Imaging

- Imaging in the TEM
- Diffraction Contrast in TEM Image
- HRTEM (High Resolution Transmission Electron Microscopy) Imaging
- STEM imaging
For high energy diffraction we make these simplifications: (1). The deviation vector is very nearly parallel to the z-axis so s is simply equal to sz. (2) the quantity \( N_z c \) is the crystal thickness, t. (3) we can ignore the widths of the diffracting columns along x and y, a useful expression for the shape factor intensity as a function of the scalar deviation parameter, s, is

\[
S \ast S(s) = \frac{\sin^2(\pi st)}{(\pi s a_z)^2}
\]

\[\therefore \text{the intensity for the diffraction, } I_g, \text{ is}\]

\[I_g = \left| \psi \left( g, s \right) \right|^2 = \left| F_g \right|^2 \frac{\sin^2(\pi st)}{(\pi s a_z)^2} \]
summary

• The intensity of diffracted wave, $I_g$, depends on the deviation parameter, $s$, and the thickness, $t$.

• The diffraction contrast in a BF or DF image is then the variation of $I_g$ for the diffracting columns located in the $x,y$-planes of the sample.

• In two-beam conditions, the intensity of the transmitted beam, $I_0$, and diffracted beam, $I_g$, are complementary. With the incident intensity normalized to 1:

$$I_0 = 1 - I_g$$

• This **kinematical theory is valid when** $I_g << I_0$, i.e. $s$ is large and the diffraction remains weak through the depth of the sample.

• When $s \to 0$, the kinematical results is incorrect. A dynamic theory is necessary.
Extinction Distance from Howie-Whelan equation

From dynamic theory

\[ I_g = \left( \frac{\pi}{\xi_g} \right) \frac{\sin^2 \left( \pi \cdot s_{\text{eff}} \cdot t \right)}{(\pi \cdot s_{\text{eff}})^2} \]

This equation is valid even when the diffracted beam is strong and kinematical theory is not valid.

The effective deviation parameter, \( s_{\text{eff}} \), defined as:

\[ s_{\text{eff}} = \sqrt{s^2 + \xi_g^{-2}} \]

The extinction distance, \( \xi_g \), defined as:

\[ \xi_g = \frac{\pi V}{\lambda F_g} \] where \( V \) is the volume of the unit cell,

\( \lambda \) is the electron wavelength, and \( F_g \) is the structure factor for the diffraction \( g \).
The Phase-Amplitude Diagram

• Construction

• Application

1. A polar representation of a complex number of modulus unity

2. Using this polar representation to develop a graphical scheme, known as “phase-amplitude diagram” to evaluate the diffracted wave

\[ \psi\left(\vec{g}, s\right) = F_g \sum_{r_g} e^{i2\pi \vec{s} \cdot \vec{r}_g} \]

3. The diffracted wave as seen above equation is a sum of vectors in the complex plane, and each one represents the relative phase factor of a diffracted wavelet.

4. These phase factors of wavelets are summed from unit cells at progressively greater depths, \( r_g \), in the specimen.
Polar representation of a complex number, \( \exp(i \vartheta) \), on the unit circle

\[ z = \cos \vartheta + i \sin \vartheta \]
The diffracted wave

\[
\psi(g, s) = F_g \sum_{r_g} e^{i2\pi s \cdot r_g}
\]

Incident wave, \( \psi_0 \)

When the Laue condition, \( \Delta K = g \), is satisfied exactly, and \( s = 0 \), all of the wave scattered by each layer of the specimen add in phase with each other. This situation is richer when there is a deviation from the Laue condition and \( s \) is equal to zero.

Wavelets diffracted from unit cells at increasing depth, \( r_g \), and \( s \) along the vertical \( z \) direction.
The diffracted wave

\[ \psi(\mathbf{g}, s) = F_g \sum_{r_g} e^{i2\pi s \cdot r_g} \quad (a) \]

Vectors represents the individual terms, \( F_g \exp(i2\pi s.r_g) \), in equation (a), \( \psi(s) \)
The diffracted wave

\[ \psi(g, s) = F_g \sum_{r_g} e^{i2\pi s \cdot r_g} \quad (a) \]

\[ \psi(s) = e^{i0} + e^{i2\pi 1 s \cdot r_1} + e^{i2\pi 2 s \cdot r_1} + e^{i2\pi 3 s \cdot r_1} + e^{i2\pi 4 s \cdot r_1} \]

