Advanced Metallurgical Thermodynamics

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Lecture No. # 25

Kinetics of Eutectoid transformation

We started talking about eutectoid, last two three classes. We will try to continue today and then go into martensitic transformation, because that is a kind of an offshoot of eutectoid transformation that you get. So, this one aspect about eutectoid transformation, which we have been seeing yesterday is about, what is the kinetics of it? Yesterday, we considered it partially, last class.

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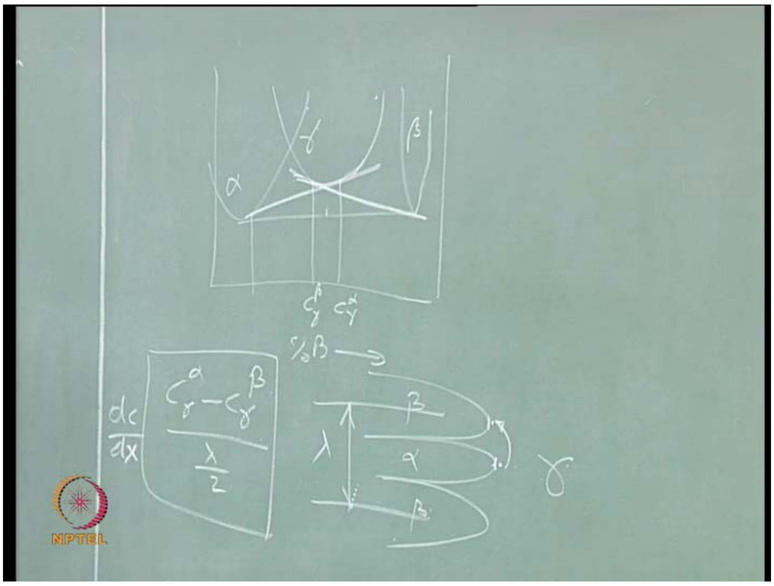

Let us try to continue this. Though it is slightly off from our class, as you all know that eutectoid transformation as long as you take a composition of that sort. We have seen that the driving force that you get for such a transformation, which can be seen from a free energy composition diagram, which we have seen.

Let us say, you take a case like that, that you have a certain driving force at a given temperature below the eutectoid transformation. This under cooling, which translates to this driving force is actually drives the whole reaction. These we have seen it even in the eutectic; so I thought let us just recapitulate once again. Because we if you look at eutectoid transformation as as some phase; let us say call it as alpha, beta and this is beta coming out of a gamma something. We can take it as an iron carbon system or any other system. If you take it, we see that this growth of this lamellae depends on the the atoms joining from gamma to both beta and alpha right.

So, when atoms from gamma come and join the beta and we know that, beta is let us say B rich phase, alpha is A rich phase. So, you need excess B atoms to come and join beta and excess A atoms to come and join alpha and then, you will see the whole thing will grow far. This how a gamma can supply excess atoms of B to beta and excess atoms of A to alpha is what you can see, again if you draw a common tangent between gamma and alpha and gamma and beta. So, if you look at this, call this as gamma; this as alpha; this is beta and simply draws a common tangent between, let us say I am drawing a common tangent between gamma and alpha.

That means, the moment I have gamma in equilibrium with alpha. What is the composition of gamma, which is just in front of alpha? The composition of gamma which is just in front of alpha, which is in equilibrium with alpha, is given by this common tangent composition. Am I right? So, where gamma is in equilibrium with alpha; let us not bother about. What is the composition of alpha in equilibrium with gamma? But if you look at, what is the composition of gamma in equilibrium with alpha, this is what you will get. And similarly, if you look at, what is the composition of gamma in front of beta?

Again we draw another common tangent between the beta and the gamma and you get another composition here. We call this as C gamma alpha; we call this as C gamma beta right and we can see now, that the composition of gamma in equilibrium with alpha has more B atoms; if you say percentage B is increasing in this direction. So, here you have more B atoms than here. As a result, you will see that the B atoms will start moving in this direction and that gradient, if you want to considerate that gradient can be given by C gamma alpha minus C gamma beta divided by lambda by 2; because this whole thing is lambda. Am I right? That is what we call it as interlamellae distance.

The center of one beta to center of another beta is what the interlamellae distance and this distance, that we are talking about is half of that interlamellae distance. So, if that is the case, this is what is the dc by dx. Am I right? You put this x as small. Again certain notations one has to remember, x capital usually means mole fraction. So, put x as small let us say. Infact some people preferred to write x here rather than C. For composition either you can use C or x depending on whatever you want, as long as you are consistent in one of the notation. You can use x or C there.

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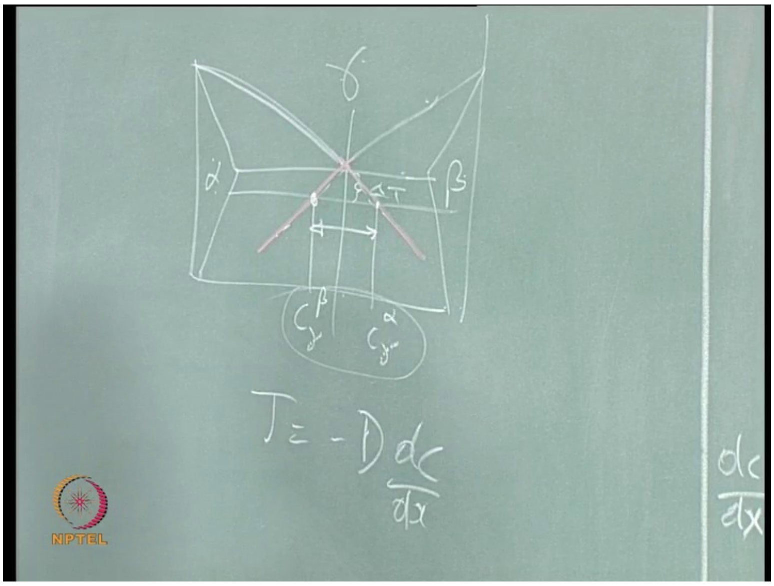

So, if this is the case, then we know the the flux is J is minus D dc by dx. So, again coming back to what we have done in eutectic is exactly similar. So, it is this which drives this and at a given temperature, if this eutectoid transformation is happening, if I say that I have taken this alloy gamma brought it to this temperature and holding it there at that temperature, till the transformation is over. If I am saying that at that temperature these two are fixed; that means, this is fixed and that if it is fixed and if the lambda is fixed, lambda basically depends on the under cooling.

