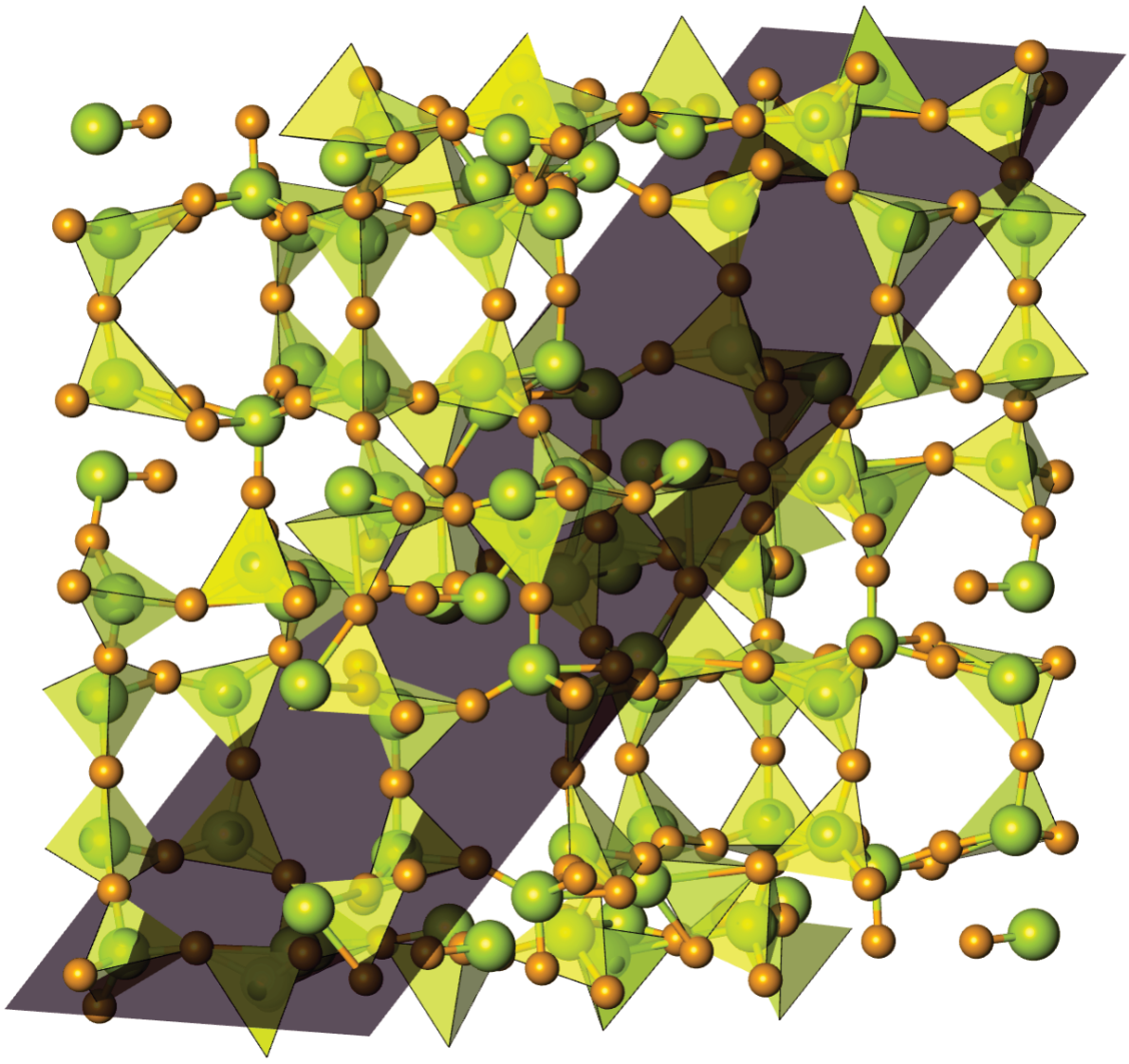
Tempas User Manual

Total Resolution LLC



Contents

Installation

Installing the Tempas Application

Drag the Application Tempas to your application folder

If you already have a previous universal version of MacTempasX/Tempas running properly on your machine, there should be nothing else to do. You are done. If not...

Installing the USB Dongle Driver

Tempas uses a hardware copy protection key which must be installed on your computer. If you already have installed a key for use with CrystalKit, you do not need a second key to run Tempas and you can proceed to the next paragraph describing how to activate the key for running Tempas. Just plug the USB key into an open USB slot on your computer, keyboard or display.

In order to communicate with the USB dongle that contains the license, you must install the driver for the dongle.

There are two different USB dongles that may be supplied. There is a Black/Grey dongle and a Green dongle.

a) Black/Grey Dongle. Execute the application "SentinelSystemDriver" and following the instructions.

b) Green Dongle. Mount the disk image Sentinel_Runtime and follow the instructions.

When Tempas is run for the first time it will put up its installation screen. Enter your name and affiliation as appropriate together.

If you have just changed your computer or installed a new clean version of the macOS or Windows, you must run the appropriate installation program once more to install the driver.

Without the driver in place, the program will not recognize the hardware key and Tempas will run in demonstration mode.ore will install. If you

encounter (under macOS) the message that the software cannot be installed, you must specifically allow it to be installed through the "Security & Privacy" section of the "System Preferences" application. Likewise you may find that the macOS is blocking the running of the kernel extension "Sentinel.kext" which is required for communicating with the USB dongle. Hence you may find that you need to allow the kernel extension to load in the "Security & Privacy" section. A similar thing may happen under Windows where one specifically need to allow the installation of the software.

Introduction to Image Simulation

The best High Resolution Transmission Electron Microscopes (HRTEM) have a resolution approaching 1 Å which sometimes leads to the erroneous conclusion that using an electron microscope, all atoms in a structure can be resolved. However, it is not the inter-atomic distances that matter, but rather the projected distances between atoms seen from the direction of the incident electron. In order to obtain interpretable results, it is necessary to orient the specimen such that atomic columns are separated by distances that are of the order of the resolution of the microscope or larger. This is a condition that very often is difficult to satisfy and often limits the use of the HRTEM to studies of crystals only in low order zone-axis orientations.

The HRTEM image is a complex function of the interaction between the high energy electrons (typically 200KeV - 1MeV) with the electrostatic potential in the specimen and the magnetic fields of the image forming lenses in the microscope. Although images obtained from simple mono-atomic crystals often show white dots separated by spacings that correspond to spacings between atomic columns, these white dots fall on or between atomic columns depending on the thickness of the specimen and the focus setting of the objective lens (O'Keefe et al.,1989). Fortunately, in many cases it is only necessary to see the general pattern of image intensities to gain the desired knowledge. However, in general, the image can be best thought of as a complex interference pattern which has the symmetry of the projected atomic configuration, but otherwise has no one-to-one correspondence to atomic positions in the specimen. It is because of this lack of directly interpretable images that the need for image simulation arose.

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Image simulation grew out of an attempt to explain why electron microscope images of complex oxides sometimes showed black dots in patterns corresponding to the patterns of heavy metal sites in complex oxides, and yet other images sometimes showed white dots in the same patterns (Allpress et al.,1972). This first application was therefore to characterize the experimental images, that is to relate the image character (the patterns of light and dark dots) to known features in the structure.

Most simulations today are carried out for similar reasons, or even as a means of structure determination. Given a number of possible models for the structure under investigation, images are simulated from these models and compared with experimental images obtained on a high-resolution electron microscope. In this way, some of the postulated models can be ruled out until only one remains. If all possible models have been examined, then the remaining model is the correct one for the structure. For this process to produce a correct result, the investigator must ensure that all possible models have been examined, and compared with experimental images over a wide range of crystal thickness and microscope defocus. It is also a good idea to match simulations and experimental images for more than one orientation.

The simulation programs can also be used to study the imaging process itself. By simulating images for imaginary electron microscopes, we can look for ways in which to improve the performance of present-day instruments, or even find that the performance of an existing electron microscope can be improved significantly by minor changes in some instrumental parameter. Alternatively, based on imaging requirements revealed by test simulations, we can adjust the electron microscope to produce suitable images of some particular specimen, or even of some

particular feature in a particular specimen.

In order to simulate an electron microscope image, we need firstly to be able to describe the electron microscope in such a way that we can model the manner in which it produces the image. As a first step, we can consider the usual geometrical optics depiction of the transmission electron microscope (TEM).
Figure 1 shows such a diagram of a TEM operated in two distinct modes, set up for microscopy (a), and for diffraction (b). In microscopy mode we see that the TEM consists of an electron source producing a beam of electrons that are focused by a condenser lens onto the specimen; electrons passing through the specimen are focused by the objective lens to form an image called the first intermediate image (I1); this first intermediate image forms the "object" for the next lens, the intermediate lens, which produces a magnified image of it called the second intermediate image (I2); in turn, this second intermediate image becomes the "object" for the projector lens; the projector lens forms the greatly-magnified final image on the viewing screen of the microscope. In microscopy mode, electrons that emerge from the same point on the specimen exit surface are brought together at the same point in the final image.

At the focal plane of the objective lens, we see that electrons are brought together that have left the specimen at different points but at the same angle. The diffraction pattern that is formed at the focal plane of the objective lens can be viewed on the viewing screen of the TEM by weakening the intermediate lens to place the microscope in diffraction mode (b)

Consideration of the description of the electron microscope in figure 1 shows that the projector lens and the intermediate lens (or lenses)

merely magnify the original image (I1) formed by the objective lens. For the purposes of image simulation we can reduce the TEM to three essential components; (1) an electron beam that passes through (2) a specimen, and then through (3) an objective lens (fig. 2).

Our next step in describing the electron microscope for image simulation is to move from the geometrical optics description of the TEM to a description based on wave optics. In this description of the microscope we examine the amplitude of the electron wave field on various planes within the TEM, and attempt to determine how the wave field at the viewing screen comes to contain an image of our specimen.
By treating the electrons as waves, and considering our simplified electron microscope (Figure 2), we see that there are three planes in the TEM at which we need to be able to compute the (complex) amplitude of the electron wave field.

(1) The image plane

Working backwards, we start at our desired information, the electron wave field at the image plane; this wave field is derived from the wave field at the focal plane of the objective lens by applying the effects of the objective aperture and the phase changes introduced by the objective lens.

(2) The focal plane of the objective lens:

In turn, the electron wave field at the focal plane of the lens is derived from the wave field at the exit surface of the specimen by a simple Fourier transformation.

(3) The specimen exit surface:

In order to know the exit-surface wave field, we must know with which

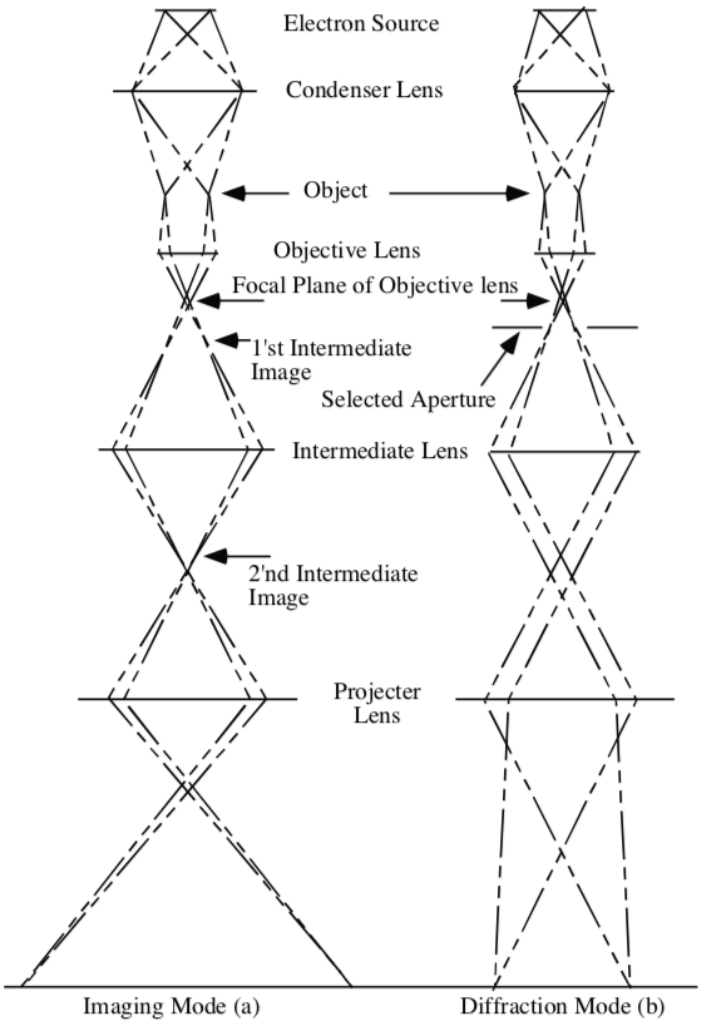

Figure 1. Geometrical optics representation of the TEM in imaging mode (a), and diffraction mode (b)

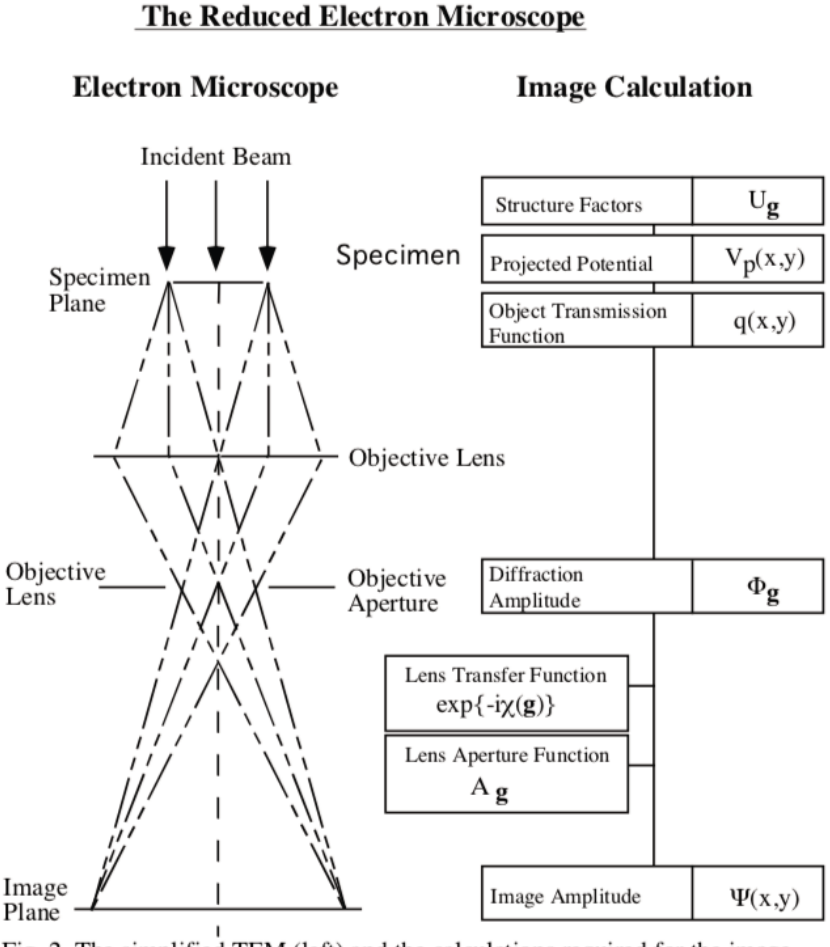

Fig. 2. The simplified TEM (left) and the calculations required for the image simulation (right). The three principal planes are marked.

physical property of the specimen the wave interacts, and describe that physical property of our particular specimen.

Cowley and Moodie (1957) showed that the interaction of an electron beam with a specimen could be described by the so-called multi-slice approximation, in which electrons propagate through the specimen and scatter from the crystal potential, the electron scattering is described by the so-called phase-grating function, a complex function of the potential, and the electron propagation is computed with a propagation function dependent on the electron wavelength. Since then there have been numerous formulations of the multi-slice approximation derived from the Schrödinger equation.

The problem of simulating images thus becomes a problem of computing the electron wave fields (wave function) at three microscope planes. Currently the best way to produce simulated images is to divide the overall calculation into three parts:

(1) Model the specimen structure to find its potential in the direction of the electron beam.

(2) Produce the exit-surface wave field by considering the interaction of the incident electron wave on the specimen potential.

(3) Compute the image-plane wave field by imposing the effects of the objective lens on the specimen exit surface wave.

Each of these steps will be covered in the next sections. However, because of space constraints, it is impossible to cover everything in great depth. For detailed derivation, the reader is encouraged to read the many excellent texts on the subject.

The Theory of Image Simulation

The specimen is a three dimensional objects consisting of a huge number of atoms. From a modeling point of view, it is necessary to reduce the number of parameters to a more manageable number. For crystalline materials described by a repeat of perfect unit cells this is easily accomplished. The unit cell in this case is defined by the lattice vectors A, B and C where A and B are in the plane the specimen perpendicular to the electron beam and C is in the main direction of the incoming electrons. A,B and C are related to the normal lattice vectors a,b, and c depending on the orientation of the specimen. The specimen is thus reduced to M number of unit cells, where M*C is equal to the thickness of the sample, giving in the end a 2D image which covers the area given by A and B.

In the case of a defect structure which no longer can be modeled as a small repeating structure, it is necessary to limit the extent of the calculation by defining a supercell which contains the defect. The resulting image obtained from the calculation will contain artifacts which arise from limiting the structure at arbitrary boundaries and care must be taken to ensure that the image gives a faithful representation of the area of interest.

The entire electrostatic potential of the specimen is now defined by one unit cell with axes a,b, and c, angles alpha, beta and gamma, and N atoms with coordinates (xi,yi,zi). For simplicity, we use the nomenclature of the crystallographic unit cell even though we are referring to the transformed unit cell (A,B,C) as described above.

The electrostatic potential in the crystal can be written

$$\phi(\mathbf{r}) = \int d^3\mathbf{r}' \, \frac{\rho(\mathbf{r}')}{|\mathbf{r} - \mathbf{r}'|} \qquad 1$$

where r(r), the charge density is:

$$\rho(\mathbf{r}) = \sum_{\substack{all \\ atoms}} \rho_i(\mathbf{r} - \mathbf{r}_i) \qquad 2$$

with the sum extending over all atoms i at positions ri, each giving rise to a charge density

$$\rho_i(\mathbf{r}) = Z_i e\delta(\mathbf{r}) - e|\psi_i(\mathbf{r})|^2 \qquad 3$$

ψ

where Zi : atomic number, e: electronic charge, Ψ: the quantum mechanical many electron wave function for the atom.

The potential f(r) is described by its Fourier transform F(u) through the relationship

$$\varphi(\mathbf{r}) = \int \Phi(\mathbf{u}) e^{-2\pi i \mathbf{u}\cdot\mathbf{r}} d\mathbf{u} = \sum_{\mathbf{H}} \Phi(\mathbf{H}) e^{-2\pi i \mathbf{H}\cdot\mathbf{r}} \qquad 4$$

since because of the periodicity of the unit cell, F(u) is non-zero only when u = H = ha*+kb*+lc* , H being a reciprocal lattice vector.

The potential F(H) is given as a sum over all atoms in the unit cell

$$\Phi(\mathbf{H}) = \sum_{\substack{all \\ atoms}} f_i^{el}(\mathbf{H}) e^{2\pi i \mathbf{u}\cdot\mathbf{r}_i} = \frac{e}{4\pi^2\varepsilon_0} \sum_{\substack{all \\ atoms}} \frac{Z_i - f_i^x(|\mathbf{H}|/2)}{\mathbf{H}^2} e^{2\pi i \mathbf{u}\cdot\mathbf{r}_i} \qquad 5$$

where the electron scattering factors fi_e^1 and the x-ray scattering factors f_i^x have been calculated from relativistic electron wave functions and

parameterized. They can be found in various tables (Doyle and Turner, 1968) and are in use by most image simulation programs such as SHRLI (O'Keefe at al., 1978), NCEMSS (Kilaas, 1987) and EMS (Stadelman,).

Taking into account any deviation from full occupancy at a particular site and the thermal vibration of the atom, the Fourier coefficients of the crystal potential from one unit cell is calculated as:

$$\Phi(\mathbf{H}) = \sum_{\substack{unit\ cell \\ atoms\ \ i}} f_i^{el}(\mathbf{H}) Occ(\mathbf{r}_i) \exp[-B_i \mathbf{H}^2] e^{2\pi i \mathbf{H}\cdot\mathbf{r}_i} \quad (6)$$

B: Debye Waller factor; Occ(r_i : The occupancy at position r_i)

The interaction between an electron of energy E and the crystal potential f(r) is given by the Schrödinger equation

$$[-\frac{h^2}{8\pi^2 m}\nabla^2 - e\phi(\mathbf{r})]\Psi(\mathbf{r}) = E\Psi(\mathbf{r}) \quad (7)$$

where m is the relativistic electron mass and h is Planck's constant.

Before entering the specimen, the electron is treated as a plane wave with incident wave vector k_0, $k_0 = 2\pi/\lambda$, so that the incident electron wave is written

$$\Psi_0(\mathbf{r}) = \exp\{i(\omega t - 2\pi \mathbf{k}_0 \cdot \mathbf{r})\}$$

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It is useful to define the quantity V(r) which will loosely be referred to as the potential as:

$$V(\mathbf{r}) = \frac{8\pi^2 me}{h^2}\phi(\mathbf{r})$$

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The Schrödinger equation above cannot be solved directly without making various approximations. Depending on how the problem is formulated, one can derive the most common solutions to the electron wave field at a position T within the specimen.

The Weak Phase Object Approximation (WPOA)

In the Phase Object Approximation (POA) (Cowley and Iijima, 1972), the phase of the electron wave function after traversing a specimen of thickness T is given as

$$\Psi(x, y, z = T) \approx \Psi(x, y, z = 0)\exp[-i\sigma V_p(x, y)T]$$

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with

$$\sigma = 2\pi me\lambda\left[1 + \frac{eE}{mc}\right]\Big/h^2$$

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where Vp(x,y) is the average potential per unit length. The specimen is

considered thin enough so that electrons only scatter once and are subject only to an average projected potential. In the weak phase object approximation, the exponent is considered much less than one, so that the electron wave function emerging from the specimen is:

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$$\psi(x,y,z=T) \approx \psi(x,y,z=0)(1-i\sigma V_p(x,y)T)$$

The WPOA only applies to very thin specimens of the order of a few tenths of Å, depending on the atomic number of the atoms in the structure (Gibson, 1994). The FT of the wave function gives the amplitude and phase of scattered electrons and in the WPOA one has:

$$\Psi(\mathbf{u}) = \delta(\mathbf{u}) - i\sigma V_p(\mathbf{u})T$$

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where u is a spatial frequency.

Again, for periodic crystals, $Vp_(u)$ are non-zero only for frequencies u=H where H is a reciprocal lattice vector in the crystal.

We will now use V to mean Vp. Thus for single electron scattering and when the Fourier coefficients V(H) are real (true for all centro-symmetric zone axis), the WPOA illustrates clearly that:

i) Upon scattering, the electron undergoes a -90° phase shift.

ii) The amplitude of a scattered electron is proportional to the Fourier coefficient of the crystal potential.

The Bloch Wave Approximation

In the BWA the electron wave function of an electron with wave vector k is written as a linear combination of Bloch waves b(k,r) with coefficients e (Howie, 1963). Each Bloch wave is itself expanded into a linear combinations of plane waves which reflect the periodicity of the crystal potential.

$$\psi(\mathbf{r}) = \sum_j \varepsilon^{(j)} b^{(j)}(\mathbf{k},\mathbf{r}) = \sum_j \varepsilon^{(j)} \sum_{\mathbf{g}} c_{\mathbf{g}}^{(j)} \exp[-2\pi i(\mathbf{k}_0^{(j)} + \mathbf{g})\cdot\mathbf{r}]$$

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The formulation above gives rise to a set of linear equations expressed as

$$[k_0^2 - (\mathbf{k}^{(j)} + \mathbf{H})^2] c_{\mathbf{H}}^{(j)} + \sum_{\mathbf{H}'} V(\mathbf{H}') c_{\mathbf{H}-\mathbf{H}}^{(j)} = 0$$

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which needs to be solved. Detailed derivation of the Bloch wave approximation can be found elsewhere.

Characteristics of the Bloch wave formulation are:

- Requires explicit specification of which reflections g are included in the calculation.
- Easy to include reflections outside the zero order Laue zone.
- Very good for perfect crystals, not suited for calculating images from defects.
- The solution is valid for a particular thickness of the specimen.
- Allows rapid calculation of convergent beam electron diffraction patterns.
- Includes dynamical scattering.

The MultiSlice Approximation

The multi-slice formulation (Goodman and Moodie, 1974 & Self et al., 1983) i s by far, the most commonly used method of calculating the electron wave field emerging from the specimen. Although it does not as easily include scattering outside the zero order Laue zone as the BWA, the multi-slice formulation is more versatile for use with structures containing any kind of defects, either they be point-defects, stacking faults, interfacial structures, etc. The multi-slice solution gives the approximate solution to the electron wave function at a depth z+dz in the crystal from the wave function at z. In the multi-slice approximation one has:

1 6

$$\psi(x,y,z+dz) \approx \exp[-i\sigma dz\nabla_{x,y}^{2}]\cdot\exp[-i\sigma\int_{z}^{z+dz}V(x,y,z')dz']\psi(x,y,z)$$

Thus starting with the wave function at z=0, one can iteratively calculate the wave function at a thickness n*dz, by applying the multi-slice solution slice by slice, taking the output of one calculation as the input for the next. Equation 16 is solved in a two step process.

The potential due to the atoms in a slice dz is projected onto the plane t=z, giving rise to a scattered wave field

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$$\psi_1(x,y,z+dz) = \exp[-i\sigma\int_{z}^{z+dz}V(x,y,z')dz']\psi(x,y,z) \equiv q(x,y)\psi(x,y,z)$$

The function q(x,y) is referred to as the phasegrating.

Subsequently, the wave field is propagated in vacuum to the plane

t=z+dz, according to

$$\psi(x,y,z+dz) = \exp[-i\sigma dz\nabla_{x,y}^2]\cdot\psi_1(x,y,z)$$

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The last equation represents a convolution in real space and is solved more efficiently in Fourier space (Ishizuka and Uyeda, 1977), where the equation transforms to

$$\Psi(\mathbf{H},z+dz) = \exp[-i\pi\lambda dz\mathbf{H}^2]\cdot\Psi_1(\mathbf{H},z) = p(\mathbf{H},dz)\cdot\Psi_1(\mathbf{H},z)$$

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where $\Psi(\mathbf{H},z)$ are the Fourier coefficients of $\psi(x,y,z)$. $p(\mathbf{H},dz)$ is called the propagator.

The multi-slice formulation is a repeated use of the last two equations and will give the wave field at any arbitrary thickness T of the specimen. If the slice-thickness is chosen as the repeat distance of the crystal in the direction of the electron beam, only the zero order Laue reflections are included in the calculation as the unit cell content is projected along the direction of the electron beam. Three dimensional information which involves including higher order Laue reflections can be included by reducing the slice thickness (Kilaas et al, 1987).

Sampling Criteria

Any numeric calculation must be performed for a limited set of data points (x,y) or reciprocal spatial frequencies **H**. Working with periodically repeated structures; if the lateral dimensions of the unit cell is a and b, which we for simplicity make orthogonal so that the axes are associated with an orthogonal x,y coordinate system, then for a given sampling interval dx=dy, we have

$$N = \frac{a}{dx} \quad ; \qquad M = \frac{b}{dy}$$

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defining the calculation to a grid of N*M points. The sampling interval automatically restricts the calculation in reciprocal space as well. The maximum reciprocal lattice vector for orthogonal axes is given as

$$\mathbf{H}_{\max}^2 = \left| h_{\max}\mathbf{a}^* + k_{\max}\mathbf{b}^* \right|^2 = \left(\frac{N}{2a}\right)^2 + \left(\frac{M}{2b}\right)^2$$

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Because most implementations of the multi-slice formulation makes use of Fourier transforms, the calculation grid N and M is adjusted so that both are powers of 2. This is because Fourier transform algorithms can be performed much faster for powers of 2 rather than arbitrary dimensions. This results in uneven sampling intervals dx,dy when a ≠ b. In order to not impose an arbitrary symmetry on the calculation, a circular aperture is imposed on the propagator. In practice, this aperture is set to 1/2 of the minimum of (hmax, kmax) as defined above in order to avoid possible aliasing effects associated with digital Fourier transforms. The sampling must be chosen such that the calculation includes all (or sufficiently enough) scattering that takes place in the specimen.

After the electron wave field emerge from the specimen, it is subjected to the varies magnetic field of the lenses that form the imaging and magnification part of the microscope. Of these lenses, only the first lens, the objective lens, is considered in the image formation calculation. Since the angle with which the electron forms with the optic axis of the lens varies inversely with the

magnification, only the aberrations of the objective lens are important. The remaining lenses serve to just magnify the image formed by the objective lens. The effects of the lens which normally are included in the calculation are spherical aberration, chromatic aberration and lens defocus. Two-fold and three-fold astigmatism, including axial coma, are considered correctable by the operator although they can be included in the equations.

Without any aberrations, no instabilities and with the specimen in the focal plane of the objective lens, the image observed in the electron microscope would be am magnified version of

$$I(x,y) = \left|\psi(x,y,z = exitplane \quad of \quad specimen)\right|^2 = \psi_e(x,y)\psi_e^*(x,y)$$

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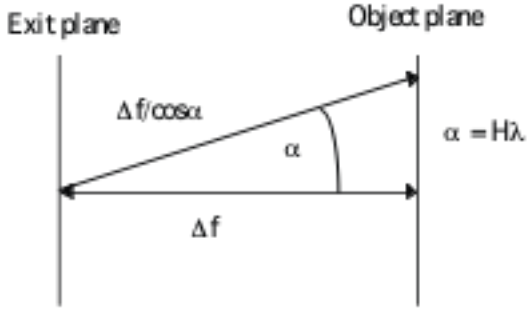
Objective Lens Defocus

Consider an electron traveling from the plane defined by the exit surface of the specimen to the plane given as the plane of focus for the objective lens. This distance is referred to as the objective lens defocus Δf.



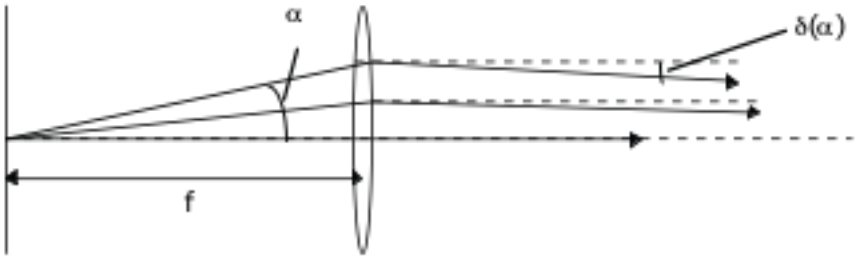
The electron traveling along the optic axis will have a path length of Df while an electron that has been scattered an angle a=Hl, will travel a distance Δf /cosα. This can be expressed as a phase difference

$$\frac{2\pi}{\lambda}\left(\Delta f/\cos\alpha - \Delta f\right) \approx \pi\lambda\Delta f\mathbf{H}^2 \qquad (23)$$

Spherical Aberration

Electrons crossing the optic axis with an angle a at the focal plane of the objective lens should form parallel paths emerging from the lens.

e



However, the spherical aberration of the lens causes a phase shift relative to the path of the unscattered electron (a=0) which is written as (Scherzer, 1949):

$$2\pi/\lambda * 1/4\, C_x\alpha^4 = 1/2\,\pi C_s\lambda^3\mathbf{H}^4 \qquad (24)$$

If there were no other effects to consider, the image would be obtained as follows:

Calculate the wave field emerging from the specimen according to

one of the approximations.

- Fourier transform the wave field which gives the amplitude and phase of scattered electrons.
- Add the phase shift introduced by the lens defocus and the spherical aberration to the Fourier coefficients.
- Inverse Fourier transform to find the modified wave function.
- Calculate the image as the modulus square of the wave field.
-

However, there are two more effects that are usually considered. Variations in electron energy and direction.

Chromatic Aberration / Temporal Incoherence

Electrons do not all have exactly the same energy for various reasons. They emerge from the filament with a spread in energy and the electron microscope accelerating voltage varies over the time of exposure. The chromatic aberration in the objective lens will cause electrons of different energies to focus at different planes. Effectively this can be thought if as rather than having a given defocus f_0, one has a spread in defocus values centered around f_0. The value f_0 is what is normally referred to as Δf as indicating defocus. The images associated with different defocus values add to make the final image. Assuming a Gaussian spread in defocus of the form

$$D(f - f_0) \propto \exp[-\frac{(f - f_0)^2}{\Delta^2}] \quad 25$$

gives:

$$I = \int |\Psi(f - f_0)|^2 D(f - f_0) df \Rightarrow \Psi(\mathbf{H}) \rightarrow \Psi(\mathbf{H}) \exp[-1/2(\pi\lambda\Delta\mathbf{H}^2)^2] \quad 26$$

This states that each Fourier term (diffracted beam) is damped according

to the equation above.

Beam Divergence / Spatial Incoherence

The electron beam is not an entirely parallel beam of electrons, but form rather a cone of an angle α. This implies that electrons instead of forming a point in the diffraction pattern form a disk with a radius related to the spread in directions. As for a variation in energy, the images formed for different incoming angles are summed up by integrating over the probability function for the incoming direction. It turns out that this also leads to another damping of the diffracted beam (Frank, 1973) so that:

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$$I(\mathbf{r}) = \int |\psi(\mathbf{r},\alpha)|^2 D(\alpha) d\alpha \quad \Rightarrow \quad \Psi(\mathbf{H}) \rightarrow \Psi(\mathbf{H}) \exp[\pi\alpha\lambda(C_s \mathbf{H}^2 \lambda^2 + \Delta f)]^2$$

The Final Image

Equation 26 and equation 27 are only valid when the intensities of the scattered beams are much smaller than the intensity of the central beam. Thus the image results from scattered beams interfering with the central beam, but not with each other. This is referred to as linear imaging. Although the formulation is slightly more complicated in the general case, the expressions above give sufficient insight into the image formation. Image simulation programs do however include the more general formulation which include non-linear imaging terms (O'Keefe, 1979). Each Fourier component is damped by the spread in energy and direction and the image is formed by adding this to the recipe in section 4.2

The Contrast Transfer Function CTF

When reading about HRTEM, it is impossible not to encounter the expression "Contrast Transfer Function". Loosely speaking, the CTF of the microscope refers to the degree with which Fourier components of the electron wave function (spatial frequencies) are transferred by the microscope and contribute to the Fourier transform of the image. Although the CTF only holds for thin specimen and linear imaging, it is often generalized and wrongly applied to all conditions. However, the CTF does provide insight into the nature of HRTEM images. In order to derive the expression for the CTF, we start by calculating the image intensity as given by the Weak Phase Object approximation. In the WPOA:

$$\Psi(x,y,z=T) \approx 1 - i\sigma V_p(x,y)T \quad (28)$$

and

$$\Psi(\mathbf{H}) = \delta(\mathbf{H}) - i\sigma V_p(\mathbf{H})T \quad (29)$$

Applying the phase shift due to the spherical aberration and the objective lens defocus which we will call Φ(H), we get that the FT of the wave function is (for simplicity V = $Vp_{)}$:

$$\Phi(\mathbf{H}) = \delta(\mathbf{H}) - i\sigma V(\mathbf{H})e^{i\chi(\mathbf{H})}A(\mathbf{H}) \quad (30)$$

where A(H) is the damping terms arising from partial coherence.

The FT of the intensity is now given as

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$$I(\mathbf{H}) = FT(\psi \cdot \psi^*) = \sum_{\mathbf{H}'} \Psi(\mathbf{H}')\Psi^*(\mathbf{H}-\mathbf{H}') \approx$$
$$\sum_{\mathbf{H}'} \left(\delta(\mathbf{H}') - i\sigma A(\mathbf{H}')V(\mathbf{H}')e^{i\chi(\mathbf{H}')}\right)\left(\delta(\mathbf{H}-\mathbf{H}') - i\sigma A(\mathbf{H}-\mathbf{H}')V(\mathbf{H}-\mathbf{H}')e^{i\chi(\mathbf{H}-\mathbf{H}')}\right) \approx$$
$$\delta(\mathbf{H}) + 2\sigma A(\mathbf{H})V(\mathbf{H})\sin\chi(\mathbf{H})$$

The last result is very useful and it leads to the frequently used concept of the Contrast Transfer Function (CTF). The CTF is defined as A(H)*sinχ(H) The equation above states that each reflection H contributes to the image intensity spectrum with a weight that is proportional to the CTF. Figure 3. shows a plot of a CTF including sinχ and the damping curves. When sinχ(H) = -1 for a large range of frequencies H, which is the condition referred to as Scherzer defocus[11], the image can be thought of as:

$$I(x, y) \approx 1 - 2\sigma U(x, y)$$

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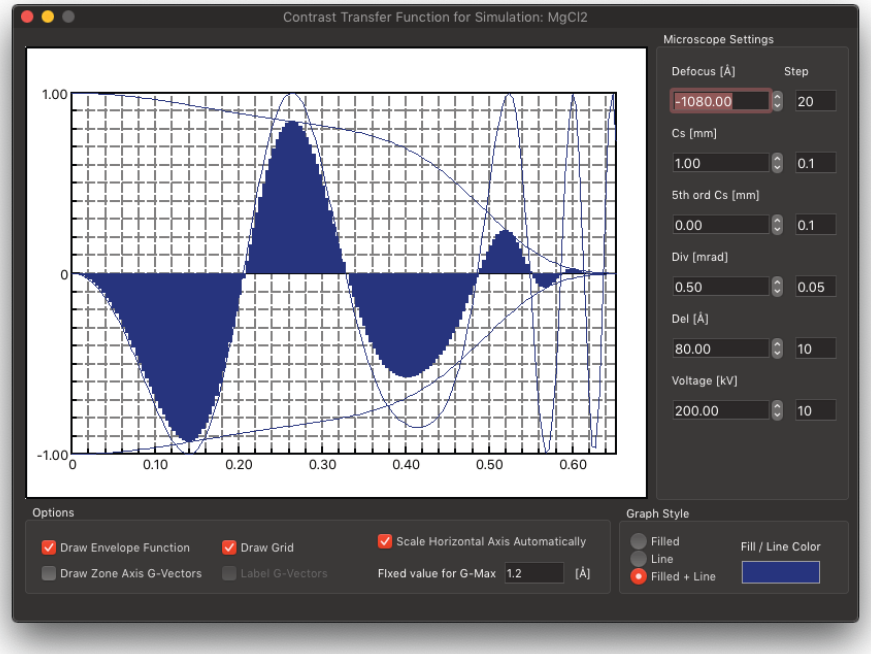
where U(x,y) is a potential related to the original crystal potential, but keeping only the Fourier coefficients related to frequencies transferred by the microscope. The equation above shows the often used rule of thumb. For thin specimens, under Scherzer imaging conditions, atoms are black.

Figure 3. Plot of the Contrast Transfer Function for a 200kV microscope with the parameters indicated.



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Introduction to Image Simulation In Tempas

Since the simulation of a HRTEM phase contrast image can be subdivided into independent calculations involving the structure, the scattering process and the imaging process, Tempas allows one to invoke these independent calculations separately through the "Calculate" menu.

Full Calculation

This command will start the calculation from the required starting point and proceed to calculate final images.

Projected Potential

generates the crystal potential that produces electron scattering from the structural data, unit cell dimensions, symmetries, and atom positions, occupancies, and temperature factors.