The phase-amplitude diagram is a graphical sum of complex exponentials, and is equivalent to equation (a). The total diffracted wave, \( \psi(s) \), is constructed by adding the vectors tail-to-head as shown below figure. The angle between successive vectors in the sum is \( 2\pi s \cdot r_1 \). If \( s=0 \), all the vectors add colinearly and a strong diffracted is constructed.
The intensity of diffracted wave is

\[ I = \psi^* \psi = \text{Re}(\psi)^2 + \text{Im}(\psi)^2 \]

I is real. The intensity does not depend on the orientation of the wave in complex space, but only its modulus.
Diffraction Contrast from Strain Fields

Here is a list of the important variables in both kinematical and dynamical diffraction theory:

- $F\left(\mathbf{g}\right)$: Structure factor of unit cell (length of vectors in PAD)
- $\xi_g$: Extinction length; $t$: Specimen thickness (number of vectors in a PAD)
- $\Delta k$: Diffraction vector ($\Delta k = \mathbf{g} + s$); $g$: Reciprocal lattice vector
- $s$: Deviation parameter; $r$: Atom centers (affects angle of vectors in PAD)
- $R$: Atom centers in a perfect crystal ($R = r_g + r_k$, $r_g$ and $r_k$ are lattice and basis vectors respectively)
- $\delta \mathbf{r}$: Displacement off the atom center ($\mathbf{r} = R + \delta \mathbf{r}$)

Spatial variation in these variables (e.g. an $x$-dependence) can produce diffraction contrast in an image. For example:

- $F$: $dF / dx$ causes chemical (compositional) contrast;
- $t$: $dt / dx$ causes thickness contours; $s$: $ds / dx$ causes bend contours
- $\delta \mathbf{r}$: $d\delta \mathbf{r} / dx$ causes strain contrast
Consider the displacements in atom positions, $\delta r$, caused by strain fields around defects. We decompose $r$ into components from the lattice vectors, basis vectors, and distortion vectors:

$$r = r_g + r_k + \delta r$$

Then the diffracted wave is

$$\psi_g = \frac{\psi_0}{\tilde{\xi}_g} \int_{-t/2}^{t/2} e^{i2\pi \left( s \delta \rightarrow - g \delta \rightarrow \right)} \, dz$$
Edge Dislocation

• There are two pure dislocations, edge dislocation and screw dislocation.

• An edge dislocation is formed by inserting an extra half-plane of atoms in the upper half of the simple cubic crystal. This extra half-plane terminates at the core of the edge dislocation line.

• As shown in figure, a circuit of 5x5x5x5 atoms, known as Burgers circuit, does not close perfectly when it encloses a dislocation line. It does enclose in a perfect cubic crystal or not including the dislocation line.

• The vector from the end to the start of the circuit is defined as the Burgers vector of the dislocation. It is perpendicular to the dislocation line for edge dislocation.
Screw Dislocation

• Burgers vector is parallel to the dislocation line in screw dislocation.

• Around the core of a screw dislocation, the crystal planes form a helix.

• As shown in figure, when we complete a Burgers circuit in the x-y plane, the vector from finish to start lies along z-direction.
Diffraction contrast from dislocation strain fields in TEM images

Strain fields around defects cause the displacements, $\delta \vec{r}$, in atom positions $r = r_g + r_k + \delta \vec{r}$. The diffracted wave is

$$\psi_g = \psi_0 \int_{-t/2}^{t/2} e^{i2\pi \left( sz - g \delta \vec{r} \right)} dz \ldots \ldots (a)$$

• When analyzing diffraction contrast mechanism in TEM imaging, we can consider either variations in $\delta r$, or variations in $s$. This is essentially a choice of reference coordinates.

• The reference coordinates could be a set of sites on a perfect lattice, where we consider displacements, $\delta r$. Alternatively, we can set all $\delta r$ to zero, but consider the bending of crystal planes in the vicinity of defect, causing a spatial variation of $s$.

• This alternative approach is more amenable to a qualitative analysis with Ewald sphere construction and PAD, and is the easiest way to understand the diffraction contrast from dislocations.
Diffraction contrast from dislocation strain fields in TEM images

- As discussed before, the diffraction planes are distorted in the vicinity of the dislocation.
- Near the dislocation line, the diffracting planes bend toward the Laue condition (smaller $s$), or bend away from the Laue condition (larger $s$).
- For the column of crystal on the left side of the dislocation line, near dislocation $s>>0$, and the vectors in PAD form a tight radius.
- For the column to the right of the dislocation line (s is smaller), the vectors in PAD add co-linearly in the regions of the dislocation line.