It is under cooling, which decide the lambda. And once the lambda is fixed, the concentration gradient is more or less fixed and once the concentration gradient is fixed and D is also fixed for that particular temperature. You would see this grows at a particular rate. So, the growth rate is dependent on the temperature; that means, the under cooling in two ways. One, the under cooling decides the D; the under cooling also decides the dc by dx. Because the under cooling is what? Decides this and how do I see that? I can see it easily. If I extrapolate this extrapolate this, this is what? This composition.

If I call this as gamma, if I call this as alpha and if have some beta here, let us say. Yes, this is the composition of gamma in equilibrium with beta. Because it is that, that solves line or that transformation line which is gamma to beta transformation line which is what I am extrapolating. So, this line is actually gives you a composition of gamma in equilibrium with beta at all temperatures. So, because I am extrapolating that to a lower temperature, then I can say that this is nothing but C gamma beta and similarly, this would be C gamma alpha.

Because this is this is the line which talks about composition of gamma in equilibrium with alpha. Am I right? So, if you use the same logic as we have used in eutectic, so this difference between these two decides the delta C here and because that decides delta C, in principle we can say the dc by dx is decided by that. And once you identify that, you can say that the growth rate of this eutectoid depends on that. But keeping in mind that, there are two I know points, two parameters here, which kind of counteract each other. One is that the dc by dx in principle increases with under cooling.

In dc by dx also, there are two. One is dc, another is dx; dx here is lambda by 2. What happens to lambda by 2 as I increase under cooling? It decreases and what happens to delta C? Increases. So you see that, the numerator is increasing; the denominator is decreasing. Am I right? So that means, this fraction increases with under cooling. But at the same time, the D again diffusivity; as I increase the under cooling, diffusivity should go down. So, this growth rate is again a combination of the temperature, which is where diffusivity decides that and the dc by dx.

So, a combination of these two will tell you that, the growth rate will reach a maximum somewhere at in the intermediate temperature range. It will not be highest near at the eutectoid transformation temperature in principle, growth rate will be 0; because dc is 0. Because we can see this difference at the eutectoid temperature is difference is 0. That means, C gamma alpha minus C gamma beta is 0 at this temperature and it is that, which is driving which is the part of gradient. And if there is no gradient, there is no diffusion.

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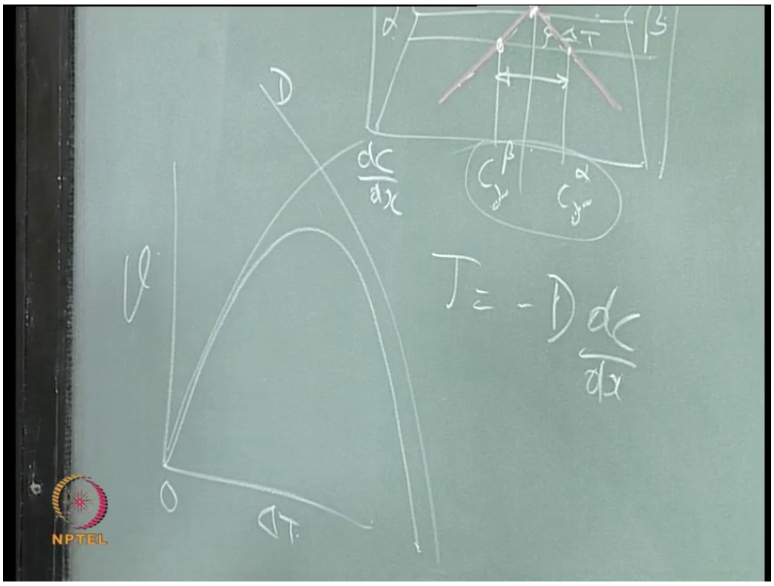

So, as a result you will see that at the eutectoid you will see, if I now plot the growth rate G as a function of under cooling. You would see the under cooling could be something like this. At 0 under cooling, what I mean by 0 under cooling? At the eutectoid transformation temperature, this will be 0. And it would again go down, when you go to higher under coolings; because the temperature is going down, where the D is going down. Though this is increasing, so that is why again it would be a combination of two curves; one, which is nothing but diffusivity. So, diffusivity is decreasing with increasing under cooling and the dc by dx is increasing with increasing under cooling.

So if I say, this is dc by dx and this is diffusivity, one is increasing with increasing under cooling; another is decreasing with increasing under cooling. So as a result, a combination of these two, you would see the growth rate will be like this. This is growth rate. I am sorry G is not free energy; so growth rate G whatever you want to call it; you call it as velocity v, growth velocity. Infact, incidentally people also use G in solidification as a gradient. Temperature gradient is also given the same term so same term, you get in different different connections.

So, if you call this as a growth velocity, so growth velocity would be having a similar curve. So, because we are talking about this; I thought, you should also have an idea of this. Now, in addition we also talked yesterday about, in the last class about the morphology of alpha. You see, if you take an alloy of this particular composition, we also said yesterday that if I take this alloy under cool it below this temperature. You directly get a pearlite or bainite depending on where you are with respect to that temperature. So, if I am doing it here; let us say, I am closer to the eutectoid temperature.