Exit wave functions(s)

generates the electron wave field at the specimen exit surface; it uses the projected potential combined with information about the accelerating voltage of the electron microscope, the specimen thickness and tilt. The computation algorithm is the multi-slice approximation.

Image(s) normal calculation

generates the image intensity at the microscope image plane; the effects of the objective lens phase changes and resolution-limiting aberrations are included via parameters like defocus, spherical

aberration, incident beam convergence, spread of defocus, and the position and size of the objective aperture.

Image Plane wave functions(s)-

generates the electron wave function at the imaging plane in the microscope. This is equivalent to the application of the Contrast Transfer Function to the Fourier transform of the electron wave function at the exit surface of the specimen followed by an inverse Fourier transform. The calculation of the image plane wave function is used for comparing with the electron wave function found by the use of electron holography. The remaining commands in the Calculate menu will be covered under the Menus chapter.

Thus "Projected Potential" calculation considers only the specimen structure, "Exit wave functions(s)" calculation treats the interaction of the specimen with the electron wave, and the "Image(s)" calculation simulates how the wave leaving the specimen interacts with the lens system of the electron microscope. Once a simulation has been made, any additional simulation will usually not require a full re-calculation; any change in microscope parameters will not affect the results of the "Projected Potential" and "Exit wave functions(s)" calculations, and only Image(s) will need to be re-run; any change in microscope voltage or in specimen thickness and tilt will not affect the output of "Projected Potential", but "Exit wave functions(s)" and "Image(s)" will need to be re-run. Of course, any change in the specimen structure will require the re-running of all three sub-programs.

Tempas generates and stores various files in the course of a simulation. The 6 possible data files are:

(1) <structurename>.at stores all the structure and microscope

information needed to run the simulation. This information is derived from user input and the supplied data files. In particular, the string "structurename" is a unique name for the structure, input by the user when creating the structure file. This is an editable file of type 'TEXT'.

(2) <structurename>.pout is the result of running the projected potential routine from the information stored in <structurename>.at; it contains the specimen potential in the direction of the electron beam. This is a BINARY file of type Real 4. The first 80 bytes consists of record information and the data starts at byte 80. The first line of data contains the data for the bottom line of the "image" since the coordinate system for Tempas is at the lower left corner of the image/unit cell. Thus if the data is imported into a program for viewing, the image will appear flipped.

(3) <structurename>.mout is the result of running the multi-slice routine using the data in <structurename>.pout with those in <structurename>.at; it contains the exit-surface wave function at one or more selected specimen thicknesses. This is also a BINARY file with the same structure as <structurename>.pout, except for the fact that the data is complex, pairs of numbers (real and imaginary). The data starts at byte 80 and the file can contain more than one exit wave function.

(4) <structurename>.iout is the result of running the image formation routine to apply the effects of the microscope parameters in the <structurename>.at file to the exit-surface wave; it contains one or more images ready to be displayed. This again is a BINARY file with data starting at byte 80 and the file can contain more than one image. Data is Real 4

(5) <structurename>.hout is the result of calculating the image plane electron wave function(s) instead of calculating the simulated images. The data is complex, pairs of numbers (real and imaginary). The data starts at byte 80 and the file can contain more than one image plane exit wave function.

(6) <structurename>.aout contains the complex amplitudes of several diffracted beams at one-slice increments in specimen thickness. The beams are specified by the user, and can be plotted as a function of specimen thickness.

In addition, two "print" files are produced (but rarely printed) just in case additional information about a computation is required by the user. These files are:

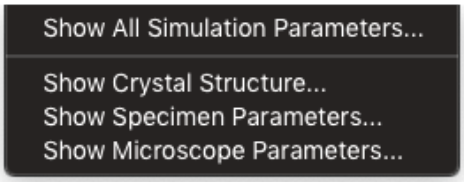
(7) <structurename>.p_prnt contains information about the way in which the "Projected Potential" subprogram processed the <structurename>.at data to produce the specimen potential.

(8) <structurename>.m_prnt contains information about the way in which the "Exit wave functions(s)" subprogram processed the <structurename>.pout data with the <structurename>.at to produce the exit-surface wave; that is, it contains information from the multi-slice computation.

Running an Image Simulation in Tempas

The first step in running a simulation is generating the structure input file. This is done through "New Normal Structure"... in the FILE menu. This generates a default cubic structure. Use this template to modify the data to fit your structure.

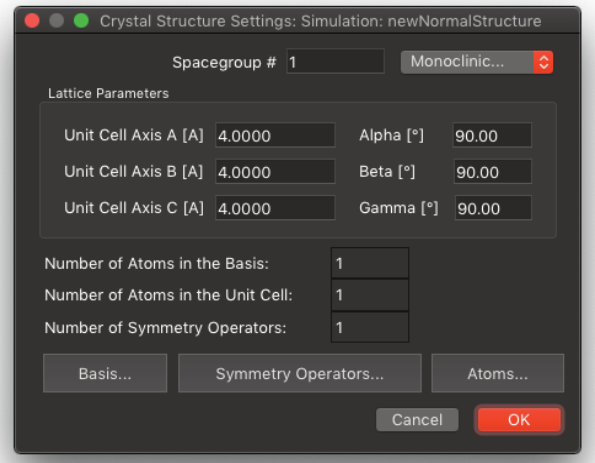
Under the Menu "Parameters", there are the following options.



The various menuitems bring up their own dialogs. The "Show All Simulation Parameters" encompass the other three. However the various aspects such as Crystal Structure, Specimen Information and Microscope Information can be viewed independently through its own menuitem.

Most of the items in the Dialogs are meant to be self explanatory.

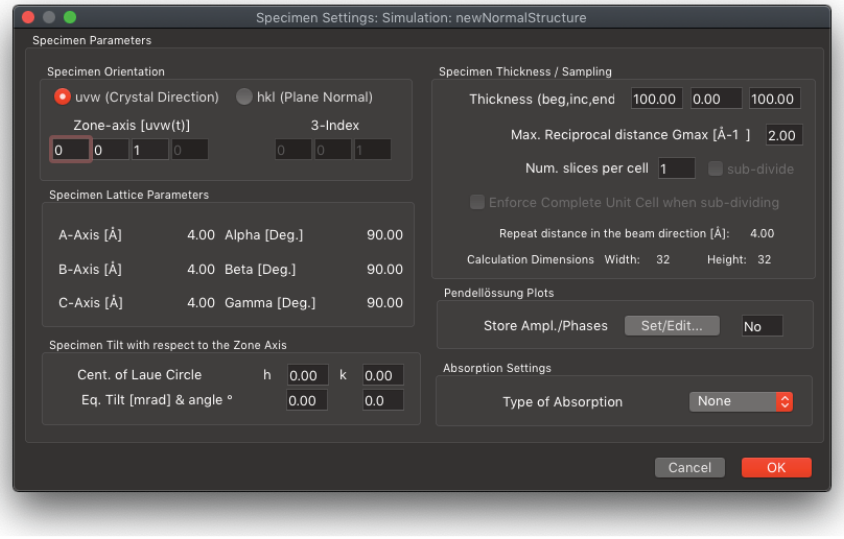
Show Crystal Structure... brings up



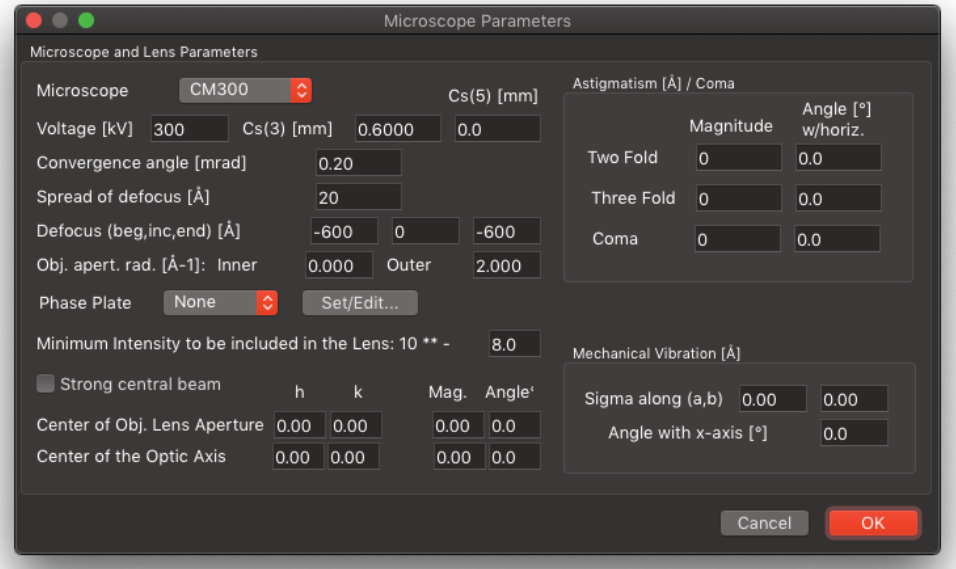
The Crystal Structure is defined through one of the 230 spacegroups which gives the symmetry of the unit cell. There are 3 axes A, B and C and corresponding angles between the axes Alpha, Beta and Gamma. Alpha is the angle between B and C, Beta is the angle between C and A, and Gamma is the angle between A and B.

The content of the unit cell is given by the Atoms in the Basis and the Symmetry Operators. Each spacegroup has a set of symmetry operators, P1 being the simplest with a single symmetry operator x,y,z. The position of an atom is given by a fractional coordinate. Thus the atom at position x,y,z is at the position r = xA + yB + zC in the unit cell. Each symmetry operator operates on the atoms in the basis (motif) and produces (possibly) a
"new" atom at a symmetry related position. As an example, the 2 symmetry operators "x,y,z" and "x+1/2,y+1/2,z+1/2" takes the atom in the basis at position (0.2,0.2,0.2) and gives as output the 2 atoms (0.2,0.2,0.2) and (0.7,0.7,0.7).

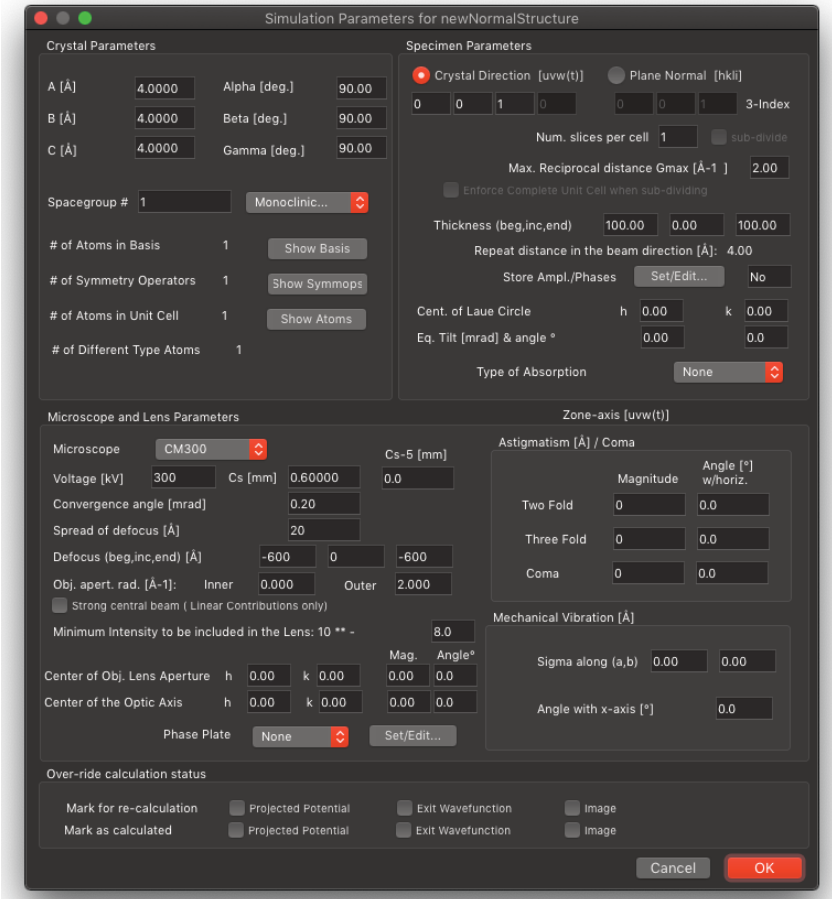
Show Specimen Parameters... brings up



Show Microscope Parameters... brings up



S*how All Simulation Parameters...* brings up



a, b, c,
alpha, beta, gamma

These are the unit cell dimensions in Ångstrøm units, and the unit cell angles in degrees. Tempas will automatically set the angles depending on the spacegroup, if possible. The program will also automatically set lattice parameters depending on the spacegroup. Thus if the user chooses a cubic system, b and c are set equal to a

Space group#

Tempas generates symmetry operators for the any one of the 230 space groups when selecting the number or the symbol of the space group (as listed in the International Tables for Crystallography). By clicking on the pop up menu "Space Group" one can choose one of the 230 spacegroups by first selecting the type of crystal-structure, i.e. hexagonal or cubic. The user can choose one of the spacegroups by clicking on the symbol for the spacegroup or by entering the number for the spacegroup.

The input also allows for choosing the second setting for a specific spacegroup if one exists. If no space group is required, one should use the space group P1 (1), in which case the only symmetry operator is x,y,z. Additional symmetry operators can be entered by opening the dialog displaying the symmetry operators.

Show (Basis atoms)

Use this button to bring up the dialog window that enables the input of the atoms in the basis.

Number of Atoms in the Basis

This value is the number of independent atom positions in the basis or asymmetric unit of the cell. When operated on by the symmetry operators, the basis generates all the atom positions within the cell. This value is never modified by the user since the program always recalculates this number depending on the data entered.

Show (Symmetry Operators)

The symmetry operators are automatically created by specifying the

spacegroup. By clicking on this button, a window displaying the symmetry operators are shown.

Show (Atoms in Unit Cell)

The atoms in the unit cell are automatically created by the operation of the symmetry operators on the atoms in the basis. The number of atoms is given and by clicking on the button "Show", a window displaying the atoms in the unit cell appears.

Number of different atoms

This value is the number of different types of atoms in the specimen structure; difference is due to a different atomic number or a different Debye-Waller factor. The correct value is calculated by Tempas and displayed.

Zone Axis

Specimen orientation in relative real space axes units.

Number of slices per unit cell

For unit cells with large repeat distances in the beam direction, moderate values of Gmax may allow the Ewald sphere to approach the so-called pseudo upper-layer line that the multi-slice allows at the reciprocal of the chosen slice thickness. In this case Tempas will sub-divide the slice into two or more sub-slices. How this is done depends upon the potential setting chosen in the Option menu.If 2-D calculation is set and the checkbox "sub-divide unit cell" is NOT checked, the projected potential is calculated for the entire unit cell in the zone-axis orientation and is used n times to cover the unit cell. If 2-D calculation is set and the checkbox "sub-divide unit cell" is checked, n-potentials are calculated from the atoms within each sub-layer and used to cover the unit cell. If 3-D calculation is set, n-potentials are calculated by appropriate integration of

the potential from all atoms in the unit cell.

Gmax

The maximum value (in reciprocal Ångstrøm units) of any scattering vector to be included in the multi-slice diffraction calculation. This value imposes an "aperture" on the diffracted beams included in the dynamic scattering process. It should be large enough to ensure that all significant beam interactions are included. The default value is 2.0. Tempas will compute phase-grating coefficients out to twice Gmax in order to avoid aliasing in the multi-slice calculations.

Specimen Thickness

The thickness of the specimen foil is entered as a beginning thickness, an ending thickness and an incremental thickness. All numbers are in Ångstrøm units. A series of thicknesses represented by the upper and lower bounds and a thickness step; e.g. 100 50 250 will cause Tempas to store the exit wave function at specimen thicknesses of 100Å to 250Å in steps of 50Å (a total of four thicknesses).

Store Ampl./Phases - Set...

Clicking this button allows a number of diffracted beams to be selected for plotting of their intensity and phase variation as a function of specimen thickness. The reflections to be tracked are determined by entering the hkl values for the reflection. Only 10 reflections can be tracked this way.

Center of the Laue Circle

Specimen tilt is specified by entering the center of the Laue circle in units of the h and k indices of the projected two-dimensional

reciprocal-space unit cell. Equivalently the tilt angle and azimuthal angle can be specified instead. The new indices and their relationship to the original reciprocal cell is found in the data file <structurename>.p_prnt

Type of Absorption

Absorption can be included in the program by introducing an imaginary projected potential.

Microscope

The type of electron microscope used to generate the imaging parameters. Predefined microscopes are shown in the popup menu together with one undefined microscope. If a predefined microscope is used, Tempas provides values for Cs, the spherical aberration coefficient of the objective lens (in mm.); Delta, the halfwidth of a Gaussian spread of focus due to chromatic aberration (in Ångstrøm units); Theta., the semi-angle of incident beam convergence (in milliradian). If the type of microscope is unknown to Tempas, the above values must be entered separately (We will see later how to define a new microscope).

Voltage

The electron microscope accelerating voltage in kilovolts.

Objective Lens Defocus

The defocus of the objective lens is entered in Ångstrøm units with a negative value representing underfocus (weakening of the lens current). As for the specimen thickness parameter, the input is a range specified by the upper and lower bounds and an increment.

Min. Intensity in the lens

This specifies a cutoff in intensity for which a beam is included in the calculation. Normally the default is adequate and saves computation time.

However, for large structures containing defects, diffuse scattered beams are very weak and a lower cutoff may be needed in order to compute the correct contrast..

Strong Central Beam

If this box is checked, only linear terms to the image intensity contribute. Normally both linear and non-linear terms are included in the calculation..

Cs, Spherical Aberration

The spherical aberration of the objective lens in mm.

Convergence Angle

This is the spread in angle for the cone of incoming electrons depending on the condenser lens aperture. The angle is given in mrad.

Spread of Defocus

This is the effective spread in defocus which results from the distribution of energies of the imaging electrons and the chromatic aberration of the objective lens. The unit is Å.

Aperture Radius

The radius of the objective aperture is specified in Å-1

Both an inner aperture and an outer aperture can be specified. Normally the inner aperture would be zero.

Center of objective Aperture

The center of the objective lens aperture is defined in units of h and

k of the two dimensional reciprocal space unit cell, as for the Laue circle center. Again, tilt angle and azimuthal angle can be specified instead.

Center of the Optic Axis

The center of the optic axis of the electron microscope is specified in terms of the h and k indices of the two-dimensional reciprocal-space unit cell, just as for the Laue circle center and the aperture center.

Two-fold astigmatism

The two fold astigmatism of the objective lens and the angle with the x-axis. The magnitude is given in Å.

Three-fold astigmatism

The two fold astigmatism of the objective lens and the angle with the x-axis. The magnitude is given in Å.

Coma

The coma of the objective lens and the angle with the x-axis. The magnitude is given in Å.

Mechanical Vibration

This simulates the effect of a slight vibration of the microscope. One finds that often the simulated images show details that are not present in the experimental data regardless of other imaging conditions. This may be due to image degradation caused by microscope vibration or other effects not included and thus one can introduce a slight mechanical vibration in an attempt to create more realistic simulated images. It is possible to specify an anisotropic vibration by introducing the amplitude in two perpendicular directions with the diagonal of the ellipse at an angle with the a axis (as in the unit cell viewed in the zone axis orientation).

Windows

This chapter explains the windows of Tempas, the information presented in each and how one interacts with the contents of the windows.

Log Window

This window shows the current status of the program indicating the number of phasegrating coefficients calculated, the current slice number being calculated, the current image being calculated etc.

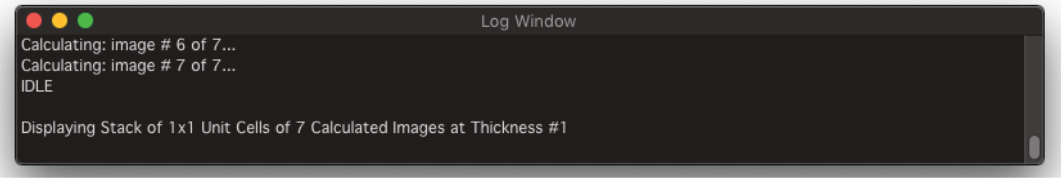


Image Control Window

This window is used to control the display of selected image(s). The active image can be the image in a image window or the selected image in a Tempas Simulation Display window.

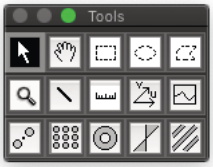
If the image selected is an ImageStack or a 3D Image, the slice selector will become active and this control can be used to cycle through the individual images.

The black and white values of the current selected image is shown and can be changed by typing in new values. The contrast and brightness can be changed by using the appropriate sliders. An image can be shown on a logarithmic scale which is the default for images in frequency space (reciprocal space). The line in the graph represents how input image values are mapped to output display values. Thus an image can be pseudo colored

by choosing a color from the color bar with the color picker tool selected and "depositing" this color in the vertical gray scale bar showing the display values.

The histogram of the current image is shown and black and white values can be chosen by clicking and dragging to select a region of the histogram. To invert the display, click in the "Invert" button. Similarly the image is reset to the original values through the "Reset" button. This window is also used to set the color of a particular atom species and the color of lines and text. To choose a color, the Color Picker Tool must have been chosen.

Tools Window



The following tools are currently defined:

Pointer Tool

Used for general moving around objects in the display window. If an object is selected and the "Option" key is held down while dragging an object, a copy is made of the object. In an image window, the pointer tool will also act as a hand tool if nothing is selected.

Selection Tools

There are 3 different selection tools. Rectangular, Oval, and Polygonal selections. These tools are used to select a portion of an image or for the rectangular selection tool objects in the Display Window for several possible operations such as copying, cutting, histogram computation etc. To select an area, click at a point in the image and drag the cursor while the mouse button is pressed. In the main Tempas window, all objects intersecting the selection rectangle will be selected.

There are some constraints that can be used with the rectangular and oval selection tool. Using the SHIFT key with the Rectangular Selection Tool forces the selection to be square. Using the OPTION key forces the selection to be powers of 2. With the Oval Selection Tool the SHIFT key forces the selection to be a circle. When the selection is an ellipse, the ellipse can also be rotated using the handles at the corners of the selected ellipse.

With the polygon selection tool, **double clicking** in a point to be used in the selection will close the path

Magnifying Glass

When selected the cursor turns into a magnifying glass which can be used to zoom in on a selected part of the display. Each time the mouse is clicked in the image window, the image is zoomed by a factor of two. By holding down the Option key while clicking, the image will be zoomed out by a factor of 1/2 for every click. Double-clicking the magnifying glass returns the image back to normal size.

Of course one can also use the scroll wheel or the trackpad to zoom up and down using scrolling or pinching movements.

Line Tool

This tool is used to draw lines on the display. If the Shift key is down, only vertical or horizontal lines will be drawn.

Ruler Tool

Use this tool to measure distances in an image. An image can be calibrated from the menu command under Process after a line is drawn using the ruler tool. The length of the ruler and its orientation will show in the INFO window. If the image is calibrated the length will display in both pixels and also in the calibrated units.

Lattice Tool

This tool is used to place a Lattice on an image. The lattice is normally refined to fit a list of peak positions.

Once a lattice is placed on an image it can be moved around by placing its origin to the desired location.

The length and the direction of the lattice vectors can be moved by dragging the anchor points. Clicking in the endpoint with the **Option** key down will add 1 to the number of lengths the lattice vector represents. Option clicking with the **Shift** key will decrement by 1 the number of lattice vectors the vector represent.

If there is a peak list on the image a very common action is to refine a lattice to the set of peak positions. This is done by placing the origin on one of the peaks and the lattice vectors on another set of peaks that should define a lattice. Either by right clicking in the selected lattice or by using the menu commands, the lattice can be refined in such a way that the root mean square deviation of peaks to the lattice is minimized.

Trace Tool / Rotate Tool

This tool button serves double purpose. When an image is selected, it will operate as a trace tool. When an object can be rotated, it will function as a Rotate Tool. This tool is used to get a line trace for the line drawn with the Trace Tool being the current tool. The integration width can be changed by double clicking on the trace-line or choosing "Edit Object" when the line is selected.

When used as a line trace, a window showing the plot of the trace will pop up. The trace in the window is correlated to the trace on the image. Using the pointer one can click and move in the Trace Window showing a marker detailing the actual underlying data as well as the position of the marker will also be shown on the actual

line on the image.

Masking Tools

The last 5 tools are masking tool normally used in reciprocal space, but they can be used in real space as well. The masks are

a) Spot mask. A reflection and its conjugate is selected.

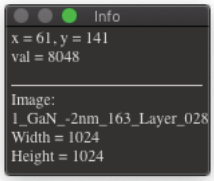
b) Lattice mask. A mask defined by two lattice vectors.

c) Band Pass mask. This mask is defined by an inner and an outer circle.

d) Wedge mask. Defined by two lines.

e) Line mask. Defined by a line and a single lattice vector.

All these masks can be transparent or opaque, meaning they work on the region within or outside of the mask. The mask parameters can be edited by double clicking on the mask or selecting the mask and choosing "Edit Mask" from the "Process" menu. The number of lattice spacings for the vector(s) for the lattice mask and line mask can also be changed by clicking in the end point of the vector with the "Option" key down. Each click increments the number of lattice spacings to the endpoint by one. Holding down the "Shift" key and the "Option" key decreases the number of lattice spacings by one.



Info Window

This window shows the current position of the cursor within the image window and the intensity of the underlying pixel. When dragging a rectangle, the dimensions of the rectangle are shown. Line lengths and angles are also displayed. Image statistics is displayed in this window when invoked through the "Statistics" in the "Process" menu.

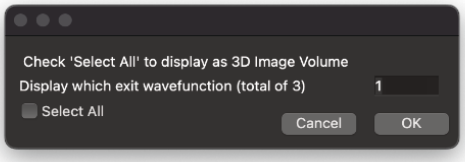
Control Window

Use this window to define which part of the calculation to display. The choices are:

Projected Potential - Essentially the output of the projected potential routine. There is a one to one correspondence between the points in the projected potential and those in the image if displayed under equivalent conditions. If there are multiple potentials that can be displayed, one is offered the option to display a specified potential or all of the displayed as a Stack of Images, see next paragraph.

Exit wave function - This is the output of the multi-slice component of the program and shows the distribution of electrons as they emerge from the bottom of the specimen, or at a predefined depth in the specimen. By holding down the Option key when selecting the button, one can select to display either the magnitude squared (default), the complex amplitude or the complex phase of the electron wave function at the exit surface of the specimen.

If there are multiple exit wave functions calculated, one is offered the following dialog to either select a specified one or all of the as a stack of images, strictly speaking a 3D Image volume. The specified image can be controlled by the arrow keys when the imagestack is selected, or using the slice control in the Image control Window.



Diffraction Pattern - Select this option to display the diffraction pattern for one of the selected specimen thicknesses. This is a dynamical diffraction pattern including multiple scattering in the specimen.

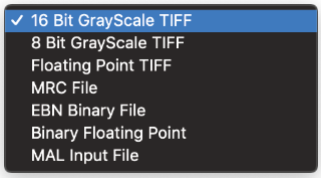
Image - When selected, one of the calculated images becomes the source of the operations defined by clicking in the Operand Window. By holding down the Option key when selecting the button, one can select to display either the image intensity (magnitude squared, default), or if the image plane wave function(s) has been calculated, the complex amplitude or the complex phase of the electron wave function at the image.

Fourier Transform - Use this to perform a Fourier Transform on the selected source. Operating on the Projected Potential will yield the structure factors, operating on the Exit wave field will yield the diffraction pattern and operating on the image will give the Power spectrum of the image.

#Unit cells

Use this to specify the number of unit cells that should be displayed. The input determines the number of cells in the a-direction and b-direction.

Display - Before the result of operating on a selected source is displayed in the image window, Display must have been clicked. Choosing the source and operations only selects the functions to be performed. When Display gets activated, the functions get executed.



->**File** - This will allow for output of the numeric values of images, amplitudes and phases to a file. Options allow for writing the data in various formats as shown. MAL input file is a binary 2 Byte unsigned int format where the data is formatted to fit a given size image (512x512, 1024x1024) with a specified sampling in Ångstrøm.

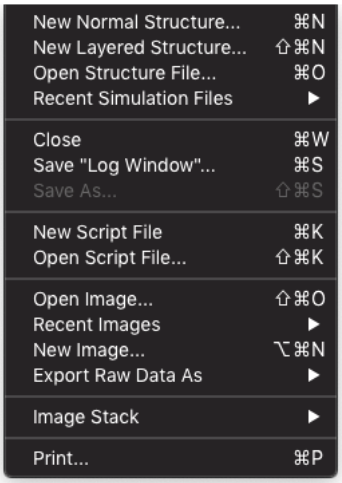
Cancel - Use this button in case the wrong sequence of commands was chosen or anything else was entered wrong. This cancels the set functions.

Tempas Simulation Display Window

Each Simulation object (file) will have its own display window. This is where most of the results of a calculation are displayed. All objects such as images, diffraction patterns, unit cell drawings, etc in this window are all objects which can be manipulated. Objects are moved with the pointer tool. If the Option key is held down, a copy of the object will be created. The objects respond to gestures, such that pinching or two finger "scrolling" will zoom the objects. The magnifying tool operates on each object separately. Several objects can be selected and copied onto the clipboard and can be pasted into other applications. Images in this window if double-clicked will open up in their own separate window. The selection tool will select any object which intersects the selection rectangle. Each object have different properties. An image can be scaled directly by dragging its corner. A kinematical SAD pattern can be rotated by using the rotation tool when the object is selected and the mouse-down event takes place in one of the diffraction spots. A drawing of the atomic structure can be magnified and rotated into different viewing directions. Most of the objects will have a contextual menu associated with it.

Menus

Many of the functions in Tempas are run from one of the Tempas menus, including the multi-slice calculation. In addition, most options are set from one of the menus. This is a list of the currently available menus and a description of their function.

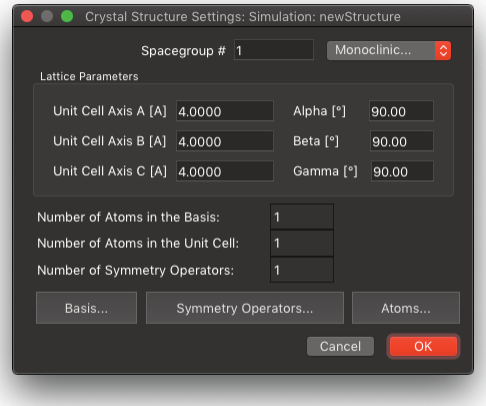


File Menu

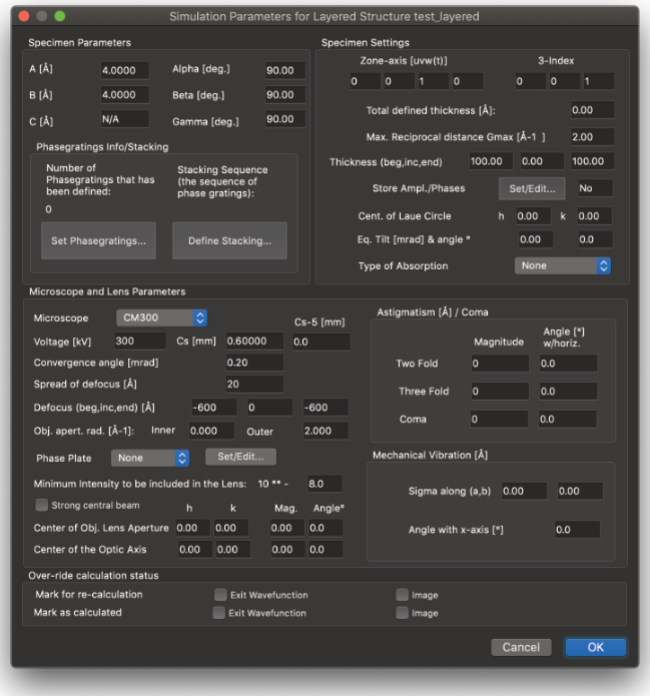
This menu contains the following commands:

New Normal Structure...

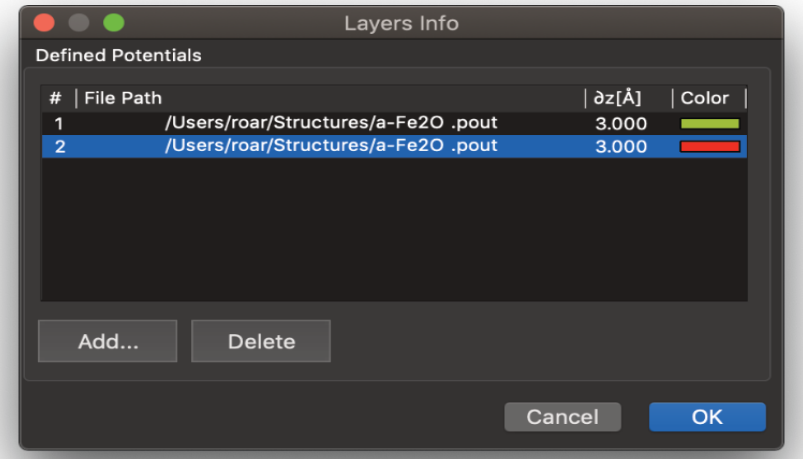
*C*reate a new structure file. A name is prompted for before input is made. Enter a unique structure name, the program will append the extension .at. Make sure the filename does not have a period in it, except for used with the extension. This creates a new structure with default values for all the parameters. The default structure is a monoclinic structure with 1 atom in the basis and 1 symmetry operator. To change the crystal unit cell, choose "Show Crystal Structure..." in the Parameters Menu. Change the input to reflect the structure that you want to create



New Layered Structure...

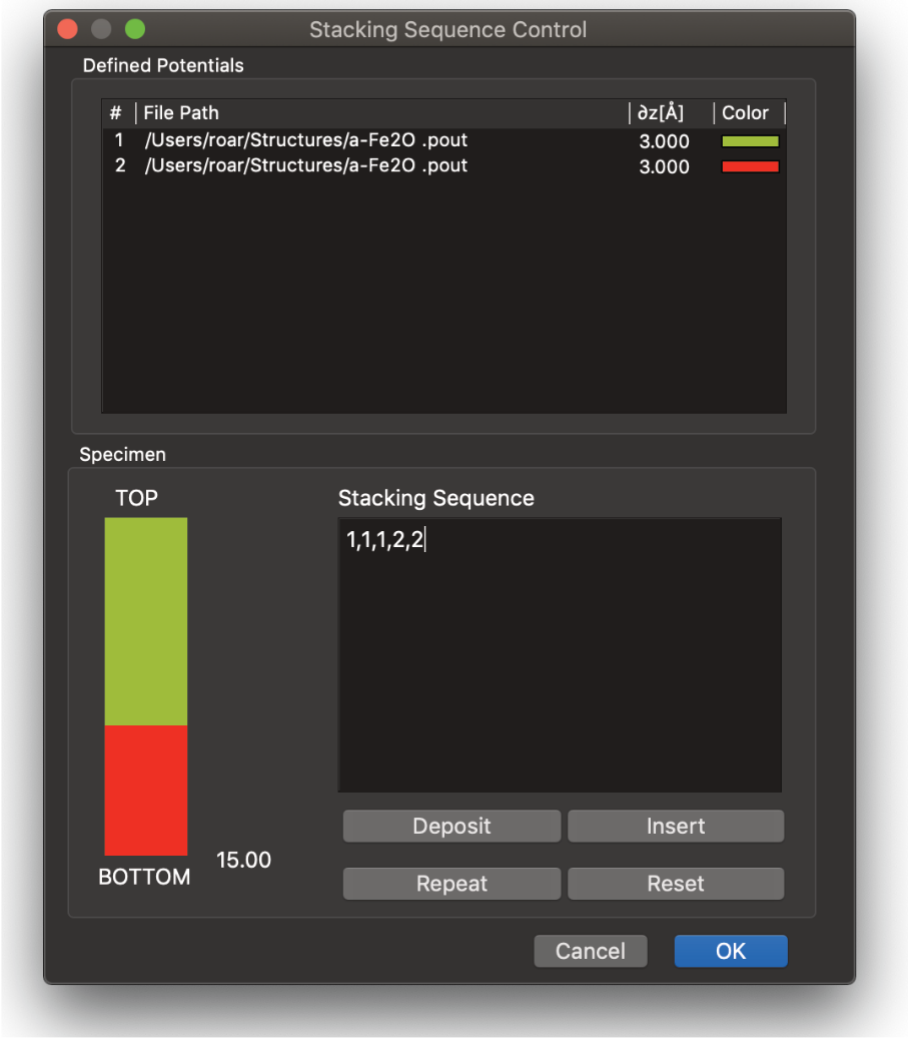


Create a new layered simulation file. A name is prompted for before input is made. Enter a unique structure name, the program will append the extension .lay. A layered structure is characterized by being made up of a sequence of pre-calculated projected potentials. Thus a layered simulation file does not contain atomic positions. Only the lattice parameters A and B and Gamma have meaning for a layered "structure". The buttons "S**et Phasegratings**" and "**Define Stacking**" are used to choose the different projected potentials and to define the sequence that makes up the entire specimen.



Set Phasegratings will set up for choosing precomputed "pout" files which should all have the same lattice parameters a and b.

Define Stacking will help set up the sequence of phasegrating files that makes up the specimen, which brings up the following dialog.



Open Structure File...

Open an existing structure or a layered file. The standard file open dialog is presented and only files of the type 'TEXT" with the extension ".at" or "lay" are displayed as selectable. The name of the

display window will change to reflect the name of the current structure.

Recent Simulation Files Menu

Shows the list of recently opened simulations and lets the user select the file to open.

Close

Close the file, image or window currently selected

Save...

Save the current information depending on the current active window. If the active window is a simulation window, the action is to save the structure file.

Save As...

Save the current structural information or image. For a simulation, do not append the extension as Tempas will automatically choose the correct extension.

The next 2 menu commands shown in the image shown of the file menu do only show up if the Scripting Module has been acctivated as part of the licensing.

Open Image...

Open an image. Supported images are currently tiff files and binary files. RGB tiff files and compressed tiff files are

not supported. Binary files can be of integer or float types (real or complex) with different length and byte order.

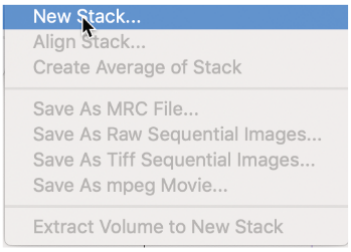
Recent Images Menu

will show the list of recently opened images.

New Image...

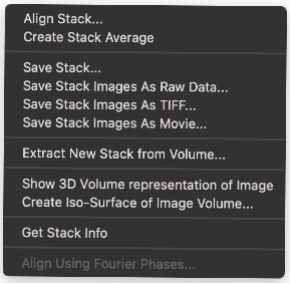
Create a new image of specified size and content..

Image Stack Menu



Allows for the creation of an image stack from a sequence of images. When an image stack is active in an Image Window, the actions in the menu are used to operate on the image stack.

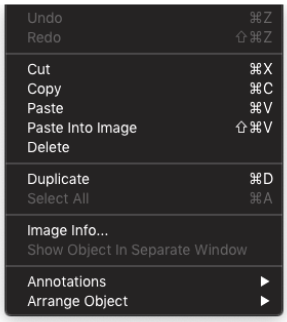
The Image Stack will also have a contextual menu associated with it.



Print...

Invokes the OS Print command for the active window.

Edit Menu



Undo

Undo / Redo the last operation. These operations do not currently work in Tempas.