Qualitative distortions of crystal planes near an edge dislocation, with corresponding Ewald sphere constructions.
The corresponding PAD from a region of perfect crystal and the two vertical columns near the dislocation region.

- The strongest diffraction is from column to the right of the dislocation line (s=0).
- For column too far to the dislocation core, the variation in s approaches the value of perfect crystal, as seen in Fig. (a).
- Fig. (b) shows the diffracted intensity profile versus column position, indicating the dislocation contrast is asymmetrical if s is not equal to zero.
- When you use EDX to examine the chemical segregation to the dislocations, you should know how to locate the dislocation core.

- So images of dislocations are not images of the dislocation core, but are images of the strong variations in s in bent crystal to one side of the dislocation core, about 300-3000 nm away the core.
• Schematic representation of the formation of the diffraction contrast image from an edge dislocation in a crystalline lattice showing the different contrast around the dislocation region.

• The strain fields around the dislocation core could be sufficient to tilt the lattice planes locally into a Bragg angle. A particular set of lattice plane around the dislocation core is strongly diffracted.

• In two-beam condition, BF imaging removes this diffracted beam by OA. So BF image shows the dark contrast in dislocation region.

Inserting OA to exclude the diffracted beam and to form BF image
Dislocation imaging mechanism

Mechanism of diffraction contrast: the planes to the RHS of the dislocation are bent so that they closely approach the Bragg condition and the intensity of the direct beam emerging from the crystal is therefore reduced.
3-D diffraction contrast for dislocation lines

Beam intensity

$$I_0 + I_g = 1$$

Direct non-diffracted (energetic) beam from “perfect crystals” - gives just bright background

Weak beam forming a dark band or line (on a white background - Bright Field Image)
(a): Dislocation in strong diffraction contrast in a metal foil.

(b): Two parallel rows of dislocations extending from top to bottom of the crystal with a thickness about 200 nm.

(c): BF image shows a projected image of (b) and dark contrast of dislocation lines.
The g.b Rule for Null Contrast

If the Burgers vector of the edge dislocation is perpendicular to the active diffraction vector, i.e. $g \cdot b = 0$, there is no diffraction contrast the dislocation. This dislocation is invisible.

- The g.b ruler can be used to determine the Burgers vector of a dislocation (see Lindoors, Phil. Mag. 24, p709, 1971).
- In practice, a dislocation is invisible when $|g \cdot b| \leq 1/3$. 

Edge dislocation g.b rule

- Assume that $\Delta k$ points into the plane of the paper:
  - $\Delta k \cdot b = 0$
  - $\Delta k \cdot b \neq 0$

3-d view

flat

front view

side view

s is constant

s changes near dislocation

bowed out
Schematic diffraction contrast edge dislocations with lines and $b$ in the plane of the sample
Illustration of the using $\mathbf{g}\mathbf{b}=0$ rule to determine the Burgers vector of dislocations
The **g.b** Rule for Null Contrast of Screw Dislocation

- Columns of a crystal near a screw dislocation core.
- The dislocation line is nearly perpendicular to the plane of the paper.
Screw dislocation \( \mathbf{g} \mathbf{b} \) rule

- If the Burgers vector of the screw dislocation is perpendicular to the active diffraction vector, i.e. \( \mathbf{g} \cdot \mathbf{b} = 0 \), there is always no diffraction contrast the dislocation. This dislocation is always invisible.

- For screw dislocation, the displacements \( \delta \mathbf{r} \) are always parallel to \( \mathbf{b} \).
Schematic diffraction contrast screw dislocations
Example of \( g \cdot b \) rule for screw dislocation

- Images of a perpendicular array of screw dislocation in a niobium-stabilized steel, together with the DP in each case.

- only one set of dislocation, \( b=1/2[110] \), is visible. The other set of dislocation, \( b=1/2[1-10] \) is out of contrast, since \( b \) is normal to the operating \( g \) vector (111).

- for \( g=(020) \), two sets of dislocations are visible.
Dislocation loops, dislocation cluster, and dislocation pairs etc.

The interstitial loops with $b=\langle 0001 \rangle$ in zinc

Fe-3\%Si alloy thin foil parallel to (011) slip plane showing the dislocation dipole trails (A) and dislocation jogs (D).
Al-3.5%Mn alloy quenched from 550 °C showing the concentric loops formed by dislocation climb mechanism.

Fig. (a) and (b) shows the dislocation morphology of Fe-3.5% Si alloy subjected to deformed and annealed processing.