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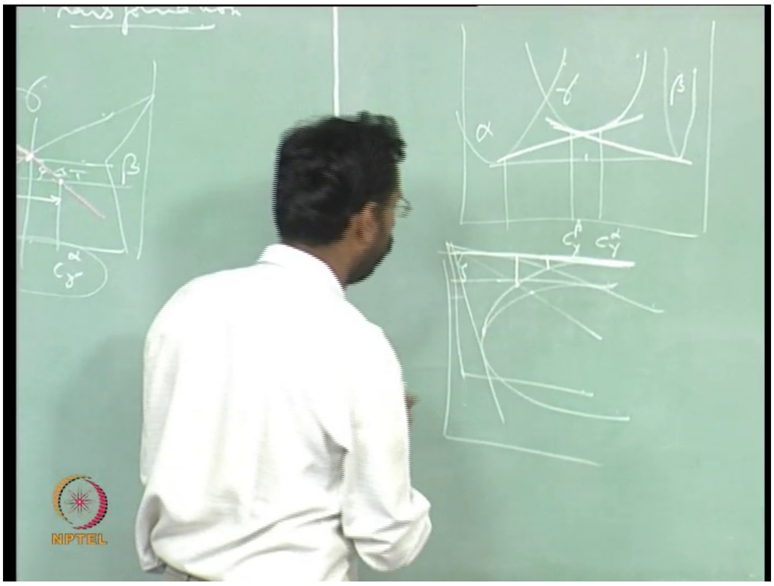

So that means in a TTT diagram, you are closer here to the eutectoid temperature. So, because you are closer to the eutectoid temperatures this particular alloy, if I under cool it below this particular temperature, then it would give me a pearlite rather than give me bainite. But if I take this alloy which is farther from the eutectoid composition; this alloy, if I can under cool ofcourse under cooling that alloy to that temperature is also not easy. Because the under cool extent of under cooling, that I need to provide is very large.

If I can somehow provide that under cooling, then you would direct and hold it at that temperature below that particular temperature. You would end up in a bainite and that is what you see. When I under cool it, hold it there; I would end up bainite there. Though there is proeutectoid phase, no proeutectoid phase will come out during this under cooling. Because we are cooling it fast, we are not allowing time for the proeutectoid phase. So, instead of any proeutectoid phase coming, straight away you come to a region.

And that is why we said, if your cooling rate is higher than what is decided by, where this proeutectoid component reaches the TTT curve of the eutectoid. And if your cooling rate is higher than that, obviously you will never get a proeutectoid phase. If your cooling rate is smaller, then you will cut both these; then you will get a proeutectoid phase. And depending on, where it is cutting and what is the under cooling that you are achieving, the proeutectoid phase also the finest of the proeutectoid phase is going to be decided.

For example, if I use a cooling rate like this with respect to the proeutectoid transformation temperature, what is called A 3? Now with respect to that, if I consider the under cooling, the under cooling for this is smaller; the under cooling for this is larger. So, because the under cooling is larger here, if I am using this cooling rate instead of that cooling rate, here I will get finer alpha, We are now talking about the proeutectoid phase. So, you would get a finer alpha here, instead of what you get here. So, that is why whatever we have talked in the last class about, what is the difference between annealing and normalizing.

If we extent this two, a hypo eutectoid alloy; in a hypo eutectoid alloy you will see another difference that, you will have a coarser alpha in a hypo eutectoid steel, if you do annealing; you will finer alpha, if you do a normalizing. That is that is the second thing that we talked about it. The proeutectoid phase amount will keep on decreasing as you increase the cooling rate; that is what we have seen there. You will see that the amount of proeutectoid phase will keep on decreasing, until you reach a cooling rate faster than this. Let us say, you have a cooling rate we can even consider a case, where I think in the last class I showed you that.

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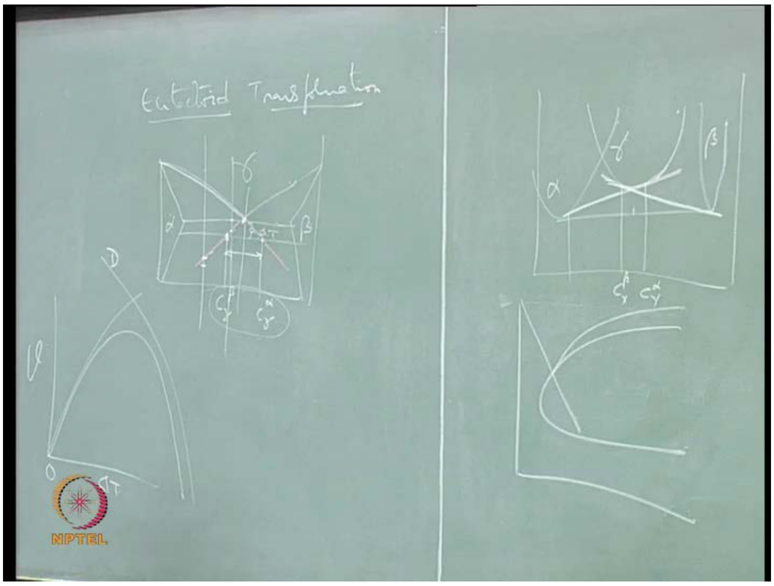

So, if you have cooling rate something like this, then you would only get eutectoid phase. In principle, one can provide you have the CCT diagram; one can talk about what is the amount of eutectoid phase? And the higher the cooling rate, that is the reason why by looking at the microstructure, if you if you want to find out, what is the composition of alloy? Infact in B.tech labs possibly would have seen people asking you, see the microstructure of the steel and tell me what is the composition? In principle, if it is a plain carbon steel assuming that, it is a plain carbon steel that.

In principle, we can do this provided we are we have annealed microstructure. If you have a normalized microstructure, it will not represent the composition of the alloy. Basically, because whatever you get from the lever rule; because basically you are going to use the lever rule and from the lever rule, find out the composition of the alloy. So, if you have seen the microstructure, certain area fraction you assume it to be equivalent to a volume fraction. Convert it into a weight fraction and finally, put it find out what will be the alloy composition corresponding to that weight fraction using lever rule.