Cut

Cut the selected Object or the Selection made by the selection tool.

Copy

Copy the selection or the selected object.

Paste

Paste the content of the paste buffer into the display window. The source for the paste can be an image cut out from another application or through the cut/copy commands of Tempas. If the object is an image, the image will be pasted into the display window if it is currently selected or into a separate image window if not.

Paste Into...

Paste Into will change depending on the active window. Normally Paste Into will place the object on the clipboard into the currently active Simulation Display Window. It can also place an image into an active Image Window.

Delete

Clears the selected objects made by the selection tool

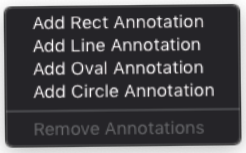
Duplicate
Duplicate will duplicate the selected object, making a copy of the object and displaying it.

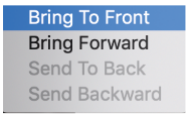
Select All
Select all objects in the display window or an entire image.

Object/Image Info...
Shows the information associated with a selected object if the object is editable. The displayed dialog box will depend on the object being edited.

Show Object in Separate Window
This will take an object in the Simulation window and display the object in its own Window. There may be additional options available with repect to how the object is displayed

Annotations
Create annotations to be placed on an image



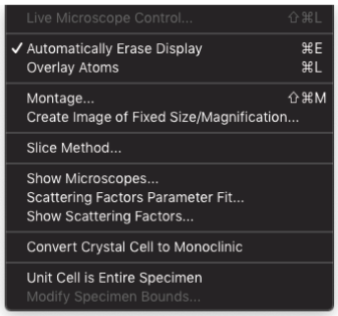


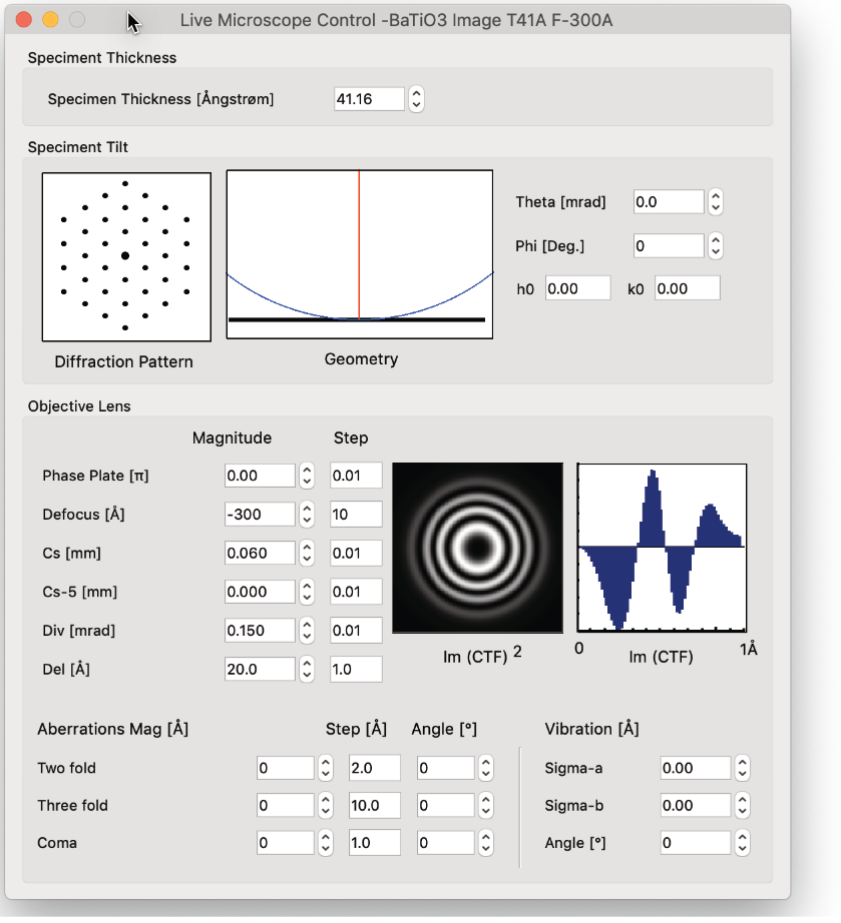
Arrange Object
Will arrange objects in the active Simulation Display window in terms of their stacking order.

Options Menu

Live Microscope Control...

When a calculated image is selected, this command can be invoked to bring up an interactive window for changing the calculation parameters for this image. Changes in the parameters are reflected live as long as the calculation time is reasonable.





Automatically Erase Display

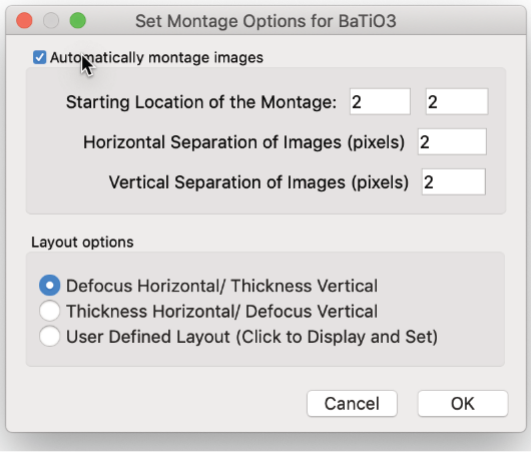
Toggles whether previous content in the active Simulation display window is erased when a new object is added to the window.

Overlay Atoms

If set, the atom positions will be drawn in as circles on top of images. The circles are scaled to the atomic radius and the color is the color set for that atom species.

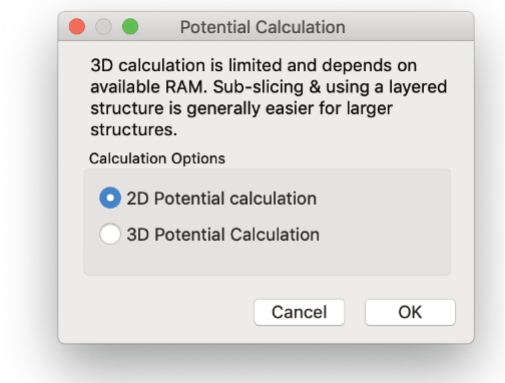
Montage...

Brings up a dialog box, allowing the user to select automatic montage of a series of images, the position of the series of images and the number of pixels to leave between images.



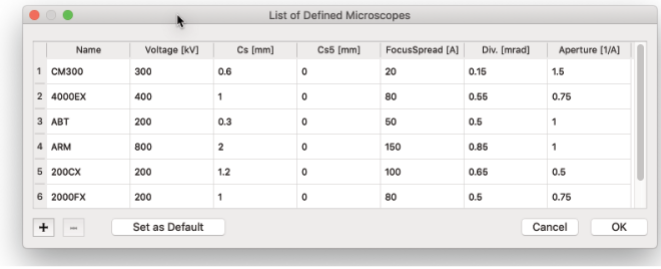
Slice Method...

Allows the user to select the option to perform a three dimensional calculation of the projected potential by summing over the third dimension (l) in reciprocal space.



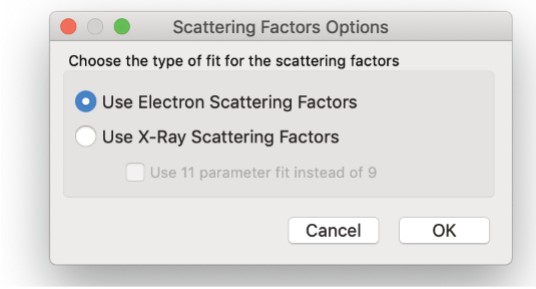
Show Microscopes...

Displays a dialog, showing the user which microscopes are known to Tempas. The default parameters associated with a known microscope can be changed by the user and a new microscope may be made known to Tempas. Additional microscopes can be added and existing ones deleted. A selected microscope can be set as tbe default microscope when such is needed.



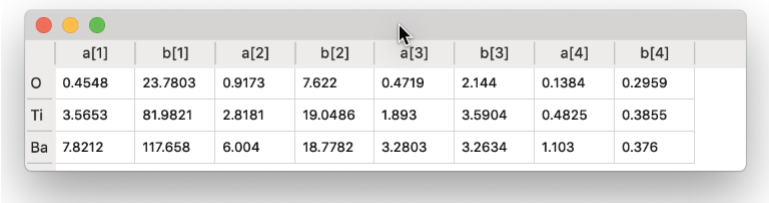
Scattering Factors Parameter Fit...

Tempas can use either the 8 parameter fit for the Electron Scattering Factors or the 9 parameter fit for the X-Ray Scattering factors. The menu item text will reflect the current setting.



Edit Scattering Factors...

Brings up a table of the fitting parameters. The entries are editable.



	a[1]	b[1]	a[2]	b[2]	a[3]	b[3]	a[4]	b[4]
O	0.4548	23.7803	0.9173	7.622	0.4719	2.144	0.1384	0.2959
Ti	3.5653	81.9821	2.8181	19.0486	1.893	3.5904	0.4825	0.3855
Ba	7.8212	117.658	6.004	18.7782	3.2803	3.2634	1.103	0.376

Convert Crystal Cell to Monoclinic

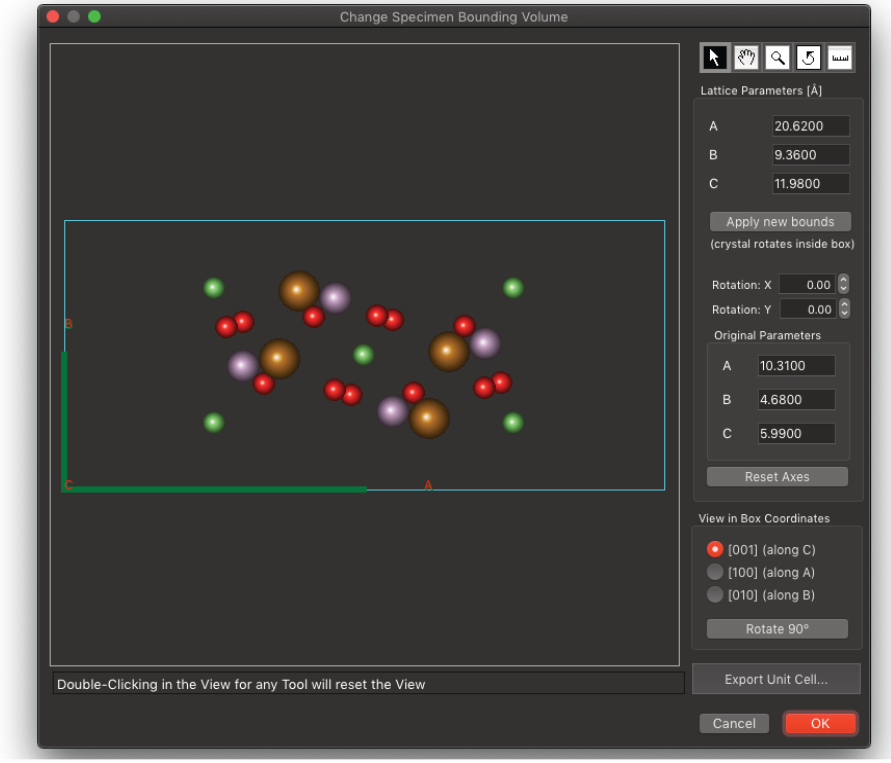
This will covert the current crystal structure to have the symmetry group P1, consisting of a single symmetry operator “x,y,z”. All the atoms in the structure will be moved into the basis of the crystalline cell. The structure will not change, except for the fact that there will be no symmetry operators applied to the basis apart from the identity operator.

Unit Cell is Entire Specimen

When this option is set, the calculation treats the unit cell as a non-repeating structure such that the entire specimen is represented by a single unit cell with the thickness of the specimen as the thickness of the unit cell.

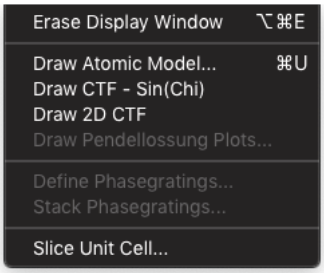
Modify Specimen Bounds

When the unit cell is the entire specimen, this will allow the user to trim the bounding volume for the cell in the specified zone axis orientation.



Commands Menu

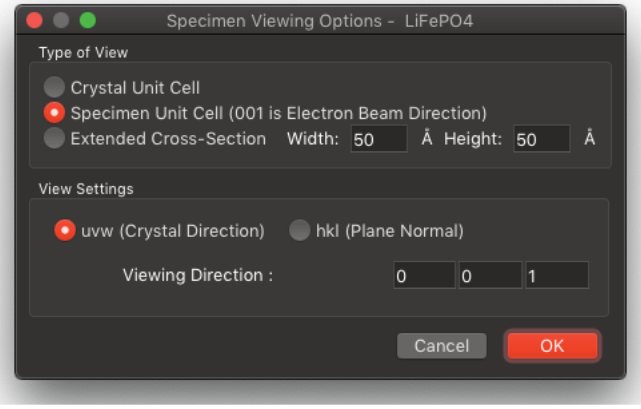
Erase Display Window



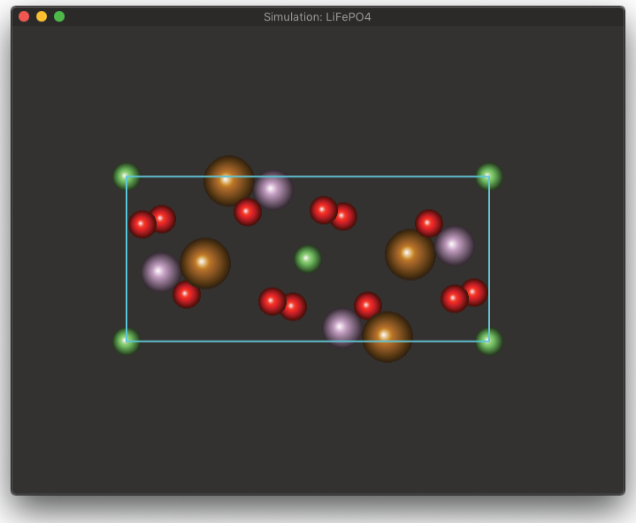
Erases the active Tempas Simulation Display window.

Draw Atomic Model...

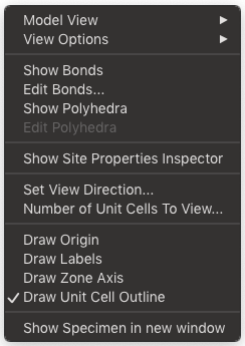
Displays a dialog box, from which the user can select to display the original or transformed unit cell from any direction, including perspective view. The transformed cell corresponds to the unit cell that Tempas uses in the multi-slice calculation. To view the cell as "seen" by the electrons, the transformed (new) unit cell should be viewed in the 001 orientation. It should be noted that the viewing direction is in units of the real space unit cell axes. One can also view a cross-section of the material in a given direction. A dialog box allows the user to specify the field of view in Å for the two directions.



The following windows shows the specimen drawn as it would be used in the multi slice simulation calculation.



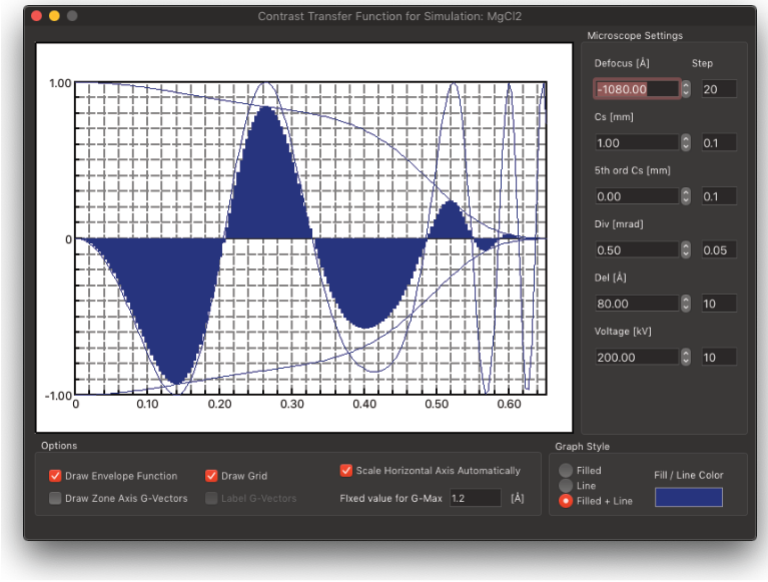
The crystal / specimen model also has a contextual menu that control some of the features associated with showin the model. By right-clicking on the model (when it is selected), the following contextual menu is available.



Draw CTF -Sin (chi)

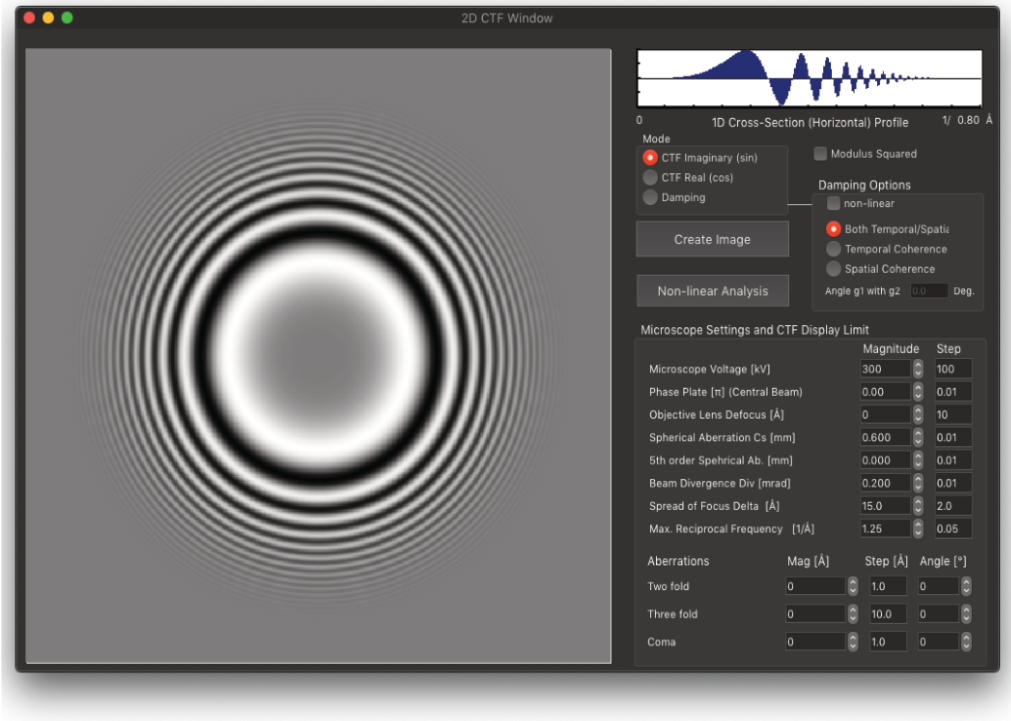
Draws the Contrast Transfer Function for the current microscope values. The original microscope values are taken from the structure

data, but the user is free to change the values associated with the CTF independent of the values used in calculating the image. Clicking in the CTF will show a bar with the values of the CTF and the resolution. The bar moves with the mouse.



Draw 2D CTF...

Draw various representations of the 2D contrast transfer function, for both linear and non-linear imaging.



Non-linear image contributions can be examined in detail through the non-linear analysis button.

Draw Pendelløssung Plots...

In case the user has selected to store a set of diffracted beams for plotting of amplitudes and phases as a function of specimen thickness, this brings up a dialog box allowing the user to set the plotting conditions. One can choose to have the amplitudes or the intensities plotted as well as the phases of the diffracted beams. Each reflection can be plotted by itself, or several reflections can be superimposed on the same plot. Instead of plotting the values, the values can also be written to a file for further manipulation or inspection.

Define Potentials...

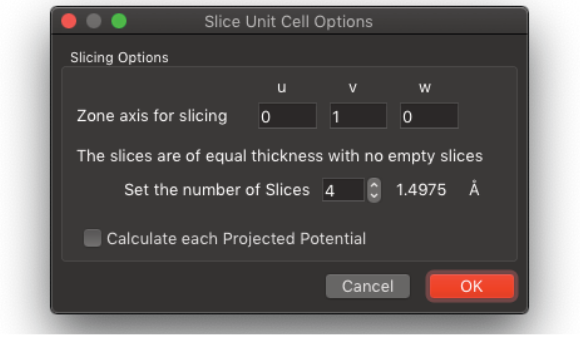
This allows the user to specify which potentials to be used in a layered structure.See Chapter 9 for a more detailed instruction on how to create a layered structure.

Stack Potentials...

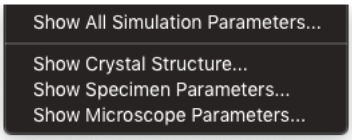
This allows the user to specify the sequence of potentials that should be used in the multi-slice calculation. This applies only to layered structures. See Chapter 9 for a more detailed instruction on how to create a layered structure.

Slice Unit Cell...

Use this option to subdivide a structure into separate layers for use in a layered structure calculation. The direction perpendicular to the slices and the number of slices must be specified.

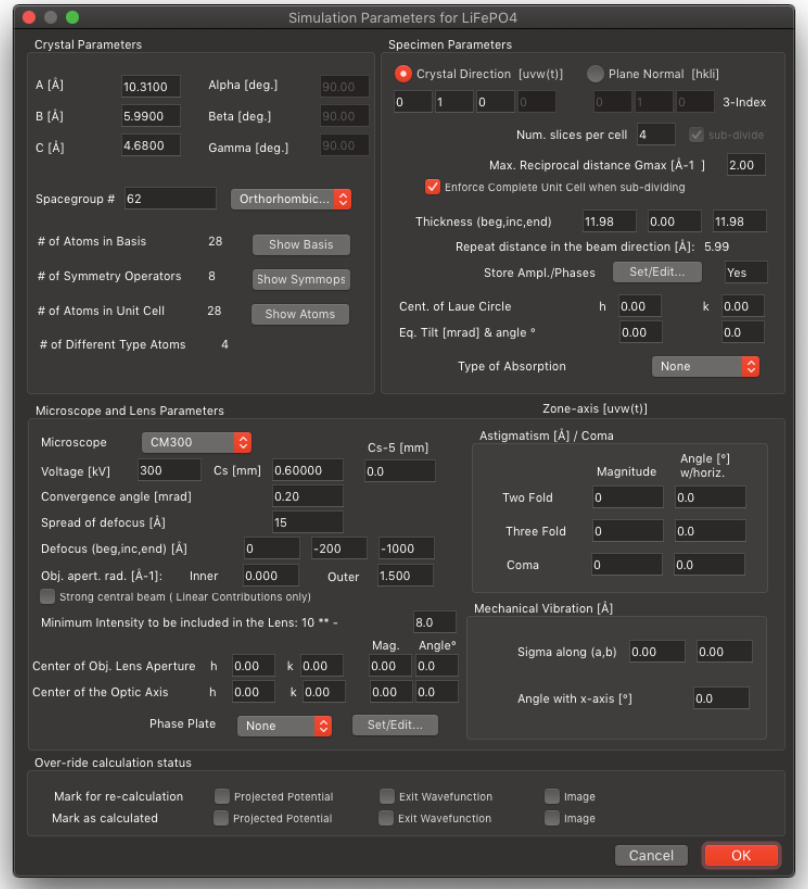


Parameters Menu



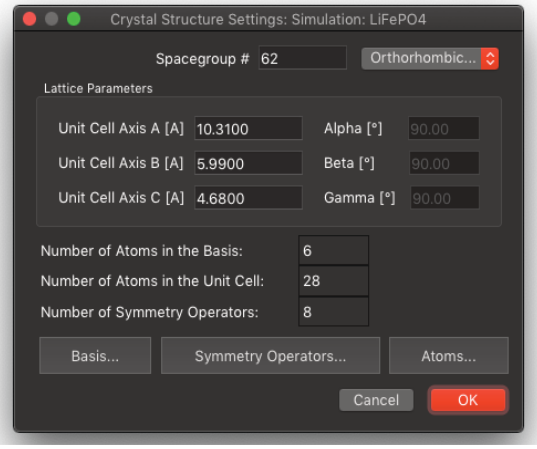
Show All Simulation Parameters...

This brings up a dialog box showing the current conditions for the simulation. The values are taken from the input given to the New... command in the FILE menu. The parameters can be changed as to bring about a new simulation.



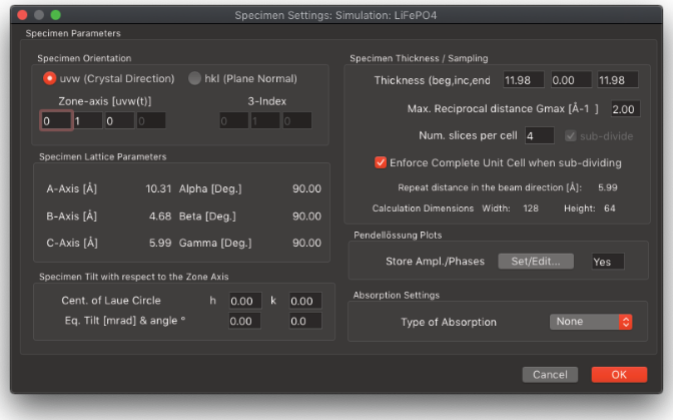
Show Crystal Structure...

Shows in a new Dialog Window all the parameters of the Crystal Structure that is being used. It shows the lattice parameters, such as the length of the aces A,B and C, as well as the associated angles alpha, beta and gamma. In addition it fives access to all the atoms that are part of the motif of the unit cell, the space group and the symmetry operators that are part of the space group.



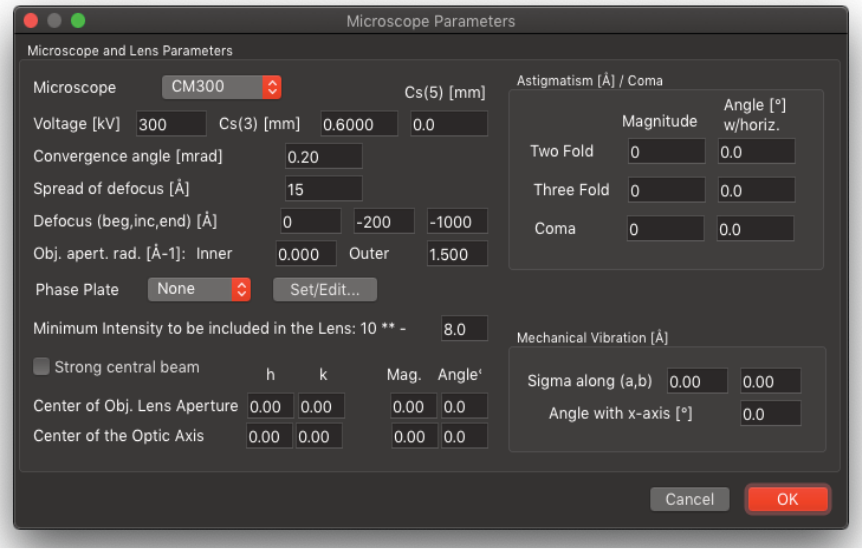
Show Specimen Parameters...

Shows all the parameters associated with the specimen. The specimen is derived from the crystal unit cell based on the direction of the electron beam..

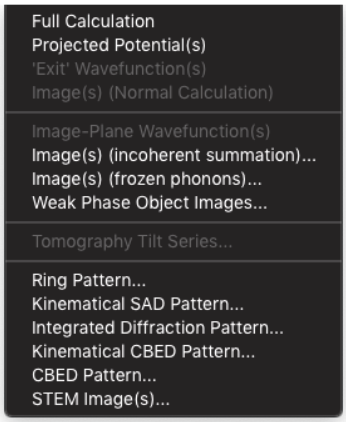


Show Microscope Parameters...

This brings up the list of symmetry operators either associated by the space group or entered manually by the user. The symmetry operators can be edited, and new ones may be added to the list or existing ones deleted.



Calculate Menu



The active commands in this menu depends on the current status of the calculation. If the simulation has already been carried out for the current set of parameters, then no commands will be active. If a change has been made or the file is a newly created structure file, the commands showing which subprograms needs to be run are shown active.

Full Calculation

Use this command if you would like the program to run the multi-slice calculation to its end starting from the point required by the last change made to the simulation parameters.

Projected Potential

Only calculate the projected potential(s)

At the end of the calculation the potential(s) will be displayed in the simulation window in the form of an image displaying 2x2 unit cells. Very large unit cells will show only a 1x1 unit cell image.

Exit wave functions(s)

Only calculates the exit wave-functions(s)

At the end of the calculation the exit wave(s) will be displayed in the simulation window in the form of an image displaying 2x2 unit cells. Very large unit cells will show only a 1x1 unit cell image.

Image(s) (normal calculation)

Calculates the image by using using partial coherence envelope

functions and the lens contrast transfer function.

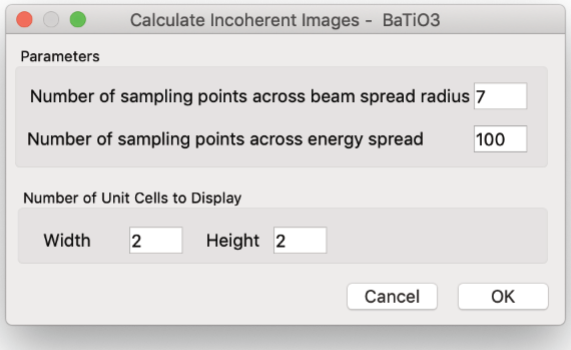
The image(s) will show in the same fashion as other calculation output. If the calculation involves a through thickness-through focus series, the output will be in the form of several image stacks, one for each thickness, containing as many images as there are focus values.

Image-Plane wave functions(s)

Calculates the wave-function at the image plane. This would be equivalent to what would be reconstructed using holography.

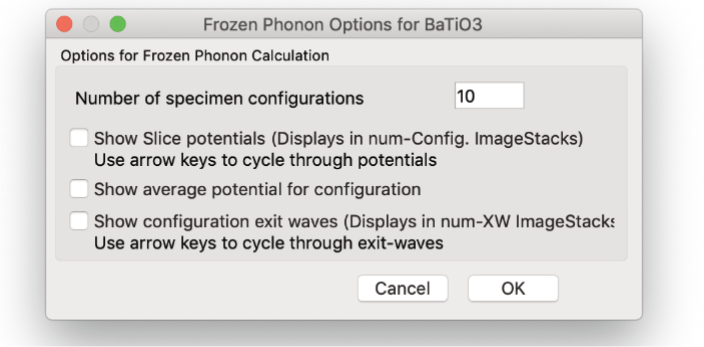
Image(s) (incoherent summation)...

Calculates the image(s), but sums up incoherent images produced by electrons from different incoming directions and different energy.



Image(s) (frozen phonons)...

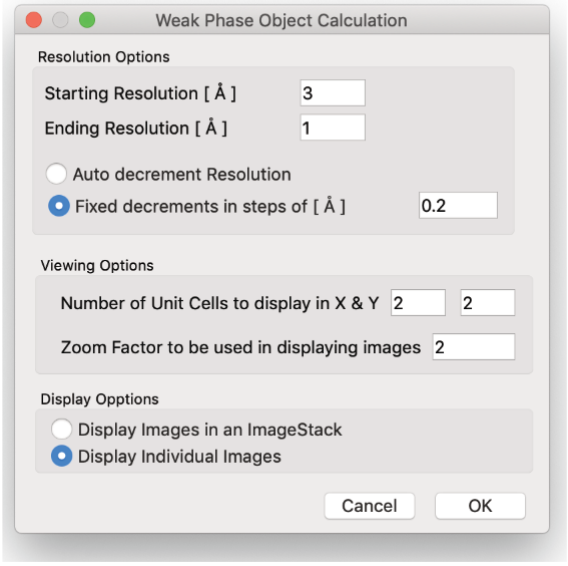
Calculates the image by summing over images calculated from different atomic configurations consistent with random displacements of atoms according to their debye-waller factors.



Weak Phase Object Images

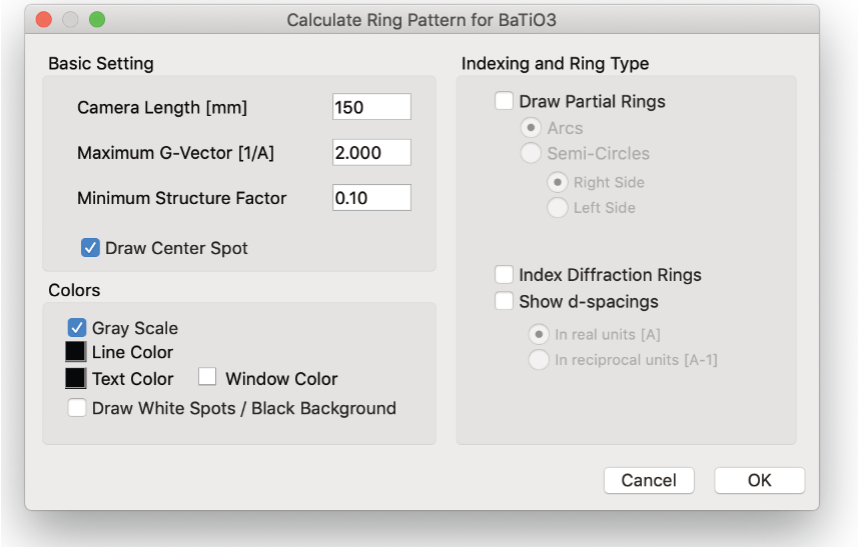
is a separate module that allows the calculation of images that would be produced in the case of an ideal Scherzer lens and validity of the weak phase object approximation.

The "WPO" calculation is discussed more in detail elsewhere.



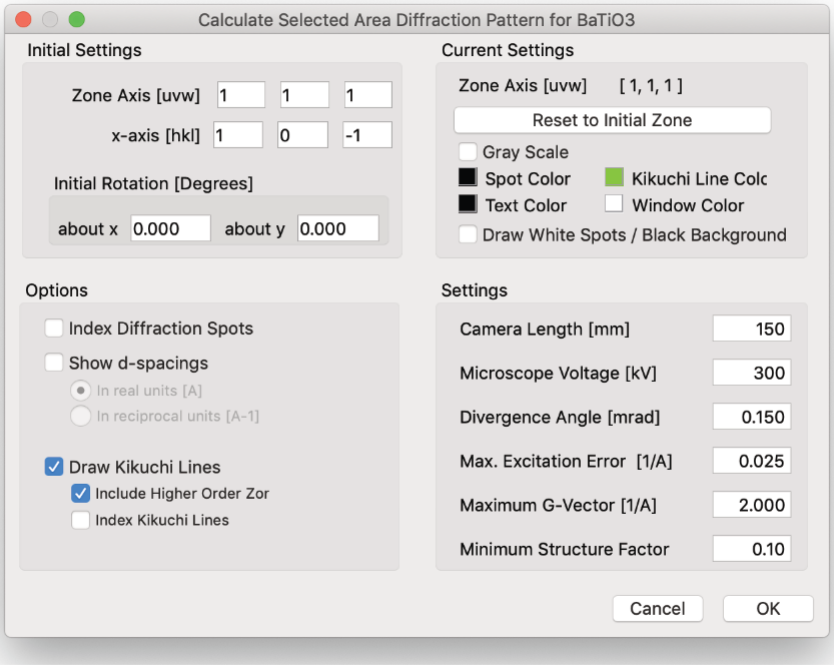
Ring Pattern...

calculates the diffraction pattern by adding up the intensities for each tilt angle within the cone of incident electron directions.



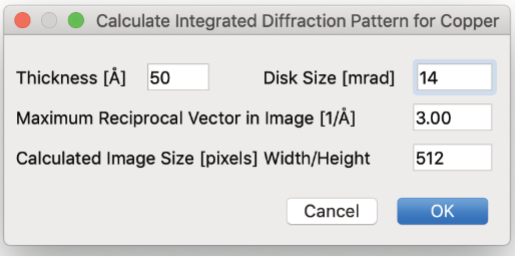
Kinematical SAD Pattern...

calculates the diffraction pattern by adding up the intensities for each tilt angle within the cone of incident electron directions.

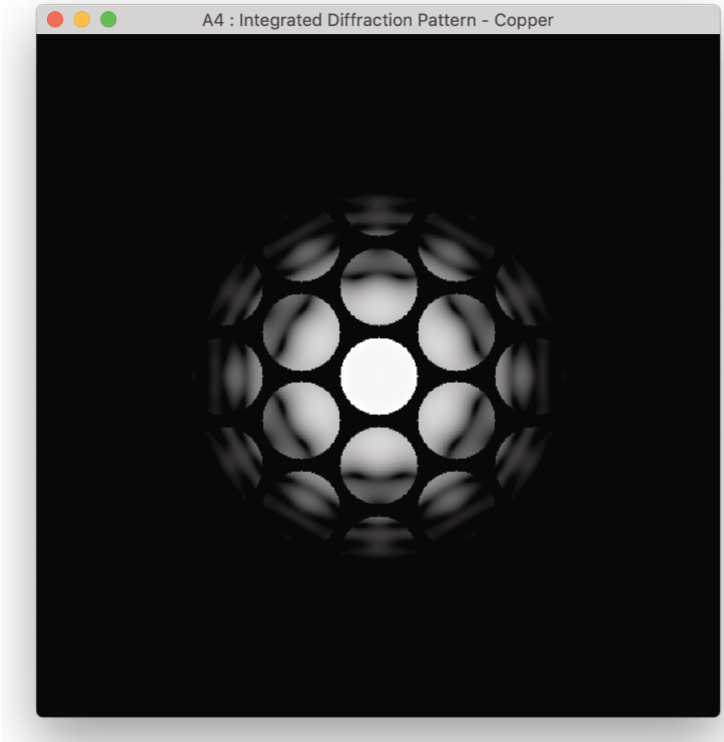


Integrated Diffraction Pattern

calculates the diffraction pattern by adding up the intensities for each tilt angle within the cone of incident electron directions.

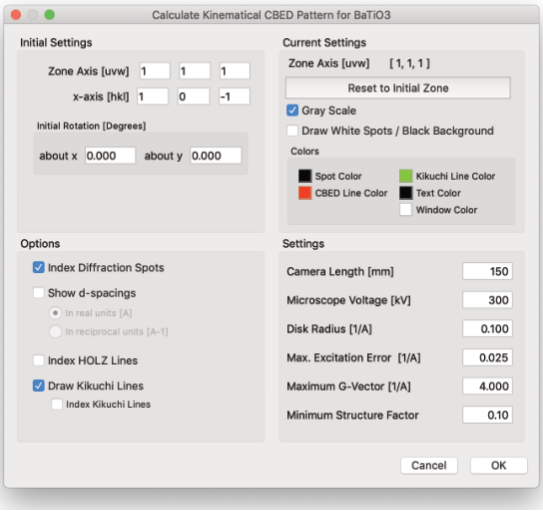


The computed pattern based on the parameters above gives.

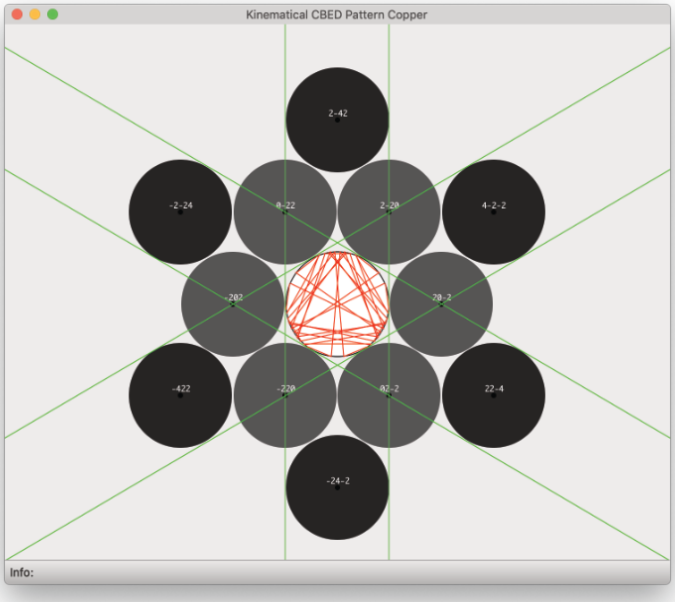


Kinematical CBED Pattern...

