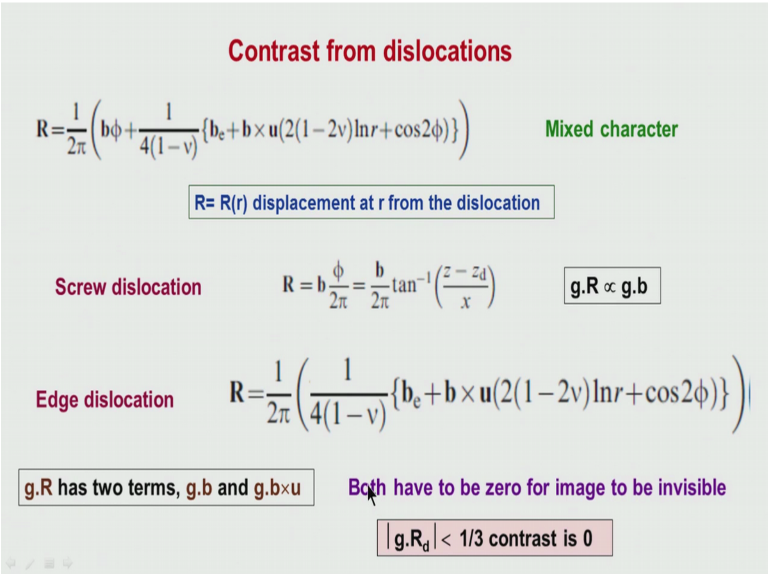

So, essentially that is the way R is defined. How is that R defined? Normally in the case of dislocation, R is defined as when we have screw dislocation, R will be v by 2 pi into theta. We take it to be a cylindrical coordinate, right.

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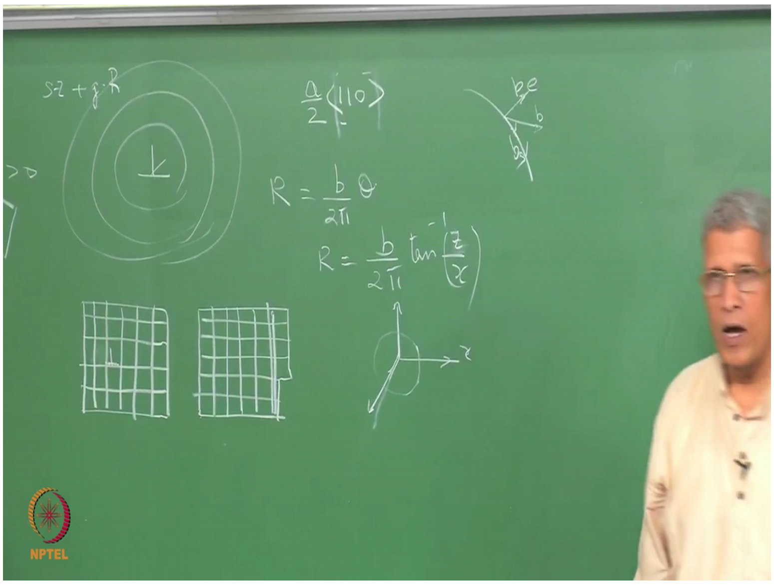

What will be the value of theta? This angle by which the cylindrical coordinates, otherwise in x and y coordinates if we represent it, R will be equal to b by 2 pi tan inverse. In these cases, it will be z by x. It will turn out to be that is if this is x z is in this direction and this direction is y and if the burgers vector is in this direction, then if we move around like this, how the displacement will be taking place. Is it clear? If this is the line direction of the dislocation, if I start from one point if a take burgers circuit clockwise if I move, like this every distance which I move, the atom is also displaced slightly in this direction as one circuit is being 1 2 pi. We complete and reach this point, the atom which was there original position, it has shifted by burgers vector, right. That means, that every atom is getting displaced.

Essentially this is what is a one general dislocation. That is what we call it as mixed dislocation, where it has got a screw component and an edge component which is associated with the dislocation. Then, R will b 1 is b s into 1 phi plus two terms comes one corresponding to the burgers vector of the edge dislocation one more term and another term which comes is corresponding to b cross u, that is line direction cross product of b which will come that is for an edge dislocation is going to be there.

So, this is how the displacement is defined. So, for a dislocation which is a mixed character, you have studied all the burgers vector of screw dislocation and edge dislocation mixed dislocation. No, see suppose a dislocation is bend like this and this is the line direction. Suppose the burgers vector is in this direction, it is a perfect translation vector. How you can find out the components if you take a tangent in this direction and perpendicular to it. We can find out a component of this in this direction and a component in this direction, the component which is perpendicular to the line direction that is the edge component, the component which is perpendicular in the line direction is going to be the screw component and with respect to this burger vectors, one can calculate what is going to be the strain field around them. That part I think has been covered in defects in material course that I am not taking it up here. So, that means that similarly for a edge component also, we can define like that strain. So, both if you write it together, this is the way an expression will come. This you can look into books where all these derivations are given.

Now, let us look at the case, where it is screw dislocation, yes.

Student: What is v?

Student: bd.

Bd is the burgers vector of the screw component in a mixed dislocation. When it is there like here, if we take this component, this will be and this will be bs.

Student: b cross u is 0.

B cross u is this burgers vector. This is inclined with respect, the line direction is b cross u. I will just show you it is important in a movement for a screw dislocation. What happens is the edge component is 0. So, it is only this factor b will come, the first term only will come. So, this can be defined in this way, ok.

What is essentially important if we look at it is that R is if we know the burgers vector of the dislocation, R can be written as proportional to b by 2 pi. The others depend only on the distance correct, this tan inverse factor. So, this g dot R is proportional to, we can write it as g dot b plus this term, this is a term which is going to be Re if we write it this way. Suppose you assume that for the dislocation, the burgers vector of the dislocation is such that g dot b becomes 0, then what happens?

Student: Dislocation is not visible.

The dislocation will not be visible, though this term is going to be there that g dot b equal to 0 is sufficient to tell that the dislocation need not be there, but when that dislocation is visible, then not only this b, this full term is going to contribute to the intensity, right. Is this clear? So, instead of writing g dot R, we can use g dot b also as a criteria to image dislocations, that is we choose g such that for some gs, we find that the dislocations become visible. If we know what all the gs which we used to find out, the dislocation become invisible if we take the class product of that and that will give information about the burgers vector correct, ok.

In an edge dislocation, what happens is this term, the burgers vector of the screw dislocation that becomes 0 because that does not exist because the dislocation is then this term will be there b e plus b cross u is also going to be there and in this case, if we take g dot R, there are two terms, g dot R equals 1 is that g dot b plus g dot b cross u. Both the terms will be there. So, if the dislocation, if g dot b equals 0, but b cross u is not equal to 0, then the dislocation still be visible. You understand that? That is a criterion for a edge dislocation. It is just not g dot b equals to 0 alone is not sufficient. So, that contrast is called as a residual contrast. I will come to in a minute and normally we know that when g dot b value itself, we said the defects vector if it is less than 1 by one-third as such I had mentioned earlier that when we look at the phase factor, the value turns out to be very close to 0. The contrast may not be visible. That is also the condition which is going to be there.

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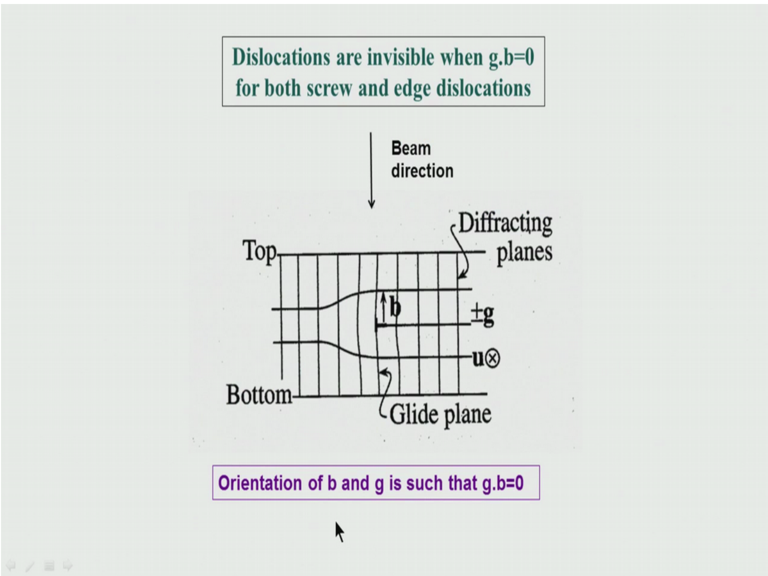

So, we have to use values, such that this term g dot R d is sufficiently high, not very small. Is this clear? So, the condition essentially we use here is just g dot b equals 0. Instead of g dot R for a screw dislocation and for an edge dislocation, not only that g dot b equal to 0, g dot b cross u also has to be 0.

You look at what does it physically mean here. I had shown an edge dislocation burgers vector is parallel to beam direction. These are all the one planes which are going to be there. Extra plane is going to be here and g vector is in this direction. G vector is perpendicular to the beam direction.