This is how, you have to actually calculate. If you want to really calculate, based on the microstructure what is the alloy composition? Because the microstructure only gives you the area fraction. It will not give you weight fraction and what the lever rule gives you only weight fraction. It does not give you any area fraction or volume fraction. Is not it? So, you need to correlate these things. So that finally, you will get the weight fraction and that, if you want to do it. The lever rule can be applicable only under equilibrium conditions, because it is coming from the equilibrium phase diagram. Is it not?

So, that means you need to use a very slow cooling conditions to be able to use a lever rule. So, if I have used a normalizing and got microstructure normalized steel microstructure, it will not correspond to… Let us say 0.6 percent carbon steel if I take, in principle lever rule tells me how much of ferrite? 0.6 percent carbon steel; how much is the ferrite? 25 percent; it is one fourth; three fourth. Because 0.6 by 0.8 that is what it is. So, if it is about 75 percent of eutectoid mixture and 25 percent of ferrite. And if I now take the same steel and do normalizing 0.6 percent carbon, you will not see 25 percent of ferrite.

You may see 15 percent; you may see 20 percent depending on the cooling rate that we have adopted. And if I see 20 percent and go back and calculate from the lever rule and then, say 20 percent. Because it is 20 percent, its composition is let us say 0.5 or so. You are doing an error. The actual composition if you do chemical analysis, it will not correspond to that. The reason is I have taken normalized steel. In a normalize steel, the volume fraction of the ferrite phase will not correspond to what is decided by the phase diagram.

So, these are the minor issues, which you have to be very clear, when you want to actually utilize the microstructures to understand the composition; because many people may not directly have analysis facilities chemical analysis facilities to do a composition. So, we tend to do simply see the microstructure and get an idea of the composition and also one more thing that you need to know is the microstructure also is decided by… If the alloy is the not a binary but a presence of alloying elements present. The moment you have alloying element present, the whole phase diagram changes.

The phase diagram is driven by the presence of ternary and quaternary alloying elements. For example, you must have seen, when you are talking about alloy steels, how chromium changes the phase diagram, how nickel changes the phase diagram. Particularly, we must have seen the gamma zone shrinks, when you add chromium. Basically, because it stabilizes ferrite; it does not stabilize austenite. So, it dissolves into it wants to dissolve into ferrite; it does not want to dissolve into austenite. So, austenite shrinks infact around 18 percent or so. The austenite region is very very small. What is the austenite region in our phase diagram?

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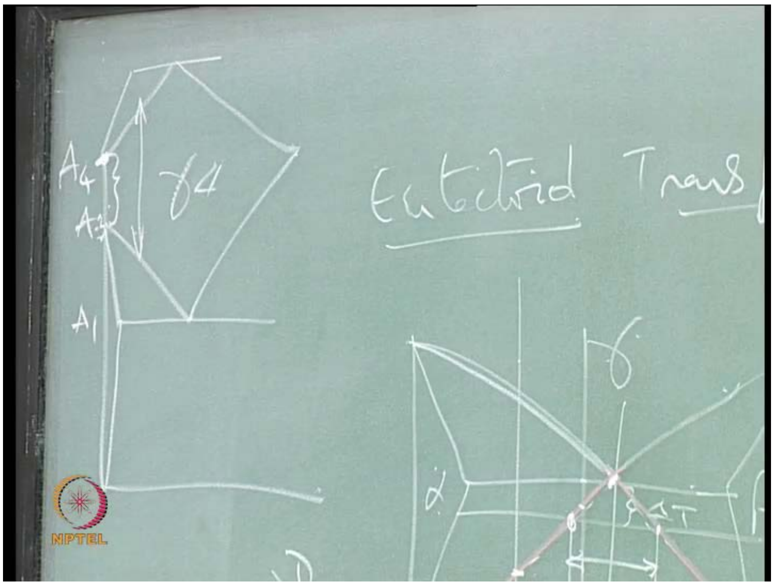

Austenite region, if I want to draw; so if I have this is the one; this is austenite domain and what do you call this temperature? What is it called? What happens there? Gamma to delta. What is gamma to delta transformation temperature denoted as?

A 4.

A4, it is A 4 temperature; this is A 3, this is A 1, and obviously, there must be A 2 somewhere. So, that A 2 is the magnetic transition. So, that A 4 and that is what you can actually see, how carbon stabilizes the austenite. The moment you add carbon, you see A 3 is decreasing; A 4 is increasing.

As I add more and more carbon, you see the A 3 is decreasing and the A 4 is increasing. That means, a pure iron gamma is stable only between these two 9, 10 to 14 1400; whereas, in around 0.8 percent carbon are so you could see it extend from here to here. Ofcourse, this is 0.18; this is 0.8. So, there is a difference. So, there is scale its difference. So, once I know that, basically these two are increasing tell you that this field is increasing and that means, whenever a phase field expands. That means that particular element stabilizes that structure.

And this is basically, because the carbon wants to be in a FCC structure rather than in a BCC structure. So, that is why the alpha is tries to depress; gamma, I mean delta also it tries to depress. So, the stability range of the delta gets reduced and the stability range of alpha gets reduced. So, that you get gamma. So, that clearly tells you that carbon is a gamma stabilizer and the same thing now, if you add the chromium you will see that this whole thing shrinks. This whole gamma field phase shrinks to a small domain like this, around 18 percent or so or 20 percent, if you add.

So, that is how a nickel if you add it, this goes down further which is the austenite stabilizer. So, any austenite stabilizer will keep on shifting it down. So, you can know from that, how each one stabilizes; anyway. So, these things also are decided by the composition, you need to know about it. The second issue is, if I take this alloy and then, I am cooling it. The alpha that is forming here, what would be the morphology of that alpha? People usually see that, whether the alpha forms there or alpha forms here.