Will calculate the Kinematical CBED pattern for the given input parameters. The size of the disks (convergence angle) can be adjusted by dragging the perimeter of the central disk. The shade of the disk is given by the structure factor of the given scattering vector and the HOLZ lines in the central disk are computed from the geometrical description of the lines.



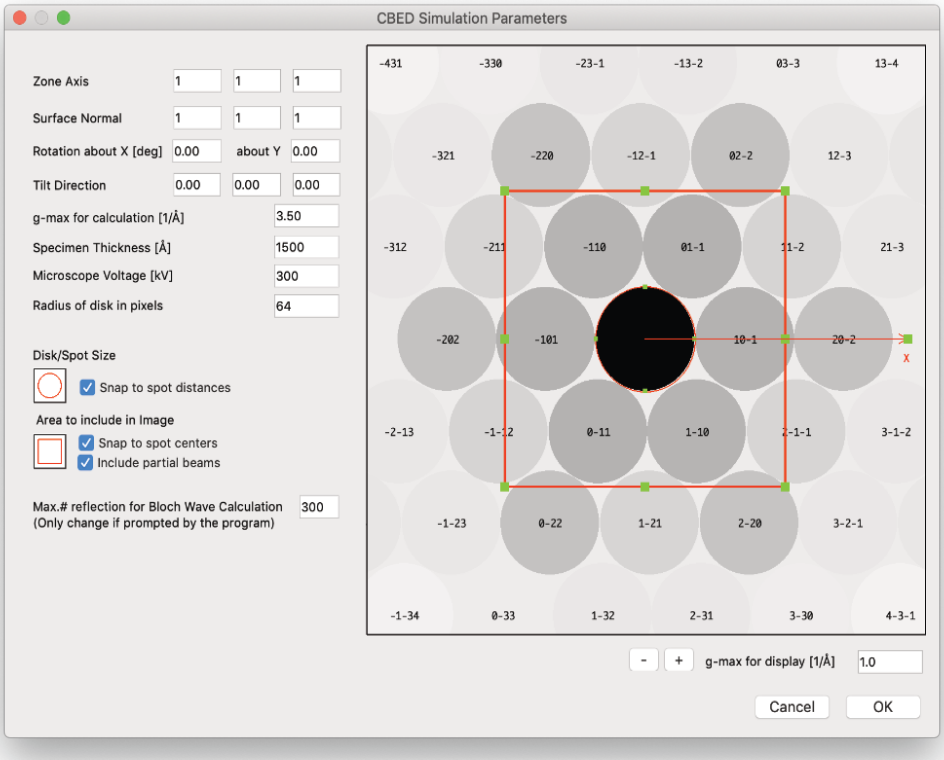
A typical Kinematical CBED pattern is shown below.



CBED Pattern...

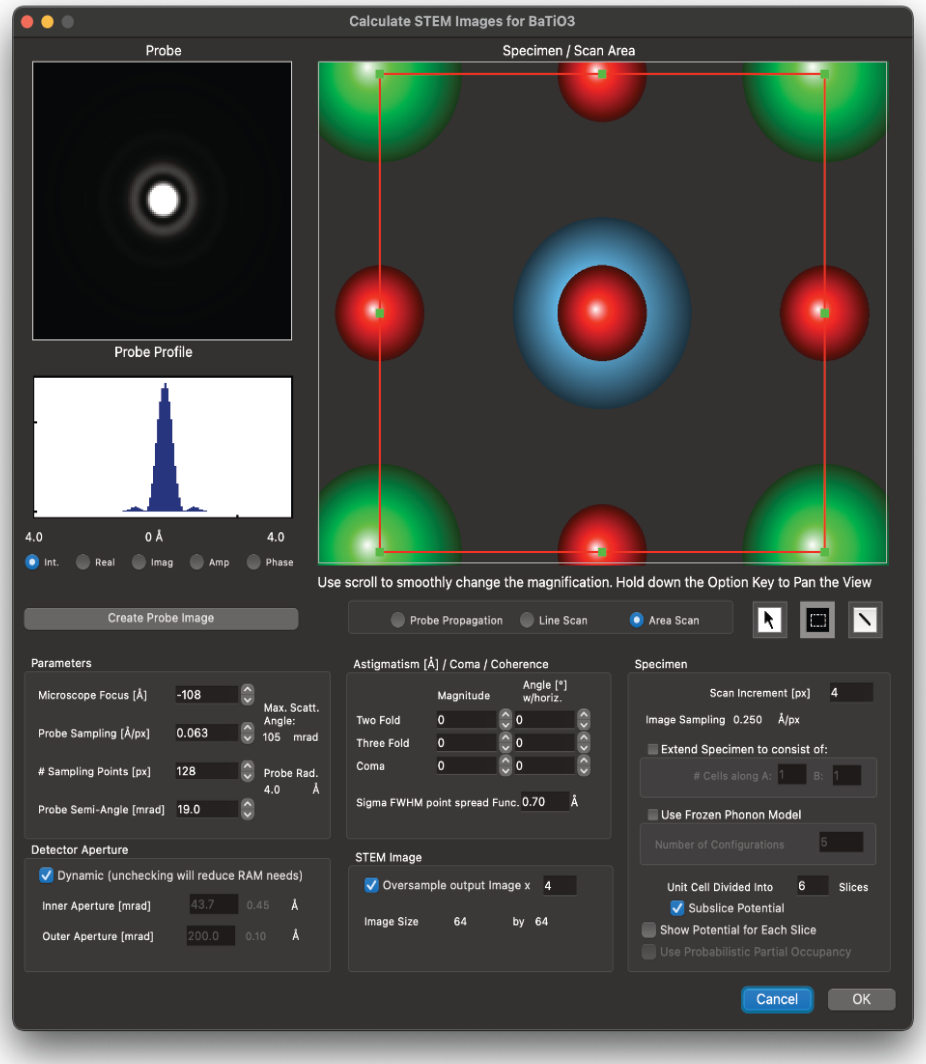
Will calculate the dynamical CBED pattern for the given input parameters using the Bloch-Wave approximation.

The x-axis of the pattern can be set by dragging the green selection rectangle of the x-axis arrow to a diffraction spot. The pattern will rotate so that the "selected" spot is along the x-axis of the pattern. The part of the pattern that will be shown in the calculated image is that which is within the red "Image" rectangle which can be sized and dragged around. The spot-size (convergence angle) is modified by changing the size of the central disk



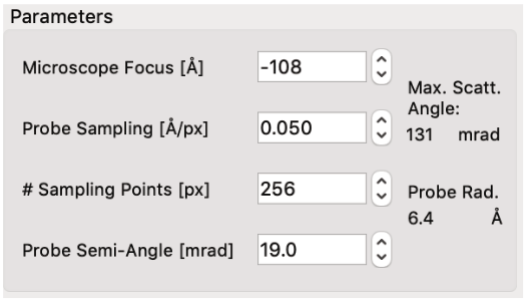
STEM Image...

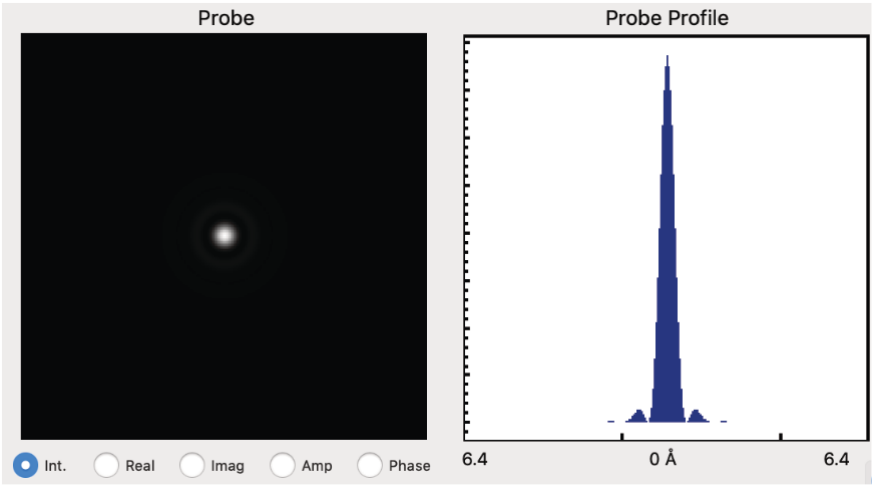
Will calculate the STEM image for the given input parameters using the by doing a multi-slice simulation for each position of the electron probe. The image intensity for that position is the integrated intensity that falls within the annular detector. The dialog below shows the input parameters that determine the calculation of the STEM image(s).



All the parameters in the two dialog-sections below define the probe and its shape. The aberrations are obvious input parameters and represent the microscope conditions. Under ideal circumstances the aberrations are all zero.

!





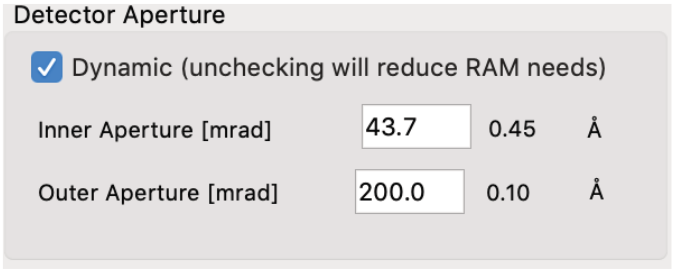
Based on the microscope parameters such as the voltage and spherical aberration, Tempas sets the focus and the probe angle that optimizes the resolution. The user is of course free to set the focus and probe angle that reflect the experimental setup.

The point spread function is an empirical value that limits the resolution of the microscope as it smears out the data and eliminates the high frequency "noise" that originates in the calculation. It becomes a Gaussian low-pass filter as it convolutes the values at each pixel with a response function (Gaussian).

The probe sampling sets the maximum frequency that is included in the calculation as is reflected by the maximum scattering angle in milliradians. The size of the probe is a function of the number of sampling points and the actual sampling interval in Angstrom. One needs to make sure that the scattering angle is such that there are electrons being scattered onto the detector aperture.

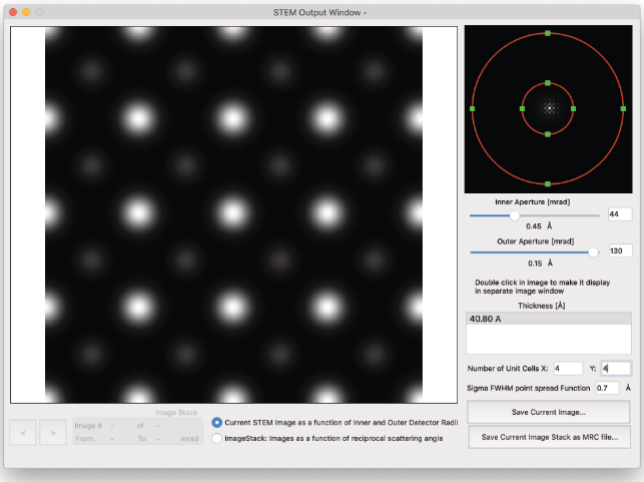
!

The detector aperture is normally given by the physical setup and should reflect this.

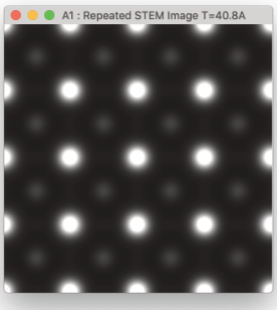


If the checkbox "Dynamic" is checked, the program will output a window (see below) which will allow the user to see how the image vary as a function of the detector configuration (inner and outer apertures).

!



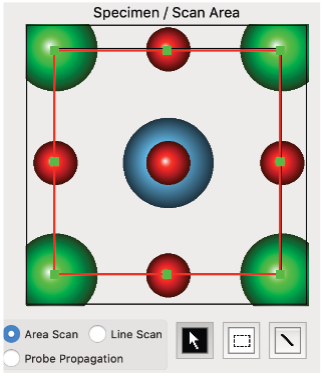
If the box is not checked, the output will be just a single image based on the fixed inner and outer aperture of the annular detector.



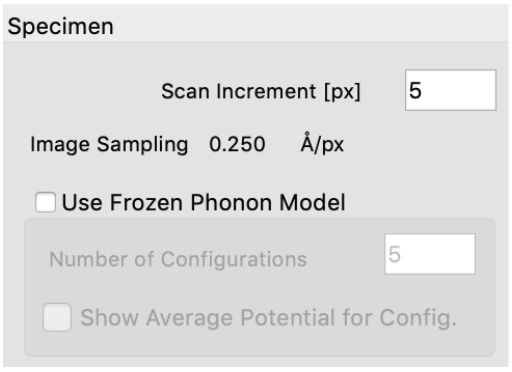
As the sampling needs to be fine enough so that the calculation extend to high g-values (scattering angles), it gives a better resolution than what the microscope is capable of (hence the point spread function introduced earlier).Thus the specimen does not need to be sampled at the same interval. In this case, the specimen is only sampled at every 5 points of the sampling of the potential.

The STEM calculation can be performed on the entire specimen area or a selection of the defined specimen. The area is given by the selection in the specimen area of the dialog. In addition, the calculation can also be done along a line segment. One can also choose to calculate the spreading of the electron beam through the specimen.

!



In order to smooth out the data (the image), the calculated image is oversampled.The oversampling of the image, combined with the point spread function result in a smooth image at a resolution reflecting the actual resolution of the microscope. Of course one has to be careful not to sample to coarsely.



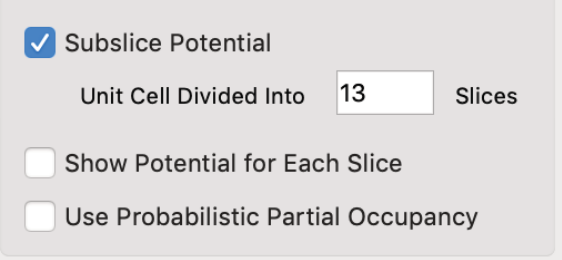
!

The output image is given a dimension of N by M times the unit cell used

for the calculation. The unit cell used in the calculation should be the smallest repeat cell as anything greater just increases the calculation time unnecessarily.

This following setting together with the option to use a Frozen Phonon model determine how the potential is calculated.

The lower checkbox and subsequent value set how many slices the unit cell is divided into.



!

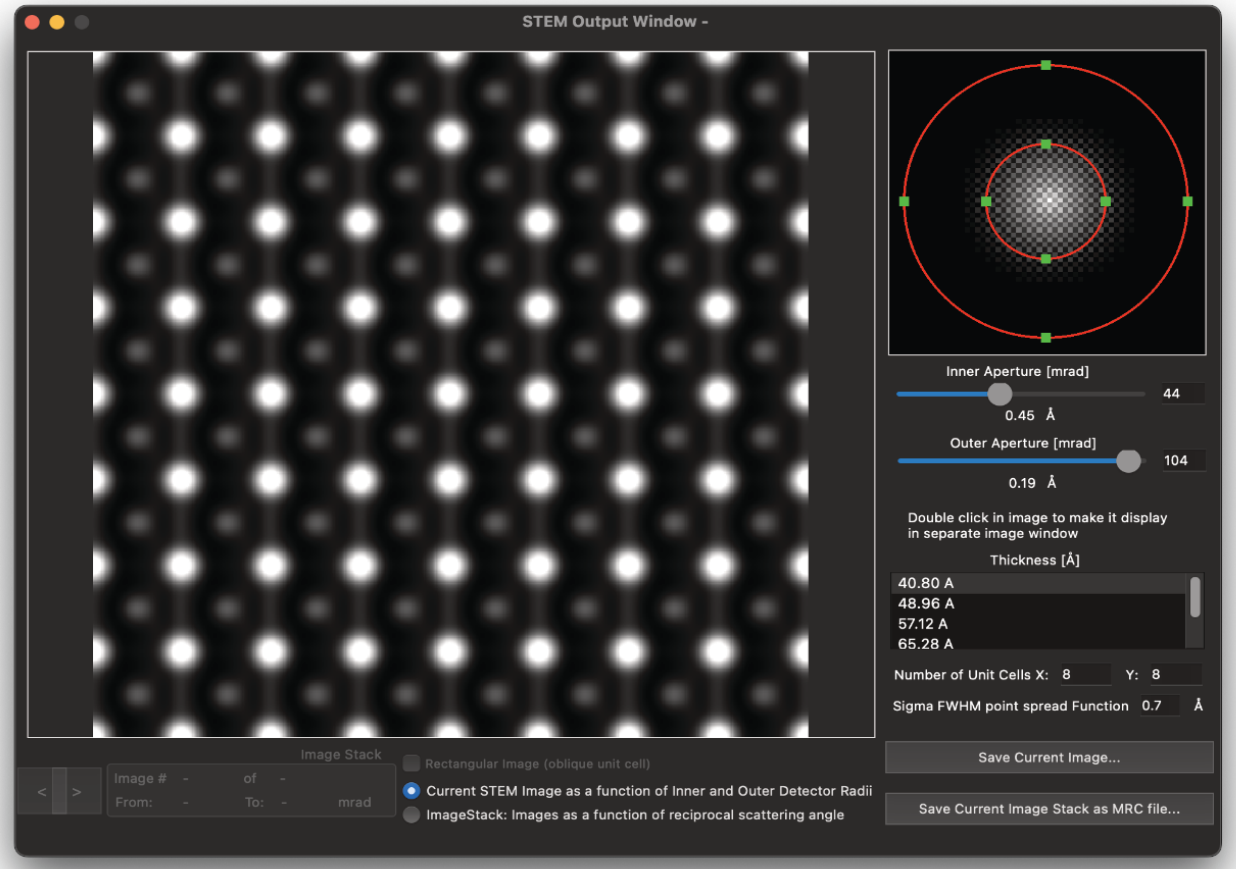
It the option "Use Frozen Phonon Model" is checked, all the atoms in each slice will be randomly displaced following a gaussian distribution with a sigma reflecting the debye-waller factor of the atom(s). The number of configurations will set how many times the multi-slice calculation is performed for the given thickness. Each time, the atoms in every slice is displaced ; each time a new image is produced. At the end, the images are all added up (for each configuration) to produce the final image.
!

Clearly, the calculation time increases as the number of configurations increases. Likewise it is more time consuming to displace atoms and calculate new potentials for every slice. One has to realize that a full multi-slice calculation for the total thickness is performed for every sampling point of the specimen.

As the calculation is spread out over the number of available CPU cores, the calculation time depends on both the core count and the

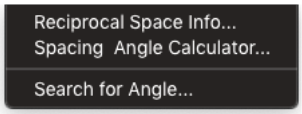
amount of available RAM.

Given that there is sufficient RAM to calculate and keep the data for different values of the detector aperture, the output window will be as following…



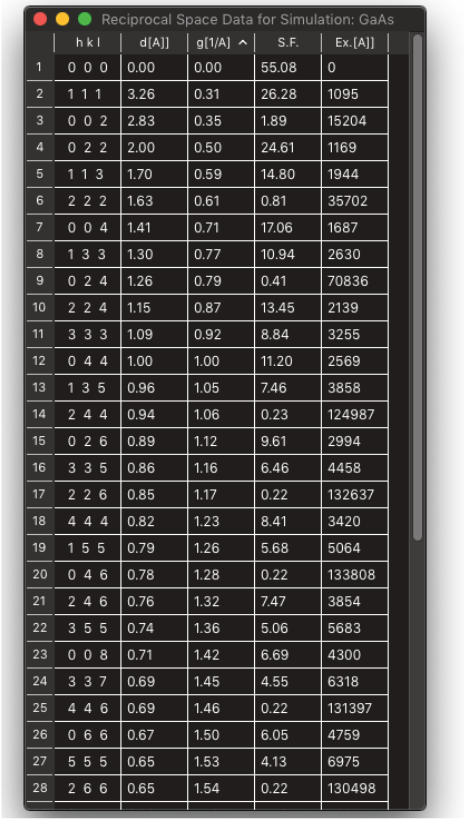
Tables Menu

The current operations in this menu are:



Reciprocal Space Info...

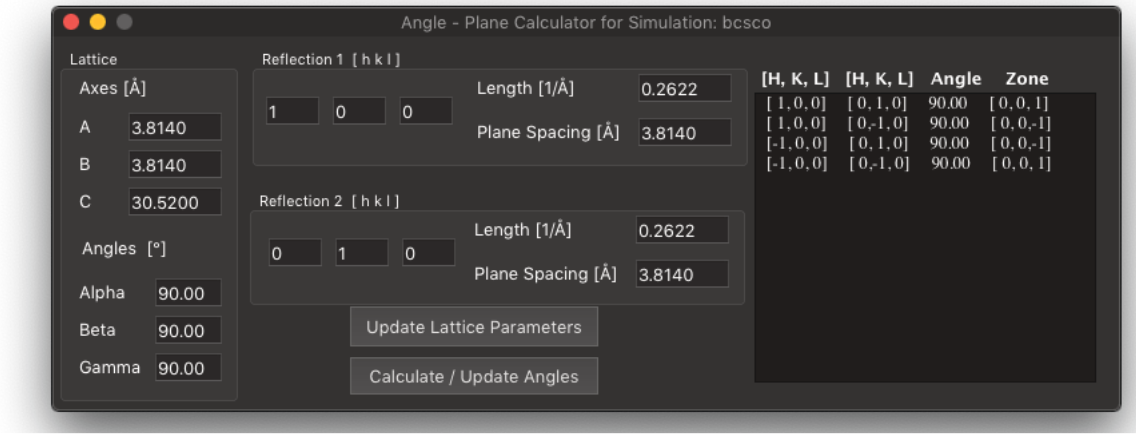
Displays information about reciprocal space data for the current structure. The data can be sorted on the reciprocal vector or the structure factor/extinction distance.



	h k l	d[A]]	g[1/A]	S.F.	Ex.[A]]
1	0 0 0	0.00	0.00	55.08	0
2	1 1 1	3.26	0.31	26.28	1095
3	0 0 2	2.83	0.35	1.89	15204
4	0 2 2	2.00	0.50	24.61	1169
5	1 1 3	1.70	0.59	14.80	1944
6	2 2 2	1.63	0.61	0.81	35702
7	0 0 4	1.41	0.71	17.06	1687
8	1 3 3	1.30	0.77	10.94	2630
9	0 2 4	1.26	0.79	0.41	70836
10	2 2 4	1.15	0.87	13.45	2139
11	3 3 3	1.09	0.92	8.84	3255
12	0 4 4	1.00	1.00	11.20	2569
13	1 3 5	0.96	1.05	7.46	3858
14	2 4 4	0.94	1.06	0.23	124987
15	0 2 6	0.89	1.12	9.61	2994
16	3 3 5	0.86	1.16	6.46	4458
17	2 2 6	0.85	1.17	0.22	132637
18	4 4 4	0.82	1.23	8.41	3420
19	1 5 5	0.79	1.26	5.68	5064
20	0 4 6	0.78	1.28	0.22	133808
21	2 4 6	0.76	1.32	7.47	3854
22	3 5 5	0.74	1.36	5.06	5683
23	0 0 8	0.71	1.42	6.69	4300
24	3 3 7	0.69	1.45	4.55	6318
25	4 4 6	0.69	1.46	0.22	131397
26	0 6 6	0.67	1.50	6.05	4759
27	5 5 5	0.65	1.53	4.13	6975
28	2 6 6	0.65	1.54	0.22	130498

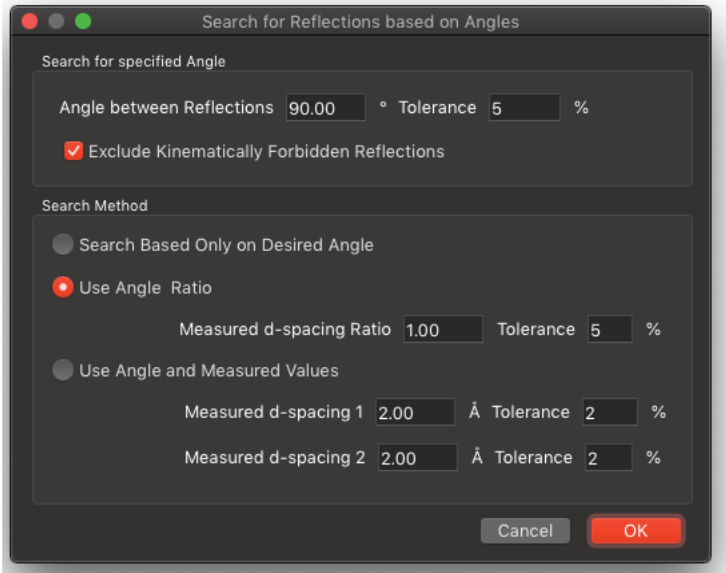
Spacing and Angle Calculator...

Shows the lengths and d-spacing for a type/family of reflections. Two reflections can be displayed and the zone in which a set of reflection exist and the angle between them is shown in the right part of the window.The reflections can be typed in and when the “Calculate?Update Angles” button is clicked, the angles between the family of reflections and the zone in which they occur will show to the right. The length of the vectors and the corresponding plane spacing is updated live when typing in a new reflection. The lattice parameters can be changed and invoked through the “Update Lattice Parameter” such that the calculator can be used for structures different from the one that is loaded.



Search for Angle...

One can search for reflections which satisfy a search criteria based on the ratio of the length of the vectors and/or the angle between the measured reflections. This can be used when trying to index diffraction patterns. One can give an accepted tolerance for the ratio, the measured spacings and the measured angle.

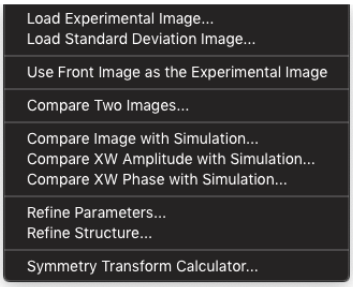


Angle Search Result

	Angle [Deg.]	|d1| [Å]	|d2| [Å]	|d1|/|d2|	Zone Axis
[-1 3 3] / [1 1 1]	48.53	1.30	3.26	0.40	[0 1 -1]
[3 3 -1] / [1 1 1]	48.53	1.30	3.26	0.40	[1 -1 0]
[3 -1 3] / [1 1 1]	48.53	1.30	3.26	0.40	[-1 0 1]
[5 3 -1] / [1 1 1]	46.91	0.96	3.26	0.29	[2 -3 1]
[5 -1 3] / [1 1 1]	46.91	0.96	3.26	0.29	[-2 -1 3]
[-1 5 3] / [1 1 1]	46.91	0.96	3.26	0.29	[1 2 -3]
[3 5 -1] / [1 1 1]	46.91	0.96	3.26	0.29	[3 -2 -1]
[3 -1 5] / [1 1 1]	46.91	0.96	3.26	0.29	[-3 1 2]
[-1 3 5] / [1 1 1]	46.91	0.96	3.26	0.29	[-1 3 -2]
[2 0 6] / [1 1 1]	43.09	0.89	3.26	0.27	[-3 2 1]
[6 2 0] / [1 1 1]	43.09	0.89	3.26	0.27	[1 -3 2]
[0 2 6] / [1 1 1]	43.09	0.89	3.26	0.27	[-2 3 -1]
[2 6 0] / [1 1 1]	43.09	0.89	3.26	0.27	[3 -1 -2]
[0 6 2] / [1 1 1]	43.09	0.89	3.26	0.27	[2 1 -3]

Quantitative Menu

This is the menu for quantitative comparison between experimental images and exit wave functions with calculated data. Structure refinement and imaging parameter refinement is also invoked from this menu.



Operating The Routines

All the routines except for the "Symmetry Transform Calculator" requires the presence of an image that is considered the experimental image. The purpose of this package is to provide the user with tools that permit the following attempts:

Quantitative comparison of experimental images with those simulated from a structure model

Automatic refinement of parameters such as the thickness of the specimen, the defocus of the objective lens, crystal tilt, aberrations, etc..

Automatic refinement of structural parameters such as atomic positions, debye- waller factors and occupancy factors for selected atoms in the structure.

The experimental image can be compared with the computed images using a number of goodness of fit criteria. A sub-area of the experimental image can also be compared to a sub-area of the simulated image. For more information on quantitative comparison methods and structure refinement see the chapter on "Introduction to Quantitative Comparison of simulated HRTEM images with experiment".

Note: All the procedures expect that the experimental image covers the exact area of that of the unit cell used in the calculation. Thus it is up to the user to make sure that the unit cell motif is extracted from the

experimental data prior to usage. The term Unit Cell is loosely used since it only refers to the size of the model used in the simulation. The experimental image does not need to be sampled equal to the simulation, since the routines will resample the experimental data to fit that of the simulation.

For parameter refinement and structure refinement, there are a number of algorithms that attempt to look for the "one" solution of parameters that maximize the fit between the experiment and the simulation. This plug-in uses an algorithm based on simulated thermal annealing which is described further in the chapter at the end of this manual. No claims are made as to the effectiveness of this method and there is no guarantee that the final solution represents the global maximum/minimum in the goodness-of-fit parameter. The effectiveness of optimization routines depend on the starting parameters. There is no recipe for setting the initial starting condition and it is necessary to develop some experience using the optimizing routine in this program. Some trial and error is a definitive part of the parameter/structure refinement. Suffice it to say, "good hunting".

Load Experimental Image

Load Experimental Image is the starting point for loading in the image to be used in the comparison. The command will bring up a standard File Open dialog with a look that depend on the version of the MacOS you are using and what else of Finder utilities you have loaded. If it is a format that does not need additional information, such as tiff, dm3, dm4, mrc, edm and ebn, the images will load normally. If there is additional information needed to load the image, Tempas will present an additional dialog to aid in the loading of the image.

The dialog will allow you to load Binary Files with different data types, byte offset and byte swap.

Once the routine has read in the file, it will display the image in a separate window.

Load Standard Deviation Image

This command allows you to read in a standard deviation image to be used together with the average experimental image for comparison with simulated images. The standard deviation image will be used in conjunction with the average image for computing c-square deviations between the experimental data and the computed data. Otherwise the input works exactly as for the loading of the average experimental image. Again, the image will show up in its own window after it has been loaded.

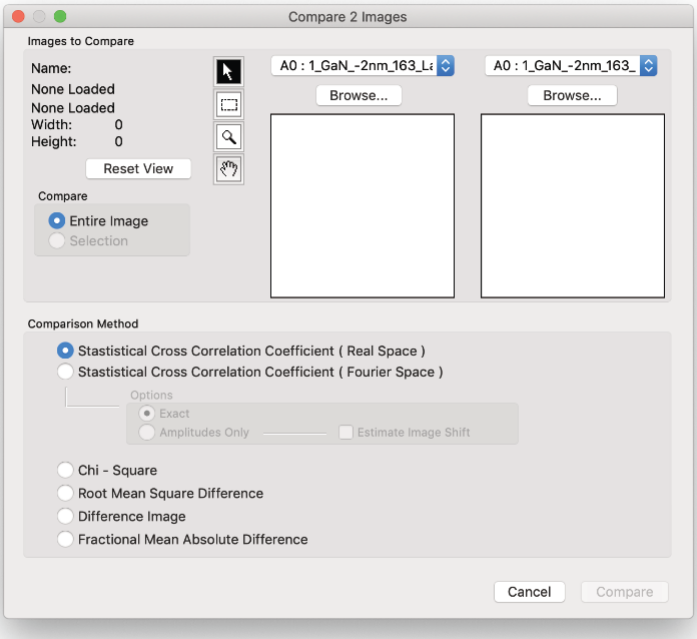
Note: If the windows are covered by other windows, it may be necessary to move these to bring the "experimental" image and the "standard deviation" image to the foreground since there is no command to bring these automatically to the front.

Use Front Image as the Experimental Image

This conveniently will use the image displayed in the active ImageWindow as the experimental image to be compared against.

Compare Two Images...

Will bring up the dialog below.



The methods for comparing etc. are described further in the next section "***Compare Image With Simulation...***".

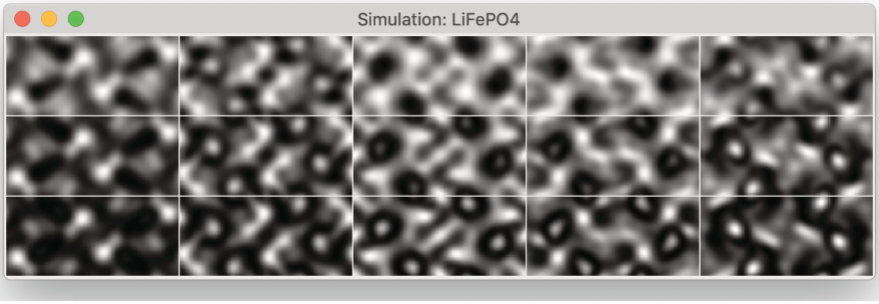
Compare Experiment With Simulation

This brings up a dialog window which shows the loaded experimental image and the standard deviation image (if loaded). The name of the experimental image together with the dimensions are shown to the left of the image.

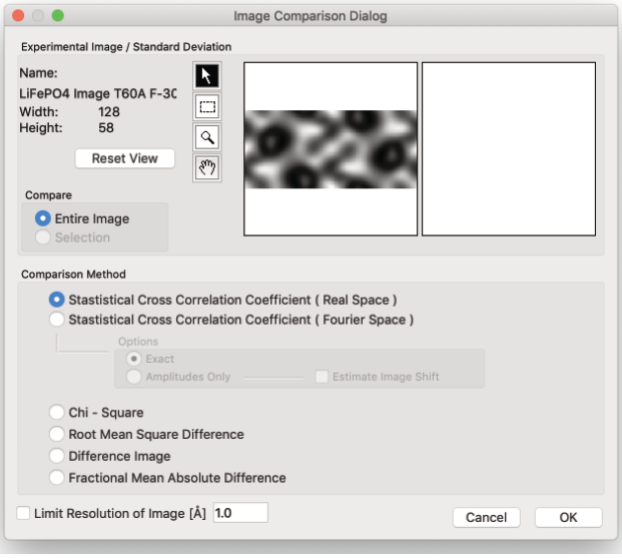
When comparing images, the experimental image will be resampled such that the sampling corresponds to the sampling of the calculated image.

The example used here will be that of a set of simulated images of LiFePO4 in the [010] zone axis orientation

The simulated images for a range of thickness and defocus are shown below.



Selecting one of the simulated images to be used as the experimental image to be compared against, and invoking **Compare Experiment with Simulation...**
will bring up.



Tools

Pointer Tool

Used essentially to ensure that no other tool is active

Selection Tool

Use this to mark an area in the image which will be used for comparison. It is okay to go outside the image, since the selection will be cropped to the actual image.

Hand Tool

Use this tool to move the image within the display area.

Magnification Tool

Use this tool to magnify the image. Holding down the Option key when clicking within the image will reduce the magnification. Holding down the shift key when clicking will increase/decrease

the magnification by a factor of 2.

Reset View

Click here to reset the view/pan/zoom of the image.

S*ELECTING AN AREA FOR COMPARISON*

The user choose to compare the entire image region or only a selection of the image.

Two radio buttons are provided:

Compare Entire Image

By clicking on this radio button, the selection area will automatically be set to be that of the entire image. The comparison area is shown as a red rectangle.

Selection

The user must first use the selection tool to mark a region of the image. Upon clicking on the radio button, the selection will be marked in red and this area will be used for comparing images. For changing an already defined regions, mark a new one and click again in the radio button.

Note: The selection does not work for the Difference Image which automatically compares the entire image.

Selecting a method of comparison

The method of comparison or goodness of fit criteria is chosen by a set of radio buttons and the choices and the corresponding description and output is given below. The goodness of fit criteria is computed for each image that is compared and displayed in a table. This table can be saved in a text file for further use.

Statistical cross correlation coefficient (CCC):

A note about the cross correlation coefficient:
The cross correlation coefficient measures similarity in the pattern between the experimental image and the computed image. Since the images to be compared are set to a mean level of 0 and normalized, any scaling of the type I(exp) = a * I(calc) + b would give a cross correlation coefficient of 1. Exact fit is given as CCC = 1 and a reverse contrast gives CCC = -1.

Real Space

This is a straight calculation of the normalized cross correlation coefficient between the experiment and the calculated image(s). For it to give meaningful results, the origin of the experimental image neeeds to coincide with the calculated image.

	-100	-200	-300	-400	-500
30	-0.3558	0.0720	0.9292	0.6541	0.2777
60	-0.2067	-0.0084	0.9996	0.6383	0.1589
90	-0.0808	-0.0946	0.8766	0.5175	0.0466

Reciprocal Space

There are two options for calculating the cross correlation in reciprocal space.

Exact

The first is an exact calculation which is equivalent to the real space CCC. The entire Fourier transform of the experimental image is compared with the Fourier transform of the simulated image and the CCC is just the reciprocal space equivalent of the calculation in

real space. Thus the reciprocal space CCC is equal to the real space CCC.

A5 : FTXCF

	-100	-200	-300	-400	-500
30	-0.3558	0.0720	0.9292	0.6541	0.2777
60	-0.2067	-0.0084	**0.9996**	0.6383	0.1589
90	-0.0808	-0.0946	0.8766	0.5175	0.0466

.

Only Amplitudes

This calculates the CCC based only on the Fourier amplitudes and (optionally) the program will try to estimate the shift between the experimental and the calculated image. This can be very useful for aligning images and to compare images when the origin of the experimental image is not known. The numbers given are the shift of the origin that should be attempted to be made on the experimental image before the experimental data is compared to the simulation using the exact formulation.

A6 : FTXCF-Amplitudes

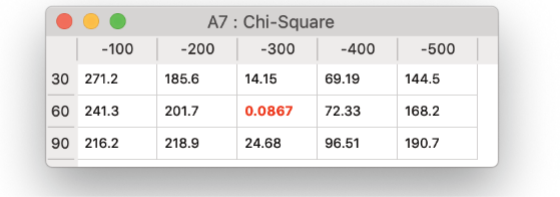
	-100	-200	-300	-400	-500
30	0.7497 0,-16	0.9194 -64,0	0.9371 0,0	0.8519 -64,-16	0.8242 0,-16
60	0.7575 -64,0	0.8630 0,-16	**1.0000** **0,0**	0.8336 -64,-16	0.7423 0,-16
90	0.7781 -64,0	0.8107 -64,0	0.8798 0,0	0.6988 0,-16	0.5795 0,-16

Chi-Square

This computes the chi-square deviation between the calculated image(s) and the experimental images, optionally using the standard deviation image for obtaining the uncertainty associated with the average pixel value.

If no standard deviation image is given, the uncertainty in the pixel value is set to 10% of the intensity in the pixel. The Chi-square goodness of fit

criteria is sensitive to the mean level of the images and the scales the calculated images so that they have the same mean as the experiment before computing chi-square. Lower values indicate better fit, with a value of 1 meaning that the experiment and simulation agree within the uncertainty of the experimental values.