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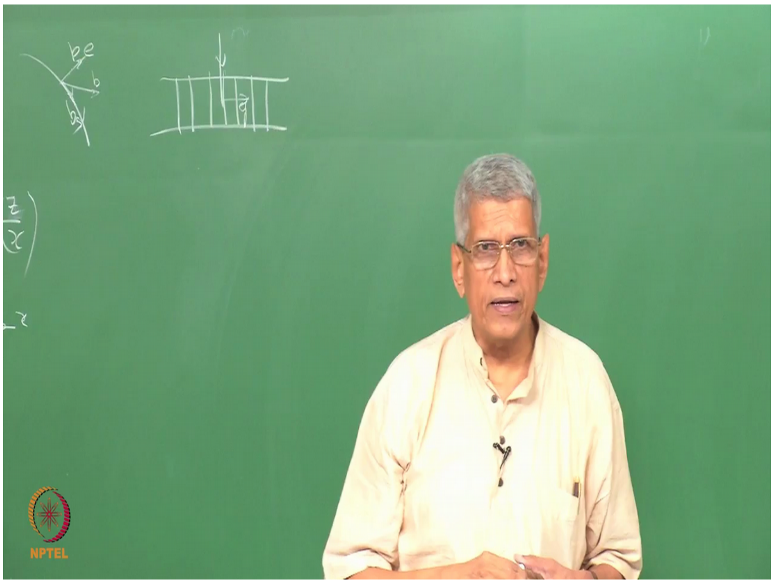

This is the g vector is perpendicular to it. That means that the planes which are parallel only are going to contribute to diffraction, correct. Is it not? So, even if I introduce an extra plane here, if you look at extra plane is there, but as far as this one closed pack line is concerned, it does not see an extra plane, the addition of the extra plane. So, since that does in c, we should not see any contrast in the image correct because g dot b condition will be equal to 0. This will satisfy this. What physically it means is the planes which we are choosing is such that the effect of extra plane is not being seen.

Now, if you see here whenever an extra plane has been added, what happens in an edge dislocation, there is a slight bending around the slip line takes place, correct. So, this is what essentially is being slight bending of the slip lines will be taking place. What will be the effect of this bending? It will be that some strain is introduced. So, that is why we put the condition that g dot b cross u should also be equal to 0. You understand that. If that condition is not being satisfied, there will be residual contrast will come because of this strain. That is all which one has to understand as far as this dislocation is concerned. Is it clear?

Then, normally when we image a dislocation, are we seeing the image of the dislocation? Actually not because when a strain field is there, you see that here an edge dislocation is being shown a parallel beam is falling onto the sample.

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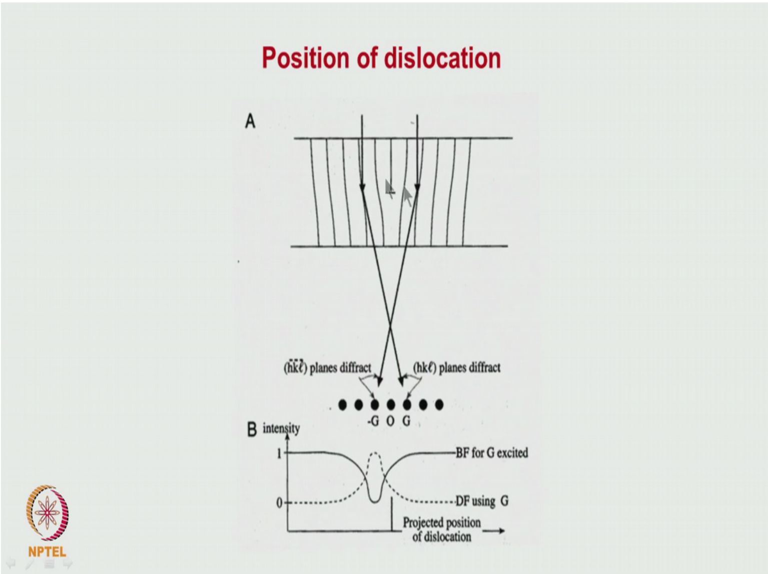

So, the beam is parallel, perfectly parallel to the planes. It does not satisfy the Bragg condition close to the dislocation line. The planes are bend. When the planes are bend because of the bending, the deviation from the Bragg condition could become just 0 for some of this condition and could satisfy exactly the Bragg condition. In such cases, what is going to happen is that from this region, there is going to be a strong diffraction. Similarly, from this region, there is going to be a strong diffraction and because of that essentially if we image with this or image with this, we are able to see as if the dislocation line is going to be here because this is where the diffraction effects are coming, right. So, we do not see the dislocation line. Though the dislocation line is here, we do not see the image here. Either we will see it here or if I use this diffraction spot to image, then I will see an image which is going to be here. You can see it, you look at that sample, you are seeing a dislocation.

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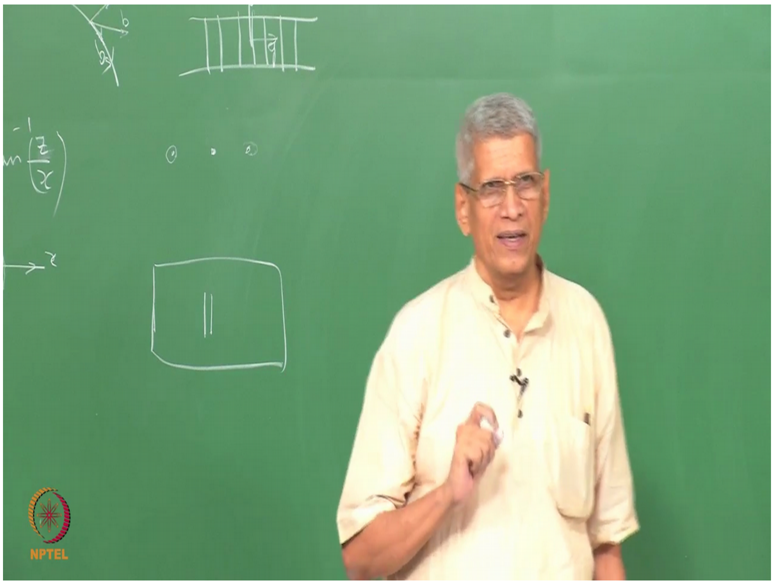

I have taken a two beam condition, where central spot is there. There are two strong spots. I put an aperture around this one and try to image it. I see dislocation line. One image I take it. Next is shift the aperture and put it here and try to image the dislocation. Then, I will see suppose I image both of them on the same plate, photographic plate, I am imaging it in one case. If I put an apertu, it would have given a line here. Another case when I image the same one, it will give as a strong line there. That means actually dislocation is in between these two. So, what you see as a dislocation line is depending upon which part of the dislocation has satisfied the exact Bragg condition. That decides this simple principle is used to distinguish between dipoles and dislocation with a same burgers vector.

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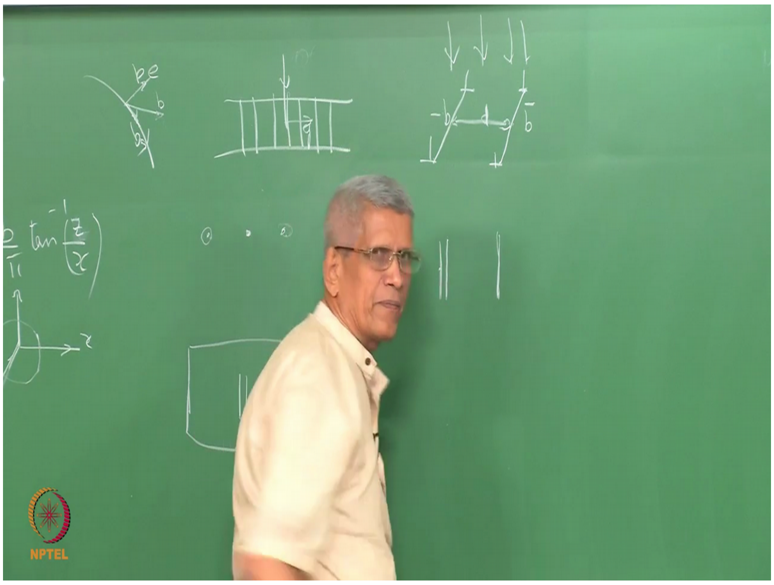

Now, I will just suppose I have a dislocation. One dislocation is there like this. There is another dislocation is there like this, there is a separation d between them. This is there in the sample electron beam is falling onto it for this condition and since, the burgers vector is same for both of them, if this satisfies for some particular value of s strong diffraction, this also will satisfy. So, because of that at one particular point, I will get the dislocation line and this line will come at another, but both are on the same side because either on one side or the other, only it will come.

Suppose this dislocation has got a burgers vector b is negative, then what will happen is the bending will be opposite. So, instead of it, the one line will come here and another line will come here, correct. So, what is essentially is going to happen is that if two dislocations of same burgers vectors are there, it does not matter which g I use. If the separation between them will remain the same, but they will be shifted in position. If I use one dislocation of the positive and another is a negative, then in one case for g, the separation will reduce and in another case, the separation will increase, but the burgers vector when we calculate it, it turns out to be the magnitude, turns out to be the same. So, looking at it we can tell that whether it is a dipole which is there or pairs of dislocations which are there which has got the same burgers vector. Yes, I will come back again.

Suppose first thing which we have to do is that we wanted to find out whether a dislocation itself is visible or not. The best is on the microscope going and looking at it.

It is going to be very difficult to identify one. So, suppose you know in FCC, we know what all the types of burgers vectors which perfect tran dislocations can have. That is what essentially is given.

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Table 7.2. Values of $g \cdot b$ for perfect dislocations in fcc crystals. After [7.4].

Plane of Dislocation	b \ g	$1\bar{1}1$	$\bar{1}11$	$11\bar{1}$	002	$0\bar{2}0$	$2\bar{2}0$
$(1\bar{1}1)$ or $(1\bar{1}\bar{1})$	$\frac{1}{2}[110]$	0	0	1	0	$\bar{1}$	0
$(1\bar{1}\bar{1})$ or $(11\bar{1})$	$\frac{1}{2}[101]$	1	0	0	1	0	1
$(1\bar{1}1)$ or $(11\bar{1})$	$\frac{1}{2}[011]$	0	1	0	1	$\bar{1}$	$\bar{1}$
(111) or $(11\bar{1})$	$\frac{1}{2}[1\bar{1}0]$	1	$\bar{1}$	0	0	1	2
(111) or $(1\bar{1}1)$	$\frac{1}{2}[10\bar{1}]$	0	$\bar{1}$	1	$\bar{1}$	0	1
(111) or $(\bar{1}11)$	$\frac{1}{2}[0\bar{1}1]$	1	0	$\bar{1}$	1	1	1

If I choose g values of this particular values, I can just take the cross product, dark product of it. I can find out for which all conditions the dislocation intensity will be visible and which condition the dislocation will be absent. This table is there. So, when we do a microscopy, we know that this g vector I had used, it dislocation has become invisible. This particular one again it has become invisible and then, we can uniquely identify what is going to be the burgers vector of the dislocation. You understand that? So, this sort of table helps a lot, ok.