For example, if I have under cool this alloy to this temperature, still you will have an alpha forming; because it has not yet intersected this. So, you will have an alpha forming, even if you have cooled it below the eutectoid temperature. If you have under cool it below eutectoid temperature not simply cooled it. So, if you have simply cooled it by that time, eutectoid transformation is already over. So, if you under cool it to that temperature, you can still get alpha. What would be the morphology of that alpha, when compared to that alpha which is coming somewhere here?

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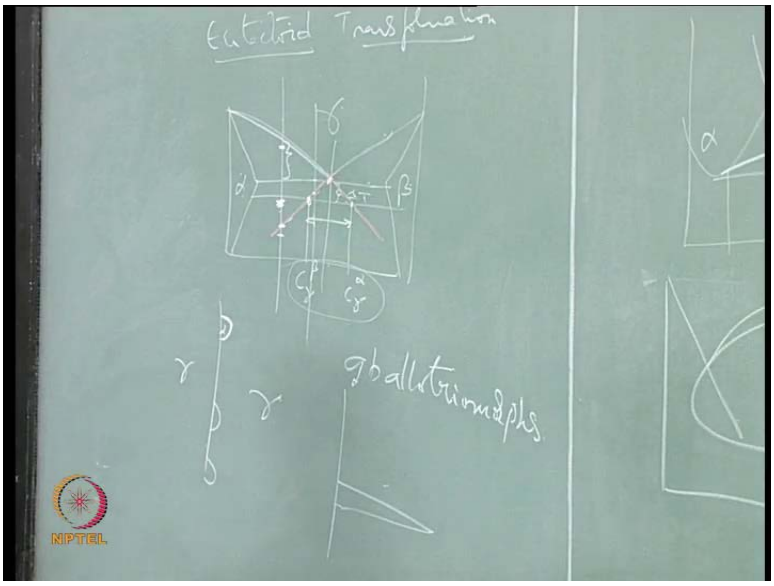

And this is what, a large number of people have studied and they see that the moment you are the transformation temperature is closer to the A 3, you will see gamma is there. A grain boundary between gamma, you see this kind of precipitation which is what is called grain boundary allotriom (( )) which are more or less spherical kind of precipitates of alpha. The moment, the under cooling is very high and let us say somewhere here, you see the alpha coming out. In such a case, you see the alpha comes out as platelets with semi coherent inter phases between the ... Here, these are all incoherent inter phase. The reason is if you look at the growth rates of a semi coherent inter phase and coherent inter phase.

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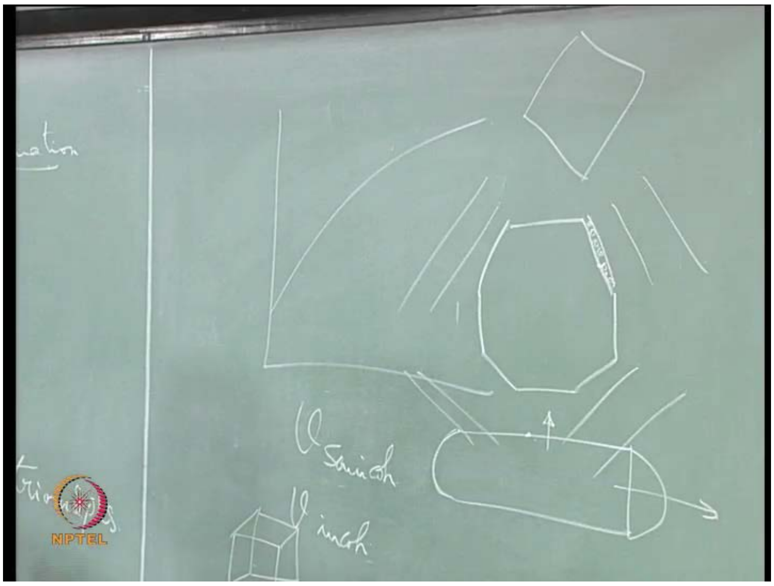

If you look at the ratio of these inter things, the rate of growth; if you think of velocity of a coherent of a semi coherent inter phase and velocity of an incoherent inter phase. At high temperature, the velocity of both of them is more or less similar; because the diffusion rates are very high at high temperatures. But at lower temperature, you would see that, the incoherent inter phase moves faster. Because if a semi coherent or a coherent inter phase has to move, it has to maintain the coherency with the matrix before it grows. For example, whenever you look at a plate, a plate this is a coherent or semi coherent inter phase; whereas, this side is an incoherent inter phase.

You will see that this grows faster when compared to this. This incoherent inter phase can grow faster. So, that is why plates grow laterally, they never grow. So, the thickness of the plate does not change during precipitation reaction. For example, if I take a aluminum copper, you get theta double prime disk. The disk thickness rarely changes much. But the disk grows laterally; because of this disk growth is much faster than this growth; because if you want to this growth to occur, then you need to maintain the coherency continuously. The continuity of the atomic planes has to be maintained; whereas, here there is no need of such thing.

So, as a result you see that, at higher temperatures if you take the ratio of… If I plot the ratio of … Yeah because whenever you are talking of coherency, coherency is on any particular plane. When nucleation is taking place, plate is nucleating on a matrix. Let us say, theta double prime or something like that nucleating on a aluminum matrix; so, the

111 plane of the aluminum matrix acts as the nucleating site. So, it is only that plane which is oriented with respect to some plane of the precipitate, whatever plane it is. And once you see that, so this is the precipitation and then if you look at both the sides, it has to maintain coherency with another plane of this.

There may not be a plane, which is perpendicular to this; which is also coherent with the matrix with the precipitate, you may not see that. So, you will see that as long as one plane is enough, there is an orientation relationship; the interphasial energy is reduced. So, over all the system wants to reduce the interphasial energy. So, it reduces; so the nucleation has taken place. Now, how about the growth? So, there is no such coherency on this sides and because there is no coherency on this side. Any of the plate like precipitate will always have incoherent component plus either a coherent or a semi coherent.

