Texture in Materials
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Module - 08
Texture evolution during phase-transformation
Lecture - 43
Various Orientation Relationships and Variants

Good afternoon everyone and this is the 3rd lecture of module 8, that is, Texture evolution during phase-transformation and we will continue to understand the orientation relationship and the texture that is that evolves during this phase transformation processes. So, this is lecture number 43, where we will understand the Various Orientation Relationship and the Variants formation and its selection.

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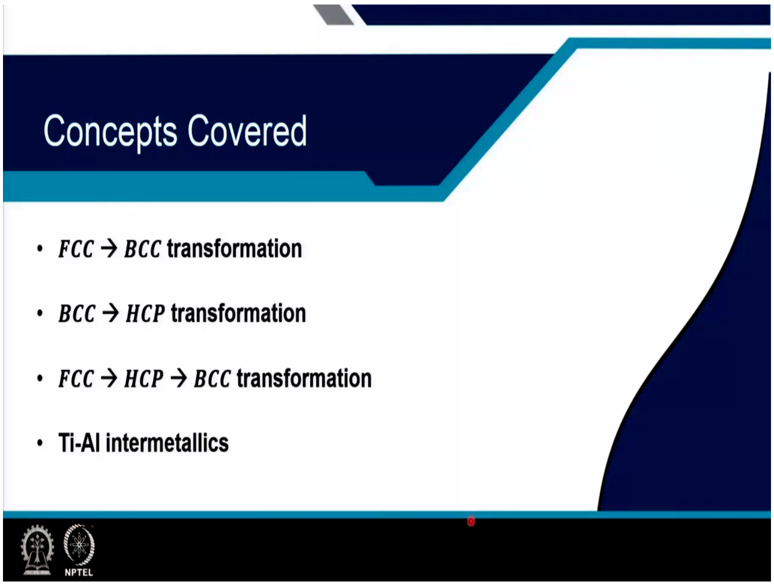

Concepts Covered

- FCC → BCC transformation
- BCC → HCP transformation
- FCC → HCP → BCC transformation
- Ti-Al intermetallics

So, the concepts that will be covered is that we will continue will understanding the FCC to BCC transformation further and we will also observed the BCC to HCP which is hexagonal close packed materials phase transformation. Then there is a new type of transformation. Of course, not new, it usually happens in materials which phase transform from face centered cubic to hexagonal close packed and then to BCC sometimes.

And, then we will take an example of titanium aluminide and see how phase transformation takes place in this particular scenario of dual phase structure.

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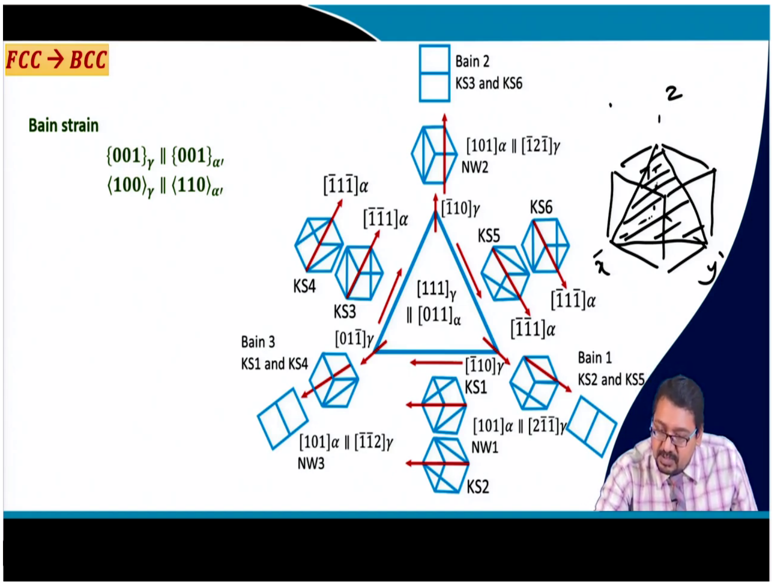

FCC → BCC

Bain strain

$\{001\}_{\gamma} \parallel \{001\}_{\alpha'}$

$\langle 100\rangle_{\gamma} \parallel \langle 110\rangle_{\alpha'}$

So, if we look into the face centered to the body centered cubic phase transformation we started with understanding the Bain strain relationship and the Bain strain basically says that the 001 plane of the gamma austenite FCC phase becomes parallel to the 001 of the alpha prime or alpha martensite.

And, on the other hand the 100 directions if we consider 001 as z, so, the 100 directions that is the x and y directions 100 or 010 directions of the gamma becomes parallel to the 110s of the alpha prime, right. And if we look into the 111 plane of the gamma phase that is FCC we can see that the 111 plane could be shown in terms of a triangle and why it is.

So, if you would like to see I will just show it with a small drawing a small diagram. So, if we draw a crystal structure something like this, right and if this is the x, this is the y and this is a z the 11 plane forms something like this a body diagonal, right. Sorry, something like this and this plane will contain two three directions, right and it contains the three 110 directions which are these, right.

So, if we look into this 111 plane we will be able to see this three 110 directions. So, one 110 and then another 110 and then another 110 out of which at least two is independent, right. Now, in case of Bain strain if it happens the relationship of Bain strain is something like this. So, it can have three different variants that can form and one variant of Bain strain is this one.