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TABLE 26.1. Different Burgers Vectors and Different Reflections Give Different g.b = n Values

g\b	$\frac{1}{6}[11\bar{2}]$	$\frac{1}{6}[1\bar{2}1]$	$\frac{1}{6}[\bar{2}11]$	$\frac{1}{3}[111]$
$\pm(1\bar{1}1)$	$\pm\frac{1}{3}$	$\pm\frac{2}{3}$	$\pm\frac{1}{3}$	$\pm\frac{1}{3}$
$(\bar{1}\bar{1}1)$	$\pm\frac{2}{3}$	$\pm\frac{1}{3}$	$\pm\frac{1}{3}$	$\pm\frac{1}{3}$
$(0\bar{2}2)$	± 1	± 1	0	0
(200)	$\pm\frac{1}{3}$	$\pm\frac{1}{3}$	$\pm\frac{2}{3}$	$\pm\frac{2}{3}$
$(3\bar{1}1)$	0	± 1	± 1	± 1
$(\bar{3}\bar{1}1)$	± 1	0	± 1	± 1

The same thing can be done for a dislocation with burgers vector which is corresponding to that of a partial dislocation, the same sort of conditions which we can find out that way we can analyze it using this table and you have done a microscopy work using different types of g vectors. You have tried to image the same area. You remember in last class I had showed that for the stacking faults, the same area when it is imaged with different g vectors, some g vectors you see the fringe contrast due to a stacking fault for some g vectors, the contrast itself disappears. The same sort of a phenomenon will occur for dislocations also.

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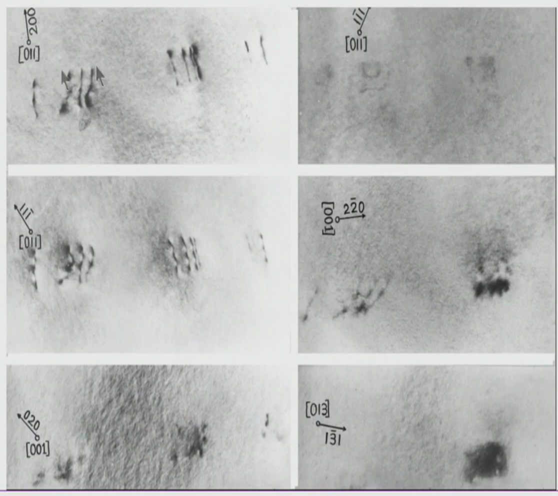

Determination of nature of dislocations by taking BF images using different g vectors and finding out g.b=0 condition for invisibility of them (g vectors marked in figure)

Here I am just showing an example, where you can see array of dislocations in a slip line, you can see here and you can see here also, but now if you see in these ones or in these one or in these vectors, the dislocations are gone out of contrast in for some g vectors. It is the same area which I had imaged. So, looking at this sort of images and then, using that table which I had mentioned which one can make on one zone, one can try to find out what is going to be the burgers vector of this dislocation. That is how the burgers vector of the dislocation is found out that specific burgers vector, otherwise we can tell that looking at it it is burgers vector is of b 1 1 0 2 type, but we do not want that value is if you wanted to find out this sort of an analysis has to be done, that means yes.

Student: Yes in case of single grain.

Everything is in single grain.

Student: But in case of fine grains.

Which one? It does not matter whichever is the grain. When you look at a dislocation, the dislocation is lying within that grain only.

Student: Because the area of diffraction spots are single grains we will get, but in case of this, there will be rings will become, no that depends upon which area you have chosen. You have to magnify the sample and make the beam fall on that single grain. That is how an analysis has to be done. Understand that.

Student: Since grain bound have some dislocation.

Which one?

Student: Grain boundary.

Yes, grain boundary dislocation also can be analyzed that dislocation has to lie on one side or the other side of the boundary.

Student: That is within the grain.

Which one?

Student: They are saying within the grain boundary dislocation.

Grain boundary dislocations are also lying one side of the grain. It is getting attached to a grain boundary whether it is coming from that side or whether it is coming from this side, how they are getting attached depending upon from which side is that getting attached. You have to use that grain for the analysis. You understand that boundary will be seeing it, but diffraction pattern from which side using it for an analysis depends that will tell you which is the side on which that is when you try to do with diffraction pattern from one side. You find that you are not able to get any meaningful information, then you have to look it from other side. Then, you find that you will be getting this trial, you have to do it. You understand that you do not have any prior knowledge. If you have a prior knowledge, we do not have to do any experiment because then we know it right for which. As I mentioned last time, you first go to a zone axis condition where all the beams are strong.

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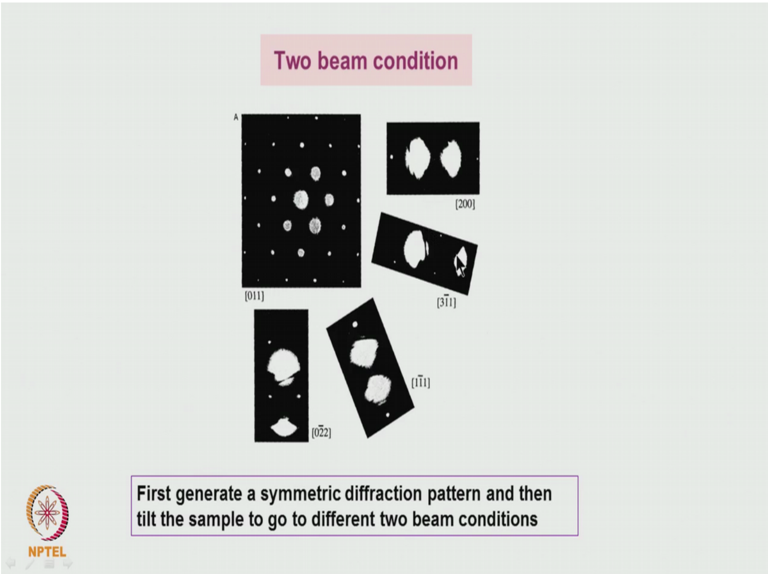

Then, by slight tilting of the sample, we can go to two beam conditions for various one. Suppose I had gone to 2 0 0 because this is that spot, then I put an aperture around this spot and try to image the bright. So, this bright field I called that it has been taken with g 2 0 0. That means, that only 2 0 0 planes are oriented for Bragg condition. So, whatever the contrast which you see in the image is because of scattering in that direction. That is what it essentially means and as I mentioned we can have that if for 2g values for which the dislocation become invisible, then that cross product if you take it, that gives the direction of the burgers vector, correct.

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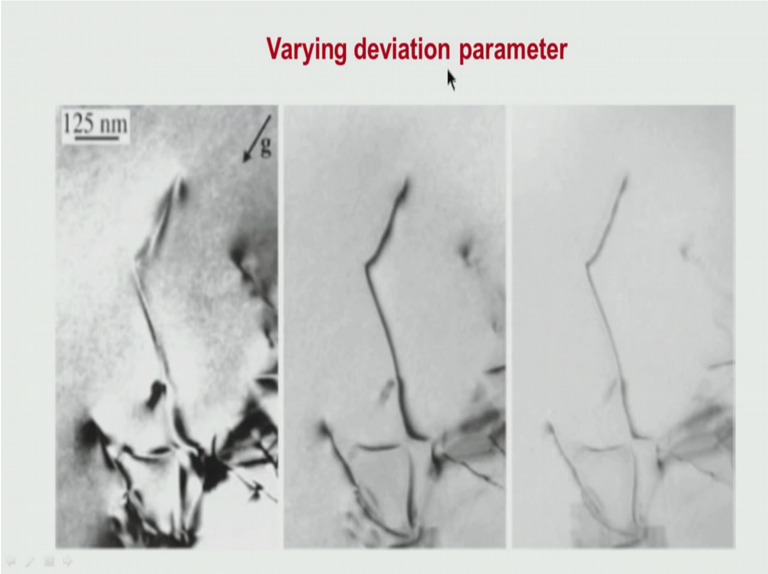

This is another factor which we have to consider. It is that when the dislocation is being seen. Suppose I vary the deviation parameter s, suppose exact. Bragg condition is being satisfied s equals 0 for same g. This is how the dislocation looks like. When g in the value of s increases, you can see that the intensity of that dislocation line is changing, right. Overall intensity changes with changing s also. So, controlling s is very important and we should also know what is the value of s which we have used it. Then, only we can quantify images. We can do simulation of images. Is it clear? The same way like we can go from s from 0 to positive side or negative side, the contrast will change. Similar to this if we go from plus g to minus g also in the diffraction vector, there will be a change in contrast which will occur.