Initially, it could be coherent; slowly it may lead up to semi coherent and finally, you would see that, this whole thing grows as a plate. This retains that, instead of getting converted into incoherent inter phase here itself. If it has grown this way, then it will finally get converted in to incoherent; instead the system would prefer a growth in this direction. That is why in many cases, you see plates coming. Even for example, aluminum silicon alloy, this silicon grows like needles. Even in the liquid to solid transformation also, you see this happening; not only solid to solid. That is because whenever you look at any precipitate.

For example, a nucleus has formed something like this or may be initially, you imagine it is like a cube. Because silicon is silicon let us say is cubic in nature; diamond cubic. So, some nucleus has formed. This nucleus has to now grow. The growth depends on the rate of atoms joining from the liquid to that silicon and not only it depends on that; it also depends on which plane of these needs less number of atoms for it to grow. Each plane here, if I consider we have talked about Woolf plot that there is shape for the nucleus. Once I consider there is a shape for the nucleus, there are so many planes which are covering that nucleus.

Out of all these planes, whichever plane needs less number of atoms per unit area for it grow; that will be the one which grows faster. That means those are the planes which are loose packed planes will grow faster than the close packed planes. The close packed planes need more number of atoms. So, at a given temperature, the liquid is supplying atoms. When the liquid is supplying atoms, the rate at which it can supply atoms depends on diffusivity. And once this diffusivity is fixed at a given temperature, then it cannot preferentially give more atoms to a particular plane and less atom to another plane.

So, the plane which receives, all the planes are receiving same number of atoms at in a given time. But this particular plane, which needs less number of atoms for it grow by a small delta x will grow. Because if you have planes like this, something like this; let us say if this plane needs less number atoms for it to grow. Once all the necessary number of atoms come and joins that, within a particular unit area. It grows by a small delta x, which is may be one mono layer. Similarly, within that particular time, the number of atoms that are required for this particular plane to completely get filled. If they have not reached, then this will not grow. Is it not?

So, like that you will see the growth morphology; sometimes will be entirely different from the morphology. So, that means this keeps on growing; these atoms keep on growing whereas, this does not grow. Finally, you will see that the growth morphology may become like this. The nucleus morphology is this; the growth morphology is this. These planes have grown; whereas, these planes have not grown. So, you can see that, the at the nucleus stage, what you see is different from after certain growth. This is because certain planes grow faster. Same thing here also, you will see some of the inter phases grow faster and some of the inter phases grow slower and that is what, you see in this particular case.

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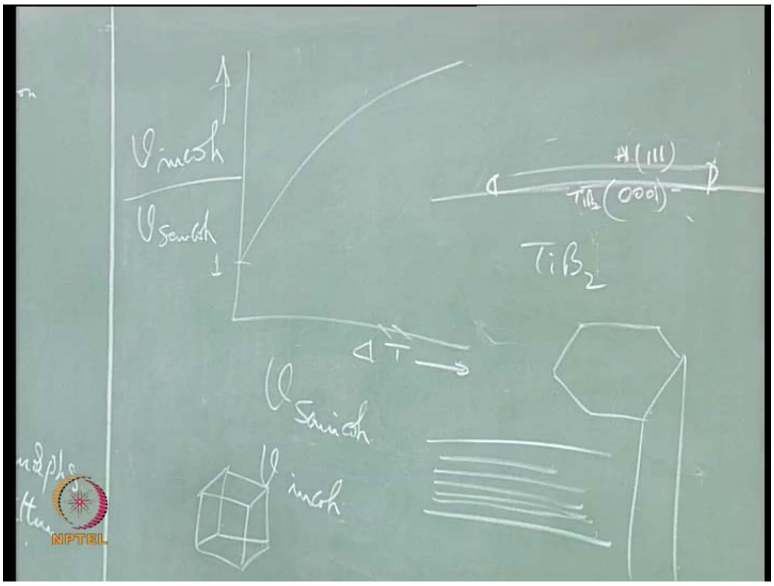

So, if I plot the velocity of incoherent by velocity of semi coherent. This ratio increases with higher under cooling. At very small under cooling, which is close to let us say 0. Then, the ratio is close to around 1. At high temperatures, the ratio is close to 1; that means both incoherent and semi coherent can grow. And once that happens, you end up in this kind of thing, spherical. All directions, the growth is like a isotropic growth whereas, here at at high under cooling, such isotropic growth is not possible. And as a result, you see that one direction it grows, which is incoherent direction.

The semi coherent directions, it does not grow and as a result, it gets elongated. So, that is what, we call it as Widman status structure. So, Widman status structures come in to picture, basically because of that. So, this is how, you would see the different types of alpha coming out of the gamma just because of the growth rates being different at different temperature; because of the necessity for a semi coherent inter phase to maintain the coherency. An incoherent inter phase need not have to maintain coherency. So, as a result it can grow faster. That is what I am saying, if you looking at a plate like this, the nucleation is only on this.

The nucleation is not here. If the nucleation is not here, that means this plane if I have certain planes in this; let us think of a case like aluminum alloys, we are Ti B 2 to nucleate aluminum, which is called inoculants. Have you heard of it? Grain refiners, we called them. So, once I add Ti B 2 to aluminum, aluminum is nucleating on Ti B 2. Where does it nucleate? Basal plane Am I right? The basal plane of Ti B 2 has the same orientation, the same atomic packing with respect to the 1 1 1 plane of aluminum. So, that means if I think there is plate of Ti B 2, on which aluminum has nucleated.

Aluminum will nucleate like this in such a way that, this is the 1 1 1 plane of aluminum on the 0 0 0 1 plane of Ti B 2. Once it has nucleated, what about the other planes? If this is aluminum, which has nucleated on it; let us say a small one which has nucleated on this particular plate. Once I think that is, it is nucleated I need I should actually draw it this way. This is nucleated on a large surface is a Ti B 2 surface, on which it has nucleated. What about this plane? This plane, which is perpendicular to 1 1 1 plane. The plane which is perpendicular to 1 1 1 plane is some other plane. Is not it? You can look at the crystallography and find out, what is that plane which is perpendicular to 1 1 1 plane?