	-100	-200	-300	-400	-500
30	271.2	185.6	14.15	69.19	144.5
60	241.3	201.7	0.0867	72.33	168.2
90	216.2	218.9	24.68	96.51	190.7

Root Mean Square Difference

This calculates the root mean square difference between the experiment and the simulation. Lower values indicate better fit with 0 being exact fit between experiment and simulation. Since the values depend on the mean level of the images that are compared, the simulation is scaled to have the same mean as the experiment.

The table below shows the result from the RMSD comparison.

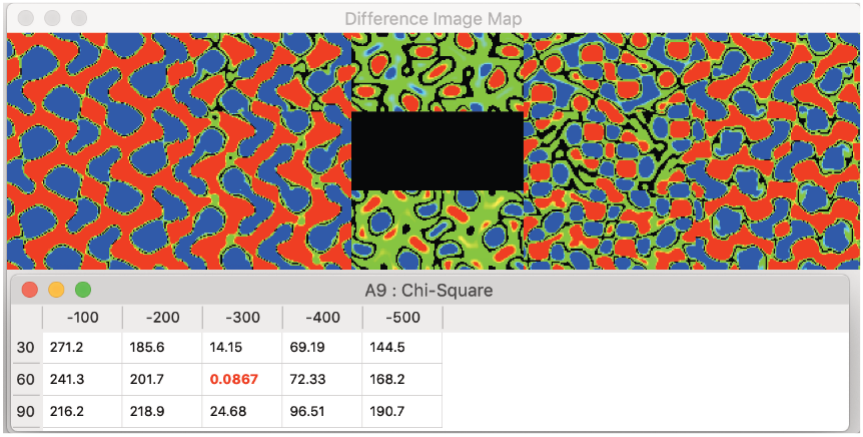
	-100	-200	-300	-400	-500
30	0.8406	0.6955	0.1921	0.4246	0.6136
60	0.7931	0.7250	0.0151	0.4342	0.6621
90	0.7505	0.7553	0.2536	0.5015	0.7049

Difference Image

There are two calculations performed in this case. One gives a difference image(s) which are pseudo colored such that where the experiment and simulation agree within one standard deviation, the pixel is black, less than 1.5 standard deviations,

the pixel is colored green, and outside the pixels are shown in shades of red or blue

depending on whether the values in the simulation are lower or higher than those in the experiment.



	-100	-200	-300	-400	-500
30	271.2	185.6	14.15	69.19	144.5
60	241.3	201.7	0.0867	72.33	168.2
90	216.2	218.9	24.68	96.51	190.7

In addition to the difference image(s), the chi-squared value is also computed for each image and shown in its own table.

Fractional Mean Absolute Difference

This calculates the fractional mean absolute difference between the experimental data and the simulated data.

A10 : FMAD

	-100	-200	-300	-400	-500
30	0.7218	0.5527	0.1448	0.3268	0.4878
60	0.6872	0.5905	0.0119	0.3060	0.5171
90	0.6465	0.6238	0.1808	0.3771	0.5422

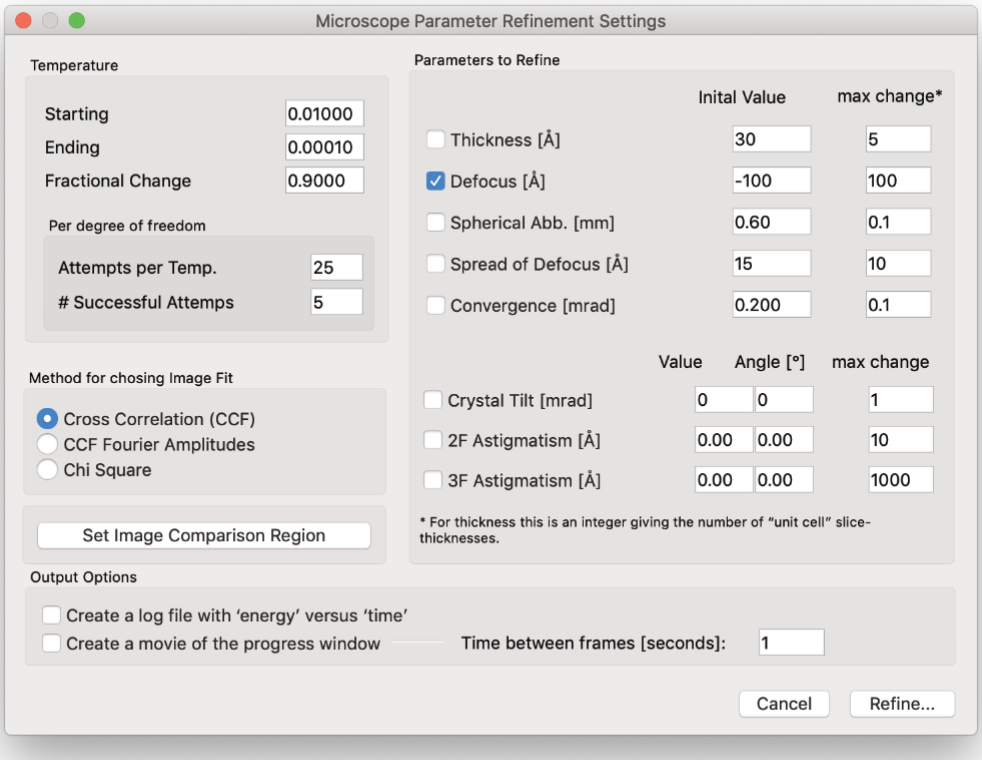
Compare XW Amplitude/Phase with simulation

works as with images, but the experimental data is compared with the exit wave-function (either amplitude or phase)

Refine Parameters

Refine Parameters... will attempt to find the image simulation parameters that produce a simulated image that gives the "best" fit between the experiment and the theory. The search routine is based on the concept of "Simulated Thermal Annealing" and requires a starting configuration of parameters, a starting temperature and an ending temperature, a maximum change in parameters and a goodness of fit criteria that measures the "Energy" of the system. For further explanation of Simulated Thermal Annealing, see the last chapter in this manual.

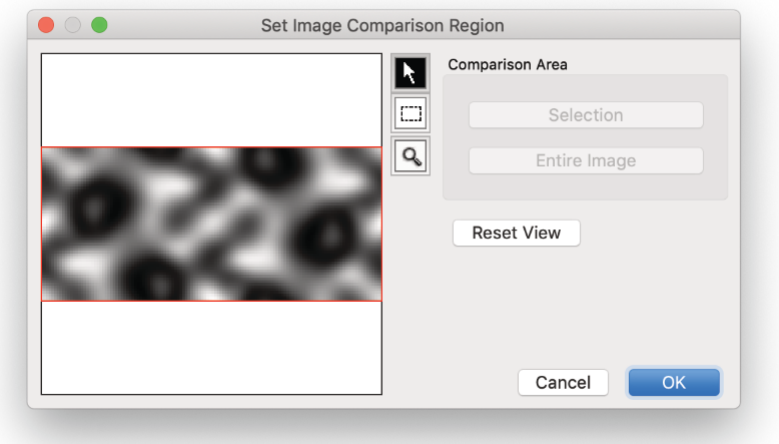
Invoking the command brings up the following dialog which is used to give the input parameters to the algorithm.



Setting Image Comparison region

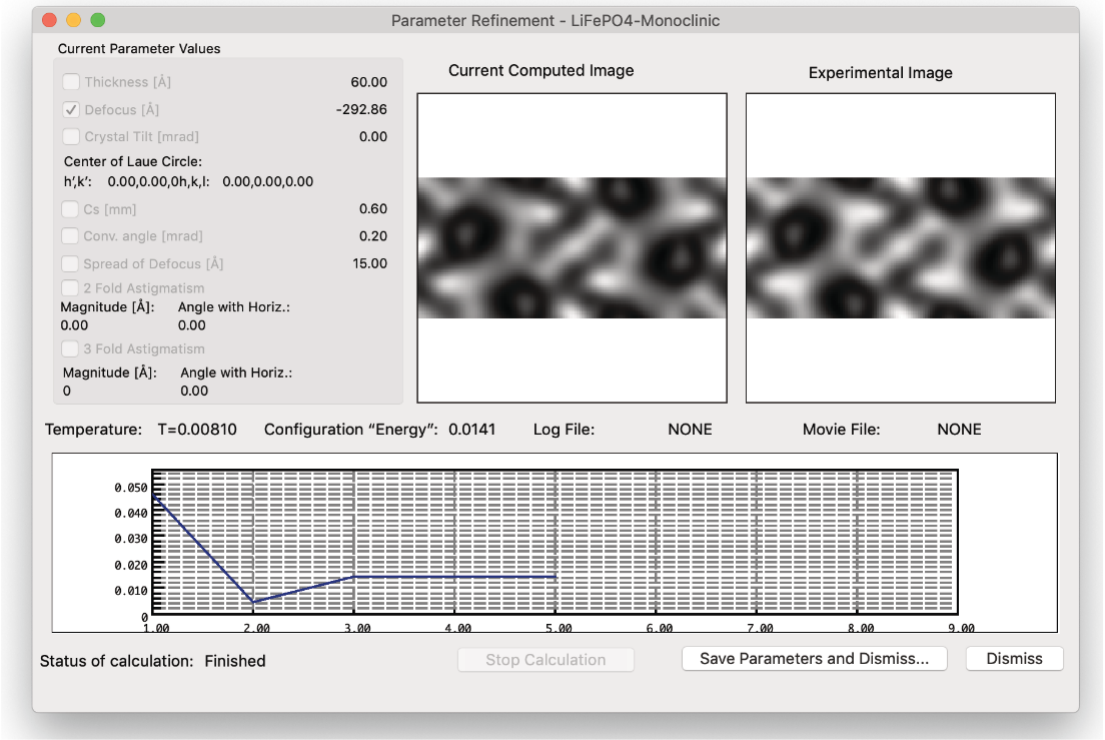
The area to be used for the comparison is set using the button "Set Image Comparison Region"

Selecting the image region is done in the same fashion as under comparing images in the previous section.



Running the parameter refinement

After selecting the area (not needed if the entire image is compared, which is the default) and selecting the parameters to be refined and clicking OK, a progress window for the parameter refinement appears. If the refinement is not progressing in a satisfactory way, the computation can be canceled by hitting the Apple-Command Key together with the Period. Optionally, a log file can be written so that the "energy" as a function of time/temperature can be plotted and also a Movie can be produced and played back. The number of frames/second for output can be set.



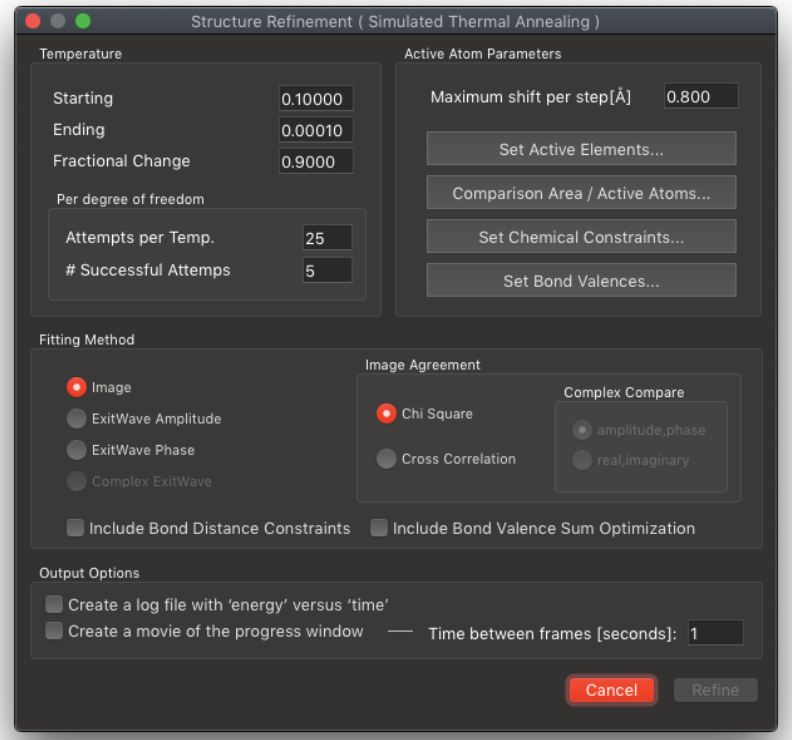
At the end of the run, the dialog box can be just dismissed and the final configuration of parameters will be discarded, or the parameters can be saved in the form of a new structure/simulation file.

Refine Structure

The structure refinement works in the same way as the parameter refinement. However, in this case it is the structure that is being varied, notably the coordinates of selected atoms and possibly debye-waller factors and occupancy. After invoking the command, the following dialog box appears.

There are several options associated with the structure refinement, such as which elements are active, which coordinates to be varied, etc.. The

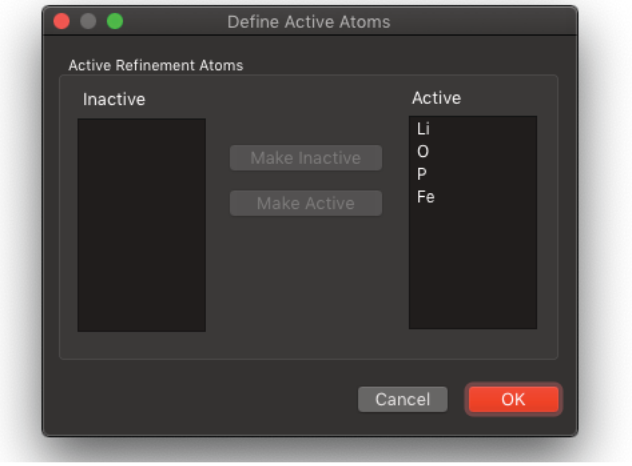
standard parameters for the simulated thermal annealing need to be specified, together with the goodness of fit parameter. Output options such as a log file and movie are identical to that under parameter refinement. In addition, it is possible to include in the calculation of the "Configuration Energy" terms that depend on selected bond distances and selected bond valence sum.



Set Active Elements

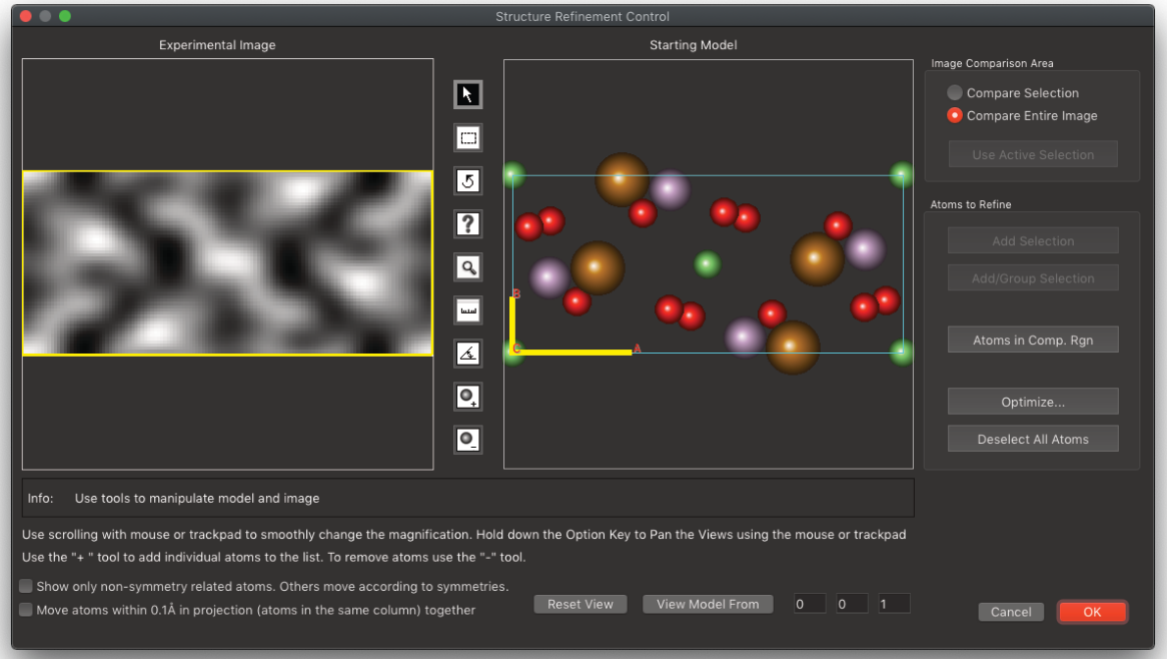
brings up the following dialog. By default all elements are active. Selecting an active element and clicking on the "Make inactive" button will move the selected element to the Inactive list. Just because an element is listed under active elements does not mean that one of its types will be used in the refinement procedure. It is

necessary to specify the area used for comparison and which atoms are to be optimized before the simulated thermal annealing is carried out.



Set Area to Compare & Active Atoms

is used to set the image comparison region (as previously explained) and to select atoms for refinement. The following window appears.



Area used for comparison

Selection

Sets the comparison region to the selection rectangle.

Entire Image

Sets the comparison region to the entire image.

Atoms to Refine

Add Selection

*I*ncludes in the list of atoms the ones that fall within the selection rectangle set in the model window.

Add/Group Selection

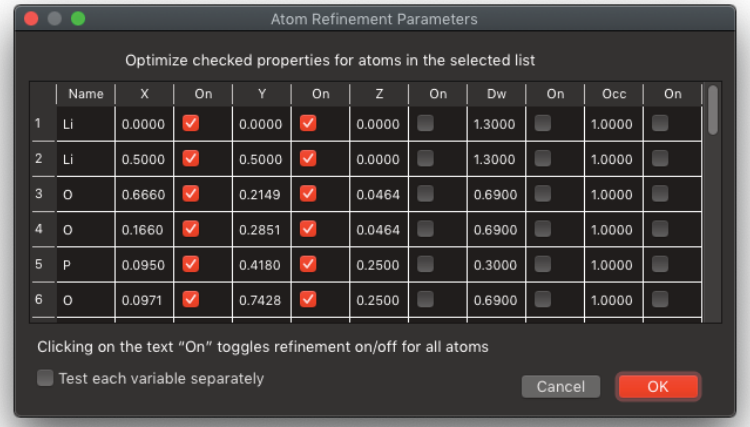
Adds the atoms to the refinement list, but constrains all the atoms in this list to move as a unit.

Atoms In Comparison Region

Sets the refinement list to include the atoms that lie within the rectangle set for comparing images.

Optimize...

Specifies which properties of the selected atoms (those in the refinement list) that are varied. The following window appears.



Normally varying the z-coordinate will not have any effect since the image simulation procedure uses a 2d projection of the atoms in the unit cell. However, in the case where symmetry operators are used, changing he z-position may result in changes in x and y for symmetry related atoms.

Each time an atom (or group) is varied, a new set of parameters for this atom/group is created and tested. If "Test each variable separately" is checked, each try will only vary one parameter. Thus if x and y are checked above, both x and y for an atom are changed at the same time unless "Test each...." is checked, in which case one time x is varied and another time y is varied.

Deselect All Atoms

clears the refinement list and allows the user to start all over, defining the set of atoms to optimize.

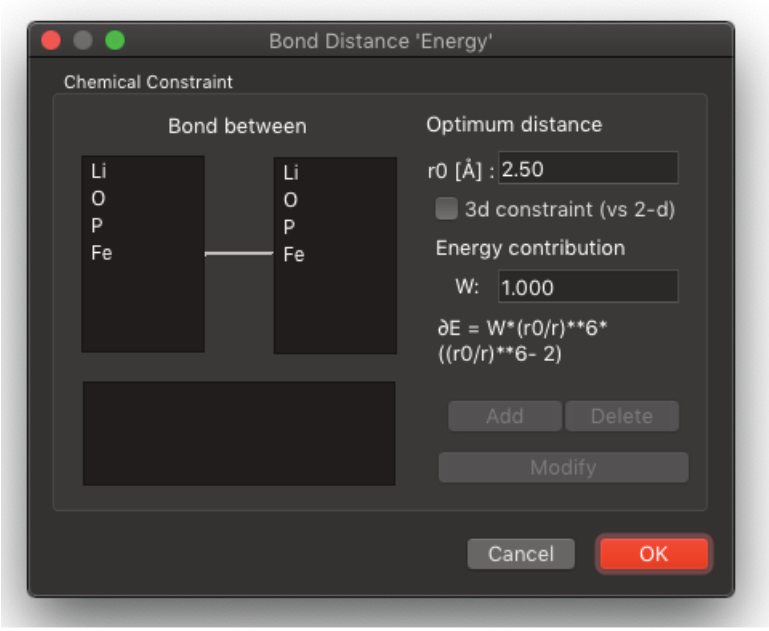
Show only symmetry related atoms/Use Symmetry elements

If the structure is a perfect crystal defined by a set of basis atoms and a set of symmetry operators, it is possible to refine positions of the atoms in the basis and to move symmetry related atoms accordingly such as to preserve the crystalline spacegroup.

Checking this options will result in only atoms in the basis to be visible in the model. Thus the atoms to be refined is selected from the basis and any change in the basis is reflected in the entire structure.

Set Chemical Constraints

brings up the following window.



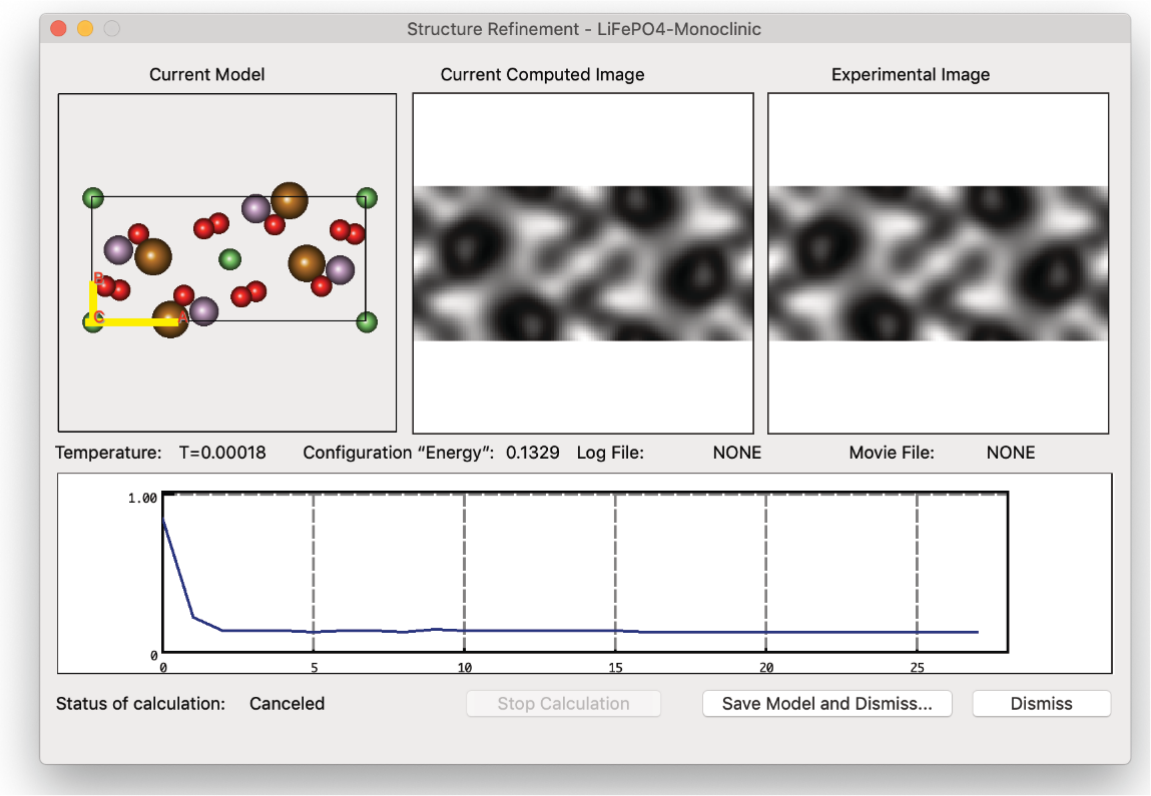
which can be used to specify a configuration energy that includes terms that depend on bond distances. It is possible to specify the optimum bond distance between two atoms, whether it is a 2d or 3d constraint, and to specify the weight of the term in the energy calculation. Even though bond distances have been defined, the inclusion of a bond-distance energy term can be turned off/on from the main "Refine Structure" window.

Include Bond Valence Sum Optimization

This allows the bond valence sum to be used to measure the energy of the configuration. It brings up the following window in which the active bonds are specified, the maximum distance between atoms for calculating the bond valence sum, the valence of the first atom in the atom-pair and the bond constant. Each "bond" has its own weight and the entire bond-valence energy terms has an overall weight.

Running the Refinement

Once the parameters are set and OK has been clicked, a progress window appears.



The current atomic configuration is shown together with the corresponding simulated energy. The energy as a function of "time/temperature" is shown in its own window and can be monitored to ensure that the system moves in a desirable fashion. It is important to understand that no specific recipe can be given to ensure that the system finds a meaningful minimum in the configuration energy. The success of the optimization depends on how far the starting configuration is from the "solution" and the choice of annealing parameters. It is not a straight forward, just run and you

get the correct answer, black box approach. An understanding of the system, a good feel for choosing a reasonable starting structure and some experience in choosing "annealing parameters" is definitely a requirement in order to have confidence in the resulting ending configuration. Experimenting with different input parameters is advised.

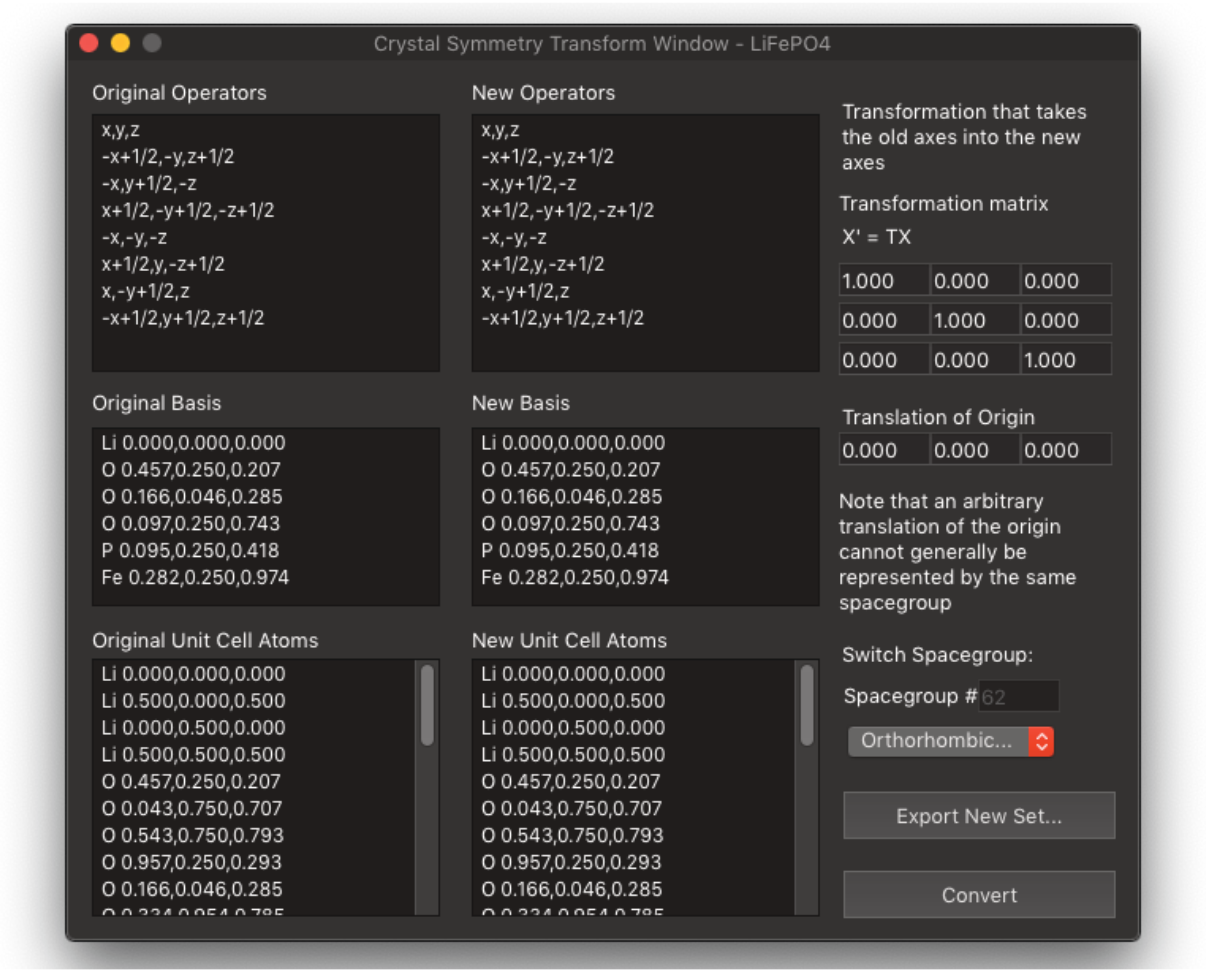
As with refining simulation parameters, it is possible to save a log file or to produce a movie of the annealing process. Saving a movie can be very useful when the refinement takes hours/days and it is necessary to see how the system varied over time. It is important to realize that a movie file grows quickly in size and that the number of frames/second should be chosen appropriately if the movie is to cover a period of many hours.

At the end of the run, the final structure can be dismissed or saved as a structure file for later use. The run can also be aborted by the normal "Command + Period" key combination.

Symmetry Transform Calculator

Choosing the "Symmetry Transform Calculator" brings up a modal dialog window that can help the user find another set of symmetry operators and basis atom positions in the case of a change in axis and origin of the unit cell. Input is the original symmetry operators which come from the spacegroup that is in use together with the original basis atoms. The new unit cell axes a',b' and c' are given by the transformation matrix T such that together with a translation of the origin specified in the old unit cell system (fractional coordinates).

Invoking the command brings up the following window.



Lists

Original Operators

This list shows the symmetry operators of the current spacegroup. Initially these are those of the crystal in use, but the spacegroup can be changed by the command “Switch Spacegroup”.

Original Basis

The basis atoms are the atoms of the current crystal.

Original Unit Cell Atoms

This list gives the atoms that are produce by the operation of the symmetry operators of the spacegroup in use on the original basis atoms.

New Operators

These operators are the result of applying the transformation operations that are given by the change in coordinate system together with a translation of the origin to the generators of the original spacegroup.

New Basis

This is the transformed basis.

Original Unit Cell Atoms

This list gives the atoms that are produced by the operation of the new symmetry operators of the spacegroup used on the new set of basis atoms.

Convert

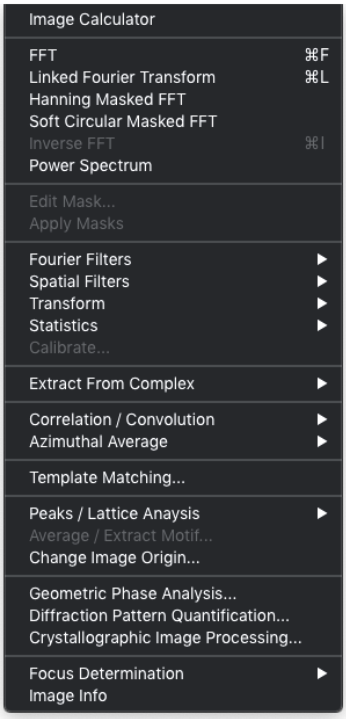
Clicking on this button initiates the computation of the transformed set of symmetry operators, the transformed basis and the new atomic positions.

Note:

It is important to realize that arbitrary input does not result in a symmetry which still can be presented by the same spacegroup with a change in symmetry operators.

Export New Set

Allows the user to export the new basis and the symmetry operators as a new structure file.

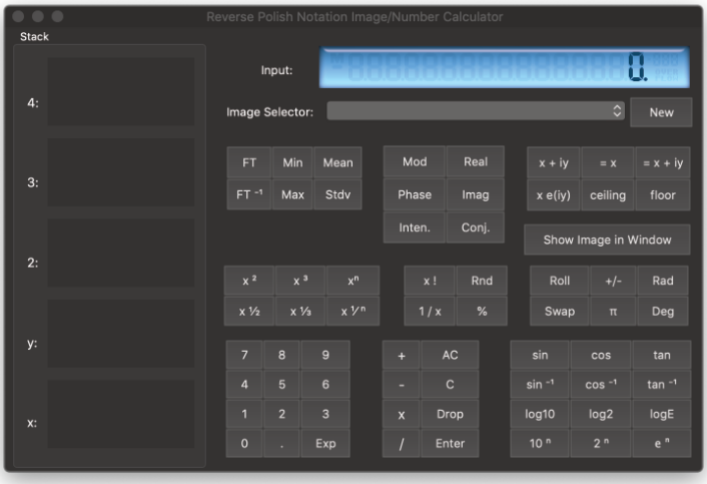


Process Menu

The Process menu is the largest menu and is the source of all image processing functions. There are menu sub-menus as well.

Image Calculator

Perform polish notation (HP style calculator).mathematics on images and numbers



FFT

Perform a Fourier transform on a 2-dimensional image. Unlike many implementations, the image is not required fo have dimensions that are powers of two. However, the image must be square.

When taking the Fourier transform from a script or invoked through the Image Calculator, the requirement that the image is square is lifted. In this case the Fourier Transform image is shown with unequal sampling in the u and v directions.

Linked Fourier Transform

The linked Fourier Transform functions in 2 different ways, depending on whether there is a selection present in the image.

S*election present...*

brings up the Fourier transform of the selection in its own window, together with the original image. If the selection us moved around in the image, the Fourier transform will update to show the transform of the selection.

N*o selection...*

brings up a new window with the transform of the original image. Any mask that is placed on the Fourier transform will be reflected on the original image "instantaneously". As the mask(s) is(are) modified, the image will be updated.

Hanning Masked FFT

Perform a Fourier transform on a 2-dimensional image. The image is first masked with a circular 2-dimensional Hanning mask.

Soft Circular Masked FFT

Perform a Fourier transform on a 2-dimensional image. The image is first masked with a circular 2-dimensional mask.

Inverse FFT

Perform an Inverse Fourier transform on a 2-dimensional image. Unlike many implementations, the image is not required fo have dimensions that are powers of two. However, the image must be square.

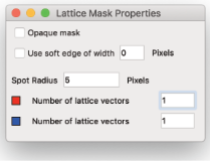
Power Spectrum

Computes the Power spectrum of a real square image.

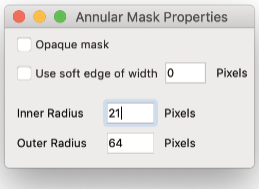
Edit Mask...

Allows the specific settings associated with a mask of a given type. Example of the properties of a lattice mask is shown below.

Lattice Mask Properties..



Annular Mask Properties...

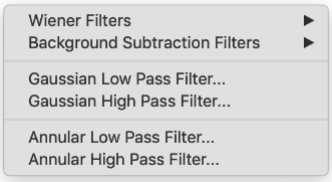


Apply Mask(s)

Applies the defined mask(s) on the image.

Fourier Filters

Contains the following filters



Wiener Filters - different strengths

The Wiener filter is an automatic filter that tries to determine the power spectrum of the "noise" and the creates an optimal filter based on the estimate of noise and signal.

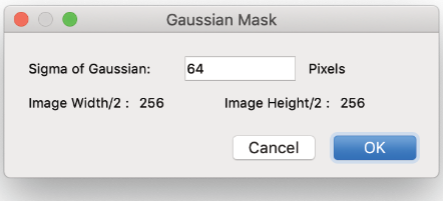
Background Subtraction Filters - different strengths

Like the Wiener Filter, this filter uses a model of the noise to "remove" the unwanted information from the image.

As the information from the crystalline structure is peaked in the Bragg reflections, one can obtain an estimate from the noise spectrum from an amorphous substrate based on the criteria that the noise is only dependent on the frequency and not the orientation. Thus the noise spectrum should be rotationally symmetric, different from the information from the crystaline part.

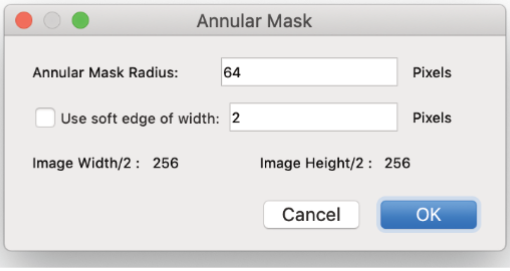
Gaussian Low Pass Filter...
Gaussian High Pass Filter...

Uses a Gaussian masked filter to eliminate part of the higher frequency components of the image (Low Pass) or the low frequency part (High Pass).'

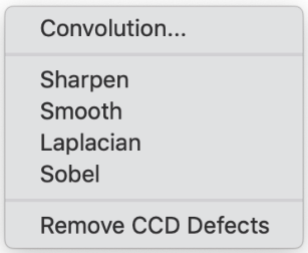


Annular Low Pass Filter
Annular High Pass Filter

Uses an annular mask to eliminate either the high or low fequency part of the spectrum.



.

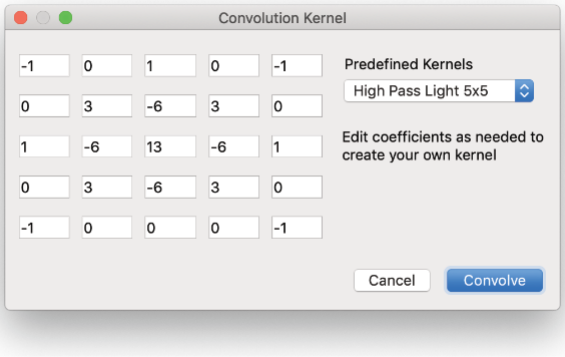


Spatial Filters

Contains the following filters

Convolution...

brings up the following dialog for defining the kernel. Predefined kernels are defined for your convenience, but you can modify the coefficients as you please.



Sharpen

Sharpens the image by applying a 3x3 Kernel with -1 in all elements except for the central element which as a value of 9.

Smooth

Replaces every pixel in the image with the average over its neighbors. The weight of the central pixel is 2, while the neighboring pixels have a weight of 1.

Laplacian

The Laplacian uses a 3x3 Kernel as in the Sharpen operation, except for

the weight of the central pixel is 8 instead of 9.

Sobel

Uses 2 separate Kernels, one for each dimension x and y.

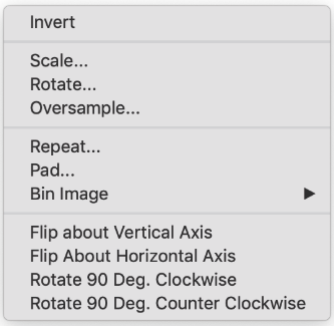
The Kernels are

$$X: \begin{matrix} -1 & 0 & 1 \\ -2 & 0 & 2 \\ -1 & 0 & 1 \end{matrix} \qquad Y: \begin{matrix} -1 & -2 & -1 \\ 0 & 0 & 0 \\ 1 & 2 & 1 \end{matrix}$$

Remove CCD Defects

This operation checks the image to see if there are pixels that are outside of the expected range of the pixel values. It treats pixels that it considers outside the range to be pixel defects in the CCD camera and "repairs" this pixels by an average value of its neighboring pixels that fall within expected values.

Transform



Contains the following operations

Invert

Inverts the image, by inverting around 0.

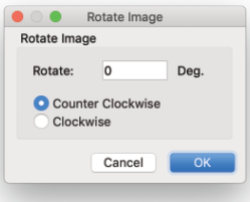
$I_{new}(x,y) = -I_{old}(x,y)$

Scale...

Scale the current image. The new image may or not maintain the previous relative sampling in x and y.

Rotate...