So, and another variant of Bain strain is this one and another variant of Bain strain is this; so, Bain 1, Bain 2 and Bain 3 as I said. So, it is in between KS1 and KS4, right and this one is in between KS2 and KS5 that is the Kurdjumov-Sachs relationship 2 and the Kurdjumov-Sachs relationship 5 whereas, this one is between the Kurdjumov-Sachs relationship 3 and the Kurdjumov-Sachs relationship 6.

So, the Bain strain relationship forms a orientation change from the parent to the product phase that is from the parent FCC phase to the product phase if we look into visualized visualize it in this manner it will form something like this, right.

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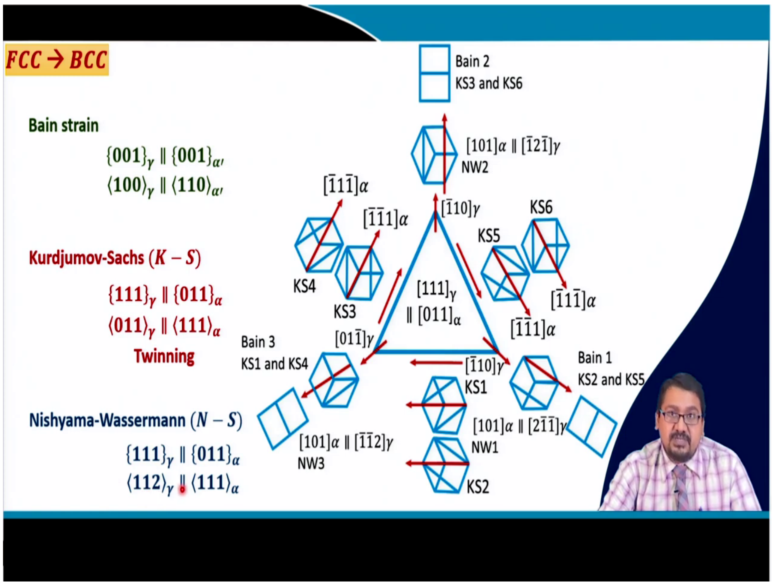

FCC → BCC

Bain strain

$\{001\}_\gamma \parallel \{001\}_{\alpha'}$

$\langle 100\rangle_\gamma \parallel \langle 110\rangle_{\alpha'}$

Kurdjumov-Sachs ($K-S$)

$\{111\}_\gamma \parallel \{011\}_\alpha$

$\langle 011\rangle_\gamma \parallel \langle 111\rangle_\alpha$

Twinning

Nishyama-Wassermann ($N-S$)

$\{111\}_\gamma \parallel \{011\}_\alpha$

$\langle 112\rangle_\gamma \parallel \langle 111\rangle_\alpha$

Now, if we look into the Kurdjumov-Sachs transformation as we have seen that Kurdjumov-Sachs has three possible alternatives for sorry, four possible alternative for the 111 plane and each 111 has three 110s. So, 3 into, 4 into 3, 12 and then twinning which makes it 24 variants. And, then let us look into 111 plane and it has three 110s and let us look into this 110.

So, 1 bar 10 so, this has one orientation relationship with this parent crystal and that is KS1 and then the twinning 1 which is KS2 and for this 011 bar that is this direction it has KS3 and then the twinning one KS4 and for this 1 which is 1 bar 1 0 this particular direction for which it is having KS5 and KS6. Now, one can see that in all this situation, the 111 plane actually becomes parallel to the 110 planes of the alpha BCC phase whereas, the 110 direction of the gamma becomes parallel to the 111s of the alpha phase that is BCC.

On the other hand, we have observed the Nishyama-Wassermann relationship; the Nishyama-Wassermann relationship is slightly different from the Kurdjumov-Sachs relationship. In this case the difference is that the direction of the gamma austenite which is FCC become is 112 which becomes parallel to the closest packed direction of the alpha.

So, apart from 110, the 112 is also the closest packed one of the closest packed direction in case of FCC mentioned. So, this orientation ideally forms at this situation where this 1 one 2 or this 1 one 2 family of direction. So, there are three 112 family of direction corresponding to a single 111 plane forms this relation which is basically you see Nishyama-Wassermann relationship 1, Nishyama-Wassermann relationship 2, Nishyma-Wassermann relationship 3.

So, for a single plane there is three relationships. So, there are four 111 planes and so, 4 into 3, 12. In case of Kurdjumov-Sachs there are 6 alternatives for a single 111 plane because there are three 110s and including twinning so, 6 into 4, 24 orientation relationship that can form from FCC to BCC.