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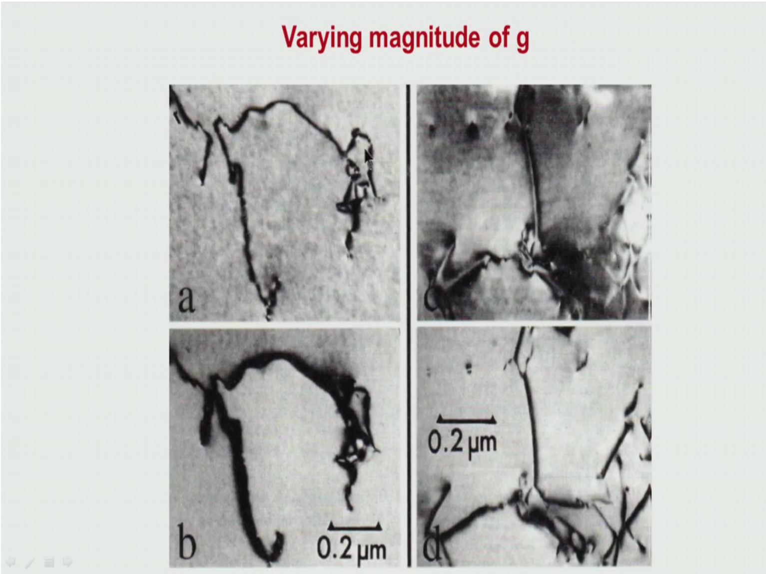

Here I am showing an example where the dislocations are visible whatever the g value which is chosen, but when I take the g dot b, the value of g dot b is going on increasing from this to this one. The g dot b has increased like what it can do. It is that suppose I had used 2 0 0 reflection.

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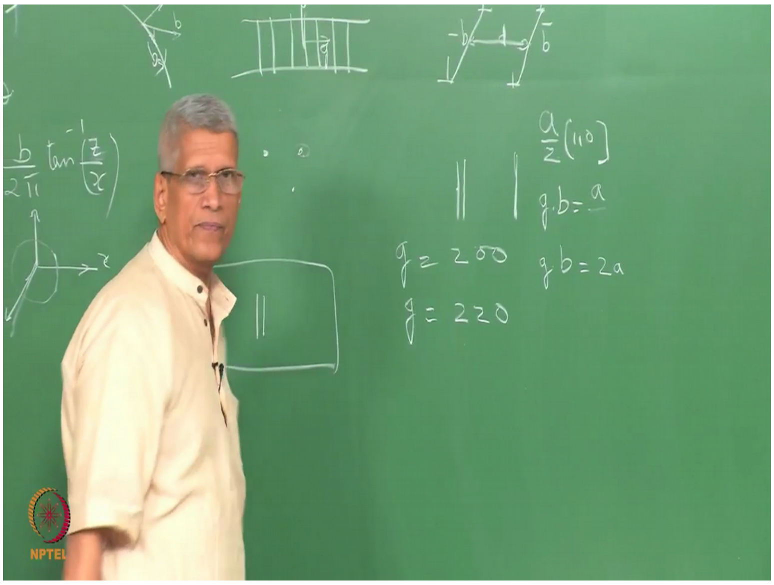

Like g equals 2 0 0, in another case g equals 2 2 0 I have used and the burgers vector of the dislocation is a by 2 1 1 0, then what it will happen g dot b in this case, what will be the dot product? It will be a by 2. It will turn out to be a correct here g dot b in the second case 2 a. That means the magnitude has increased. This is what is essentially is being shown because in both the conditions since g dot b is not 0, the dislocation is visible because this contributes to the image.

Now, you see that the width of the dislocation line itself has become very large. Similarly another case happens is that these images are taken under two beam condition. Suppose other than this one, suppose one more spot is there which is strong that will also give raise to contrast and in such cases, you get double images of the dislocation. It does not appear as one line. It appears as two lines and that is why it is happening because it is not the true image of the dislocation which we see.

What we see is one side of the dislocation. So, that is going to change because of the double image will come. Suppose you have not looked at any of these things, you are just looking at under some condition you are just taken image of a dislocation. If you look at this sort of an image, what is the interpretation which you will do? You will think that two dislocations are present whereas, actually only one single dislocation which is visible for it. That is why for correct interpretation, one should also look at the diffraction pattern and see under what condition these imaging is being done. If you do not do that, you will not be able to make correct analysis. Is it clear? It is the same dislocation only were region is being looked at, but you see that different types of contrast which they give, we will be making different types of interpretation.

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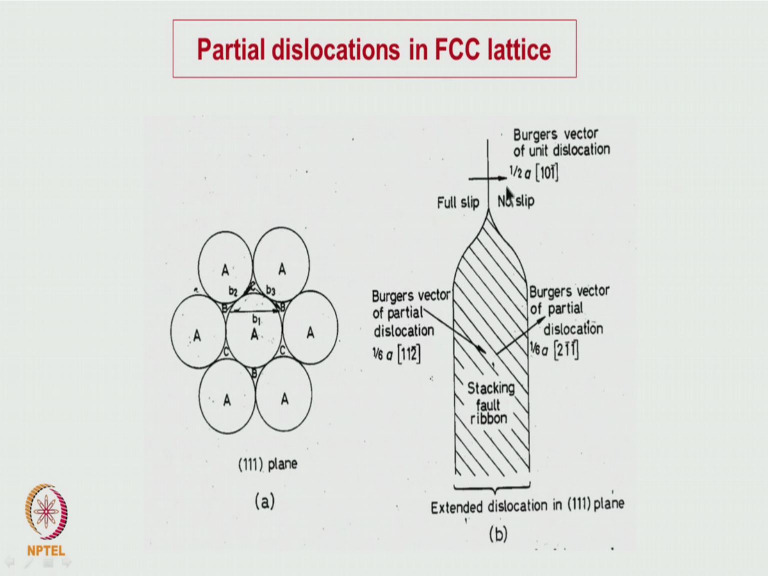

Then, this is a dislocation like a perfect dislocations can split into a partial. Then, a stacking fault will be generated, but these are bounded by the partial, ok.

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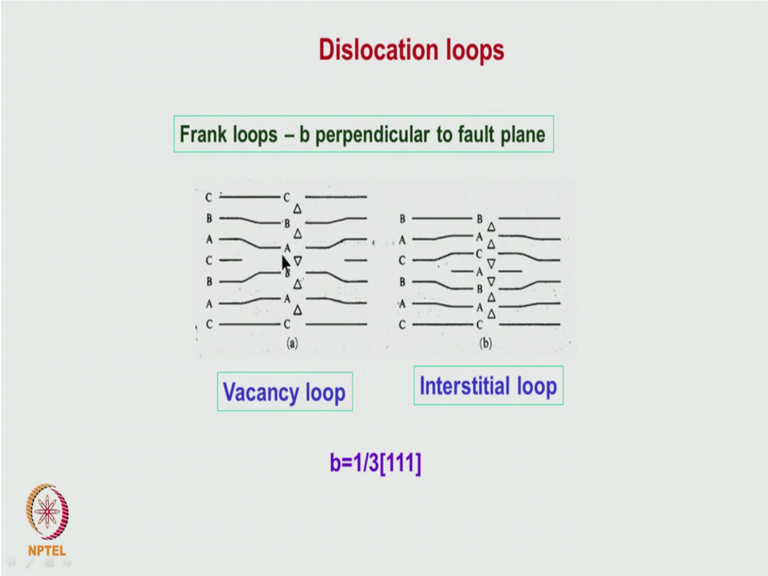

That same thing like if remove an extra layer similar to that I can create a fault, but this is surrounded by a dislocation which we call it as either a vacancy loop or an interstitial loop or we call it as a frank loop also. The general terminology which we use it and the burgers vectors will be in different directions. So, when we see a stacking fault, stacking fault could be caused due to different type of dislocations trying to find out the burgers vector of not only fault vector. We can find out by analyzing the stacking fault contrast, but imaging the dislocations and finding out conditions where the dislocations bounding the fault become present or absent, that also gives information about what is the type of dislocation which has given rise to this fault that I will show you in a moment.

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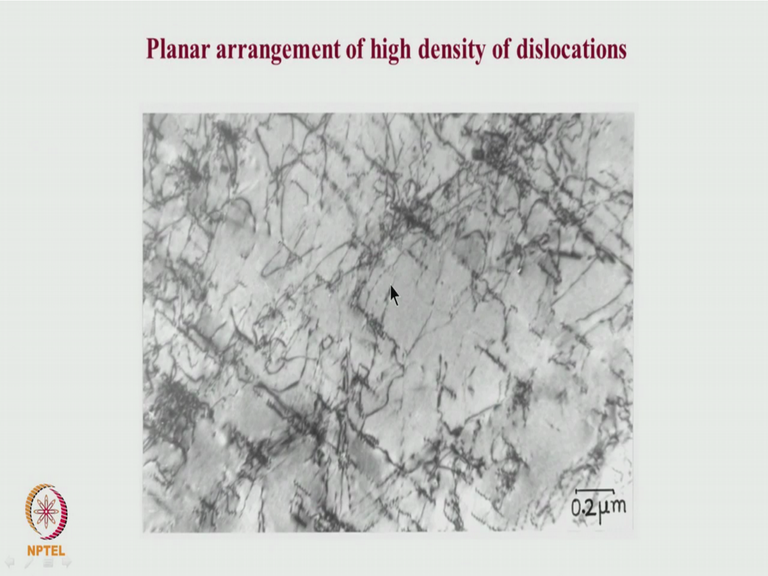

This is just a general picture to show that the sample is deformed. You can see the what is the sort of dislocation arrangement. It will be very difficult for anybody to look at all. This may I think has come with two deformation. So, essentially most of the dislocations analysis we do or with very small percentage of deformation, one percentage or less and then, we try to look or try to find out the burgers vector. The different line direction of each and every component of the dislocation in this sort of dislocation if you look at, it will take ones life time. One can spend on it looking at the sample and doing it because I know one friend of mine who has done it, published it that in that micrography has 200 segments of the dislocation. It took him about 2 to 3 years to complete that full analysis, but he did it and he found out what all type of interactions, what sort of junctions which have formed through analysis was done that requires a different type of a dedication to do that.