If you look at that particular plane that and considered what is perpendicular to the 1 1 1 plane of the T i B 2? What is the perpendicular to the 1 1 plane of T i B 2, which are basically the side planes. Basal plane and the planes, which are side; which are basically of the type of 1 1 1 1 bar 1 0 1 type of planes, you can look at them. So, if you look at those planes, that means what I am talking about is if you think of, so these planes this planes are perpendicular to this plane. These planes are the ones, which are perpendicular to this particular plane and whatever plane which is perpendicular to 1 1 1, that may not have a coherency. These two planes may not have coherency.

As a result, on the two sides of this aluminum, you may not see any coherency unless both of them are cubic. If this is cubic and that is cubic, this is FCC that is FCC, then the plane which is perpendicular to 1 1 1 plane will definitely have… Because both are FCC you will have this happening. If not, the plane which is perpendicular to 0 0 0 1 and the plane that is perpendicular to 1 1 1 may not have coherency and a as result, this particular part will be incoherent. So, it will nucleate there and then, grow; but in actually in liquid to solid transformation, the aluminum which is nucleating on the Ti B 2.

You do not see it as a growing as plates, basically because of this; because this nucleation is and growth is taking place at temperature just below the melting point. At that temperature, the diffusivities are very, very high. So, as a result ultimately aluminum grows like a grain, equal as grain on Ti B 2 particles; whereas if you look at, at low temperature where such a nucleation has taken place. At a low temperature, it is easier for it to grow in this direction rather than it to grow in that direction. If it has to grow in that direction continuously, you need to have maintenance of the coherency. Clear?

This is what, you would see in… With that more or less, we have understood eutectoid. Yes, that is not in eutectoid. Please, eutectic. Yes, tell me. The reason is, because it is a solid state. You have to consider the strain also here; you have to consider the strain also and if you consider the strain and particularly, because there you are talking of nucleation of of schematic inside the liquid. Is not it? The schematic if you consider a eutectic composition; let us say 4.3 percent composition, it is forming inside the liquid. Here, it is forming inside the gamma. If it is forming inside the gamma, then you need to consider, what is that that interphasial energy?

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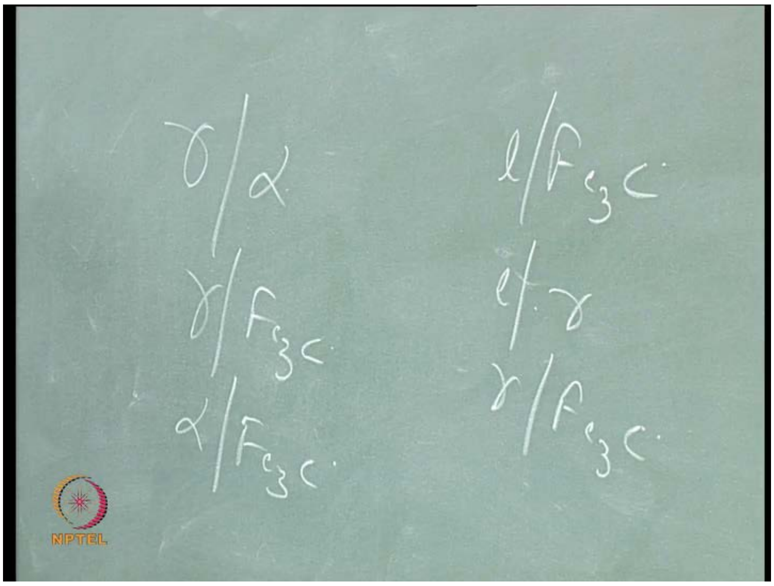

So, that means I have to consider the gamma alpha interphasial energy and gamma schematized interphasial energy and then alpha schematized interphasial energy. These are the three interphasial energies, I should bother. In that case, what I have bothered is liquid Fe 3 C interphasial energy and then, liquid whatever the alpha gamma interphasial energy and gamma Fe 3 C interphasial energy. These are the three there. So, there if you consider that this was lower and here, if you consider this is lower, that is it as simple as that. It looks at which is the lower interphasial energy and so whichever is lower, it would happen.

So and one more thing is, once it nucleates how it grows; it grows as chunky or not. It depends on the growth rate that is what I have just now said. Some of them would grow as needles. There are number of cases in eutectics, where you have needle like microstructure or rod like microstructure. Aluminum silicon is an example. There also examples of lamellar eutectics. Is not it? You have aluminum copper is an example, where you have lamellar eutectics; so, where you see this kind of a situation coming into a picture. This is lower energy, then this or this so as a result you would see that happen.

That means two solid interphasial energies are lower. In aluminum copper, two solid interphasial energies are lower. Basically, because both are you know if you consider, one is a theta phase which is actually orthorhombic and alpha is the cubic. But there is if you look at high temperature, the interphasial energies of both of them are close to each other. And as a result, they can come out as infact, theta is not even orthorhombic; it is

BCT; it is tetragonal; sorry orthorhombic is Fe 3 C. So, tetragonal structure has co-relation with a cubic structure excepting the C axis, everything else is similar. Infact you take a cubic, pull it; that is how, you get a tetragonal structure.

So, that is how you can see that, there is a possibility of alpha theta lamellar eutectic coming. That is why, you get a lamellar eutectic whereas, here you cannot get lamella eutectic; as gamma and orthorhombic, there is no co-relation, and compositions are also entirely different. As a result, you will not see a possibility of a lamellar there. Whenever when do you say interphasial energy is lower, inter phase energy lower when the structures are similar. For example, Ni 3 Al nucleating in a gamma phase, we say it forms very easily; because the structures are similar. If the structures are not similar, the interphasial energies cannot be.