Rotate the current image



Oversample

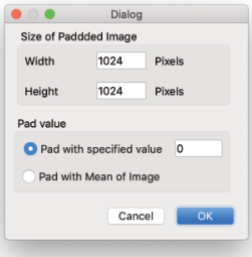
Oversampling the image by a given factor. When oversampling the image, the frequency spectrum is extended with zeros and subsequently Fourier transformed back to real space.

Repeat...

Allow the current image to be repeated n by m times. This operation makes more sense when the underlying image is periodic in x and y.

Pad...

Pads the current image out to new dimensions.



Bin Image
Bins the image by a given factor.

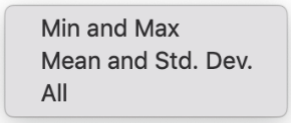
Flip about Vertical Axis
As the menuitem says..

Flip About Horizontal Axis
As the menuitem says..

Rotate 90 Deg. Clockwise
As the menuitem says..

Rotate 90 Deg. Counter Clockwise
As the menuitem says..

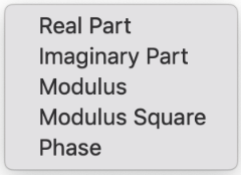
Statistics
Contains the following operations



The data is shown in the Info window.

Calibrate...

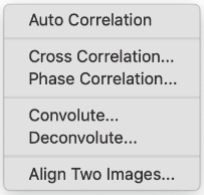
When a length have been marked by the Ruler tool, the image can be calibrated. If there is no line drawn, the menu command is inactive. The image can also be calibrated through the "Edit Object..." menu command. In that case no line needs to be drawn.



Extract From Complex

If the front image is complex, images can be created from various component of the complex values.

Correlation / Convolution



Contains the following operations

Auto Correlation

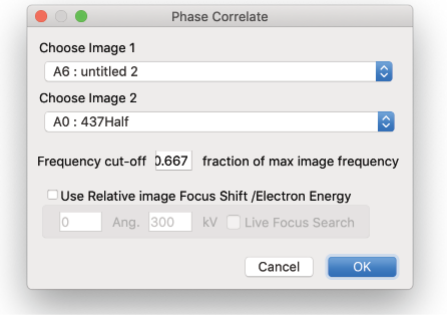
operates on a single image and calculates the auto-correlation function of the image.

Cross Correlation...

calculates the cross correlation image for two images.

Phase Correlation...

calculates the correlation function based on the phases of the two images disregarding the amplitude of the Fourier components.



Convolute...

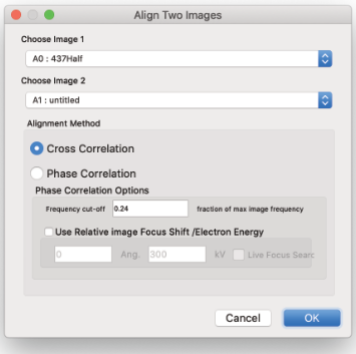
will calculate the convolution image from two images.

Deconvolute...

will produce a third image which is the first image deconvoluted with the second image.

Align...

will try to align two images by either the cross correlation image or the phase correlation image.

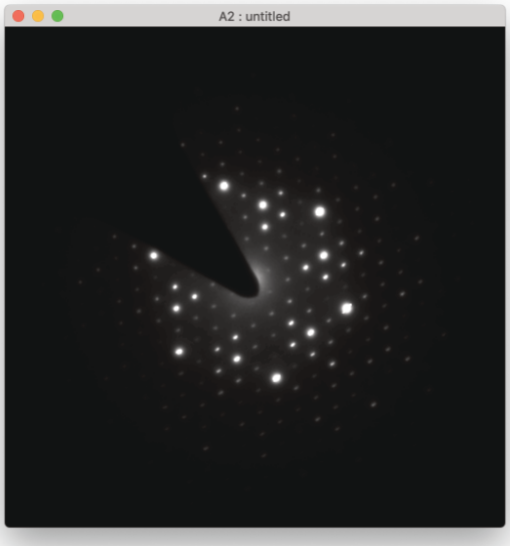


1D Image
2D Image Split Plane
2D Image

Azimuthal Average

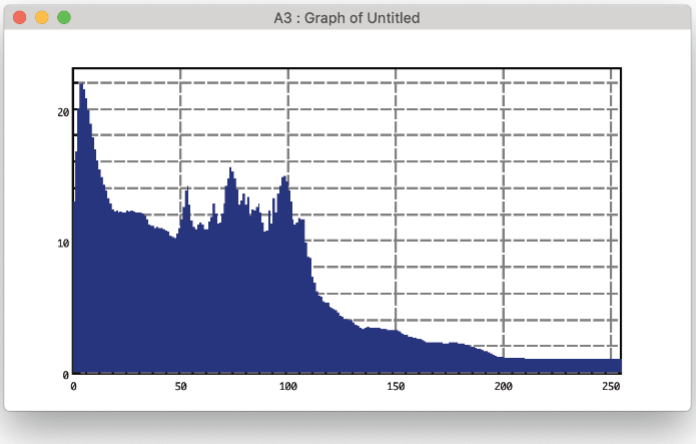
Contains the operations shown to the left.

Using the following image as a starting point, the results of the various operations are shown below.



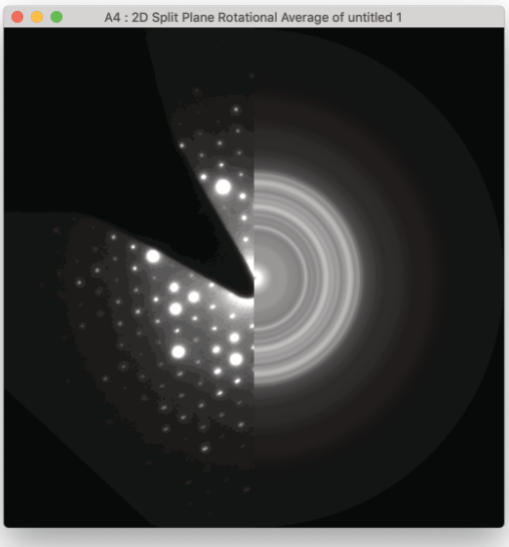
1D Image

calculates the azimuthal average and displays it as a one-dimension trace.



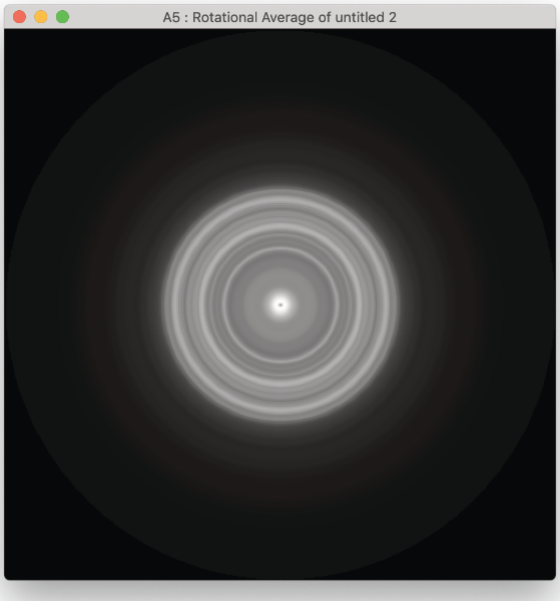
2D Image Split Plane

will display the original left half of the image and replace the right half with the 2D azimuthal average.



2D Image

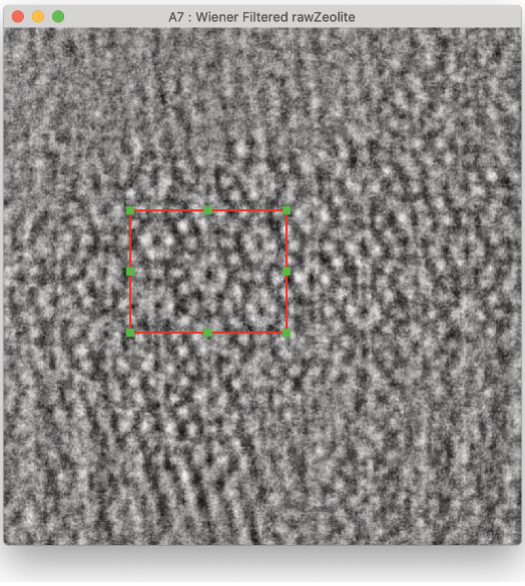
will create a 2D image from the azimuthal average.



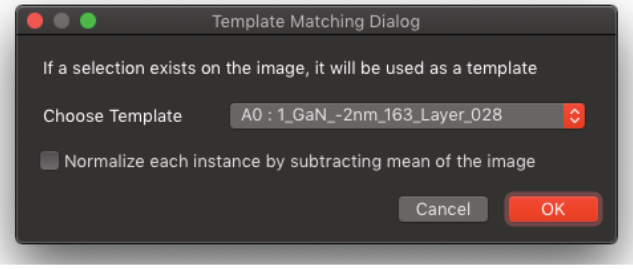
Template Matching...

will calculate an image containing the cross correlation coefficient for the template as a function of its position in the image. The cross correlation coefficient is calculated for each pixel position in the original image.

As an example the Template Matching operation will be used on an image containing the crystalline structure of Zeolite which quickly becomes is destroyed by the electron beam. The area that will be used as a template is shown be the selection. The operation will attempt to localize the regions within the image that has a resemblance to the template.

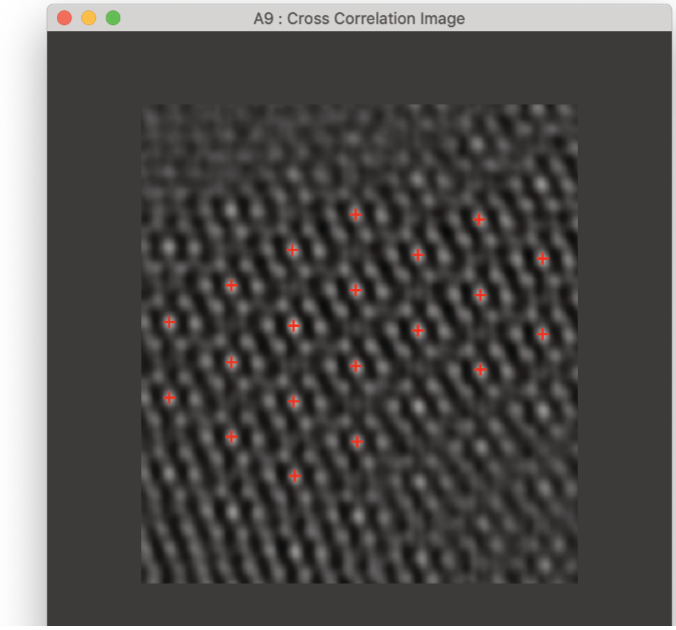


Selecting Template Matching from the menubar brings up the following dialog.

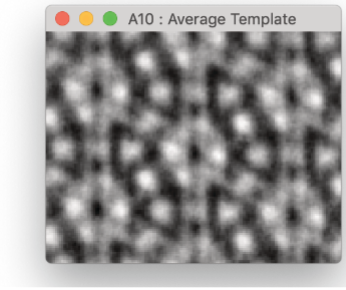


The result is a new image which will give the cross correlation coefficient between the template and the underlying region in the original image for each position of the template.

The peaks in the cross correlation image corresponds to the positions where the cross correlation coefficient between the template and the underlying image are above a given threshold.

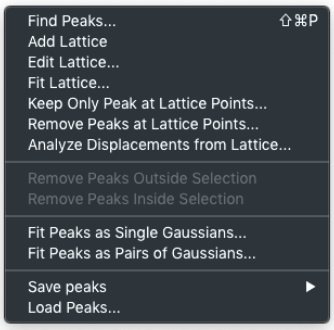


By using the peaks found above, one can calculate the average template image obtained by summing up the images at each peak position. The output produces the following average template.



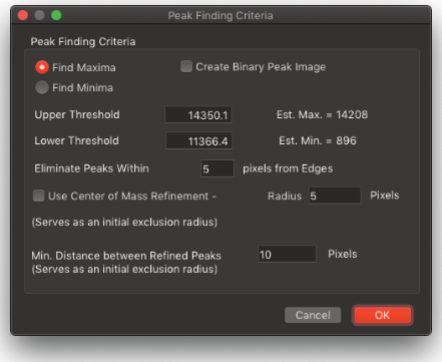
Peak/Lattice Analysis

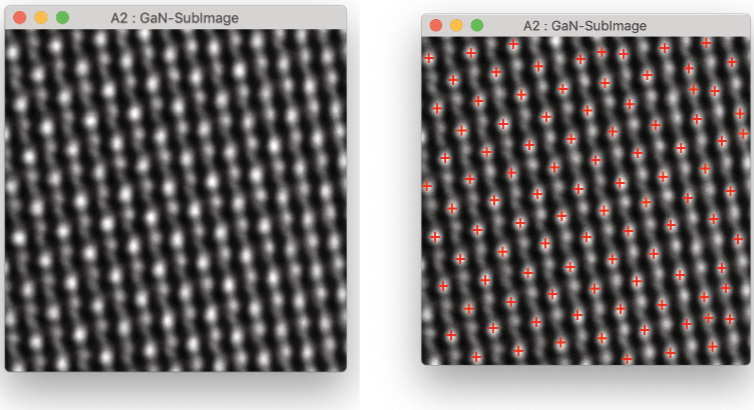
Contains the following operations



Find Peaks...

Finds all the peaks in the image which satisfy the criteria specified in the following dialog.





The peaks will be marked on the image. Changing the criteria and re-finding peaks will mark a new set of peaks, replacing the old. The peaks can be selected as a whole and deleted as for an object. The peaks can also be copied and pasted into another image (preferably of the same size). A list of peaks can be used in a displacement analysis in conjunction with an associated lattice that can be refined through a set of peak positions.

If there is a selection on the image, only peaks lying within the selection will be found. It is also possible to first find peaks within the whole image, create a selection, and then choose to keep peaks either inside or outside the selection.

Add Lattice

will place a lattice onto the image. The lattice will contain handles that will allow the lattice origin to be moved as well as the axes of the lattice.

Changing the number of lattice spacings defined by the lattice vector..

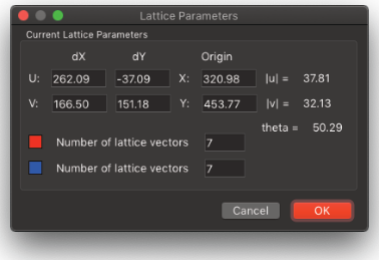
By clicking at the endpoint of the lattice vector with the Option Key

down, the number of lattice spacings will increase by 1 for every click. If the combination Option/Shift key is used, then the number of spacings will decrease by 1.

Edit Lattice..

will allow the editing of the parameters for the lattice.

This will allow all the parameters of the lattice to be set numerically.



Fit Lattice..

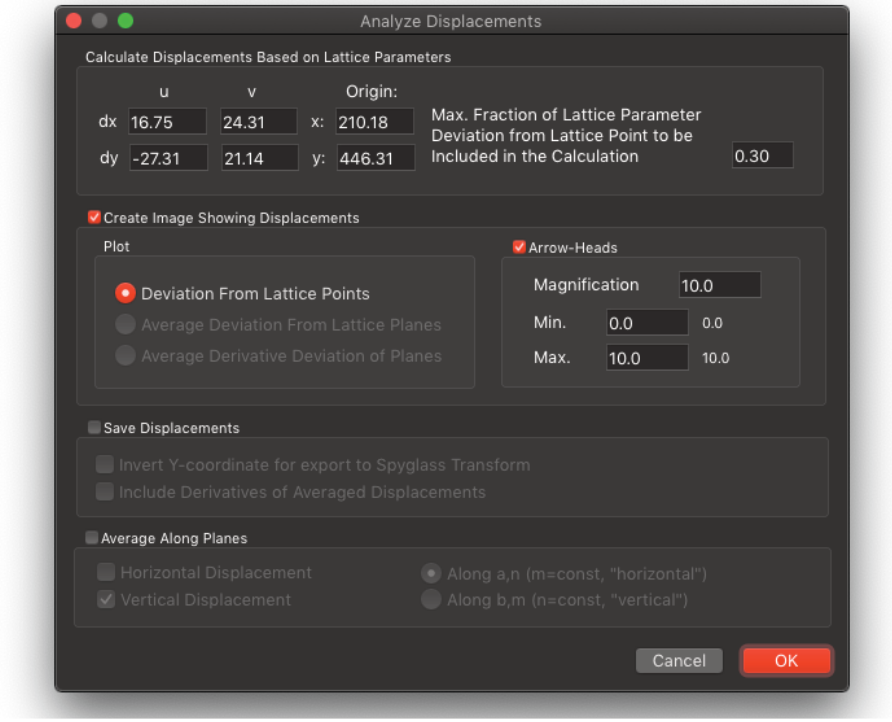
will provide the refinement of a lattice based on a least square refinement procedure using the list of peaks associated with the image in which the lattice is defined.

The lattice needs to be created with the Lattice Tool or the menu command “Add Lattice”. The origin of the lattice should be placed on/near a peak in the “center” of the peaks. The lattice is movable and the lattice vectors are adjustable. Don’t move a lattice after it has been refined. A lattice is an object and thus can be copied and pasted into another image.

If a selection rectangle exists in the image, only peaks within the selection rectangle are used to refine the lattice.

Analyze Displacements...

Allows for the analysis of displacements. The displacements are relative to a reference lattice which normally come from a refined lattice.



Geometric Phase Analysis...

allows for displacement analysis from variations around spatial frequencies..

The geometric phase calculation is described in more detail in Chapter 11.

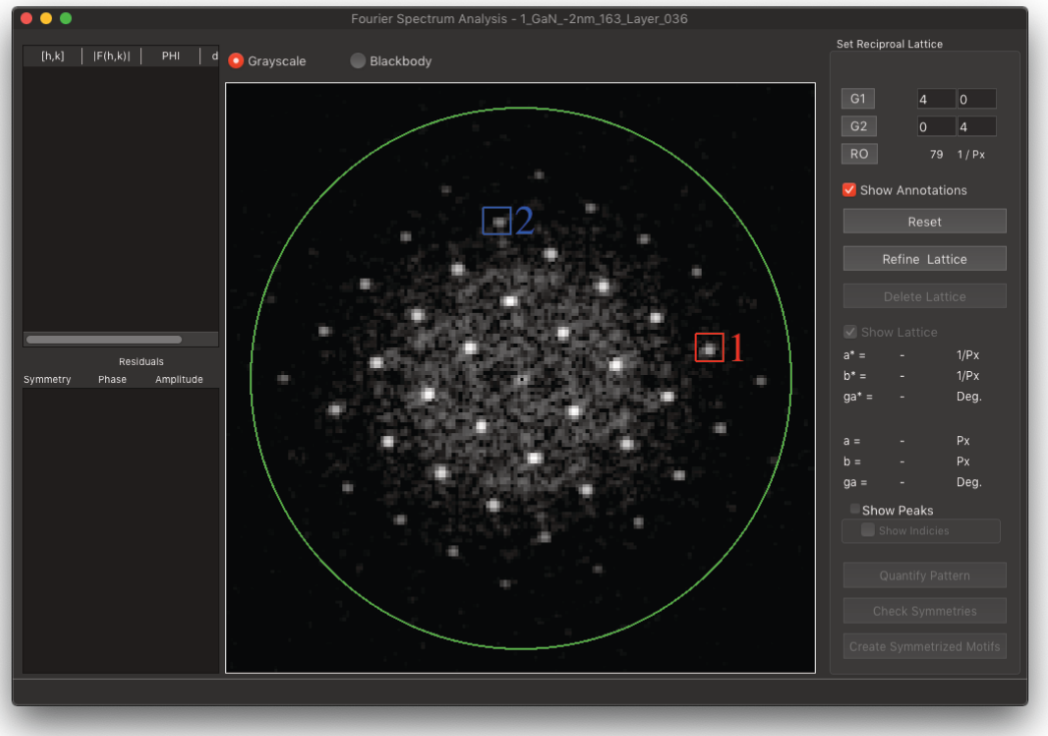
Cryst Image Processing...(Crystallographic Image Processing...)

Can be invoked on a real space image which is square or has a square selection.

Crystallographic Image Processing is based on the fact that the Fourier Transform of the image of a crystalline material exhibits a certain relationship between the Fourier coefficients in order for the symmetry of the 2D projection to belong to one of the possible plane groups which is required if the "unit cell motif" can be used to completely fill space.

Thus by taking the Fourier Transform of the Image of the crystalline material, one can calculate how close the Fourier Coefficients are to satisfy the symmetry of the various plane groups. For instance a 4-fold symmetry, plane group P4, should have 4 Fourier coefficients that are linked such that the amplitude of the 4 coefficients are the same and the Phases are all the same, being either 0 or 180°. One can check for each possible plane group what the phase differences between the Fourier coefficients of the experimental image are to those of the plane group symmetry. If one deduces that the motif of the crystal has the symmetry of a given plane group , one can then apply the symmetry of the amplitudes and phases onto the experimental coefficients and do an inverse Fourier Transform to impose the plane group symmetry onto the symmetrized motif of the crystal. This will of course give a perfectly symmetrized motif and no other information that may have been present in the image will any longer be visible.

Selecting the menu command when the active window contains a square image of a region containing crystalline material brings up the following window.



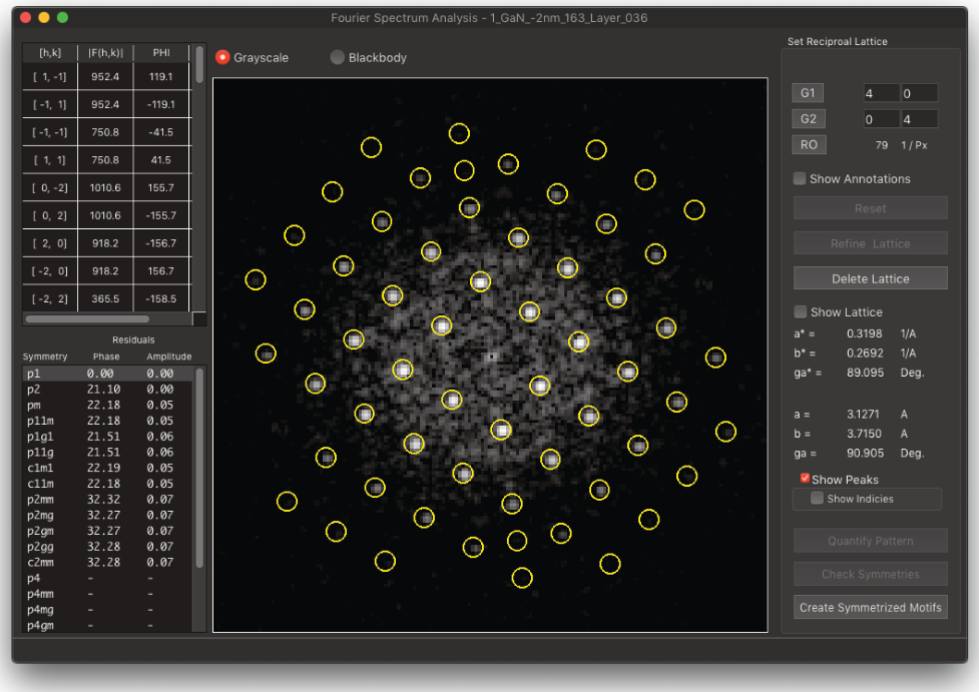
In order to get started, the Hanning Masked Fourier transform will to be calculated. Once the Fourier transform has been calculated, the reciprocal space lattice needs to be set using the a* and b* tools and clicking on two consecutive reflections that defines the reciprocal space. The number of reflections used in the lattice refinement and information extraction can be limited by the circular aperture tool.

Once the lattice reflections have been marked, the next action is to invoke the command “Refine Lattice”. This will mark the lattice which then should go through all the reflections that define the crystal. If it doesn’t you can start over.

After the lattice has been refined, click on “Quantify Pattern”.

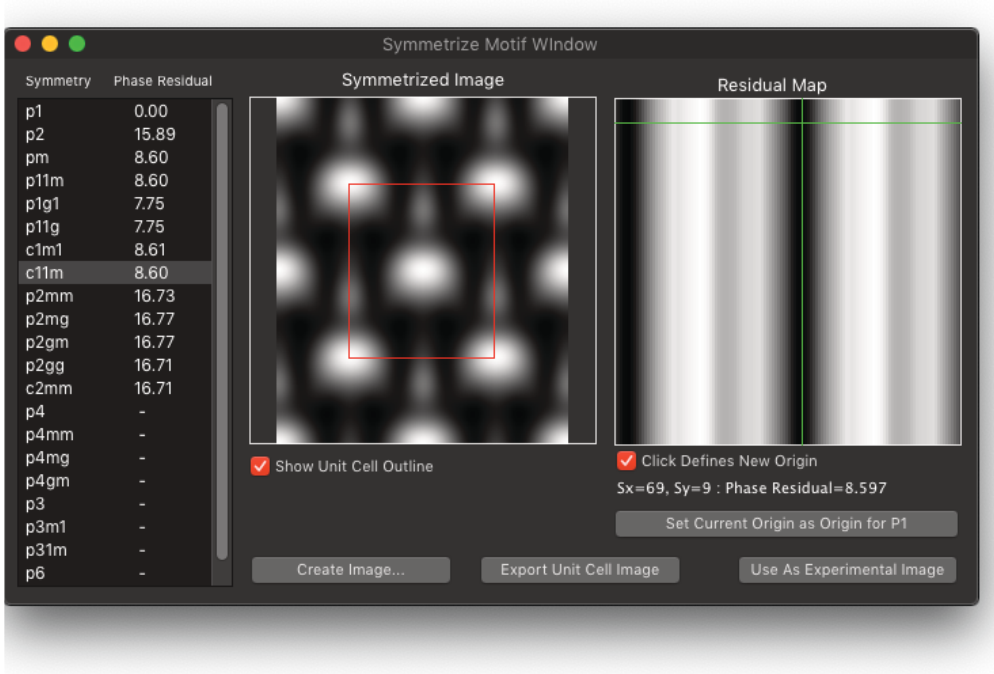
This will extract all the phases and amplitudes from the Fourier transform at the lattice points and show all the extracted amplitudes and phases of the reflections lying on the lattice peaks.

One can test for possible symmetries by invoking “Check Symmetries”.



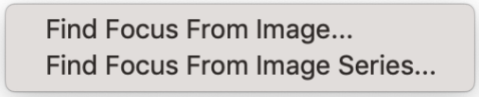
Each symmetry can be tested and imposed on the reflections to form a new unit cell motif. The origin of the unit cell motif is shown on the right and can be changed.

Once a solution has been chosen, a new image can be created with the imposed symmetry and can also be used as an image to compare with simulation.



Focus Determination

Contains the following operations



Finding focus from a HRTEM image is outlined in more detail in chapter 12

Image Info...

shows the general information regarding an image, as shown below.

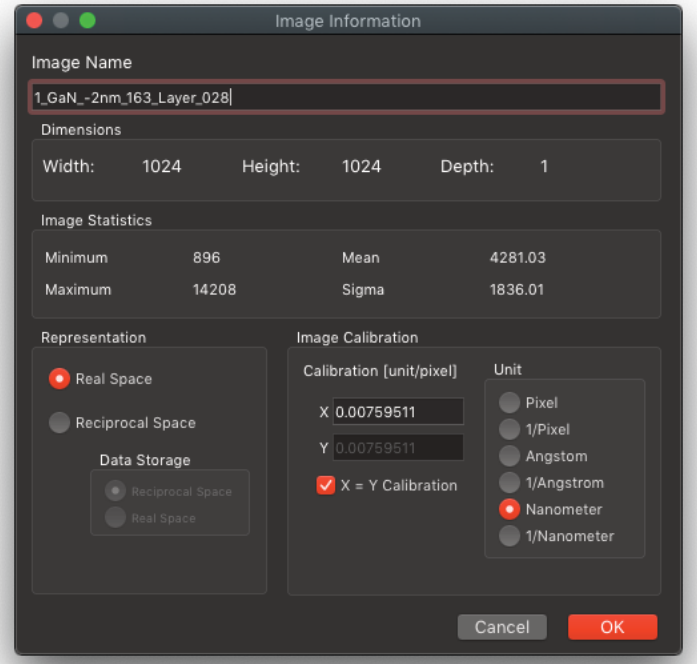
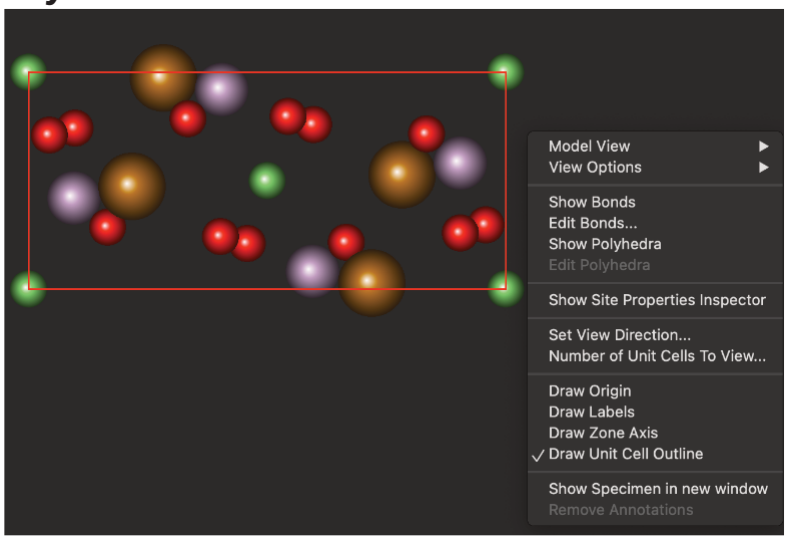


Image Information
Image Name
1_GaN_-2nm_163_Layer_028
Dimensions
Width: 1024
Height: 1024
Depth: 1
Image Statistics
Minimum 896
Maximum 14208
Mean 4281.03
Sigma 1836.01
Representation
Real Space
Reciprocal Space
Data Storage
Reciprocal Space
Real Space
Image Calibration
Calibration [unit/pixel]
X 0.00759511
Y 0.00759511
X = Y Calibration
Unit
Pixel
1/Pixel
Angstom
1/Angstrom
Nanometer
1/Nanometer
Cancel
OK

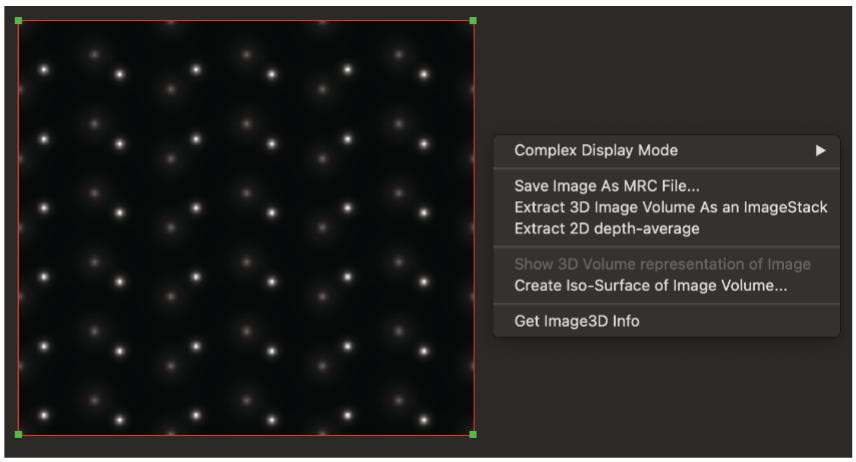
Contextual Menus

Many objects have contextual menus associated with them. A number of them are shown below in conjunction with the object they are associated with.

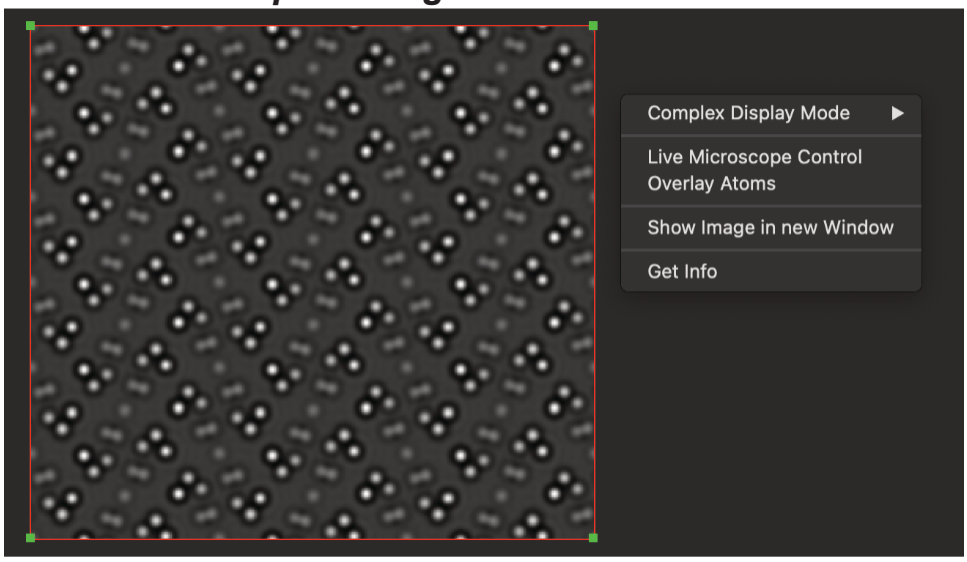
Crystal View Contextual Menu



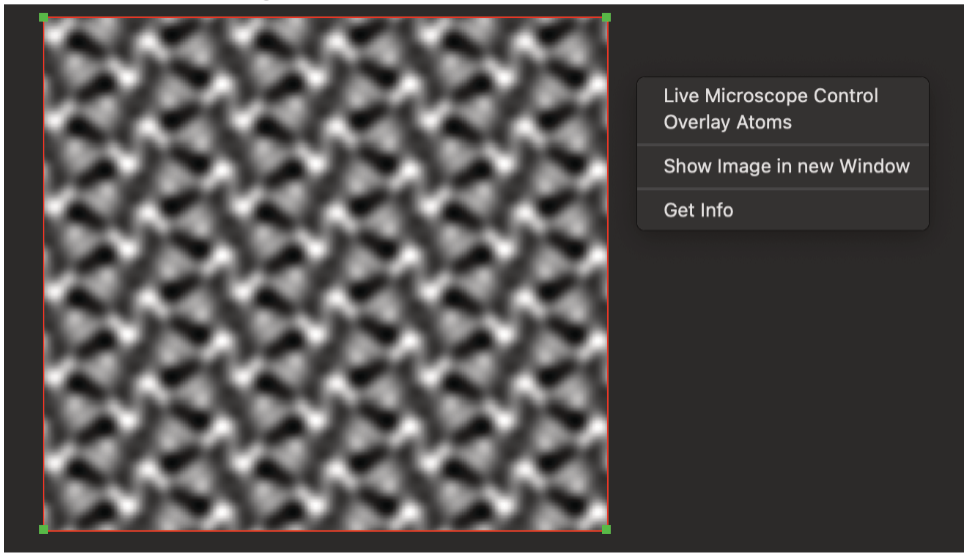
Projected Potentials as a 3D Image Stack Contextual Menu



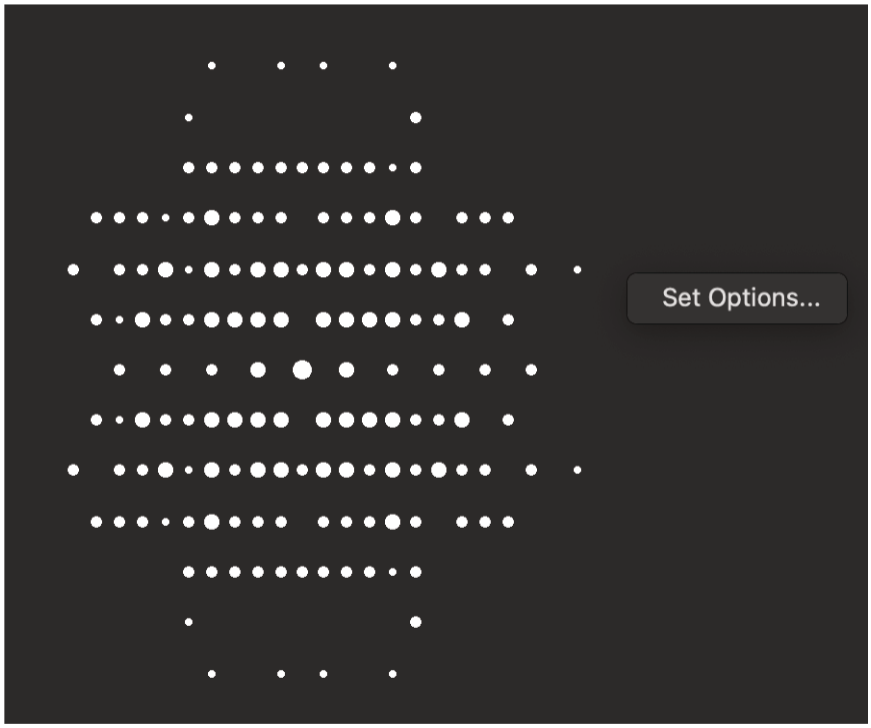
Exit Wave Complex Image Contextual Menu



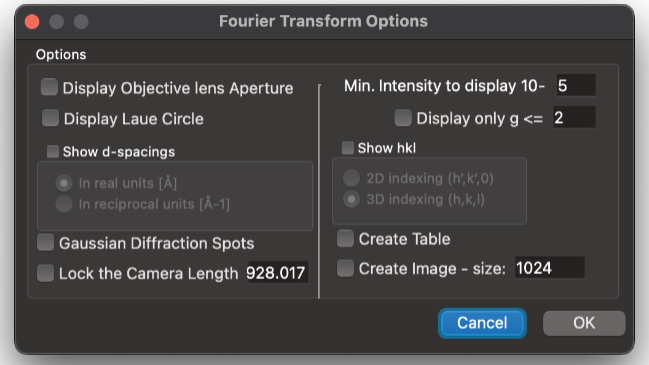
Simulated Image Contextual Menu



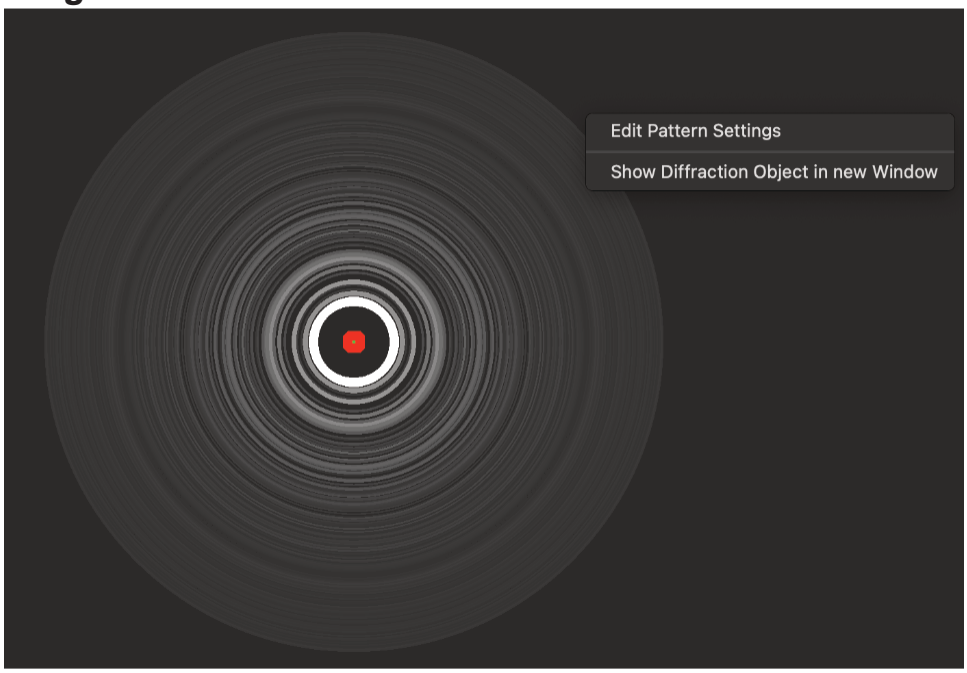
Simulated Diffraction Pattern Contextual Menu



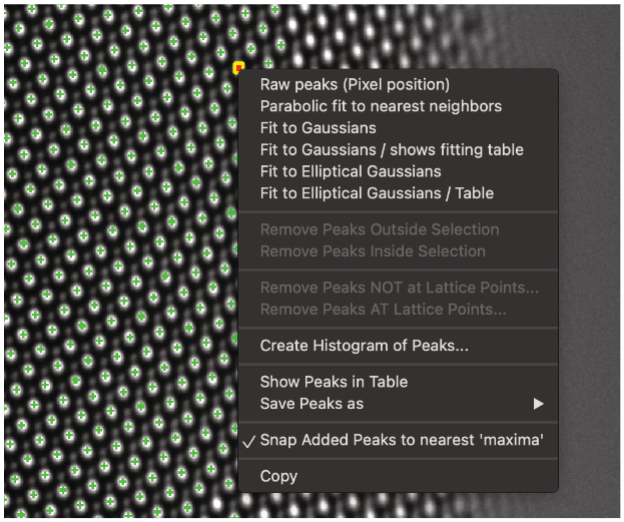
Simulated Diffraction Pattern Options



Ring Pattern Contextual Menu



Peaks Contextual Menu



File Format

The structure file created by New... in the File Menu is a file of type 'TEXT' and can be produced by a text editor. At times it is desirable to edit the file directly, rather than using Tempas to create this file. In fact the user may sometimes want to write a program to generate the data in the structure file. For that purpose in particular, the format of the structure file <structurename>.at is given below:

Line #	Parameter(s)	Meaning
1	Title	Arbitrary description of this structure
2	SpaceGroupNumber	Just as is says, one of the 230 spacegroups, (1-230).
3	a b c a b g	The lattice parameters and angles
4	Gmax	The maximum reciprocal lattice vector in the multi-slice calculation. The potential is evaluated out to twice this value, units Å-1.
5	iu iv iw	The direction of the electron beam in units of the real space crystal lattice vectors.
6	NSymops Nslices (I3d)	Number of symmetry operators, number of slices per unit cell, and a flag indicating 2d (0) or 3d (1) potential calculation only if Nslices is different from 1.