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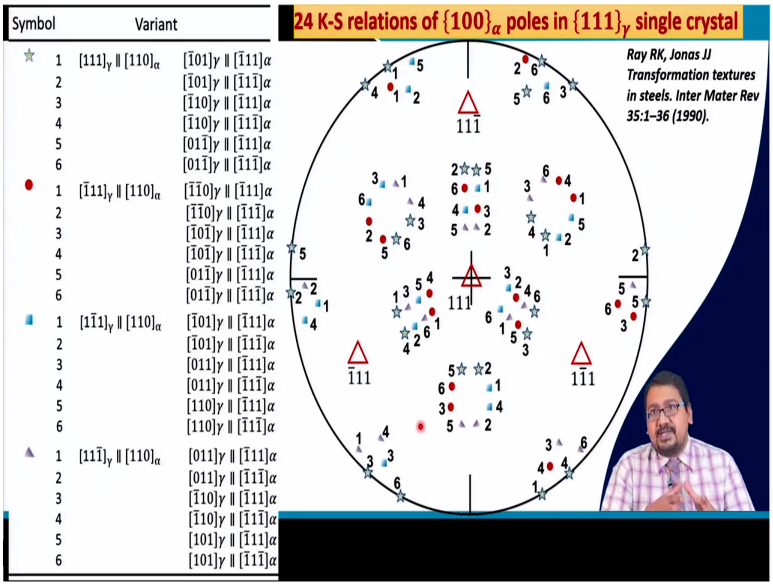

24 K-S relations of $\{100\}_\alpha$ poles in $\{111\}_\gamma$ single crystal

Symbol		Variant	
☆	1	$[111]_\gamma \parallel [110]_\alpha$	$[\bar{1}01]\gamma \parallel [\bar{1}11]\alpha$
	2		$[\bar{1}01]\gamma \parallel [\bar{1}1\bar{1}]\alpha$
	3		$[\bar{1}10]\gamma \parallel [\bar{1}11]\alpha$
	4		$[\bar{1}10]\gamma \parallel [\bar{1}1\bar{1}]\alpha$
	5		$[01\bar{1}]\gamma \parallel [\bar{1}11]\alpha$
	6		$[01\bar{1}]\gamma \parallel [\bar{1}1\bar{1}]\alpha$
●	1	$[\bar{1}11]_\gamma \parallel [110]_\alpha$	$[\bar{1}\bar{1}0]\gamma \parallel [\bar{1}11]\alpha$
	2		$[\bar{1}\bar{1}0]\gamma \parallel [\bar{1}1\bar{1}]\alpha$
	3		$[\bar{1}0\bar{1}]\gamma \parallel [\bar{1}11]\alpha$
	4		$[\bar{1}0\bar{1}]\gamma \parallel [\bar{1}1\bar{1}]\alpha$
	5		$[01\bar{1}]\gamma \parallel [\bar{1}11]\alpha$
	6		$[01\bar{1}]\gamma \parallel [\bar{1}1\bar{1}]\alpha$
■	1	$[1\bar{1}1]_\gamma \parallel [110]_\alpha$	$[\bar{1}01]\gamma \parallel [\bar{1}11]\alpha$
	2		$[\bar{1}01]\gamma \parallel [\bar{1}1\bar{1}]\alpha$
	3		$[011]\gamma \parallel [\bar{1}11]\alpha$
	4		$[011]\gamma \parallel [\bar{1}1\bar{1}]\alpha$
	5		$[110]\gamma \parallel [\bar{1}11]\alpha$
	6		$[110]\gamma \parallel [\bar{1}1\bar{1}]\alpha$
▲	1	$[11\bar{1}]_\gamma \parallel [110]_\alpha$	$[011]\gamma \parallel [\bar{1}11]\alpha$
	2		$[011]\gamma \parallel [\bar{1}1\bar{1}]\alpha$
	3		$[\bar{1}10]\gamma \parallel [\bar{1}11]\alpha$
	4		$[\bar{1}10]\gamma \parallel [\bar{1}1\bar{1}]\alpha$
	5		$[101]\gamma \parallel [\bar{1}11]\alpha$
	6		$[101]\gamma \parallel [\bar{1}1\bar{1}]\alpha$

Ray RK, Jonas JJ Transformation textures in steels. Inter Mater Rev 35:1–36 (1990).

Professor RK Ray and Professor John Jonas worked on it in 1990, to show this 24 Kurdjumov-Sachs relationship and they have they showed it forming a stereographic projection which relates the initial orientation of the gamma phase with respect to the final orientation of the alpha prime you know martensite or alpha ferrite structure.

If we look into this stereographic projection, what we are doing here is looking on a crystal of an FCC structure that is the gamma austenite having its zone axis which is 111. On the basis of the symmetry of this 111 representing the zone axis 111 for an FCC gamma austenite crystal which is at the center, the symmetry decides the positions or the poles of other 111s present in the gamma austenite FCC, right.

When the phase transformation takes place 2 alpha prime or alpha 24 way orients sorry variants form and each variant will have a certain position with respect to this 111. So, the 100 poles of the alpha BCC phase forming during phase transformation from FCC to BCC has been demonstrated in this paper and is discussed in detail.

If we look into this stereographic projection which is basically showing the 100 poles with respect to the initial crystal structure, so, it is a pole figure which represents the 100 poles of the alpha phase with respect to the parent crystal structure having 111 at the center.

One can observe that because there are 24 variants which are possible and if we calculate if we find out right we can have the 111 planes which are parallel to 110 of alpha. So, 111 of gamma parallel to 110 of alpha, 1 bar 11 the another 111 plane of gamma parallel to another 110 of alpha. So, maybe we have not written it properly but it is another 110 of alpha and then 11 bar 1 plane of gamma, the third one closest packed plane of gamma, with another closest packed plane 110 of alpha and the fourth 111 bar of gamma parallel to another 110 plane of alpha.