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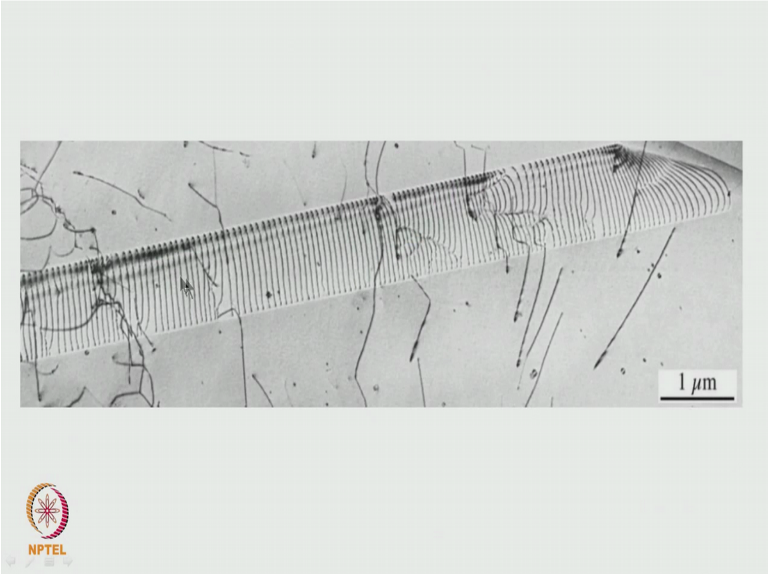

This is another example where the dislocations are moving in a planar array. Then, you can find out the burgers vector of the separation between them. You are able to see it, correct. Suppose these dislocations there is a dipole is there. Some dislocations have got a burgers vector which is plus and another has got a burgers vector which is negative, then what will happen that is the case which we considered here.

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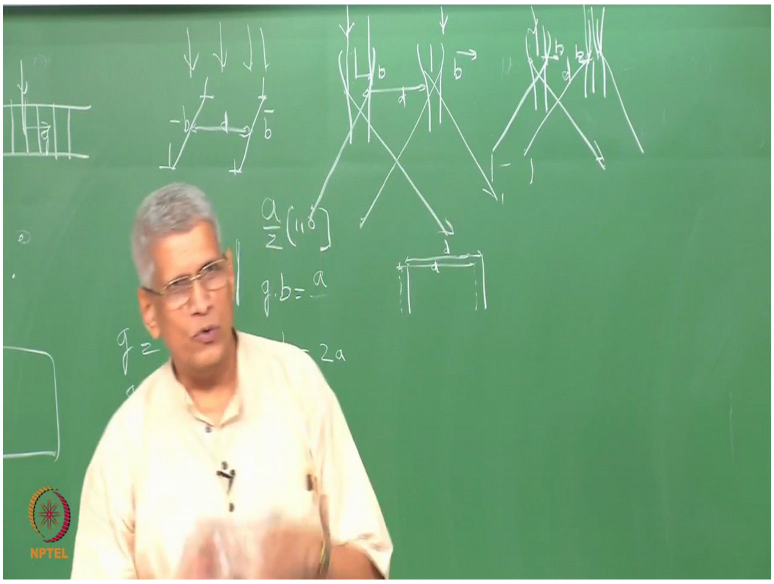

See in this case burgers vector is in the same direction, correct separated by a distance. You assume it to be d between them. Suppose the beam is coming like this and falling onto the sample and it will be scattered in this direction and in this, it is in this directions. Similarly the same thing will happen here in this direction. Here it is going to be from in this direction. So, if I use g vector which is corresponding to like in this one, I will be getting image of these dislocations as two lines like this. When I use the other g vector to image, the dislocation I will be getting, but this separation if I try to see it has not changed at all. It remains that same how do we define a dislocation with another burgers vector is like you take this case. So, here b is in this direction, here b will be in the opposite direction, correct. If these two dislocations are there, their separation is also d.

Now, when the beam enters, you see that here this will be scattered in this direction. Here what is going to happen is that when the beam enters like this from here, it will be scattered, right. So, the direction from the plane which is going because of this, what will happen in one direction for the same beam here it is scattered and in this beam, it is going to be scattered like this, here it is going to be scattered like this. So, because of that in one case from this side for the same beam and in this direction which is scattered this part. Another comes from here. This will give a separation which is going to correspond to this distance, correct. Other case, it is one this side and one from the other side, the separation will increase. So, using different gs burgers vector if you look at it, magnitude remains the same. This gives a clear indication that if the separation changes, then you can say that it is dipole which is there. It is not a dislocation with a same type of a burgers vector.

Another example which I am taking it is that so far different types of dislocations I had mentioned and we can take a coherent precipitate.

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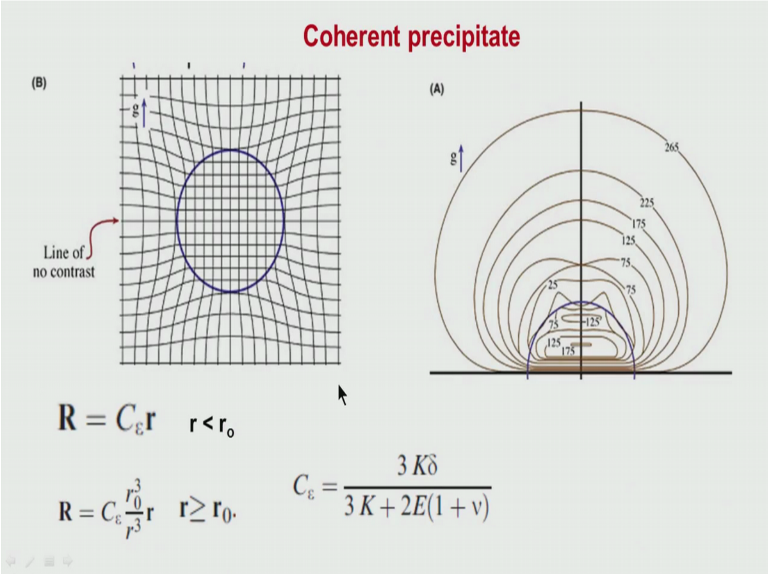

What is a coherent precipitate? That is the precipitate and the matrix has got the same lattice parameter, but the difference is very small. In such a case, you look at it. Here the planes have to bend to maintain the coherency. Any bending of plane is equivalent to generating a strain in this particular case which is assumed is that the second phase particle is so hard that no strain is accommodated within the particle. The strain is only accommodated outside in the matrix if such a case that displacement R is equal to C epsilon into r, where r is less than these all the formulas which are derived, but outside of it, this is the formula which has to be used and this is the value of the constant which is going to be there which depends upon the elastic constant of the material, but what is essentially going to happen is that when we use this sort of a formula and try to find out how the intensity will be varying.

Now you can see that here it is 0 and as we go away from it that intensity is becoming changing in as far as the diffraction is concerned exactly the opposite will happen. So, the intensity will become poor and poor, dark regions will start appearing and if you see here also, this is the direction of g in this direction. If this is the direction of g which is going to be there, you assume that this is the sample and this is the way the beam is coming, then in this particular direction you try to see it. These are all the planes which are going to different and because of that there will not be the effect of g dot b will be equal to 0 will come, but all other directions g dot b does not becomes 0.

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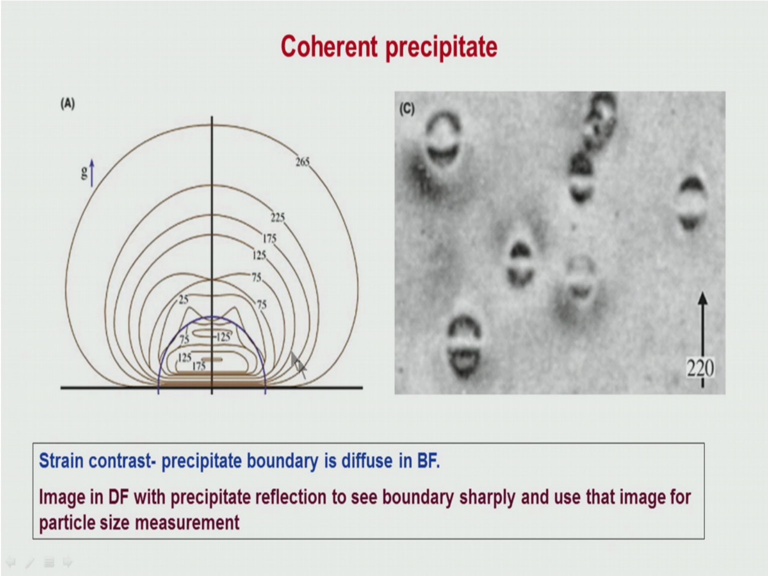

So, how it will appear now you see this it is a spherical particle which is being imaged. Now, it looks like two lobes which you can see. Now, with a line of no contrast, this is perpendicular to g. If I change another g vector which is perpendicular to this that is the same precipitate which is present in the material.

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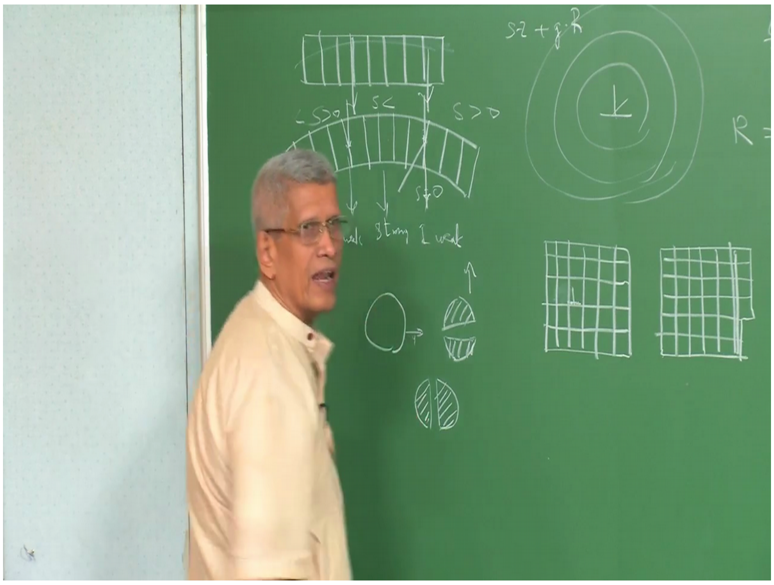

In one case, it gives an appearance of contrast like this. For another one, this same precipitate will give an appearance that is in this case, the g is in this direction. Suppose g is in this direction, the same precipitate gives a contrast like this. This is what will happen, but what all the consequences of it. Consequence is that first you are not able to make out where is the boundary of the precipitate, correct. Only a strain contrast which gradually fades away at large distances, this is the information what we get it, but looking at this contrast, one can say that there is going to be some precipitate.