(a): Approximately uniform distribution of dislocation of alloy rolled 20 percent.

(b): Formation of small sub-grain by dislocation migration in rolled material annealed 15 min. at 500 °C.
Cont’.

(c): As for (b) annealed 15 min at 600 °C showing dislocation migration forming sub-grains and well-arranged dislocation network.

(b): As for (b) annealed 30 min at 600 °C showing re-crystallization by merging the sub—grains to form large grain, and resulting grain growth.

Fig. (b) (c) show the dislocation moving procedure.
Simulated dislocation TEM image

\[
\begin{align*}
\text{w} &= 0.0; \ A = 3.2; \ z_0 = 4 \ \xi \\
\text{w} &= 0.5 \\
\text{z}_0 &= 3 \ \xi \\
\text{A} &= 1.0 \\
\text{A} &= 5.0 \\
\text{A} &= 15.0 \\
\text{BF} & \quad \text{DF} \\
\text{BF} & \quad \text{DF} \\
\text{BF} & \quad \text{DF} \\
\end{align*}
\]
Simulated dislocation TEM image
Weak-beam Dark-Field (WBDF) imaging of Dislocations

• The WBDF imaging technique is useful making sharp images of dislocation lines, resolving pairs of dislocations, and in relating images to calculations of diffraction contrast.

• To make a WBDF image, the specimen is tilted to a large, positive value of s, where diffraction is weak and the DF image is quite dark over most of the crystal.

• Only near the core of the dislocation are strains large enough to bend the crystal planes into a diffraction with s close to zero.

• The WBDF image shows the diffraction from these severely-bent planes near core of the dislocation.

• There are two difficulties of the WBDF 1). It requires accurate tilting of the specimen and electron beam; 2). The operator is often faced with taking long exposure of relatively dark screen. During this time, the specimen may drift in position, producing blurred image.
Procedure to make a WBDF image

• Orient the specimen in a good two-beam condition and excite the appropriate +g, Fig. (a).

• Tilt the incident beam until the position of the vector +g moves into the position of transmitted beam. In this case, +g becomes weak, so we call it weak beam. ([This is “amateur mistake” in obtaining an axial DF image], Fig. (c) and (d).

• After this tilt of incident beam, the transmitted beam has moved, and the strong diffraction spot is now 3g spot. This is called “g-3g” condition, Fig. (d).

• Form a DF image by placing the OA around the +g spot. This diffracted beam travel along the optic axis, so it makes a high quality axial DF image. It is a weak diffraction, so only a dim image appears on the screen.
The WBDF procedure involves the tilting the illumination beam, not the specimen. So on the screen, the diffraction spots move, but the Kikuchi bands do not, Fig. (b).

After tilting, the row of spots and the Kikuchi bands in DP are positioned as seen in Fig. (b).
Diffraction condition for a WBDF image

(a) Before tilt

(b) After tilt of 2θ

(c) A DP obtained when the specimen is tilted to a suitable orientation for WBDF imaging. Here \( g \) is a (220) reflection and \( 3g \) is strong.

So the \( g-3g \) diffraction:

\[
G_g - 3g = \frac{g^2}{k_0}
\]

or in general

\[
G_g - ng = \frac{n - 1}{2} \frac{g^2}{k_0}
\]
An example of WBDF imaging

- The dislocation at the position (x,y), introducing a displacement, $\delta r$.
- Assume sample is 1000Å.
- We can calculate $s$ for g-3g condition.
- Select A, B, C, D four points to analysis the PAD.

Strain fields around defects cause the displacements, $\delta r$, in atom positions

$r = r_g + r_k + \delta r$

the diffracted wave is

$$\psi_g = \frac{\psi_0}{\xi g} \int_{-t/2}^{t/2} e^{i2\pi \left( sz - g \delta r \right)} dz \quad \text{......(a)}$$

$$s_{g - 3g} = \frac{g^2}{k_0}$$

Or in general

$$s_{g - ng} = \frac{n - 1}{2} \frac{g^2}{k_0}$$
An example of WBDF imaging

• consider the phase factor @ δr=0, the variation of sz with z is linear as shown in Fig. (a).

• for g δr=g δr_x, δr_x is the displacement component in x direction, if g_x>0 and δr_x <0 in the right of dislocation, the change of g δr_x is shown in Fig. (b)
An example of WBDF imaging (cont')

- The change of \((sz-g\delta r_x)\) is shown in Fig. (c) for four column A, B, C, D.
- Fig. (d) shows the Ewald sphere construction for column B.
An example of WBDF imaging (cont’)

- In above PADs, ends of diagrams are marked with cross (+)

- If $s=0.01\text{Å}^{-1}$ by calculating, the column A has a PAD that wraps around 10 circles (thickness is 1000Å).