So, in this way if we see for each plane there are 6 variant, so, 6 4 times 24 variants and the position of each variants for example, we can see this is a triangle and this is the say for example, triangle and the variant number 2 is here whereas, variant number 5 is here, the variant number 3 is here, right. In case of say for example, a square the variant number 6 is here, the 3 is here, 2 is here, 5 is here, 1 is here, 4 is here and then it may come multiple times.

And, so, you see if 24 Kordjumov–Sachs relationship occurs then the single orientation of the parent grain when converts equal into 24 Kurdjumov-Sachs relationship the deformation may lead to the phase transformation following deformation may lead to form a random orientation or at least a weak orientation, right.

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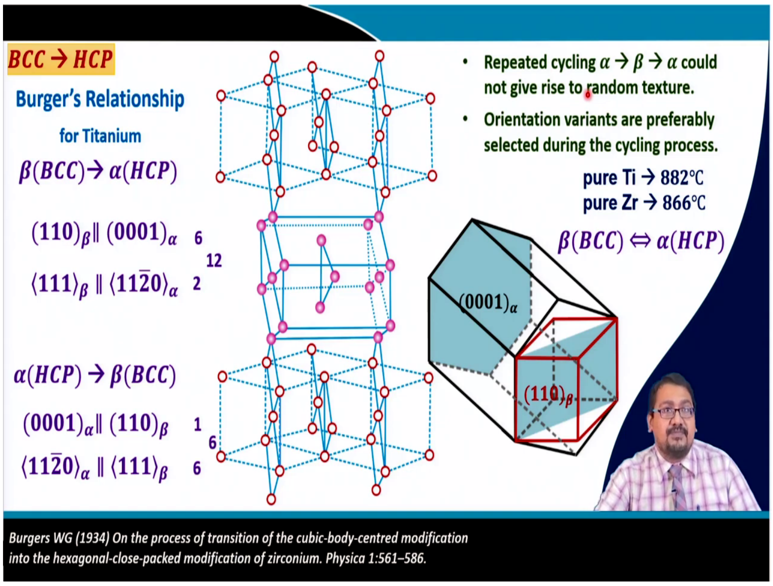

BCC → HCP

Burger's Relationship

for Titanium

$\beta(BCC) \rightarrow \alpha(HCP)$

$(110)_\beta \parallel (0001)_\alpha$ 6

12

$\langle 111\rangle_\beta \parallel \langle 11\bar{2}0\rangle_\alpha$ 2

$\alpha(HCP) \rightarrow \beta(BCC)$

$(0001)_\alpha \parallel (110)_\beta$ 1

6

$\langle 11\bar{2}0\rangle_\alpha \parallel \langle 111\rangle_\beta$ 6

- Repeated cycling $\alpha \rightarrow \beta \rightarrow \alpha$ could not give rise to random texture.
- Orientation variants are preferably selected during the cycling process.

pure Ti → 882°C
pure Zr → 866°C

$\beta(BCC) \Leftrightarrow \alpha(HCP)$

Burgers WG (1934) On the process of transition of the cubic-body-centred modification into the hexagonal-close-packed modification of zirconium. Physica 1:561–586.

But, this is usually not possible because the deformation or the intrinsic structure of the material due to previous processing will have certain stress state associated with it which leads to certain variants which are selected during the deformation and thereby not all 24 variants form, but only few 2 or 3 or sometimes only 1 variant developed leading to a strong texture from a if there is a strong texture in the parent grain it leads to a strong texture in the product phase too.

Now, the another kind of transformation is the BCC to HCP that is the body centered cubic to hexagonal close packed material transformation and this relationship is known as the Burger's relationship. And, such a kind of relationship is was observed in titanium where the beta BCC transform into the alpha HCP.

And, as you have seen that during solidification of your titanium the initial molten metal solidifies into BCC structure which is the beta structure and then it phase transforms when it cools down to BCC to HCP structure following the relationship that is 110 of beta that becomes parallel to the 001 of alpha, right.

Now, on the other hand the 111 of beta becomes parallel to 112 bar 0 of alpha. We can see that this transformation basically follows the root where the closest packed plane of the beta BCC phase becomes parallel to the closest packed plane of the alpha HCP structure whereas, the closest packed direction of the BCC becomes parallel to the closest packed direction of the alpha HCP structure.

Now, in case of the 110 plane of BCC there are 6 alternatives. So, in a crystal structure there are 6 different 110 planes and so, there are 6 alternative and in each 110 plane there are two 111 directions. So, 6 alternative for the 110 plane and 2 alternatives for the 111 direction makes a total of 12 variant possible. If we look into this figure we can see that initially the microstructure is something like this shown by the dotted lines and this dotted lines represent the body centered cubic structure and we have shown nearly 3 crystal structure 1, 2 and 3, right.

And, when we try to arrange to see how it is forming a hexagonal close packed structure we can just connect those crystals sorry, connect those atoms and make this BCC structure into an hexagonal close packed structure something like this, right. On the other hand, if we look into the hexagonal close packed structure like this with this plane which is 001 of the HCP alpha this can convert into the BCC structure like this 001 plane becomes parallel through the closest packed plane which is 110 or beta and the crystal structure may look something like this.

Now, the Burger's relationship shows that when BCC converts into HCP there could be 12 variants that can develop. On the other hand, if from the HCP structure the material is heated up, so that the phase transformation takes from HCP to BCC. Then you see that from the 001 alpha for the alpha phase that is the closest packed plane of HCP becomes parallel to the closest packed plane of BCC during the heating process and the phase transformation.