Then, because of this you will never be able to find out the actual size of the precipitate. In such cases, you have to always use a dark field with the reflection corresponding to the precipitate. If we do it, we will be able to image the precipitate clearly, otherwise coherent precipitates in bright field condition it is extremely difficult to find out the particle sizes accurately, but looking at it we can talk many things about what about the strain which is there. Lots of information which we can get, it depends upon on what information which we look for. We have to use different type of analysis. That is what I wanted to tell this. So far I have talked about how to do analysis. Now, I will tell a little bit more about what all complications which are going to be there.

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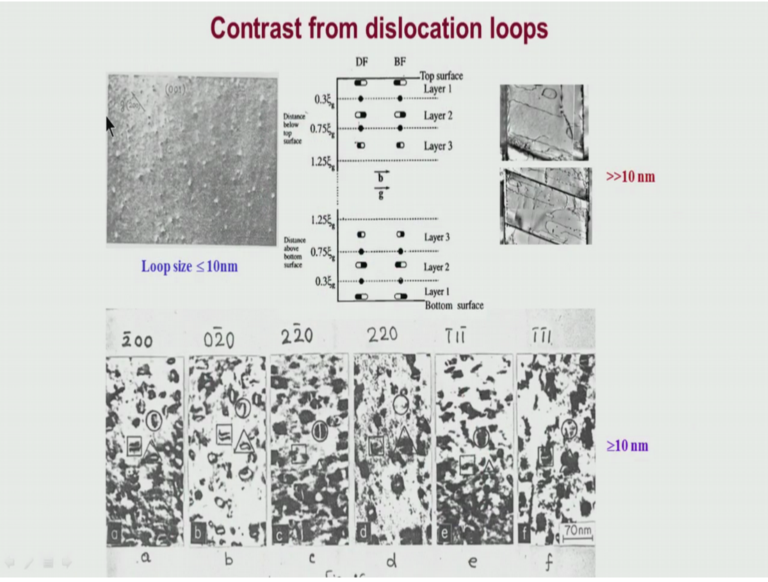

This is a case, where we have dislocation loops are present in the sample. In this one, the loop radius is 1 or 2 nanometers and when the loop radius is 1 or 2 nanometers, they give a contrast which is called as I am not going to explanation for how they appear, but I am just telling that they will appear as black and white dots. That is how they appear. The same dislocation loop when it has increased a size to close to about 10 nanometer, they give a contrast. You look here in this particular one, this looks like exactly what you have seen for the precipitate with the coherency strain. In fact, that is some similarities is there about some strain field around it because in both of them, this contrast is coming because of the strain field. So, they appear that same, but in this case it is a dislocation loop and not a precipitate.

In the other case, it is a precipitate. These are all the pitfalls which are there if you just microscopy without understanding what is happening. That is what I wanted because totally you can give a wavered interpretation can come and when the loop has grown to a very large size. Now, you see that a clear loop could be seen. You understand. So, even the contrast is a size dependent effect.

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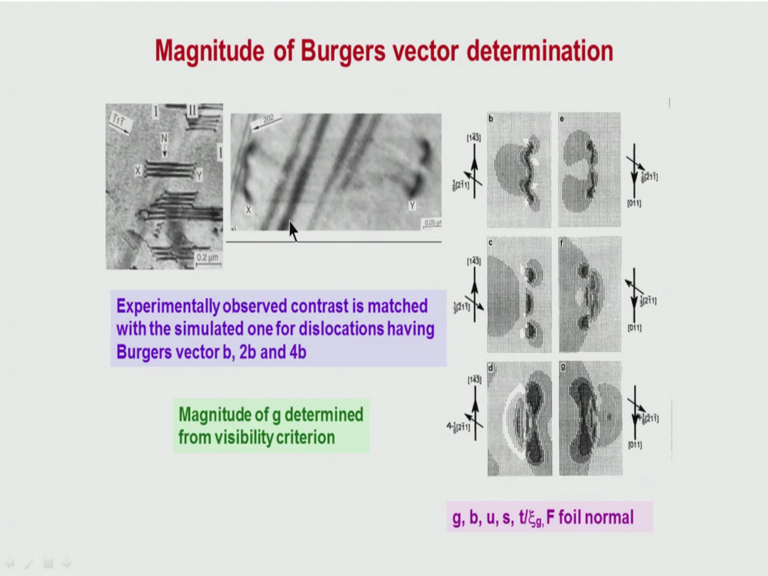

This is one example which we are seeing stacking faults in a sample. We wanted to find out the partial which is bounding, the stacking fault what is the burgers vector. This is in an ordered alloy in this particular alloy.

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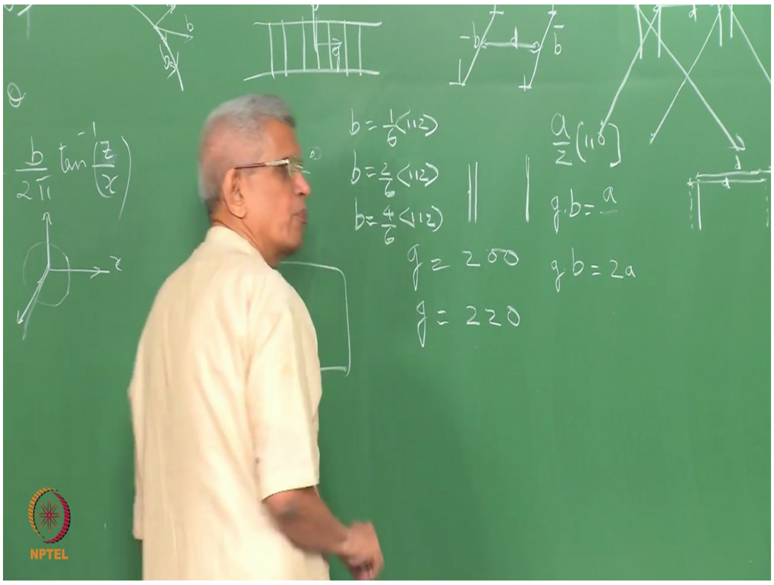

The partial could have burgers vectors b equal 1 by 6 1 1 2 b equals 2 by 6 1 1 2 b equals 4 by 6 1 1 2. All the four possible, three possibilities are there, but if you look at the burgers vector, the magnitude only changing, correct. For all these conditions, g dot b will be always equal to 0. It does not matter what the magnitude is, right. Suppose I wanted to find out which is the dislocation which is responsible for this, by looking at the g dot b condition, I will not be able to make out what the magnitude of the burgers vector is. In such a case, I have to image the dislocation and look at how the strain contrast will vary depending upon the magnitude. That is exactly what has been done.

So, we have chosen a condition where the stacking fault has become invisible. Only the partials which are bounding them can be seen. Then, we have to do conditions under which the dislocation, these partials have been imaged. We have to do a computer simulation which was done and these are all the images which we had obtained which we are showing it. So, you can see that when it is 1 by 6, the type of contrast which it appears and when it is 2 by 6, how it should appear. When it is 4 times the partial, you can see that the contrast matches with this particular value. This is how we could conclude that this is what the burgers vector, this sort of situation also will happen in many materials, real materials. You will have this sort of problem that is any ordered material you will enter into this sort of a problem, this will not happen for disordered material, but in ordered alloys, you will have this sort of a situation where the magnitudes of the partials whether it is perfect or partial can change. Then, how do you find out?

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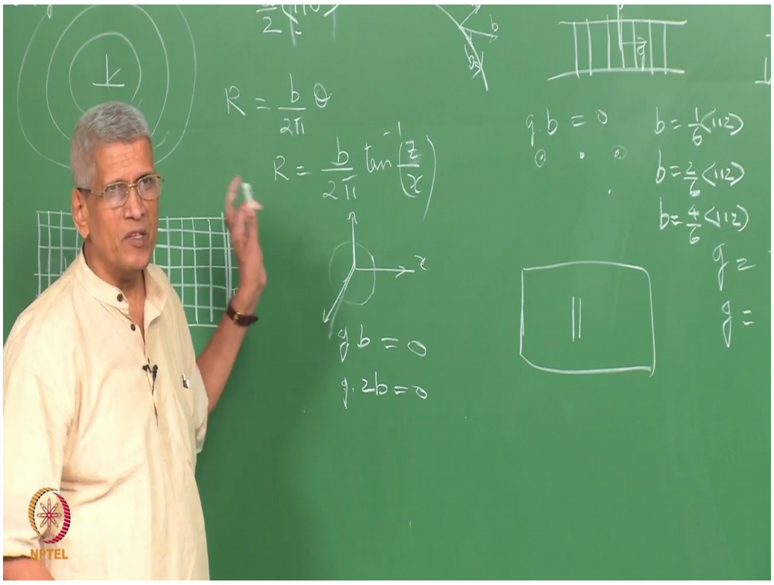

So, then you have to use because this is very simple that is suppose g dot b equals 0 g dot 2 b also equal to 0 correct, but because b remains that same. So, whether it is b or 2 b, the magnitude is different and the way the dislocations will behave and a line self energy of the dislocation everything is going to be different, the strain is being different. So, looking at g dot b equals 0, we can find out what the value of b, nature of b, but the magnitude cannot be determined. So, to determine the magnitude, quiet often you may have to do computer simulation, ok.