- Column D has an even tighter curvature for the diffracted wave near the center of the specimen.

- Column B is likely to produce the maximum diffracted wave.

- Column C is bent beyond the condition $s=0$ into $s<0$, so the PAD has a region of curvature that undergoes a reversal.

- The DF image of dislocation is brightest near the position of column B.

- The position of B with respect to dislocation core can be calculate using WBDF analysis and dislocation theory.
An example of WBDF imaging

Figure shows the dramatic improvement in resolution and contrast of the WBDF technique over a corresponding BF image.

(a): BF image in two-beam condition with strong (2-20) diffraction in Si thin foil
(b): g-3g WBDF image with weak (2-20) diffraction.
Ag$_2$Al hcp precipitates in fcc Al-rich matrix in an Al-Ag alloy.

(a): BF image with strong (-111) diffraction. It is difficult to see either the interfacial dislocation network or the dislocations interacting with precipitates.

(b): $g$-3$g$ WBDF image with (-111) diffraction. The individual dislocations in the array and at the intersection are clearly visible as thin white lines. The dislocation separated by only a few nanometers are distinguished.
Phase shifts of electron wavelets across interfaces

- Interfaces between different volumes of materials are 2-D defects.
- The internal interfaces in a bulk material include stacking faulting, grain boundaries, and anti-phase boundaries.
- At all of these interfaces, there is an abrupt change in the phase relationship between the wavelets scattered above and below the interface.
- PAD is handy for understanding the diffraction contrast from these interfaces.
Assume there is a displacement, \( \delta \mathbf{r} \), between the interfaces:

\[
\mathbf{r}_g = \mathbf{r}_g^0 + \delta \mathbf{r}
\]

so the diffracted wave

\[
\psi(\Delta \mathbf{k}) = \sum_{\mathbf{r}} f_{at}(\mathbf{r}) \cdot e^{-i2\pi \Delta \mathbf{k} \cdot \mathbf{r}}
\]

develops the phase factor

\[
e^{-i2\pi \Delta \mathbf{k} \cdot \mathbf{r}} \approx e^{+i2\pi \mathbf{r}_g \cdot \mathbf{s}} \cdot e^{-i\alpha}
\]

where \( \alpha = 2\pi \delta \mathbf{r} \cdot \mathbf{g} \)

A perfect crystal with an interface causing a displacement, \( \delta \mathbf{r} \)
• The first term in equation (a) is the typical of unit cells oriented with diffraction error, s, including those unit cells above the interfaces.

• the second term in equation (a) multiplies the phase factor for every unit cell below the interface, but not for the unit cells above the interfaces.

• the phase, $\alpha$, can be used to specify the diffraction effects of the interfaces

The diffracted wave

$$\psi(\Delta \vec{k}) = \sum_r f_{at}(\vec{r}) \cdot e^{-i2\pi\Delta \vec{k} \cdot \vec{r}}$$

the phase factor

$$e^{-i2\pi\Delta \vec{k} \cdot \vec{r}} \approx e^{+i2\pi \vec{r}_g \cdot \vec{s}} \cdot e^{-i\alpha} \quad (a)$$

where $\alpha = 2\pi \delta \vec{r} \cdot \vec{g}$
• Diffracted ray paths without interface (left) and with interface (right), Fig. (a)

• Relative orientation of phase factor vectors without interface.

• PAD for a crystal without an interface showing close circle, the diffracted wave vector is very small (close zero).

• PAD for a crystal with an interface showing an kink (large diffracted wave vector).
(a). BF image of APBs (Anti-phase boundaries) in ordered Fe3Al using a (100) superlattice diffraction.

(b). DF image

(c). Schematic of APB
Bright field and dark field images for a permutation boundary in BaTiO₃. The operative reflection is g₁₁₀ at 200 kV.
Bright field and dark field images (200 kV) of an inclined grain boundary in Ti. The grain at the top is far from Bragg orientation, whereas the grain at the bottom is in exact Bragg orientation. The thickness fringes change direction abruptly at the boundary.
Moire’ fringes

- A common situation at interfaces between crystals with nearly commensurate interplanary spacings.

- There is a periodic matching of the lattice planes between the two crystals along the interface.

- A kinked PAD is therefore expected for a diffraction plane like (100).