There is only one alternative because there is only one basal plane present in the material whereas, each basal plane has 6 alternatives of 112 bar 0s and which can develop you know the 112 bar 0 which is the closest packed direction becomes parallel to 111 which is the closest packed direction in case of BCC. So, there are 6 alternative. So, total 6 variants can form when HCP is heated to phase transform to BCC.

So, if we can think that ok if we you know cool down the material from the BCC to HCP there are 12 variants that will form and then if we heat up again and 6 variant and then we cool up and then 12 variant in this way if we do it multiple times then there is a possibility to get a lots of variant and finally, a random texture. But, as I said earlier that such kind of situation does not such kind of experimentation merely gives any change in the texture in the material or make the texture weaker.

A stronger initial texture in the parent grain leads to the you know strong texture in the product grain, and there is always a selection of certain variants that will form and it will continue to form even after repeated heating and cooling. So, this is a situation that we usually show we see in case of titanium and zirconium type material where the phase transformation temperatures for titanium is 882 degree centigrade and for zirconium it is 863 degree centigrade.

One can see by doing this very simple experiment by heating the material in a furnace and taking it to the BCC and then cooling it again and taking it to the HCP and then heating and cooling and heating and cooling to observe that is there any change in the microstructure of course, it will be, but how much it is going to affect the texture.

So, repeated cycling from you know alpha to beta to alpha or you know beta to alpha does not give a random texture. Orientation variants are preferably selected during the you know this cycling process. So, this is a work done by Burger's in 1934 and this is the you know link to the paper.

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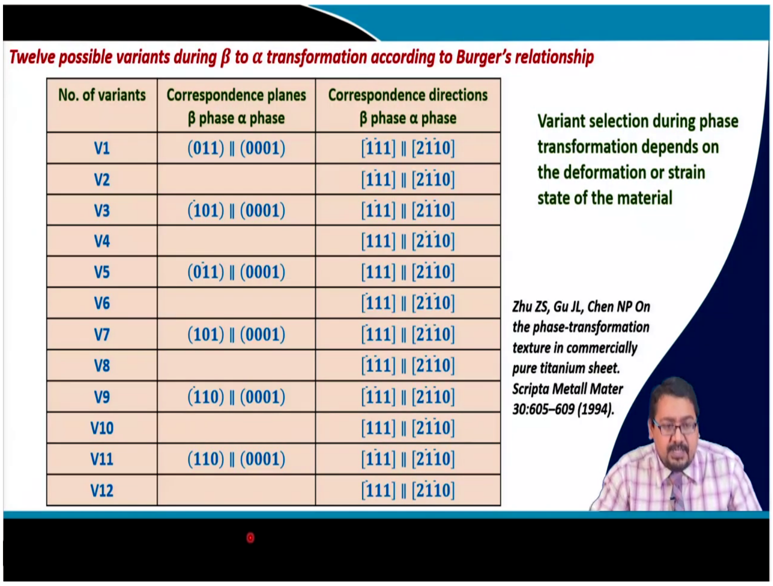

Twelve possible variants during β to α transformation according to Burger's relationship

No. of variants	Correspondence planes β phase α phase	Correspondence directions β phase α phase
V1	(011) ∥ (0001)	[111] ∥ [2110]
V2		[111] ∥ [2110]
V3	(101) ∥ (0001)	[111] ∥ [2110]
V4		[111] ∥ [2110]
V5	(011) ∥ (0001)	[111] ∥ [2110]
V6		[111] ∥ [2110]
V7	(101) ∥ (0001)	[111] ∥ [2110]
V8		[111] ∥ [2110]
V9	(110) ∥ (0001)	[111] ∥ [2110]
V10		[111] ∥ [2110]
V11	(110) ∥ (0001)	[111] ∥ [2110]
V12		[111] ∥ [2110]

Variant selection during phase transformation depends on the deformation or strain state of the material

Zhu ZS, Gu JL, Chen NP On the phase-transformation texture in commercially pure titanium sheet. Scripta Metall Mater 30:605–609 (1994).

And, as I said that there are 12 possible variants that are you know that occurs when the material transform from BCC to HCP, and there are 6 variants which forms which can which is possible when the transformation takes place from the HCP to the BCC. So, here is a table which shows the 6 12 possible variants for you know the 6 different 110 planes of BCC.

So, 011, 101, 11 bar 0, 101 bar, 110 this is 1 bar 01, this is 101, this is 1 bar 10, 110. So, various 6 110 planes are becomes parallel to the 001 which are the closest packed planes in case of the BCC and HCP and their corresponding directions are shown here. So, there could be 12 possible variant.

So, variant selection during phase transformation depends upon the deformation and the strain state in the material and therefore, it will not give a random texture if you do it repeatedly rather it will give a strong texture corresponding to the strong texture of the initial state parent phase, right. So, this is table is taken from the paper by Zhu et al in Scripta Metallurgical and Materials Engineering in 1994.