Line # Parameter(s) Meaning

7 NBasis NtypesThe number of atoms in the basis, the number of different types of atoms. A different type is associated with a different chemical symbol or a different Debye-Waller factor.

8 it symb x y z The type of atoms (a number

dw occf from 1 - NTypes), Chemical symbol, x-,y-,z coordinates in relative units of the lattice vectors, Debye-Waller factor and occupancy factor.

9 The same as line 8 for atom number 2.

10 The same as line 8 for atom number 3.

8+NBasis

MicName Cs Del Th The name of the microscope, the spherical aberration (mm), the spread of defocus (Å) and semi-angle of divergence (mrad).

9+NBasis

Voltage Accelerating voltage (kVolt).

10+NBasis

Lh Lk The center of the Laue circle in units of the h and k of the transformed reciprocal unit cell. (Real numbers).

11+NBasis

Thickness The specimen thickness or

T1,T2,DT First thickness, last thickness, increm. The commas are required.

12+NBasis

IPlot Amplitudes to be stored as for possible plotting, (YES/NO).

13+NBasis

ih ik il The indices of the reflection to be stored, or if IPlot == NO then :

Defocus Objective lens defocus or

D1,D2,DD First defocus, last defocus, increment. The commas are required.

14+NBasis

+NAmp

ApertureRad. Radius of the objective lens Aperture in units of Å-1.

15+NBasis

+NAmp

Ah Ak The center of the objective lens aperture in units of h,k of the transformed reciprocal unit cell.

16+NBasis

+Namp

Oh OkThe center of the optic axes in the same units as Ah,Ak.

17+NBasis

+Namp

s1,s2,s3 Symmetry operator number 1. An example is x+1/3,y+5/6,z+1/3. The commas are required.

.

17+NBasis

+Namp

+NSymop

istat The calculation status of this structure. For a new structure this should be 1

18+NBasis

+Namp

+NSymop

Vibration Halfwidth of mechanical Vibration in A.

Note: If different wordprocessing software is used, Microsoft Word, Write Now etc., make sure that the text file is saved at the end as type TEXT.

Sample Calculation

As an example of a calculation using Tempas we consider a BCSCO super-conductor structure. Using the structure determined by Tarascon et al (1988), we show the steps necessary to input the model structure, examine it, compute the diffraction pattern and simulated images, and display and print them.

As published by Tarascon et al in Phys. Rev. B 37 (1988) p.9382-9389, the tetragonal structure has the following parameters -

Space group: I4/mmm

Cell parameters: a=b=3.814Å, c=30.52Å, a=b=g=90

with nine atom positions in the basis:

Atom	Wyckoff notation			x	y	z	Occupancy
Ca	2a	0	0	0	1		
Sr	4e	0	0	0.1097	1		
Bi	4e	0	0	0.3022	0.87		
Bi	4e	0	0	0.2681	0.13		
Cu	4e	0	0	0.4456	1		
O(1)	8g	0.5	0	0.446	1		

O(2)	4e	0	0	0.375	1
O(3)	4e	0	0	0.205	1
O(4)	4d	0.5	0	0.25	0.065

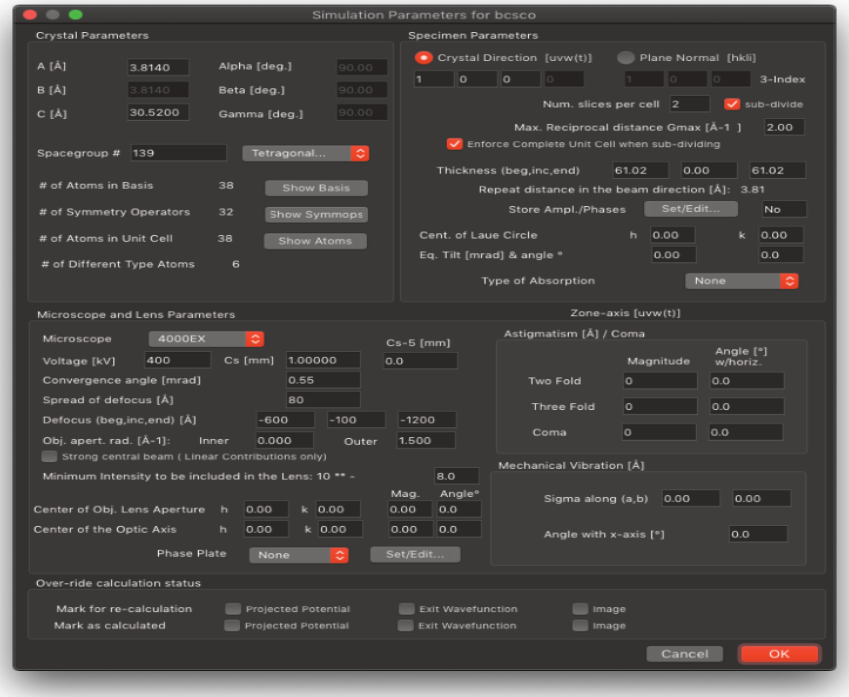
Isotropic thermal parameters for all atoms are fixed at 3.6 Å2.

To enter a new structure into Tempas, we first go to the FILE menu (Section 3.3), and select New Structure File... After entering a filename in the New File dialog, Tempas will create a default cubic structure.

Specify a filename under which to file the input data. It should be descriptive enough to be easily remembered when you need to open it later. Make sure you use no extension.

We need to change the data to reflect that of our structure. The easiest way to quickly set up the simulation is to go to the menu command "S***how All Simulation Parameters...***" under the P***arameters*** menu in the menubar.

This will bring up the following dialog where you can set up all the parameters in the calculation.



Spacegroup # 139

From the structure information, we know that the cell is tetragonal with a space group I4/mmm. From Table 6.2.1 of the International Tables for Crystallography, we find that the space group number for I4/mmm is 139. Choose the correct space group from the popup menu.

a 3.814

Enter the correct value for the lattice parameter a. In this example Tempas knows that b is equal to a for the tetragonal space group #139, and so enters b automatically once a has been set. Similarly,

Tempas puts in the correct unit cell angles, since they are defined by the space group (in this particular example). Note that cell parameters are input in Å, not in nm.

c 30.52

The value of the C cell parameter is input in Å.

Gmax(default=2)

Gmax is the size of the "multi-slice aperture" and defines how far out in reciprocal space the diffraction calculation will extend. The value of Gmax is automatically set to 2.0 reciprocal Ångstrøm units, so that Tempas will compute all of the dynamically-diffracted scattered beams out to this value, by considering all their interactions with phase-grating coefficients out to twice Gmax (a default of 4.0 reciprocal Ångstrøm units). Note that these default values (2 for the multi-slice and 4 for the phase-grating) are normally large enough to ensure that all significant contributions to the dynamic scattering are included; however Gmax is displayed in the Tempas menu so that it can be set to a larger value if greater precision is required with a structure that includes heavy atoms.

Zone Axis 0,1,0

The correct response is the set of three integers that defines the direction of the electron beam with respect to the specimen (or the specimen orientation with respect to the incident electron beam direction). In this example we choose to enter 0,1,0 in order to image the specimen down the b-axis.

Number of slices per unit cell (default=1)

This value will be computed by Tempas from the repeat distance of the structure in the beam direction and the current value of Gmax. This

number can be changed if desired (as, of course, can all the parameters entered in response to the prompts listed in this chapter).

Show Basis

Click on the command to bring up the dialog box for entering the information regarding the number of atoms in the basis. We enter the nine different atom positions listed for the basis atoms.

For each of the atoms in the basis, Tempas requires the chemical symbol, x,y,z coordinates, DW factor and occupancy factor. From the information given above, we use the following information for the nine atoms that are given in the structural basis.

Chemical Symbol Ca

x,y,z 0,0,0

Debye-Waller Factor 3.6

Occupancy 1

The data for the first atom include the chemical symbol for calcium (used by Tempas to select the correct scattering factor table), the atom coordinates, the temperature factor (or Debye-Waller factor), and the occupancy factor.

The second atom position is entered in the same way with responses of -

Chemical Symbol Sr

x,y,z 0,0,0.1097

Debye-Waller Factor 3.6

Occupancy 1

The third atom position is similar, except that the occupancy is set at 0.87 -

Chemical Symbol: Bi
x,y,z: 0,0,0.3022
B Factor: 3.6
Occupancy: 0.87

After all nine atom positions have been entered, Tempas will need the parameters of the electron microscope for which to compute the simulation.

Microscope 4000EX

If the input microscope name is listed in Tempas stored microscope data, various operating parameters will be set automatically. If the entered name is unknown to Tempas, values will need to be given for each of the operating parameters. In this example, we use 4000EX, and find that Tempas sets the spherical aberration coefficient to 1.0mm, the Gaussian halfwidth of depth of focus to 80Å, and the semi-angle of beam convergence to 0.5milliradian.

Specimen Thickness 40 20 80

The foil thickness response may be in one of two forms, either a single value in Ångstrøm units, or a construction combining a starting and ending thickness with an incremental value. The construct that we have entered requests Tempas to store diffraction results for thicknesses starting at 40Å and continuing through 80Å in steps of 20Å. That is, at specimen thicknesses of 40Å, 60Å and 80Å.

Store Ampl./Phases No

As well as storing all the beam amplitudes at specified specimen thicknesses, Tempas can store a selected few beam amplitudes at each

single-slice increment in thickness, then plot amplitude (or intensity) and phase as a function of thickness for any of the stored beams. To store beams for plotting, click on the command to enter the indices for the reflections that will be stored. In this starting example we will not be entering any information here.

Voltage (400)

The voltage would need to be entered if an unknown microscope type were selected. Since we have selected a 4000ex, Tempas will choose a value of 400keV.

Center of the Laue Circle 0,0

The pair of values specified as the Laue circle center are used by Tempas to define the direction and amount by which the specimen is tilted from the exact zone-axis orientation specified above, and, in fact, specify the center of the Laue circle in units of the h and k coordinates in the diffraction plane. Note that the values supplied need not be integers, but should not define a tilt of more than a few degrees. The default values of 0,0 specify exact zone-axis orientation.

Objective Lens Defocus -200 -200 -800

So far, we have supplied all the information Tempas requires to carry out the dynamical diffraction part of the simulation; now we input the imaging conditions. The first imaging-condition prompt is for the objective lens defocus. We choose to enter four values of defocus by specifying defocus values from -200Å to -800Å in steps of -200Å. Note that a negative value denotes an objective lens weakened from the Gaussian condition; that is, underfocus is negative.

Aperture Radius 0.5

The value for the radius of the objective aperture should correspond to the radius in reciprocal Ångstrøm units, as measured from a diffraction pattern exposed with the aperture superimposed. We will enter 0.5 to represent a value of a resolution of 2.0Å.

Center of the Objective Aperture 0,0

In order to simulate dark-field images, Tempas provides for an objective aperture displaced from the center of the diffraction pattern. As for the Laue circle center, the aperture center is defined in units of h and k. We leave the default values of 0,0.

Center of the Optic Axis0,0

To provide for microscope misalignment, or for conditions of tilted-beam imaging, the coordinates of the diffraction pattern at which the optic axis lies can be specified in the same manner as the center of the aperture. Again, we use default values of 0,0.

When all the data in the top field are satisfactory, we go to "S***how Atoms***" in the parameters menu to check that all atom parameters have been entered correctly. At this stage it is also worthwhile getting Tempas to display a model of the structure by.going to the Commands menu and selecting the menu command "D***raw Atomic Model***".

When we are satisfied that all data are correct, we run the simulation by.clicking on "F***ull Calculation***" in the C***alculate*** menu. Note that Tempas displays the current status of the calculation in the Status Window. First, Tempas computes the phase-grating for the structure (the status window shows the number of coefficients generated so far), then the dynamical

diffraction for each slice of the specimen (current slice number is shown in the Status Window), then four images are computed at each of the three specimen thicknesses that we specified (the image number is shown in the window).

Once Tempas has finished the computation, the results (diffraction patterns, images and diffractograms) can be displayed. (Also beam amplitude and phase plots if any of these has been stored).

The potential, exit wave or simulated image will automatically be displayed at the end of the calculation. Images of a thru-focus series will be displayed as an Image Stack. If there are several thicknesses as well, there will be a separate Image Stack for each thickness.

To display the images, we.go to the Display Window and select "Image", then "Display". Tempas will ask which of the 12 images is to be displayed, then display the requested image in the center of the screen. The image can be moved around with the pointer tool

To get all 12 images onto the display screen simultaneously, select the options menu and the "Montage" option. Back in the source window select "Display".

Now go back to the montage option and deselect "Montage".

To display the projected potential for comparison with images, select "Potential" in the source window, then "Display".

To display the diffraction patterns at the stored specimen

thicknesses, select “Diffr .Patt” in the source window, then “Display”. To change the size of the patterns, choose “Diffr. Patt”. from the Options Menu and choose a different camera length. The size of the diffraction spots also depend on the divergence angle set in the main parameters.

One can also interactively use the mouse to drag a diffraction spot to change the camera length and also drag the perimeter of the zero-g spot to change the convergence angle.

To display the power spectrum of one of the images, we choose “Image” from the source window. Respond by answering which image and then choose “Fourier Transform” from the operand window. Finally click on “Display” to view the power spectrum. The options for the power spectrum are the same as those for display of diffraction patterns. The circle drawn in diffraction patterns and power spectra corresponds to the objective aperture and can be turned off from the diffraction option.

The Weak Phase Object Approximation

The Weak Phase Object (WPO) approximation is a useful tool to find out what kind of information about a specific structure may be revealed at different levels of resolution, given a specific spherical aberration Cs.
The WPO approximation has already been described earlier, and some of that information is repeated here. There are two important assumptions that are made in the WPO approximation.

The wave function of the electron can be written as

$$\Psi(x,y) = 1 + i\sigma t V(x,y)$$

Where Ψ(x,y) is the electron wave function at a point (x,y) and V(x,y) is the projected electrostatic potential at the same point. Sigma is the interaction parameter between the electron and the potential of the atoms and t is the specimen thickness. This first approximation is good for very thin specimens containing light atoms.

When taking into consideration the phase shift caused by the lens X(u**)**, then

$$\Phi(\mathbf{u}) = \delta(\mathbf{u}) + 2\sigma t\,\overline{V}(\mathbf{u})\sin\chi(\mathbf{u})$$

An ideal Scherzer lens is a lens that transfers all diffracted beams with a g-vector that is less or equal to 1/resolution, and blocks all diffracted beams with a larger g-vector. In addition it adds a phase shift of 90 degrees (relative to the central beam) to all beams passing through the lens. This in addition to the 90 degree phase shift introduced by the scattering event itself (the 'i' in the equation for above) causes all scattered beams that pass through the lens to

be 180 degrees out of phase with the central beam.

Under the two assumptions above, the image intensity in Fourier space in the WPO approximation can be written as

$$\Phi(\mathbf{u}) = \delta(\mathbf{u}) - 2\sigma t\overline{V}(\mathbf{u}) \;\; ; \; |\mathbf{u}| \leqq |\mathbf{u}|_{MAX} \; ; \; \overline{V}(\mathbf{u}) = \mathcal{F}(V(x,y))$$

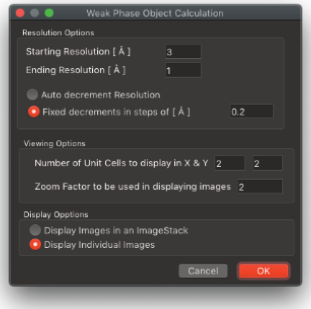
where **u**max is the highest reciprocal spacing to be included in the image.

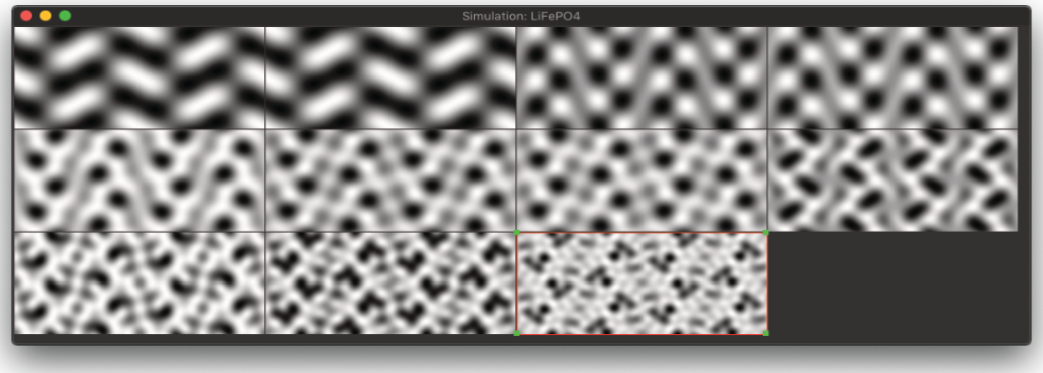
F stands for the Fourier transform. After the spatial frequencies are band-limited at a given frequency, the Image intensity can then be written as

$$I(x,y) = 1 - 2\sigma t V(x,y; |\mathbf{u}| \leqq |\mathbf{u}_{MAX}|)$$

This shows that under the Weak Phase Object Approximation and a Scherzer Lens, the image intensity is low in areas of high electrostatic potential, the location of atoms. Atoms of higher atomic number show up as larger and darker regions in the image. This type of image will often be similar in appearance to images calculated by a full multi-slice calculation for equivalent resolution for a thin specimen for Scherzer defocus.

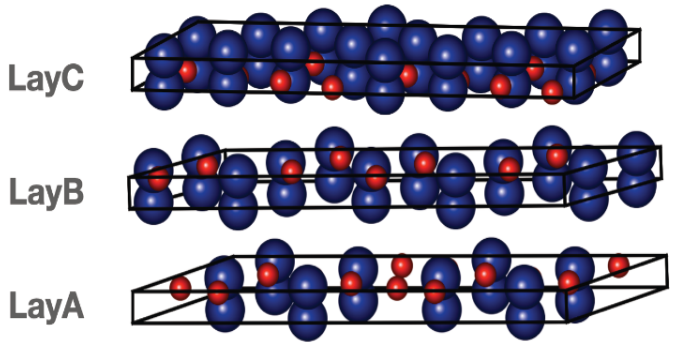
The WPO approximation is invoked from the menu bar in the same fashion as the multi-slice calculation. The input to the WPO calculation is a starting resolution in Å and an ending resolution. The steps in resolution can be fixed (user determined) or automatic. When automatic steps are chosen, the program will calculate the first image corresponding to the reflections that lie within 1/BeginningResolution and each new image will be calculated for the next set of reflections corresponding to a higher resolution until the end resolution is reached.




Simulation: LiFePO4

Creating a Layered Structure

A layered Structure is a special type of "structure" where the composition varies in the direction of the electron beam. An example of this would be a crystalline material having surface layers of amorphous material. Another example would be a crystalline structure where the repeat distance in the electron beam direction is too large for the repeat to be used as the slice-thickness and the unit cell must be sub-divided into several slices with different atomic content. As an example we will work with three layers which we will call LayA,



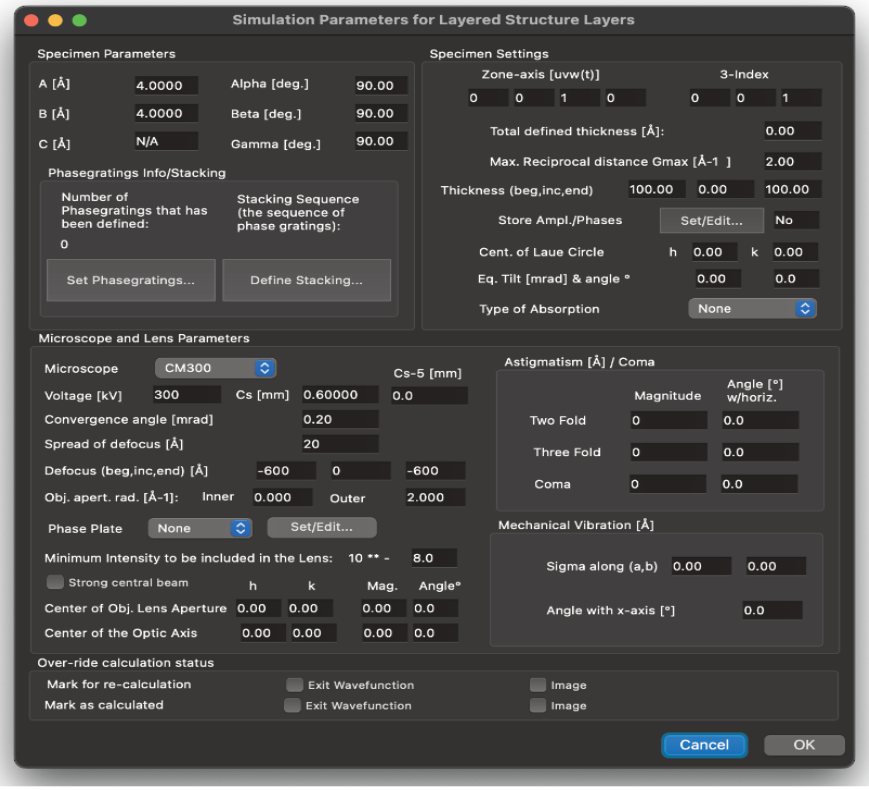
LayB and LayC. Each of these "layers" are what we would call a "single" structure. That means they are defined as a unit cell with lattice parameters and atomic content. The one thing they have in common is that the lattice parameters A and B with respect to the electron beam are the same and that we will use identical sampling in each case, see figure above.

The idea of the layered structure is that the 3 layers can be arranged in any chosen sequence to make up the total structure. The steps in creating and calculating the image for a "layered" structure are as follows.

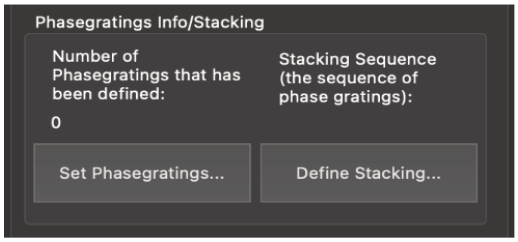
1) Define the 3 layers LayA, LayB and LayC as single structures with the same unit cell dimensions perpendicular to the electron beam (A and B).

2) Calculate the phasegrating for each structure LayA, LayB and LayC using the same value for Gmax.

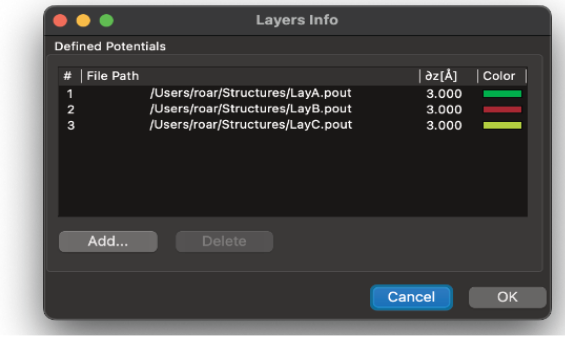
3) Now create a New Structure in Tempas using the option New Layered Structure. You will be asked to fill out information regarding the lattice parameters A and B etc. There are no input for atoms, because a layered structure has no atom information per se. Even though you are asked to fill out a specimen thickness, this value has no meaning at this time, because the content of the structure has not been defined. The values of A and B come from the structures LayA, LayB and LayC. When you create the layered structure, a default value of 2 Å-1 is supplied and you must change it in the main parameters if a different value was used in calculating the phase-gratings for LayA, LayB and LayC.



4) Once the information in 3) has been filled out, the file is created and you must define the "structural" or "phasegrating" content of the layered structure. This is done by using the command "Set Phasegratings…".

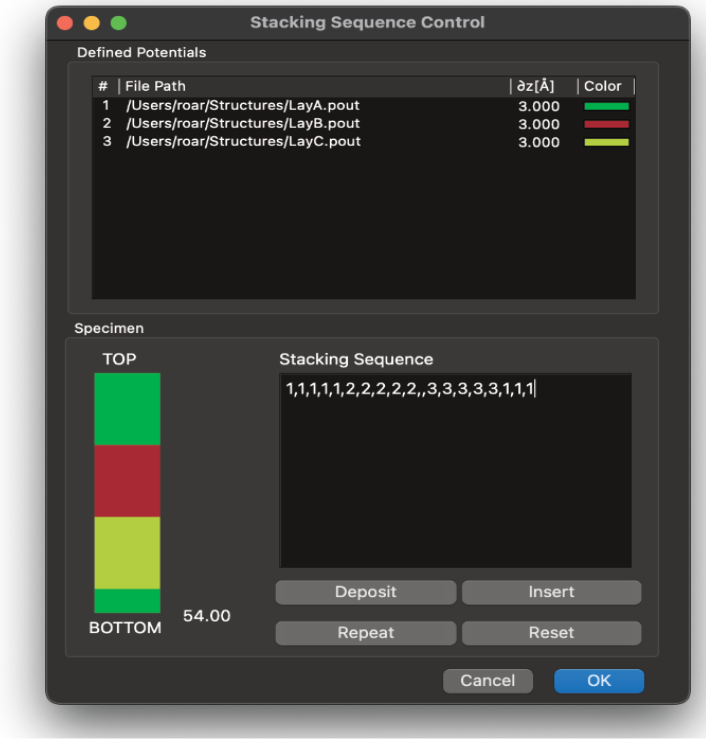


If this is a new file, there will be no phasegratings listed and the command "Add…" must be used to define the layers. You get a list of the available phasegrating files (—.pout). Double Click on LayA.pout and fill in the value for the slice-thickness that was used in the calculation of LayA.pout. Continue and do the same for LayB and LayC.



Now the program has information as to which phase-gratings it can use and the final part is to define the sequence by using the button "Define Stacking…". The sequence can be defined in different ways. One way is to type in the sequence as

1,1,1,1,1,1,2,2,2,2,2,23,3,3,3,3,1,1,1 where 1 stands for LayA, 2 for LayB and 3 for LayC. One can also use the commands to define the sequence. At all times the specimen is drawn as a colored bar at the left. Once this is done, you have defined the structure.



5) Now check the Main Parameters to see that everything is correct and finally run the calculation. The calculation will begin with multi-slice.

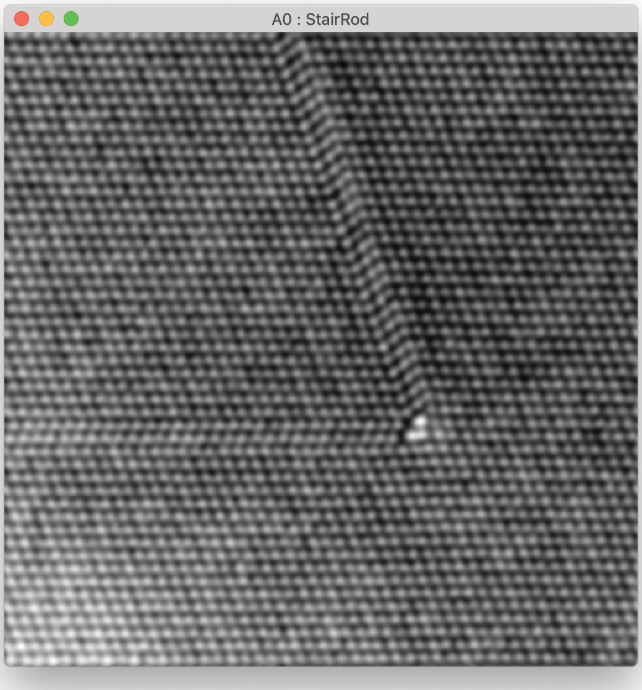
The Geometric Phase Analysis

Background

The Geometric phase routines attempt to determine displacement and strain relative to a reference lattice by analyzing the the variations around specific reciprocal lattice frequencies.

Many of the routines will be illustrated using an example of a HRTEM image of a stair rod calculation.

Invoke the "Geometric Phase Analysis" from the menu with the image to be analyzed as the front window.

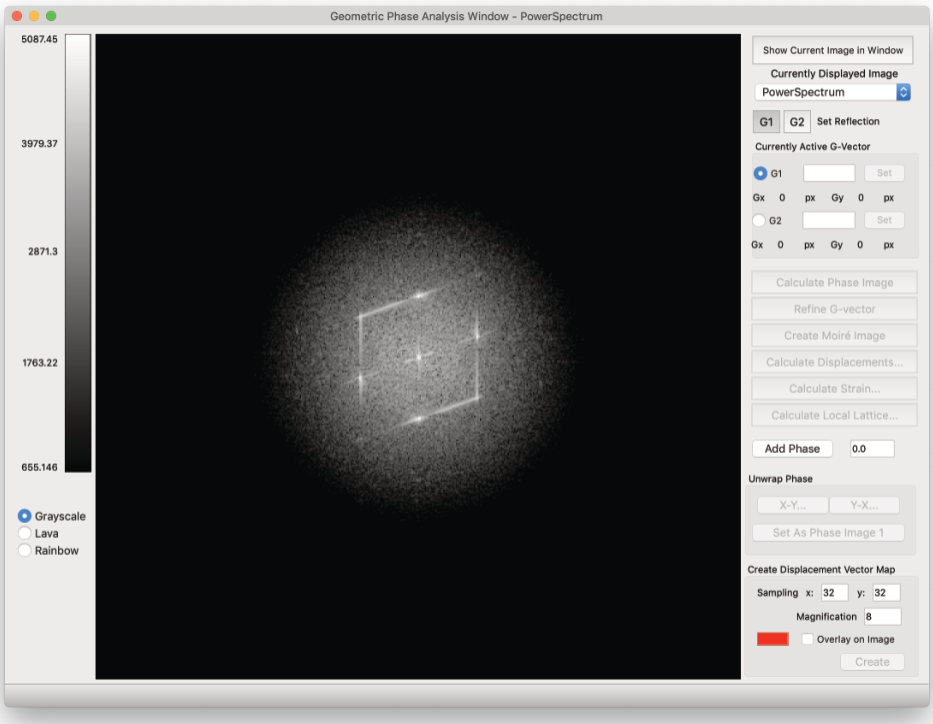


Note:

1) The image must be square.

Alternatively, a selection satisfying the criteria above must exist for the image to be analyzed.

2) By default the program automatically computes the Fourier transform of the image and sets it as the “Current Displayed Image”. The original image can be shown by selecting it from the pull down menu “Current Displayed Image”.



Create Phase Image

This routine calculates the geometric phase and other images associated with the chosen reflection.

The calculation performed is the inverse Fourier transform of the masked Fourier transform of the original image. The mask in question is a mask of the type defined below and centered on the reciprocal spatial frequency g.

giving rise to an

amplitude image A(r)

and a

phase image $P(r) = -2\pi g.\partial u(r)$

The term $2\pi g.r$ has been subtracted from the phase, which is equivalent to moving the origin to the position of the reflection g.

$\partial u(r)$ is the displacement field with respect to the lattice planes defined by the frequency g.

For calculating the local value of everywhere, this is done as:

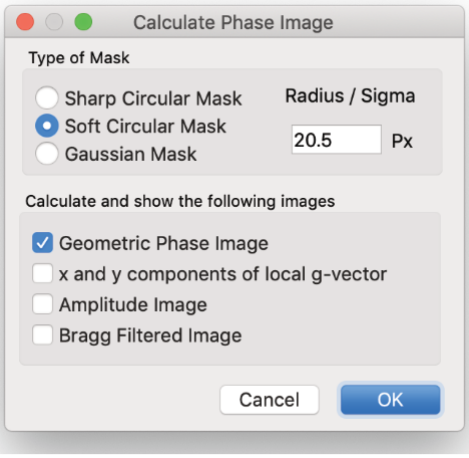
$g(r) = -\partial/\partial r\ P(r) + g$

U*sage:*

1) A G-vector must have been chosen using either the G1 or G2 tool for the Calculate Phase Image button to be activate. By clicking on a reflection a circle will be drawn around the chosen reflection, The size of the mask will be determined by the radius of the drawn circle which can be changed by its handles or by typing in the desired radius and clicking the "set" button.

2) Invoke the menu command Calculate Phase Image...

3) A dialog will give you the options available or creating the phase image.



Options:

Mask

The following choices to the type of mask can be made

- A sharp circular mask,

  M = 1, r <= R, M = 0, r > R

- A soft circular mask, using a Gaussian edge

  with a halfwidth of 1/5 of the radius

  M=1, r < 0.8R;

  M = exp{- (r-0.8R)2/sigma2}

  for r >= 0.8R, sigma = 0.2R

- A Gaussian mask, exp(-r2/sigma2)

Output

***K*eep the geometric phase image(default). Normally this is what you want.**

Calculate and display the local g-vector.

This calculates two images containing the x and y components of g(r) at each point in the image.

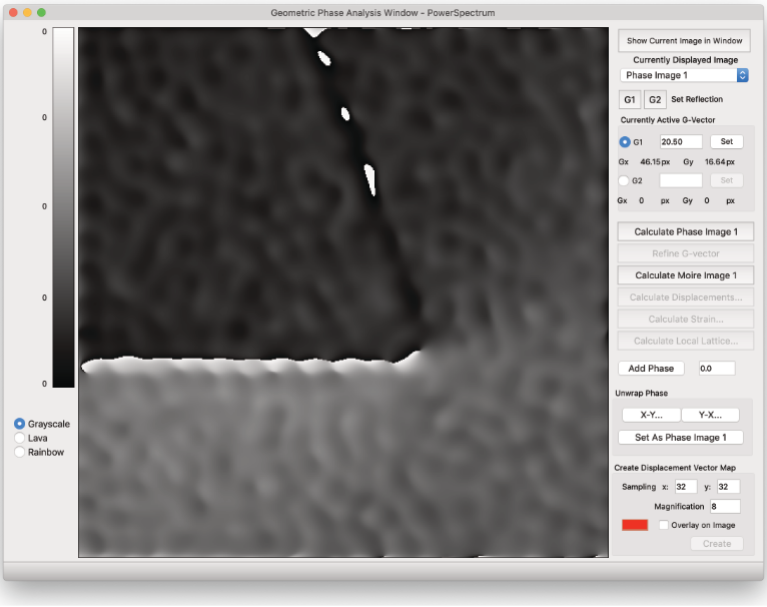
Keep and display the amplitude image.

Keep and display the masked FT from which the amplitude and phase images are calculated.

The resulting phase image is calculated by subtracting out the term

where g0 is given by the value returned from finding the position for the peak intensity in the power spectrum.

The resulting images are by default named Phase Image & Amplitude Image



Note:

1) After the phase image is calculated the next operation is to refine

the local g vector. This is performed by selecting an area in the phase image over which the lattice spacing are considered constant and invoking the menu-command Find local average g-vector. This implies that the spacings found this way becomes a reference to which subsequent calculations are related.

2) Before calculating displacements and strain, two non-colinear frequencies g1 , g2 must be defined.

3) After refining g1, the next step would be to set g2 using the G2 tool in the same fashion as the previous step, refining the local vector for the same area as when refining g2.

Refine local g-vector

Background:

In order to calculate displacements or strain one needs a reference lattice. The vector g that is found when creating the phase image in the previous command is the one with the highest amplitude and is not necessarily the one that one would like to choose for the reference lattice. In order to further refine the vector g one chooses an area with the selection tool that is the area that is to become the reference area from which to calculate an average lattice-spacing. In order to define a reference lattice one needs two vectors g1 and g2 which both will need to be refined with respect to the same area.

Usage:

Mark an area with the selection tool where you want to find the average g-vector.

Run the command “Find local average g-vector”. The command will determine if there is a residual ramp $2\pi\partial g.r$ within the selected area that should be subtracted out. The correction ∂g is added to the vector g that was found during “Calculate phase image...”. The routine can be used iteratively to refine the vector g, by using successively larger selections if the first selection needs to be small

because of phase-jumps.

The images below show the original phase image with a selection marking the area which will be used to find the average local g-vector. To the right, is displayed the result of the operation.

N*ote:*

As mentioned above, two vectors g1 and g2 are used for defining a reference lattice. Thus normally this command is used for each vector g from which a phase image is calculated before the phase image is referred to as phase image #1,2.

Add Phase

This routine adds a constant phase to the image "Phase Image". Under normal operation this command is not used. It can be used to remove discontinuities associated with wrap around effects as the phase crosses 2π

Create Moiré

Usage:

Use this routine to create a Moiré image calculated from the image "Phase Image" It is equivalent to moving the origin for the reciprocal space inverse FFT to the point |g|/M along the line from the origin to the spatial frequency g. Mathematically, the Moiré image becomes

$P(r) = 2\pi g.r/M - 2\pi g.\partial u(r)$ where M is the magnification chosen in the dialog box.

$M \rightarrow \infty$ gives the geometric phase image calculated in the "Calculate phase image(s)" command.

Calculate Displacements

Executing this command produces two images ux and uy which are the local x,y displacements with respect to the lattice defined by g1 & g2.

The command requires that the two images "Phase Image 1" and "Phase Image 2" exist together with the associated values for g1 and g2.