Infact if you see that eutectic, you would see both the phases that are coming out are cubic let us say. You would definitely invariably end up in lamellar eutectics, because interphasial energies will be similar. For example, if I take a copper silver eutectic, copper silver eutectic will definitely be a lamellar. If you go by you can see, because both are FCC and there are no inter metallic in that phase diagram at all. This is a simple eutectic phase diagram. If you want to call it as devour eutectoid, in principle bainite is a devours eutectoid. Is not it? There is no lamellar structure. So, it depends on where you are having the… yeah actually interesting question.

We do not call it actually as eutectoid; because it is not happening. Infact if that is the case, schematic formation eutectic reaction, where schematic is coming out first and the remaining liquid is becoming gamma. Can I call this as eutectic? Because both of them are not forming at the same time. So, ultimately you have to see the final microstructure, liquid giving two solids finally. So, if you look at it, the gamma giving two solids finally; one is ferrite, another is this one. Infact we talked about it even in peritectic reaction. Peritectic reaction is finished; the moment peritectic layer is formed.

There is no more peritectic. You think of this case of you know liquid plus delta giving you gamma, the moment gamma forms on the delta; where is the peritectic reaction? But at the same time, liquid and gamma will delta will vanish and finally, you will get gamma. How does it happen? We said this happens because of the diffusion. Do I call this as peritectic reaction? So, we called it as peritectic transformation and solve our problem. Is not it? So, that is what it is. So, in principle a bainite can be called as you know eutectoid. It is only schematic; it is only how you look at it.

Finally, you have; but if you look at it, the whole transformation is first the gamma is giving you alpha of a same composition, which is polymorphous type of a transformation and from polymorphous transformed alpha, schematic is precipitating out; that is what is that actual transformation that is taking place. So, in principle it is from that point of view, it is not actually eutectoid. But it is the ultimately the mixture is alpha plus gamma schematic mixture with a different morphology. Anything else? Ferrite is an active nucleant. Yes, (( )) again again it is not very difficult to understand.

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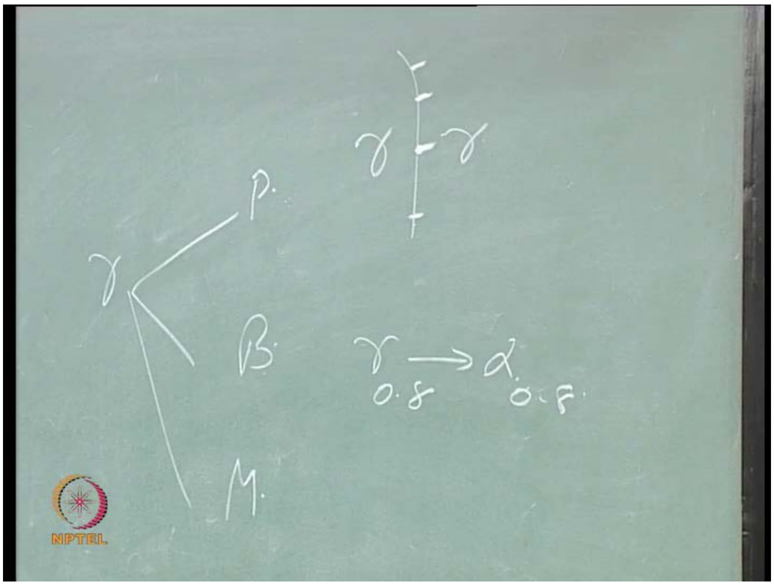

See, if you look at the two transformations, gamma giving pearlite; gamma giving bainite; in gamma giving bainite, we have under cool to such an extent that the carbon is not able to move much. Because carbon is not able move much, long range diffusion short range possibly it can still do; long range diffusion is difficult. As a result, gamma first becomes alpha of the same composition polymorphous. Here, the carbon is able to move. So, what happens is carbon would come to the grain boundaries of gamma and form a small region of enriched gamma carbon. Once that enriched carbon regions reaches that of 6.67 or 25 atoms out of 75.

Initially, the number of atoms of carbon is very, very small. It is 0.8; that is weight percent. If I convert it into atomic percent, you would get still possibly about 2 or 3; so, 3 atoms or 4 atoms of carbon in 100 atoms of iron, something like that. That is the initial state. From that initial state, more carbon atoms come from a different region and so that, that region becomes 25 atoms. Then, you can easily form nucleus. That is the reason why and carbon can happily move at a temperature closer to 723. So, that is the reason why schematize is a nucleus there; whereas, here schematize is not able to be a nucleus.

Carbon is not able to move and if you further go down, you get Martensite. Because even that small short range diffusion that we can see there is not able to happen and finally, you get a diffusion less. Tomorrow, we will talk about this. Anything else? Yes, you introduce certain let us say vacancies; there are people who have done by… Somebody may say that all are inter related. For example, if I do irradiation or if I do ball milling, I do lot of let us say I take gamma austenite powder and do ball milling. I introduce lot of defects in to it, dislocations. So, the moment there are large number of dislocations, you have what is called pipe diffusion.

Have you heard of it? Pipe diffusion is diffusion through the dislocation core. Dislocation is nothing but a row of vacancies. Is not it? So, if you introduce large number of dislocations, then the diffusivity can be increased. So, that is the reason why many in ball milling, many reactions that can happen; that usually are supposed to happen at high temperature can happen at low temperature. Another thing is you make them nano; you introduce large surfaces and reactions can happen. For example, a simple example I can give you aluminum plus Fe 3 Fe 2 O 3. If you take iron oxide plus aluminum, there is a reduction reaction.

What do we call this? Metalothermic aluminothermic reduction; thermite reaction people call it. This usually occurs above the melting point of aluminum. The same thing can occur much below the melting point of aluminum, the moment you make both of them nano. This is what, we see in ball milling. You take iron oxide and plus aluminum mixture, put it into a ball mill; ball mill it. After sometime, at room temperature you see this reaction happen at low temperatures. Just because you have enhanced the diffusivities; because surface diffusion can happen now, and dislocation diffusion through dislocations can happen; these are other ways will stop now.