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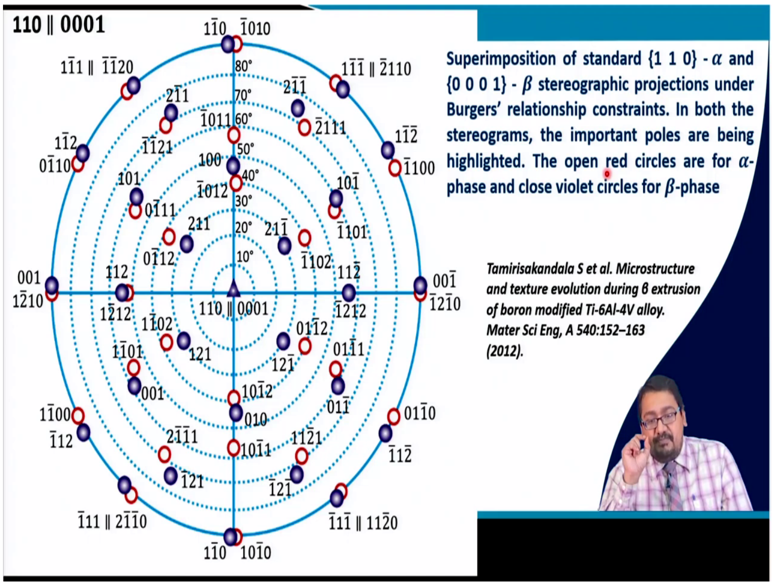

110 ∥ 0001

Superimposition of standard {1 1 0} - α and {0 0 0 1} - β stereographic projections under Burgers' relationship constraints. In both the stereograms, the important poles are being highlighted. The open red circles are for α-phase and close violet circles for β-phase

Tamirisakandala S et al. Microstructure and texture evolution during β extrusion of boron modified Ti-6Al-4V alloy. Mater Sci Eng, A 540:152–163 (2012).

So, whenever we are looking into the BCC to HCP transformation, we will see that apart from the ideal orientation relationship where we say that the closest packed planes becomes parallel to the closest packed plane in case of during the transformation from a BCC to HCP and the direction to the direction, but you will see that because of the large difference in the symmetry of the crystal structure between BCC and HCP, the positions of the transformed planes are not always exact or not always ideal.

So, here we are showing a standard 110 you know stereographic projection for the alpha phase which corresponds to the standard 001 stereographic projection for the HCP phase. So, if we look into this and if 110 remains parallel to the 001 for this situation, the other transformed you know poles are slightly deviated for all the other cases, right. So, you see that the poles corresponds to the BCC which transforms into HCP are not exactly over each other, but they are slightly deviated.

So, in both the stereogram the important poles are highlighted, the red circles are basically showing the hexagonal close packed structures poles and the violet circles are for the BCC beta phase, right. This is a work done by Tamirisakandala in paper Microstructure and texture evolution during beta extrusion of boron modified titanium aluminum vanadium alloy and is given in you know Material Science and Engineering in 2012.

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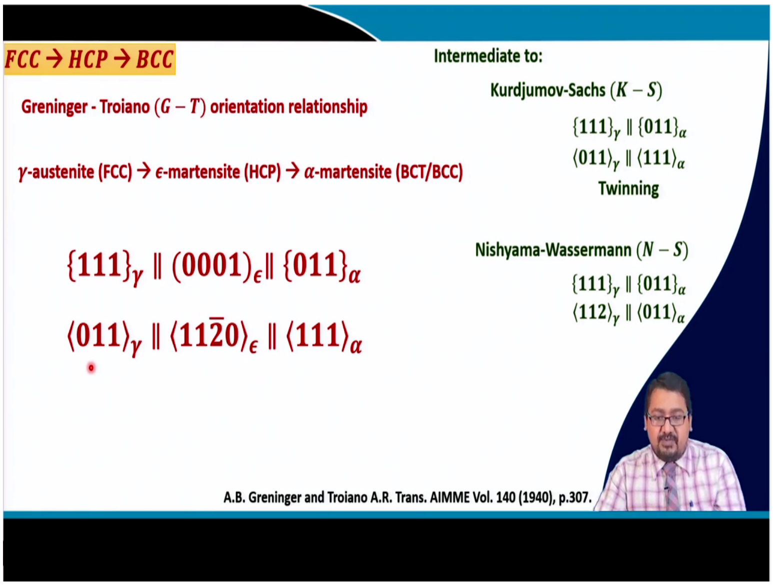

$FCC \rightarrow HCP \rightarrow BCC$

Greninger - Troiano $(G-T)$ orientation relationship

γ-austenite (FCC) → ϵ-martensite (HCP) → α-martensite (BCT/BCC)

$$\{111\}_\gamma \parallel (0001)_\epsilon \parallel \{011\}_\alpha$$

$$\langle 011\rangle_\gamma \parallel \langle 11\bar{2}0\rangle_\epsilon \parallel \langle 111\rangle_\alpha$$

Intermediate to:

Kurdjumov-Sachs $(K-S)$

$\{111\}_\gamma \parallel \{011\}_\alpha$

$\langle 011\rangle_\gamma \parallel \langle 111\rangle_\alpha$

Twinning

Nishyama-Wassermann $(N-S)$

$\{111\}_\gamma \parallel \{011\}_\alpha$

$\langle 112\rangle_\gamma \parallel \langle 011\rangle_\alpha$

A.B. Greninger and Troiano A.R. Trans. AIMME Vol. 140 (1940), p.307.