Now, I will come to the last part of it is what I mentioned about weak beam imaging. What is called as weak beam imaging is that almost all this dislocation image which we considered under two beam condition where the transmitted beam and the diffracted beam are equally strong. Another condition which is considered is where image using a condition, where the deviation from the Bragg condition is very large. When the deviation from the Bragg condition is very large, then it is similar to kinematical condition, correct. It is like imaging dislocations under kinematical condition. Why it is necessary because under two beam conditions, you saw that as g dot b value increases, the line width or the dislocations becomes very large and it is so large that suppose two dislocations are present, you will not be able to separate them and some cases where there is one dislocation is where you get two images.

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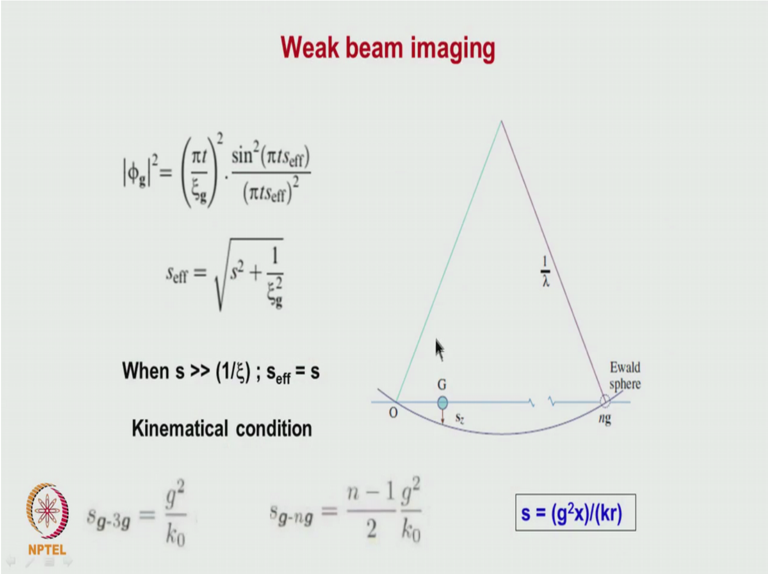

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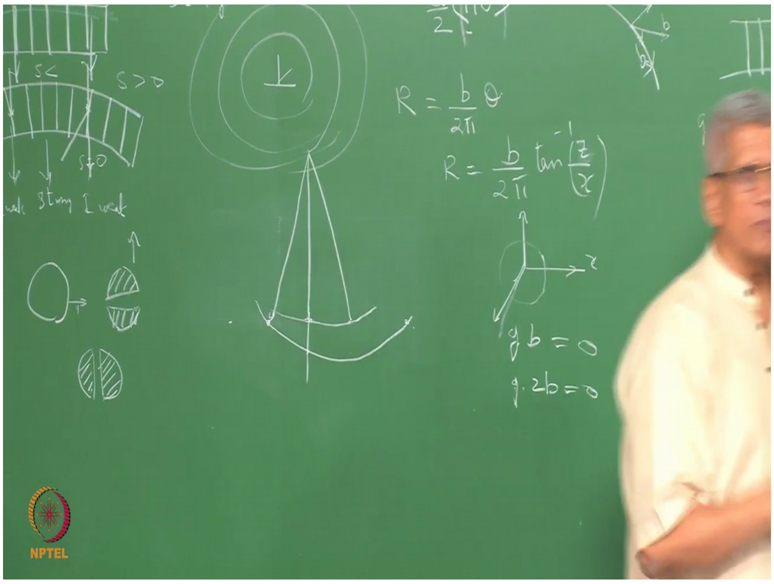

So, all sort of complications can come how we can go about it. Suppose we tilt the sample. If these are all the diffraction spots which are there, if the Ewalds sphere passes through both these spots, then this is the exact Bragg condition, correct.

Suppose I shift the Ewalds sphere to tilt the samples, so that the incident beam comes here and tilting is such that the Ewalds sphere passes through this one. Then, if I look at this spot, this is not at the Bragg condition only. Here Ewalds sphere is touching it, but with respect to this one, there is a streaking which is going to be there, correct. Its cutting is equivalent to some deviation from the Bragg condition that deviation could be very large.

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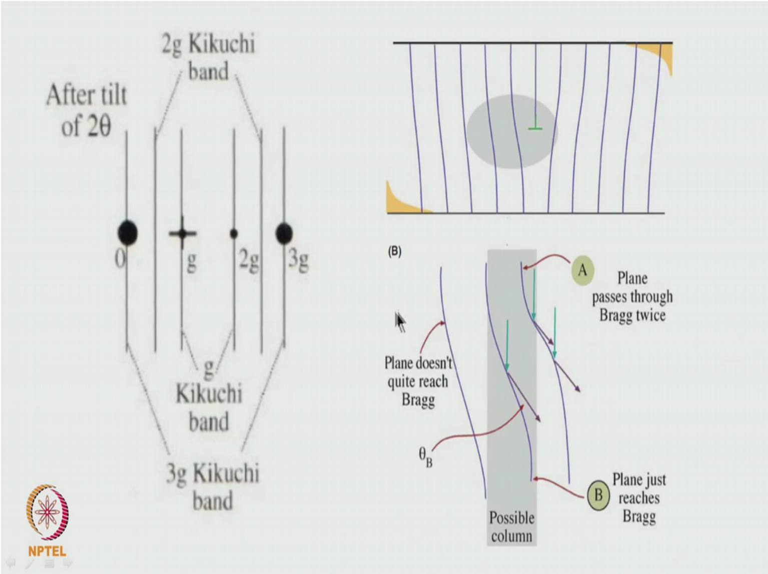

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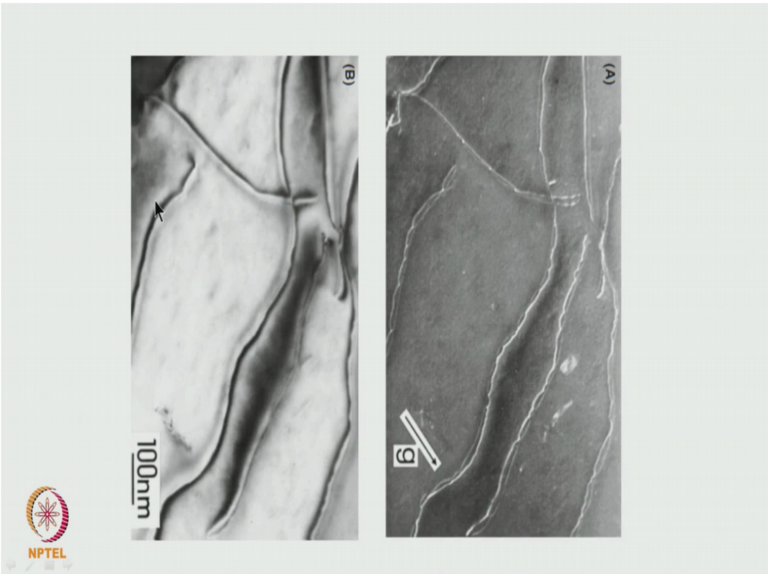

This phenomenon is used to get an image. So, what is the effect of it? Now, you see that image of a dislocation which you are seeing it here in a bright field under strong condition, the same dislocation when it is imaged by making the deviation very strong, ok. In the weak beam condition, now you can see that there are two dislocations are present here, correct whereas, if you see here are you able to see two dislocations.

Student: I did not understand how did that.

Which one? That is the make, the deviation from the Bragg condition very large for which what I do generally in a microscopy, what is done is that you got a condition where you have this is the transmitted beam, this is the g vector. Now, I tilt the beam in such a way the direct beam comes here. Then, what is going to happen is that this is equivalent to tilting the Ewalds sphere. So, the Ewalds sphere will get tilted like this and now Ewalds sphere crosses on this row of spots at one which is at a very high one, but since the sample is thin, you will have a streaking which is associated with it. So, this is exact Bragg condition. When the Ewalds sphere cuts here, this is equivalent to have using a beam which has got, which is deviated away from the strong condition. That is how it is done under this condition. Two things which happen is that the width of the dislocation line which appears in the image becomes very narrow. That depends upon the larger, the standard the deviation from the brag condition, yes it is going to be more.

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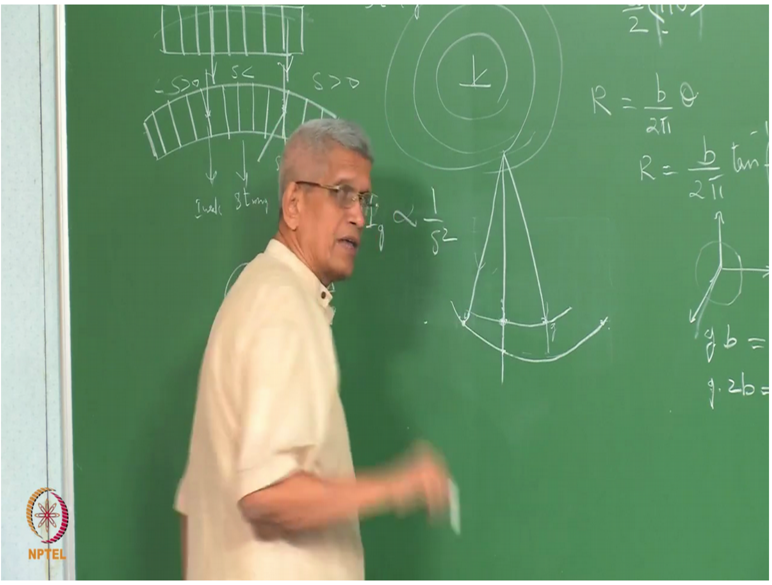

If you remember earlier we studied that i is intensity of the diffracted beam is proportional to 1 by s square. In the last class I had shown, that is more the deviation becomes, the intensity falls down. So, you have to optimize a value of s, so that the dislocations are barely visible. When they are imaged, this is the sort of the thing which you can see. So, in this though in a bright field you are not able to see whether dislocations has split into partials or not, all these things took dislocations have closed by is there in a weak beam technique. We are imaging only a very small portion of dislocation that is essentially what I had shown here by tilting that sample, so that when we deviate very from large value, the bending is such that closed to the core where bending is large, those regions satisfy Bragg conditions and others do not. So, because of that the width from which the scattering is taking place is going to be very small. That is how it happens.