- A set of fringes, known as parallel moire’ fringes, will be observed in a DF or BF image that includes the interface.

- When we tilt the sample (change s) or change the order of the diffraction (change g), we actually change the radius of the circular arc in PAD. The position is shifted, but the spacing remains unchanged.

- The fringe spacing depends on g only.
Moiré fringes reveal the presence of dislocations in a thin film of CoCa grown on a GaAs substrate. The (001) interface lies parallel to the specimen surface. Although the images contain much detail, most of it cannot readily be related to the structure of the defects.
• BF image of Moire’ fringe image in an Al-Cu-Mg-Ag alloy

• Moire fringes are often seen at small semi-coherent precipitates in thin samples.

• Parallel Moire’ fringes are often observed at the precipitate / matrix interface when the precipitate is viewed edge-on along the interface
Stack fault

- A stacking fault exists when two atomic planes are not in their proper crystallographic registry.
- In fcc crystal, the stacking sequence is ABCABC……
- Errors in this sequence such as ….ABCABABCABC….. Known as stacking fault
- Atoms across a stacking fault are shifted off their proper positions by a displacement equal to the Burgers vector of a partial dislocation, such as $b = a/6<112>$ in fcc crystal
- When a region of TEM sample contains one crystal, the stacking fault often extends from the top to the bottom.
- The TEM image of a stacking fault is typically a set of fringes that run parallel to the intersection of the fault with the surfaces of the specimen.
Geometry of stacking fault inclined in a thin sample

(a) Set of stacking faults in AISI 304 stainless steel. (b) Poles of stacking fault planes on [013] stereographic projection and relative orientation of stacking faults in (a)
Bright field and centered dark field images for stacking faults in Cu-15 at% Al, recorded at 200 kV. The operational reflection is $g_{111}$, and the faults on the (111) planes exhibit $\alpha$-fringe contrast.
Simulated $\alpha$-fringe contrast for crystals $A$ and $B$ for simulation parameters $[95, 95, 1, 1, -\frac{2\pi}{3}, 25]$. 
Contrast from Precipitates and other defects

1. Vacancies
   - Single vacancies are not visible in the TEM
   - Condensed vacancy are observed in TEM including vacancy loops, vacancy tetrahedral, and voids

![Contrast from Precipitates and other defects](image)
Contrast from Precipitates and other defects

2. Coherent precipitates

• Second phase particles embedded in a matrix can be classified as 1). Coherent particles, 2). Semi-coherent particles, and 3). Incoherent particles.

• A particle that is coherent with its surrounding matrix material has one to one matching of its crystal planes with those of the matrix, as seen in figure next slide.
in an infinite matrix containing an isotropic misfitting sphere, the displacements u are radial, and depend on the distance r from the center of the particle as
\[ \mathbf{u} = u_r = \frac{gr_0^3}{r^2} \] (outside the particle)
\[ \mathbf{u} = u_r = \varepsilon r \] (inside the particle)
\( \varepsilon \) is related to the misfit \( \delta \) between the unconstrained lattices of the precipitate and matrix by
\[ \varepsilon = \frac{3k\delta}{3k + \left( \frac{2E}{1+\nu} \right)} \]
where \( k \) is the bulk modulus of the particle,
\( E \) and \( \nu \) are the Young's modulus and Poisson's ratio of the matrix
the misfit parameter \( \delta \) is:
\[ \delta = \frac{2(a_p - a_m)}{a_p + a_m} \]
a\( _p \) is the lattice parameter of the precipitate
a\( _p \) is the lattice parameter of the matrix

the diffracted wave
\[ \psi_g = \psi_0 \int_{-i/2}^{i/2} e^{i2\pi \left( z \cdot \mathbf{g} + \mathbf{u} \right)} d\mathbf{z} \]

...(a)
(a) Predicted contrast around a symmetrical misfitting particle showing the diffracted intensity distribution.

(b) The line of no contrast which corresponds to the plane that is not distorted by the strain field of the particle

(b) Experimental image of misfitting particles in Cu-Co alloy showing strain contrast caused by precipitates.
Simulated diffracted contrast image produced by nine spherical inclusions in a 200 nm Cu foil in different diffracted vectors.
HW # 18

How would you distinguish between contrast from the following types of defects? Why does your method work?

a). Moire’ fringes and wedge thickness contours
b). Bend contours and dislocations
c). Moire fringes and dislocations

Due: Nov. 24/08

Next lecture: Nov. 12/08

High resolution TEM imaging