So, if we look further there are various kinds of transformation other than this. One of the important transformation that has been observed in case of FCC to BCC is which is the Kurdjumov-Sachs to Nishyama and Nishyama-Wassermann. Other than this intermediate to this is the FCC to HCP to BCC and this transformation is known as Greninger and Troiano orientation relationship G-T orientation relationship this transformation was given in 1940 and here is the reference to that.

So, in this transformation it has been shown that in case of certain kind of steels you know mostly austenitic stainless steel that is gamma austenite FCC, the martensite does not stands firm directly to BCT, rather it forms an epsilon martensite which is basically HCP. So, under that transformation scenario, what happens? That the closest packed plane 111 of the FCC structure becomes parallel to the closest packed plane of the HCP structure of the martensite that is epsilon martensite and, that is 001.

And, then on further cooling this epsilon martensite which is HCP transform into BCC that is the then it becomes parallel to the 0011 plane of the alpha prime or alpha, right. On the other hand, the direction that is the 011 of FCC closest packed direction becomes parallel to the 112 bar 0 direction of the epsilon martensite which is closest packed direction and which becomes parallel to 111 of the BCC martensite which is the alpha prime or alpha martensite at the lowest temperature. So, such relationship also exist.

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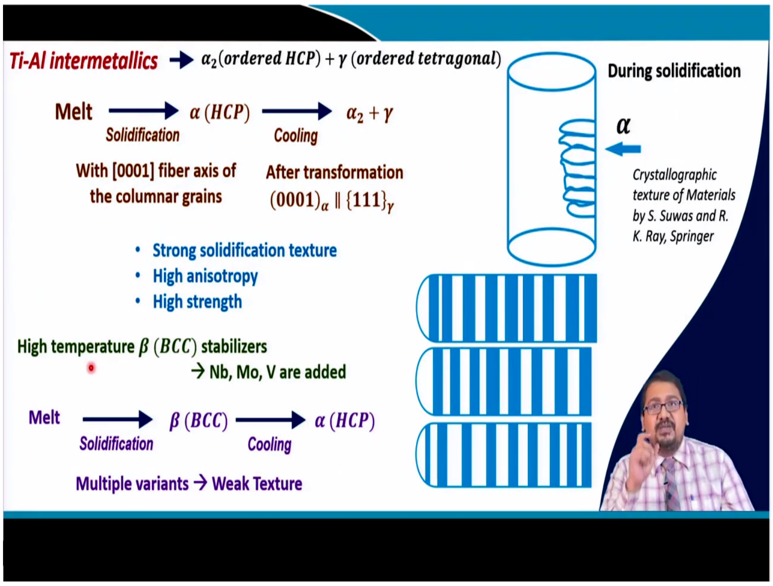

Ti-Al intermetallics → $\alpha_2 (ordered\ HCP) + \gamma\ (ordered\ tetragonal)$

Melt $\xrightarrow[\text{Solidification}]{}$ $\alpha\ (HCP)$ $\xrightarrow[\text{Cooling}]{}$ $\alpha_2 + \gamma$

With [0001] fiber axis of the columnar grains

After transformation $(0001)_\alpha \parallel \{111\}_\gamma$

- Strong solidification texture
- High anisotropy
- High strength

High temperature $\beta\ (BCC)$ stabilizers
→ Nb, Mo, V are added

Melt $\xrightarrow[\text{Solidification}]{}$ $\beta\ (BCC)$ $\xrightarrow[\text{Cooling}]{}$ $\alpha\ (HCP)$

Multiple variants → Weak Texture

During solidification

Crystallographic texture of Materials by S. Suwas and R. K. Ray, Springer

Importantly, we will give I will try to give you an example of titanium aluminide intermetallic which shows a different kind of phase transformation while the material is cooled down or after the solidification. You see, that if the material is solidified in the ingot after the chilled zone a columnar grain growth always exist in the material.

And, this is from the book Crystallographic texture of Material by Suwas and Ray, published in Springer. And, you can see that as the growth of the material while it starts to solidifies from the melt occur that is the growth of the alpha grain occur. In titanium intermetallic the melt solidifies directly as an a hexagonal close packed structure.

So, it forms directly a alpha phase which is hexagonal close packed instead of a BCC phase that occurs in normal titanium or titanium alloys. So, while it forms first HCP structure and it has a columnar growth, the columnar growth as usual will occur in the direction of the most loosely packed you know direction or the columnar growth plane will be the loosely packed plane.

So, there will be a computation of both which state that whether the growth will occur in the atom will sit on the loosely packed plane or the growth will occur in the most loosely packed direction whichever will win, but in this case you see the 001 fiber axis is shown to be parallel to the columnar grain. And, while this alpha HCP is further cooled down, so, while this alpha HCP is cooled down further then alpha 2 ordered HCP structure and gamma ordered tetragonal structure basically forms.

So, of during this transformation the alpha and the alpha 2 has the same crystal structure. So, we are not bothered about that let us consider that when alpha converts into gamma. So, the transformation takes place in such a way that the closest packed alpha HCP that is the 0001 plane becomes parallel to the closest packed direction of the ordered tetragonal gamma phase that is 111 gamma.

So, while the material transforms the material converts from alpha to alpha 2 and gamma phases, something like that while the solidification is columnar and the columnar growth is taking place in this direction, this leads to the formation of a two phase microstructure with a strong texture along the columnar direction. So, strong solidification texture forms in this region. It has therefore, high anisotropy and this material because of this texture also has a very high strength.