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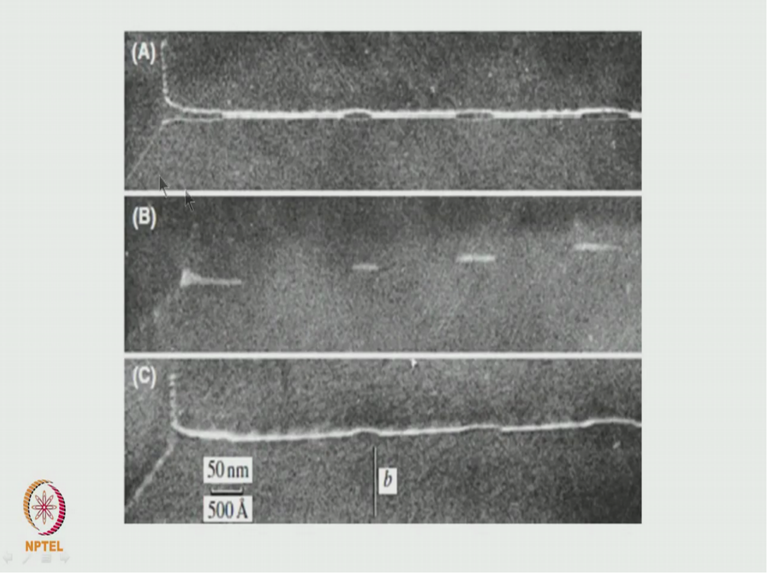

So, I will show you some images, like here you see this. Here the dislocation has split into a partial, here it has split into a partial using weak beam images. One could identify all this and here one dislocation is there. Both the partials could be seen faintly. If you look here, one partial is seen and another is not seen. Each is with respect to different type of weak beam condition which is being used. So, using this also you can find out that is where the dislocations are visible for which conditions dislocations are invisible g values. We can find out what the partials are especially where the separation between the samples are becoming extremely small.

This is in another case where in a ordered one, the domain boundary, the dislocations have split into partials. We are using weak beam conditions. They have been imaged. This is a small angle boundary where the dislocations are imaged in a weak beam condition, where you can make out that as we increase the deviation from Bragg condition, this line can become sharper and sharper.

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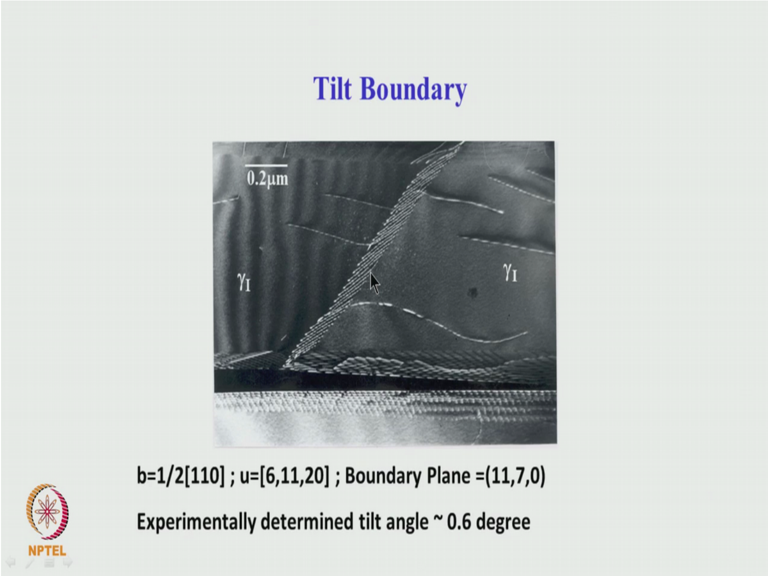

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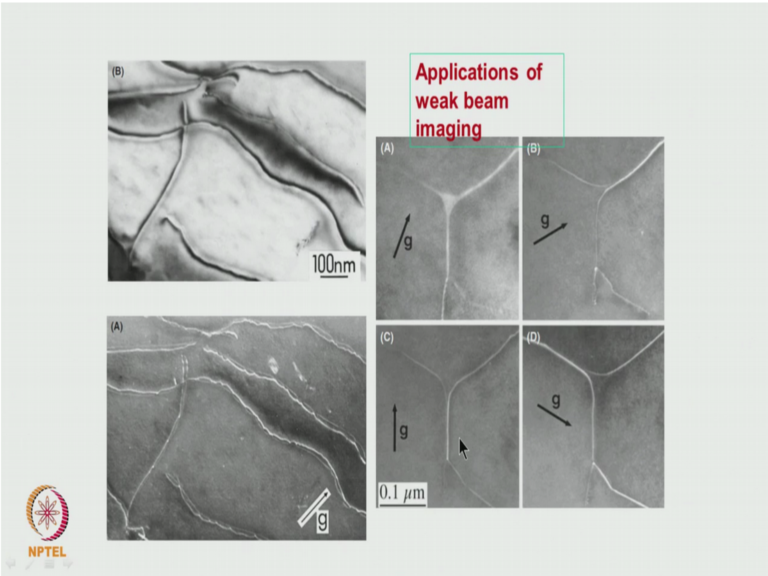

This itself is quiet sharp. This is another example I shown here. This is at node. There is a splitting of this partial. You can see that using different g values, now you can difference segment of the dislocation line becomes invisible. So, using this condition, you can analyze it all. This is how you have to do it. That means, that the region which you see it requires a lot of effort to do it, but at the end of it, you will have a complete analysis of all the characteristics of the defect which we were seeing it.

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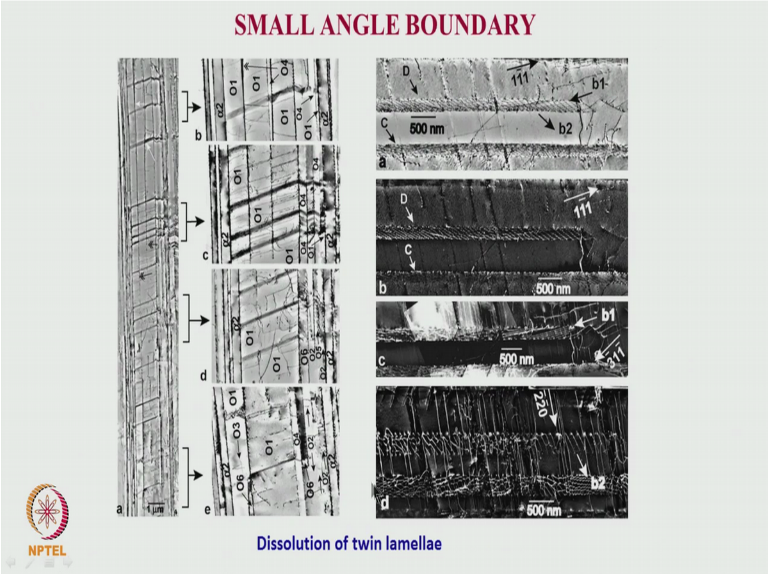

This is another one. It is only just an example. I am just giving you this is what we have taken where how the dislocations have been imaged using dislocation.

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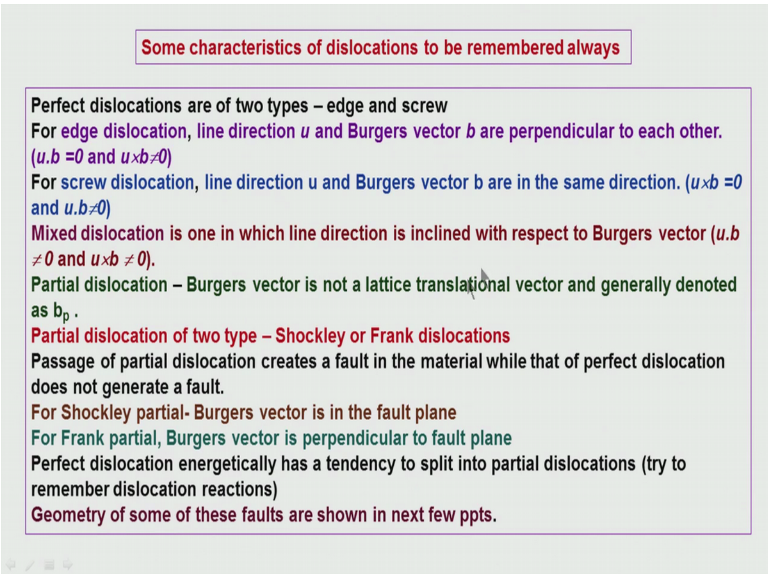

In these one, I had just given overview of what all the conditions, the methodology which has to be adopted to just get g dot b equals invisibility as well as visibility criterion which I had explained over the last one hour. Here it is clear using this techniques that is using g dot b condition as well as weak beam under some conditions, you have to do computer simulation of the image also, but what is essentially important is that every picture you take it, there should be a corresponding diffraction pattern which tells you what is the exact condition under which you have to take and if that picture is there with kikuchi pattern, we can also find out what is the deviation from the Bragg condition which has been used for the analysis. Only if those informations are available, you can do computer simulation.

You understand that to simulate the image, to get it, when you look at images which are being put on the paper and they have given this is what it is, but in a paper when you publish it, you will not be able to put all these information together. So, they will have only one picture, but to put that one picture, they might have taken hundreds of pictures over which the analysis is done and on one picture, they had explained it. So, if one wants to do serious microscopy, this is the way one can do defect analysis.

I will stop here now.