So, what it can be done that if you do not want this strong texture high temperature beta stabilizers are basically introduced in this melt. These are like niobium, molybdenum or vanadium could be added and now, instead the melt solidifies directly to hexagonal close packed structure it solidifies to the BCC structure and while it cools it forms alpha HCP.

So, it has been shown that under this situation and because of the addition of this niobium molybdenum vanadium this inoculant heterogeneous nucleation develops and the material in a liquid phase may go to the constitutional super cooling region leading to formation of multiple variants and weaker texture.

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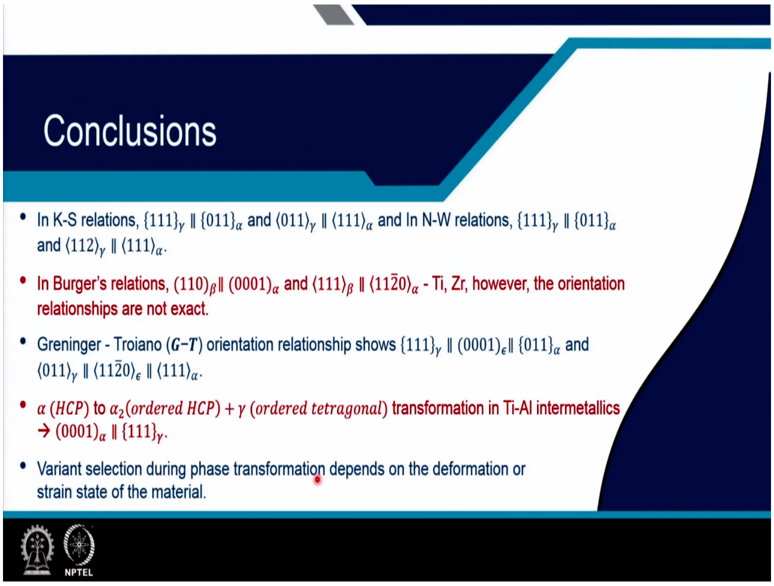

Conclusions

- In K-S relations, $\{111\}_\gamma \parallel \{011\}_\alpha$ and $\langle 011\rangle_\gamma \parallel \langle 111\rangle_\alpha$ and In N-W relations, $\{111\}_\gamma \parallel \{011\}_\alpha$ and $\langle 112\rangle_\gamma \parallel \langle 111\rangle_\alpha$.
- In Burger's relations, $(110)_\beta \parallel (0001)_\alpha$ and $\langle 111\rangle_\beta \parallel \langle 11\bar{2}0\rangle_\alpha$ - Ti, Zr, however, the orientation relationships are not exact.
- Greninger - Troiano ($\boldsymbol{G-T}$) orientation relationship shows $\{111\}_\gamma \parallel (0001)_\epsilon \parallel \{011\}_\alpha$ and $\langle 011\rangle_\gamma \parallel \langle 11\bar{2}0\rangle_\epsilon \parallel \langle 111\rangle_\alpha$.
- $\alpha\ (HCP)$ to $\alpha_2 (ordered\ HCP) + \gamma\ (ordered\ tetragonal)$ transformation in Ti-Al intermetallics → $(0001)_\alpha \parallel \{111\}_\gamma$.
- Variant selection during phase transformation depends on the deformation or strain state of the material.

Therefore, in this lecture, we can conclude that in the Kurdjumov-Sachs relationship the closest packed plane 111 of the FCC becomes parallel to the closest packed 001 of the BCC and the closest packed direction 011 of the FCC becomes parallel to the 111 direction closest packed direction of BCC. In Nishyama-Wassermann relationship the plane remains the same, but the direction now is the closest packed direction 112 of the FCC becomes parallel to the 111 of BCC.

In case of Burger's relationship the beta which is BCC structure, the closest packed plane is 110, the closest packed direction is 111 becomes parallel to 001 closest packed plane of cp HCP and 112 bar 0. The closest packed direction of HCP respectively. For titanium and zirconium it has been shown experimentally, the orientation relationship in any of this are not I am does not form exact it deviates from the ideal orientation relationship.

There is another orientation relationship which is known as Greninger and Triano orientation relationship GT orientation relationship which shows that the relationship between FC FCC to HCP to BCC structure and the relationship is the closest packed plane of FCC 111 becomes parallel to the closest packed plane of HCP that is 001 and becomes closest packed plane of BCC that is 011 as the material cools down.

On the other hand, the closest packed directions 011 of FCC becomes parallel to 112 bar 0 of epsilon martensite which is HCP becomes parallel to 111 of the BCC structure.

So, in case of titanium aluminide intermetallics, the solidification of the melt directly occurs from the melt to HCP alpha phase and then it converts to alpha 2 which is an ordered hexagonal and a gamma which is an ordered tetragonal structure, following the relationship where the closest packed plane of the high temperature HCP phase that is 0001 becomes parallel to the closest packed plane of the gamma order tetragonal structure that is 111.

Finally, variant selection during phase transformation depends on the deformation and the strain state of the material therefore, it is very difficult to obtain you know a random texture by doing phase transform by using phase transformation technique a initially strong parent texture will lead to the formation of the final strong product texture and that is all for the phase transformation for this course.

Thank you very much.