Invoking the command brings up the following dialog box.

producing the following output as the result of the operation.

Prerequisite:
Both g1 and g2 including the associated images “Phase Image 1” & “Phase Image 2” must exist prior to executing this command.

Below is shown a vector plot using the x-displacements as the length of the x-component of the vector and the y-displacement as the y-component of the vector.

The data can also be shown as a surface plot depicting the magnitude of the displacement.

Alternatively the magnitude of the displacement can be shown as a contour map as illustrated below.

Calculate Strain

This command calculates the strain with respect to the reference lattice defined by g1 & g2.

Prerequisite:
The two phase images “Phase Image 1” and “Phase Image 2” must have been calculated.

Results:
Depending on the requested output, the result will be one or more of the following

a) the Deformation Matrix
b) the Symmetric Strain Matrix
c) the Rotation Angle
d) the Principal Strain Components

The images on the next page illustrate the output produced by choosing to calculate and display the principal strain components e1 and e2. Four images are produced, the x and y components of e1

and the x and y components of e2.

P*rocedure:*

The program first calculates the deformation matrix

$$D = \begin{bmatrix} e_{xx} & e_{xy} \\ e_{yx} & e_{yy} \end{bmatrix}$$

which relates the positions of a point r' (x',y') to its unstrained location r (x,y) through the relationship

r' = (1+D)r = Sr

The matrix S is then decomposed into a symmetric matrix S' and a rotation matrix W

$$S = \begin{bmatrix} 1 + e'_{xx} & e'_{xy} \\ e'_{yx} & 1 + e'_{yy} \end{bmatrix}$$

and

$$W = \begin{bmatrix} \cos\delta & \sin\delta \\ -\sin\delta & \cos\delta \end{bmatrix}$$

with

$$\cos\delta \approx 1$$

$$\sin\delta \approx \delta = \frac{1}{2}(e_{xy} - e_{yx})$$

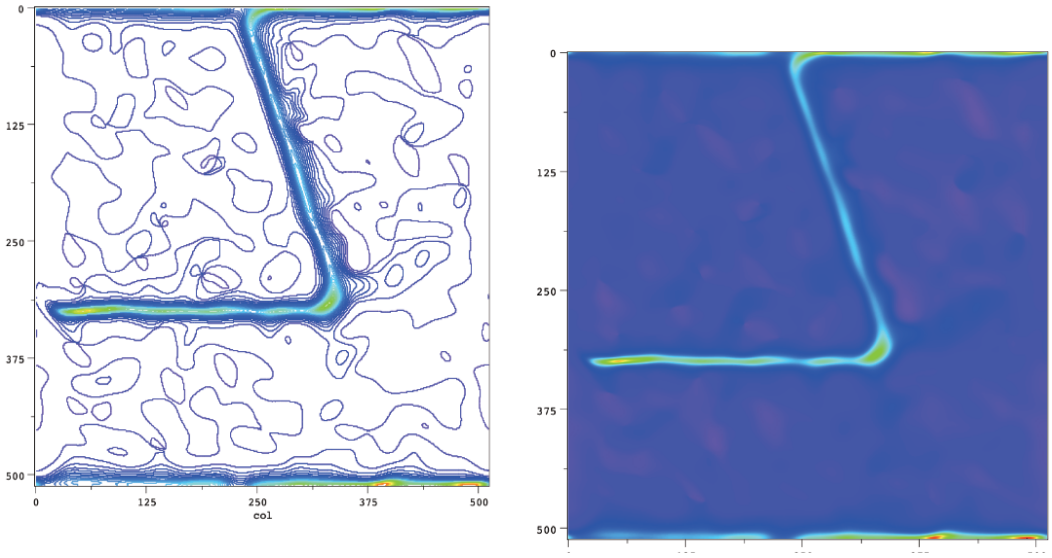
Thus we have the symmetric strain matrix

$$e' = \begin{bmatrix} e'_{xx} & e'_{xy} \\ e'_{xy} & e'_{yy} \end{bmatrix}$$

which is calculated and the three components are shown as images.

as the two vectors e1 and e2. These are also calculated and the images e1x,e1y and e2x and e2y are displayed. These are the x,y components of the two principal strain vectors respectively.

Contour map of the magnitude of one of the principal strain components



.

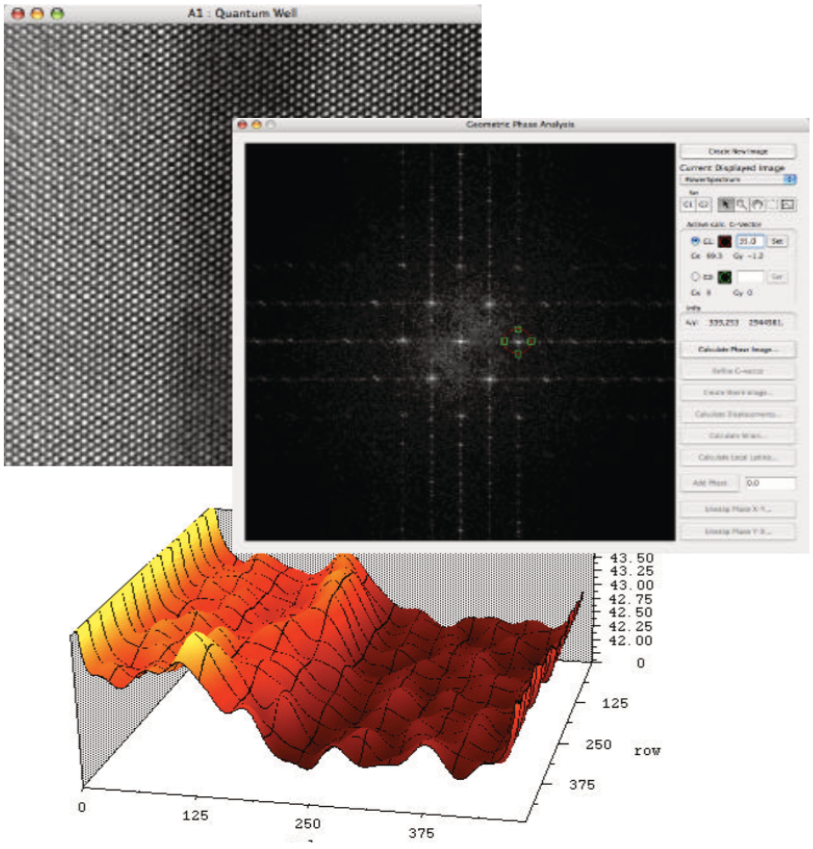
Pseudo colored image of the magnitude of a principal strain component.

Calculate Local Lattice

This command is used to calculate the local lattice parameters a and b at different points in the image. It uses the local values for g1 and g2 at each point together with the Miller indices of the two reflections to refer to a lattice given by the vectors a and b. Not that the y-coordinate can either refer to picture coordinates (0,0 top left) or a normal coordinate system (x,y) depending on what is set under Preferences.

Prerequisite:
Both vectors g1 and g2 together with the respective phase images Phase Image 1 and Phase Image 2 must have been set.

Example
The images shown below illustrate the use of the routine on a HRTEM image of a quantum well in In(Ga)N. Left image is the HRTEM image and the right image shows the power spectrum and one of the reflections used for calculating the lattice. The third image at the bottom is a surface plot of the a-lattice parameter as a function of position in the image.



Unwrap Phase (x,y)

The command "Unwrap Phase(x,y)" will try to unwrap the phase in the image "Phase Image".

The user is prompted for a starting position which is assigned the phase at that point. The routine then moves first along each row, trying to preserve continuity in the phase value. Instead of making a jump from $\pi-\partial$ to $-\pi+\partial$, the phase will take on the value $\pi+\partial$ and continue to increase or decrease when crossing phase jumps.

This command first does the rows and then the columns of the image.

Unwrap Phase (y,x)

The command "Unwrap Phase(y,x)" will try to unwrap the phase in the image "Phase Image".

The user is prompted for a starting position which is assigned the phase at that point. The routine then moves first along each column, trying to preserve continuity in the phase value. Instead of making a jump from $\pi-\partial$ to $-\pi+\partial$, the phase will take on the value $\pi+\partial$ and continue to increase or decrease when crossing phase jumps.

This command first does the columns and then the rows of the image.

Finding Focus from a HRTEM Image

Background

The amount of defocus in a HRTEM image can be determined from its power spectrum if sufficient amount of amorphous material is present. Under these conditions it assumed that the weak phase object approximation can be applied to describe the features in the power spectrum that derives from the effect of the objective lens on the scattered electrons passing through a thin amorphous material. Under the weak phase object approximation, the power spectrum for a HRTEM image is proportional to:

$$\sin^2\chi(\mathbf{u}, C_s, \lambda, f_0, \Delta f, \alpha)$$

w*here:*

$\boldsymbol{u}$ = r*eciprocal lattice vector*

C_S = s*pherical aberration coefficient*

λ = e*lectron wavelength*

f_0 = o*bjective lens defocus*

Δf = s*pread offocus*

α = c*onvergence angle*

In the absence of any astigmatism and ignoring the effect of the spread of focus (temporal incoherence) and convergence (spatial incoherence), the intensity in the power spectrum is approximately proportional to

$$\sin^2\chi(|\mathbf{u}|, Cs, \lambda, f_0)$$

where

$$\chi = \frac{\pi}{2}(C_s\lambda^3 u^4 - 2\lambda f_0 u^2)$$

This shows that there will be maxima in the power spectrum whenever

$$\chi = \frac{\pi}{2}(C_s\lambda^3u^4 - 2\lambda f_0 u^2) = n\frac{\pi}{2}$$

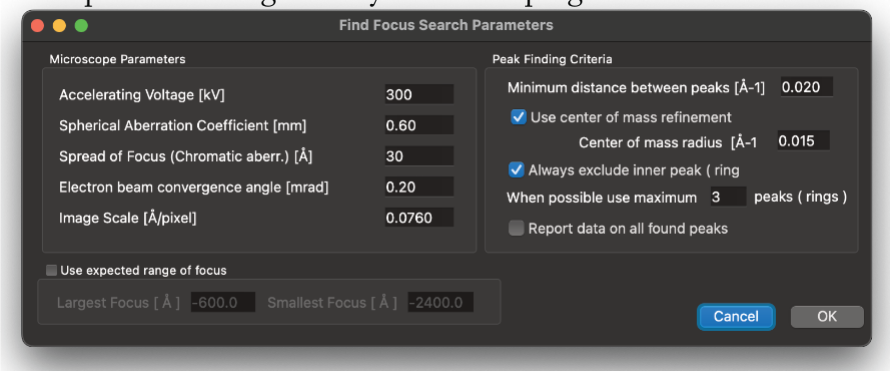
For n =odd integer

Thus if the spherical aberration is known, the focus can be determined by finding the locations of the maxima in the power spectrum, assigning indices (n) to the various maxima (rings) and solving for focus ($f0$).

With several rings in the power spectrum, the value for focus obtained from each ring will vary and the focus is expressed as the average value together with a standard deviation.

Before you do anything with these routines, you will need to specify the microscope parameters that are used in the calculation of the focus from the HRTEM image. The parameters that you will need to set are “Microscope Voltage” and “Objective Lens Spherical Aberration Constant”.

In addition, there are a number of parameters used by the program in its search for rings in the image power spectrum and the assignment of ring indices. It is recommended that you first accept the default parameters until a better understanding of the influence of the parameters is gained by use of the program.



Microscope Parameters

Voltage

This is the accelerating voltage of the microscope in kV. It is used in determining the contrast transfer function for the microscope.

S*pherical Aberration*

This is the spherical aberration coefficient of the objective lens in mm. It is used in determining the contrast transfer function for the microscope.

S*pread of Defocus*

This is the effective spread of defocus which is a due to of the chromatic aberration of the objective lens in conjunction with fluctuations in accelerating voltage, lens current and the energy of the electrons emerging from the filament. It is given in Å and is used in determining the contrast transfer function for the microscope.

C*onvergence Angle*

This is the half angle for the cone of incoming electrons onto the specimen in milliradians. It is used in determining the contrast transfer function for the microscope.

P*eak Finding Criteria*

When search for ring maxima in the power spectrum, the program uses adjustable parameters in order to exclude spurious peaks and to determine the "exact" position of a ring.

M*inimum distance between peaks*

If two peaks are closer than the specified distance in Å-1, the smaller of the two are discarded.

U*se center of mass (Intensity) within given distance.*

If the option is to use center of mass, the peak position will be refined by using the intensities within the given distance in Å-1 to obtain a new position for the peak.

R*eport all peaks found*

If this option is set, the program outputs data on all the peaks that it finds.

Focus Determination Criteria

In assigning the ring indices, the first maxima will be assigned the index 1 provided that the first ring gives a focus that lies within the range of acceptable focus values. If it doesn't, the program will move to the next maximum in the power spectrum. Following rings are given indices 3,5,7, etc.. Generally the first ring has a large uncertainty associated with the location of its maximum and the program will not use the focus value obtained from the first ring if that option is set (default).

E*xclude the first ring*

This makes the program not include the focus calculated from the first ring when it finds the average focus value.

U*se maximum # rings*

The program might find many maxima which increasing uncertainty with respect to whether they should be included in the focus determination. The program will not use any more than the number specified.

Focus Range

The program will use this range to help the algorithm determine if certain peaks are spurious and should not be used in assigning ring indices.

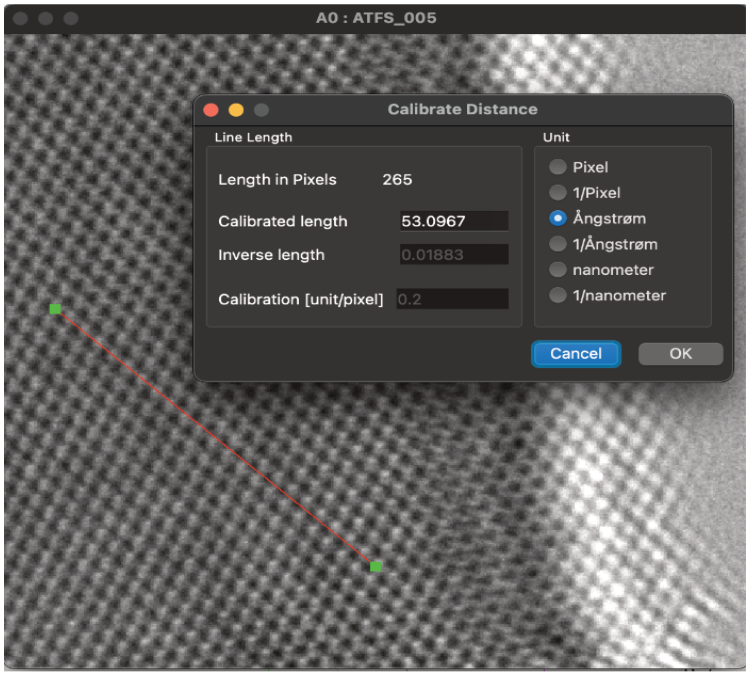
Find Focus from Image

This will attempt to find the focus value from a single HRTEM image. The accuracy to which the focus can be determined depends on many factors; the resolution of the image, the size of the Fourier transformed region (the resolution in reciprocal space), the number of rings in the power spectrum (the value of the focus) and the amount of amorphous material present in the image.

Procedure:

Unless the image is already calibrated or the calibration is known, you will first need to calibrate the image you are working on.

Use the line Ruler tool to create a line of known dimension in the HRTEM image and calibrate the line.

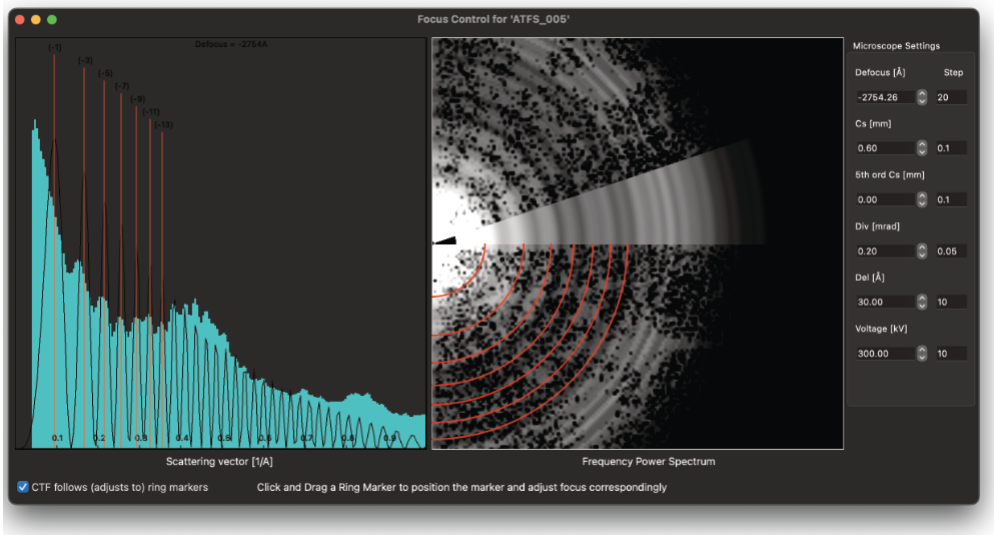


The calibration should be in Å or nm for the routine to work.

Once this is done, select an area that is square with dimensions being powers of 2 (hold down the Option key when you select the area) that contains some amorphous region. In this example, the area is a 512 square region of a 1024 image. If no selection is set, the routine will use the entire image. In this particular example, choosing a 256 square region would give insufficient sampling in reciprocal space to give good results.

Before you execute the command “Find Focus from Image”, make sure that there is only one selection set in the image.

Executing the command brings up the following result window.



The program attempts to calculate the power-spectrum associated with the amorphous content as a function of the radius (reciprocal distance). This is calculated by producing the median value for the magnitude of the power spectrum for each radius. The median is

plotted in blue-green and corresponds to the experimental data. The peaks in the one dimensional data set are found and assigned ring indices. The focus is calculated from each ring and the mean value for focus (defocus) is used to calculate the hypothetical Contrast Transfer Function CTF which is plotted as

$$\sin^2 \chi(\mathbf{u}, C_s, \lambda, f_0, \Delta f, \alpha)$$

based on the values that are listed.

The maxima for the CTF are shown by their vertical markers which can be dragged by the mouse. Click on a marker, hold down the mouse and the location of the maximum can be changed. All the maxima changes and the corresponding value for focus is shown. As an inset, the experimental power-spectrum is shown with a region showing the average data that the program uses for finding the rings. Superimposed on the experimental data are the rings corresponding to the current value for the focus. By moving the markers, the rings change position and the user can manually try to obtain the best fit for the value of focus.

HOLZ Interactions & Sub-slicing

With suitable algorithms, it is possible to include in the diffraction calculation the effects of out-of-zone scatterings, or non-zero (or higher-order) Laue zone (HOLZ) interactions. Basically, there are four ways to produce the set of phasegratings (or projected potentials) that describe the "multi-sliced" crystal. For structures with short repeat distances in the beam direction, the simplest method is to use one slice per unit cell. For structures with large repeats in the beam direction, several methods may be used, three of which rely on sub-dividing the slice into "sub-slices". Any of the four methods can be used in Tempas.

Identical slices with only one sub-slice per unit cell repeat distance

A multi-slice computation in which every slice is identical contains no information about the variation in structure along the incident beam direction, and includes scattering interactions with only the zero-order Laue zone (ZOLZ) layers. For structures with short repeat distances in the beam direction such a computation is adequate, since the Ewald sphere will not approach the (relatively distant) high-order zones.

Identical sub-slices with n sub-slices per unit cell repeat distance

For structures with large repeats in the beam direction, a method of sub-dividing the slice is required in order to compute the electron scattering with sufficient accuracy. The simplest, but most approximate method, is to compute the projected potential for the full repeat period then use 1/n of the projected potential to form a phase-grating function that can be applied n times to complete the slice. This method avoids interaction with any "pseudo-upper-layer-line" (Goodman and Moodie, 1974), but ignores real HOLZ layers.

Sub-slices based on atom positions

An improvement on sub-dividing the projected potential is to sub-divide the unit cell atom positions. In this procedure the list of atom positions within the unit cell is divided into n groups depending upon the atom position in the incident beam direction. From these sub-sliced groups, different projected potentials are produced to form n different phase-gratings, which are applied successively to produce the scattering from the full slice.

Sub-slices based on the three-dimensional potential

A further improvement on sub-dividing the atom positions, is to sub-divide the three-dimensional potential of the full slice, since an atom with a position within one sub-slice can have a potential field that extends into the next sub-slice. Rather than compute a full three-dimensional potential and then integrate over appropriate sub-slices (a 128x128x128 potential would require over two million samples to be stored), it is possible to derive an analytical expression for the potential within the sub-slice z0 ± ∂z projected onto the plane at z0 (Self et al., 1983). It is possible to apply this method routinely to structures with large repeats in the beam direction, thus generating several different phase-gratings for successive application, and even to structures (perhaps with defects) that are aperiodic in the beam direction and require a large number of individual non-repeating phase-gratings (Kilaas et al., 1987).

Tempas sub-slicing

While ensuring that the calculation remains sufficiently accurate, Tempas will normally choose the simplest (and quickest) method of specifying how slices are defined for any particular combination of specimen, zone axis, accelerating voltage, and maximum g. To this end, the user can choose to neglect HOLZ interactions if these are judged to be unimportant. If HOLZ interactions are important, then the user should select the “3D-Potential Calculation” radiobutton in the Options menu, rather than “2D-Potential Calculation”.

When a two-dimensional calculation is selected, Tempas will use one slice

per cell if the cell repeat distance in the beam direction is small. If the repeat distance is too large for one slice per unit cell, Tempas will avoid pseudo-upper-layer-lines by producing n identical sub-slices.

When a three-dimensional calculation is selected, (3D-Potential Calculation activated), Tempas uses a sub-divided three-dimensional potential when the repeat distance is large, and defaults to one slice per cell if the distance is small enough. Note that the number of sub-slices per unit cell can be forced to be greater than one by setting it explicitly in the Parameter menu; this will ensure that any HOLZ interactions are included even for small repeat distances. Of course, if the repeat distance is very small, leading to a distant HOLZ in reciprocal space, both the calculation and the experiment it is modeling will interact only very weakly with the HOLZ reflections.

Use of the Layered Structure option to produce the scattering from a structure that is layered or aperiodic in the incident beam direction is effectively an application of the method of sub-slicing based on atom positions. Thus the user could create a number of sub-slices by assigning selected atoms to different structure files, then forming a phasegrating for each sub-slice, and using the Stack Phasegratings command to specify how the sub-slices are to be used to describe the specimen structure. This is the suggested method to try first if upper Laue layers are to be included or 3-dimensional effects are important as it is much faster than using a complete 3D calculation.

Other methods

Van Dyck has proposed other methods to include the effects of

HOLZ layers, including the second-order multi-slice with potential eccentricity (Van Dyck, 1980) and the improved phase-grating method (Van Dyck, 1983). Tests of these procedures show that the extra computation involved in using potential eccentricity may be worthwhile, but that the improved phase-grating method diverges too easily to be useful.

Goodman P, Moodie AF (1974) Numerical evaluation of N-beam wave functions in electron scattering by the multi-slice method. Acta Cryst. A30, 322-324.

Kilaas R, O'Keefe MA, Krishnan KM (1987) On the inclusion of upper Laue layers in computational methods in high resolution transmission electron microscopy. Ultramicroscopy 21, 47-62.

Self PG, O'Keefe MA, Buseck PR, Spargo AEC (1983) Practical computation of amplitudes and phases in electron diffraction. Ultramicroscopy 11, 35-52.

Van Dyck D (1980) Fast computational procedures for the simulation of structure images in complex or disordered crystals: A new approach. J. Microscopy 119, 141-152.

Van Dyck D (1983) High-speed computation techniques for the simulation of high resolution electron micrographs. J. Microscopy 132, 31-42.

Structure Refinement Through Matching of Experimental and Simulated HRTEM Images

Introduction

The goal of performing simulation of HRTEM images is to compare these with the experimental data in order to determine the structure. In practice this means that various models are proposed and that images are calculated until a match is found. At that point, the structure is presumed to be known (atomic positions and atomic numbers) with some given uncertainty. Alternatively, one starts with a given model and varies the model in a systematic fashion searching for a global maximum in the fit between experiment and simulation. This entails that one needs an efficient method to compare the experimental and the calculated image. It also requires knowledge of the uncertainty in the measurement (image intensities in the experimental image) and a way to relate this uncertainty to the uncertainty in chemical composition and atomic positions. This area of quantitative electron microscopy is fairly new and most images are still compared visually. However, it is an active area of research and many techniques from statistics are just now beginning to be used in HRTEM.

Acquiring quantitative data

In order to extract quantitative information from electron micrographs, the data must be represented by a set of numbers. Usually, images from TEMs are brought into a digital representation by one of the following methods.

i) Recording the image on a photographic plate and using a scanner to convert the film density into numbers which are stored in a computer.

ii) Recording the image on an image plate.

iii) Recording the image on a slow scan CCD camera with readout of deposited charge into a computer.

The first approach yields data that is not directly comparable to computer calculations because of the non-linear response of the film. It is however possible to calibrate the response of the film based upon a sequence of controlled exposures using varying exposure times and mapping the resulting scan values versus electron dosage[1]. The image plate and the CCD camera both yield numbers that are linear with respect to the electron dosage and only require a scaling of the data in order to compare to computed values[2]. There has been much discussion about the relative merits of the various recording media above and each has its own advantage. The CCD camera is currently limited to 2K by 2K pixels, although it may be possible to go to 6K by 6K by using multiple chips in the near future. Since its Modulation Transfer Function (MTF) can be characterized, it is straight forward to use deconvolution to compensate for the drop in high frequency response due to spread of electrons and due to spill over of charge to neighboring cells[3]. The image plate has many of the advantages of the CCD camera and covers a larger image area. However, the imaging plate is not gaining as much popularity as the CCD camera. Many laboratories are now starting to do much of the recording on CCD cameras, while still retaining the use of film.

Pre-processing of data

Once the image data has been converted to numbers, any necessary processing or transformation of the data can take place. The required pre-processing of the data depends on the nature of the information that is sought and thus there is no one optimal

method, but rather a number of possible options.

Sampling and resampling of data

If the image is distorted over the image field of view, either by the action of the imaging system or the recording system, the data can be re-transformed by a warping transformation of the image. This can be done if the transformation can be determined by imaging a perfectly crystalline material and noting deviations from where the atoms are known to be and where they are imaged[4]. On some systems, i.e. the Gatan Imaging Filter, the distortions are measured by recording the image of a square grid of circular holes.

An image of a crystalline material can be resampled onto lattice relative coordinates, such that the unit cell dimensions are represented by an integer number of pixels and commensurate with the dimensions of the final image. This will eliminate streaking in the Fourier transform of the image which is due to the truncation of the image by the edges on boundaries that do not represent a periodic continuation of the image. Streaking can be reduced by multiplying the image with a circular mask. The mask is represented by a circle of pixels with value 1 up to a specified radius and then falling off gradually to 0 within 5 to 10 pixels close to the borders of the image. A side-effect of masking is the increase in noise in the Fourier transform which is discussed below.

If only the image of a single unit cell of the crystalline material need to be determined and compared with the image obtained through an image simulation calculation, the image of the unit cell can be resampled onto the coordinate system and sampling interval used in the computation. This is equivalent to determining the matrix M defined through the equations

$$\mathbf{a}_S = M\mathbf{a}_e \qquad 1)$$

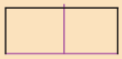
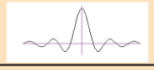
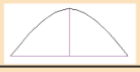
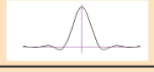
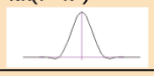
$$\mathbf{b}_S = M\mathbf{b}_e$$

Two steps are necessary; the rotation /scaling required to make the lattice base vectors identical; and secondly the determination of a common origin.

Finding the common origin between experimental and calculated image is determined by cross-correlation between the simulated and the experimental image[5].

Fourier transforms and masking

The Fourier transform of HRTEM images of crystalline materials provides useful information about lattice spacings and can also be used to compare experimental Fourier amplitudes with theoretical calculations. Because the image being transformed is rarely a periodic function in W (width) and H (height), the Fourier transform of the image of a pure crystalline material is the convolution of the Fourier transform of a perfectly periodic signal (the crystal) with the transform of a window the size of the image dimensions, making a Bragg peak take on the shape of the transform of the window.

name	Window Function	Transform	falloff	rel. noise level
none	1 	$\frac{\sin(\pi k)}{\pi k}$ 	$\frac{1}{k}$	1
cosine	$\frac{\pi}{2}\cos\left(\frac{\pi x}{2}\right)$ 	$\frac{\cos(\pi k)}{1-4k^2}$ 	$\frac{1}{k^2}$	1.23
von Hann	$1+\cos(\pi x)$ 	$\frac{\sin(\pi k)}{\pi k(1-k^2)}$ 	$\frac{1}{k^3}$	1.5

The use of a mask changes the transform of the window and can be used to make the peak profiles decay faster, but at the expense of increasing the noise level. This is illustrated above, showing the effects of employing masks on a 1-dimensional signal[6]. This also has an effect on locating peak positions in order to determine lattice spacings and on the estimate of the amplitude of the Fourier component. The standard error in both estimates increases as a function of applying a mask, with the cosine window being a good compromise.

Noise reduction

In addition to reducing noise, it is also important to have an estimate for amount of noise present and to quote a signal to noise ratio. From two equivalent regions, the noise can be estimated from obtaining the cross-correlation coefficient for two regions. Given a cross-correlation coefficient ccf, the signal to noise ratio can be estimated as

2) $$S/N \approx \frac{CCF}{1-CCF}$$

In order to reduce noise and to obtain a statistical average of the image of a single unit cell (motif), the positions of individual motifs can be determined by cross-correlation. Once these are found, statistically equivalent regions can be averaged to find the average motif and to determine the signal to noise ratio associated with individual pixels as a function of position within the unit cell. This determines a standard deviation for each pixel i and can be used to set confidence levels associated with matching of the experimentally averaged image with a calculated image[7].

3) $$\sigma^2(i) = \frac{1}{M-1}\sum_{j=1}^{M}(I^2{}_j(i) - < I(i) >^2)$$

where M is the number of equivalent regions being averaged.

Using a low pass filter to perform a smoothing of the image may be effective depending on the noise level present, particularly when averaging over statistically equivalent regions can not be performed. Smoothing helps the eye see features more clearly; but has the disadvantage that it causes correlation between image pixels, which may distort the significance threshold of simulation mismatch criteria.

Averaging can also be performed through symmetrization which is to average the motif with copies of itself to which symmetry operations known to be present are performed. This will reduce noise levels by a further factor of when M symmetry related copies are averaged, but may also just disguise defects in imaging conditions.

Matching experimental and simulated images

There are a number of various ways to measure similarity or mismatch between two images. Below are a few of these[8].

The mean square difference:

4) $$D^2 = < (I_1 - I_2)^2 > = \frac{1}{N} \sum_i (I_1 - I_2)^2$$

The Root Mean Square Difference:

5) $$D_{rms} = \sqrt{D^2}$$

The mean modulus difference:

$$6) \qquad D_{mmd} = < |I_1 - I_2| > = \frac{1}{N} \sum_i |I_1 - I_2|$$

The Cross-correlation Coefficient:

$$7) \qquad CCF = \frac{\sum_i (I_1(i) - \overline{I}_1) * (I_2(i) - \overline{I}_2)}{\sqrt{\sum_i (I_1(i) - \overline{I}_1)^2 \sum_i (I_2(i) - \overline{I}_2)^2}}$$

The brackets <> all indicate the mean of the enclosed quantity.

In each of these equations, the sum is over all the pixels i in the image and N is the total number of pixels. The cross-correlation coefficient above is a normalized coefficient where the images are normalized to zero mean.

The CCF (which measures similarity rather than difference) can also be interpreted as the cross-product between two n-dimensional vectors (n being the number of pixels in the image). In that case, one can associate an angle with the CCF, , in the general interpretation of an inner product between two vectors as with the angle being .This angle is zero for identical images. If the images are normalized to zero mean and unit length as in the definition of the normalized cross-correlation coefficient above, the angle is 180 deg. for a reversal in contrast between the two images I_1 and I_2.

Significance and Noise

Each of the above criteria must be tested for the significance of the measured value.

D^2 can be compared to the mean square intensity (or intensity deviation due to noise) in either image.

D_{rms} can be compared with standard deviation of the intensity in either image

A good way to test for the mismatch between two images, is to use a statistical measure for the probability of two images being equal given knowledge of the noise in the images. If one assumes Gaussian uncorrelated noise for each pixel in the experimental image, the optimum statistical measure is given by

$$8) \quad \chi^2 = \frac{1}{N}\sum \frac{(I_1(i) - I_2(i))^2}{\sigma^2_e(i)}$$

where N is the number of pixels in the image[9]. The value is the standard deviation associated with the pixel i and can be found as described above from a number of equivalent regions. If an experimental image Ie is compared to a calculated image Ic and there are M adjustable parameters in the calculation, the equivalent expression becomes[7]

$$9) \quad \chi^2 = \frac{1}{(N-M)}\sum \frac{(I_e(i) - I_c(i))^2}{\sigma^2_e(i)}$$

A mismatch by one standard deviation adds one to the sum in the expressions above and a value of χ^2 of 1 implies that the two images are identical within the uncertainty given by the noise. The expected value for statistically equivalent images consisting of N points is 1 and random deviations from this value by more are considered unlikely.

Writing

10) $$\chi^2 = \frac{1}{N}\sum \frac{(I_1(i) - I_2(i))^2}{\sigma^2(i)} = \frac{1}{N}\sum f^2(i)$$

leads to the definition of a Residual Image [10]

which is used to visualize and to quantify the (mis)match between two images. It has the advantage that instead of presenting a single number for how well two images match, it is a two-dimensional mapping of the local fit. Thus a difference image will more clearly reveal areas of greater mismatch. The optimum match is still defined by minimizing χ^2.

It is important to note that the fitting parameters can also be applied to the Fourier transforms of the images which sometimes will lead to a reduction in the number of the data-points to be compared[11]. In the case of images of crystalline material containing no defects, the Fourier components will be non-zero only for frequencies corresponding to Bragg-reflections of the lattice, although this is strictly only true if the motif has been averaged over many repeating regions and resampled onto lattice coordinates such that streaking due to discontinuities at the boundaries is eliminated. The complex values for the Fourier coefficients take the place of the image intensities.

It is interesting to note that the use of different matching criteria can lead to slightly different values for optimized parameters[12].

Adjusting for different means and contrast levels

Since absolute values for image intensities are not known and an experimental image may be linearly related to a calculated image, a useful way of normalizing the image intensities is to subtract the mean and divide by the standard deviation. This ensures that D2 = 0 for linearly related images and a value of around 2 for unrelated data.

Similarly, the Cross-correlation coefficient will lie in the range from -1 to 1, taking the extreme values when the two images are linearly related and being near 0 for unrelated data.

Another approach is to scale the images to the same mean. This is done as follows

11) $$I_{calc} = \frac{I_{calc}}{\langle I_{calc} \rangle} \langle I_{\exp} \rangle$$

where the calculated image is scaled to the mean of the experimental image.

In order to understand how the mean value, contrast and image pattern affect the image matching criteria, it is useful to consider how the Root Mean Squared Difference can be separated into three terms[13].

12) $$D_{rms} = \sqrt{\langle (I_1 - I_2)^2 \rangle} = (\langle I_1 \rangle - \langle I_2 \rangle)^2 + (\sigma_1 - \sigma_2)^2 + 2(1-\rho)\sigma_1\sigma_2$$

where

13) $$\sigma_{1,2} = \sqrt{\langle I_{1,2}{}^2 \rangle - \langle I_{1,2} \rangle^2}$$

and

14) $$\rho = \frac{\langle I_1 I_2 \rangle - \langle I_1 \rangle \langle I_2 \rangle}{\sigma_1 \sigma_2}$$

The first term measures the difference in the mean of the two images and vanishes if both images are normalized to the same mean value. The second term measures the difference in contrast between the two images, while the third term (where is the same as the normalized cross correlation coefficient) measures the difference (similarity) in the pattern of the two images. Thus it is important to note that the normalized cross correlation coefficient

CCF only measures similarity in patterns and ignores variation in contrast and differences in mean levels. It is generally found that most of the mismatch between experimental and computer simulated images is due to the difference in contrast[14]. The difference in contrast can be an order of magnitude and the cause is generally attributed to the following factors.

misalignment
- specimen vibration
- inelastic scattering
- specimen damage
-

There is however an ongoing debate as to the nature of the discrepancy in contrast as calculations indicate that the factors above are not sufficient to resolve the disparity. A possible explanation is that there is a general background in experimental images that is not accounted for.

Effect of noise on matching criteria

In order for two images to be considered equal, we need to consider the effect of the uncertainty or error in the matching criteria due to noise and the parameters determining the image.

A study of the effect of noise on the cross-correlation factor reveals that in the presence of noise, the cross-correlation coefficient CCF for the two images I1 and I2+h, where h represent random noise superimposed on image I2, can be written as[13]

15) $$CCF(I_1,I_2,\eta) = CCF(I_1,I_2) \Big/ \sqrt{1+\frac{\sigma^2(\eta)}{\sigma^2(I_2)}} = CCF(I_1,I_2)/\sqrt{1+\theta_n^2}$$

with

16) $$\vartheta_n^2 = \frac{\sigma^2(\eta)}{\sigma^2(I)}$$

The effect on the hyper-angle is in the small angle approximation

17) $$\vartheta(I_1, I_2, \eta) = \sqrt{\vartheta^2(I_1, I_2) + \vartheta_n^2}$$

If two images are identical except for a small error in one of the image-formation parameters (defocus, thickness, etc.) the error in the angle is proportional to the parameter error. The error in the angle due to independent parameter errors is

18) $$\theta = \sqrt{\sum \vartheta_n^2}$$

Typical mismatches in CCF (pattern matching) due to parameter errors are

Parameter	**Error**	**theta(mrad)**
Noise	0.06	
Composition	±0.03	0.02
Thickness	±2nm	0.2
Defocus	±15nm	0.4
Beam Tilt	<1.5mrad	0.8
Astigmatism	<15nm	0.2
Crystal Tilt	<2mrad	0.6
Beam Diverg.	<0.3mrad	0.1
Focal Spread	<5nm	0.15
Vibration	<0.04nm	0.2

Chi-Square or Chi-based criteria

Although all the methods above measure either the match or mismatch between two images, the important questions is not to what degree do they match, but how well do they match given systematic and non-systematic errors. Thus the fitting parameter must take into account the statistical nature of the data and the accuracy to which we know the data-points. Thus the fitting parameter should depend on a maximum-likelihood (probability) model and be a measure of the probability that A is equal to B, given knowledge of the probability distribution of the data-points. In the presence of Gaussian distribution of uncorrelated noise, each data point has a Gaussian probability distribution with the noise in one pixel uncorrelated to the noise in adjacent pixels, which leads to a χ^2 criteria. The criteria takes into account the number of adjustable parameters and the error in each data point.

As mentioned above, any data point lying one sigma away from the expected value will add 1 to the sum in χ^2.

Similarly, any data point which has only 1% probability of being measured given A = B, adds a value of 6.63 to the sum in χ^2. Thus values of χ^2 greater than about 6 states that there is less than 1% probability that A is equal to B.

The fitting parameter depends on the model of the distribution of data-points due to statistical noise with a Gaussian distribution of uncorrelated noise leading to the criteria. However, it is important to determine the statistical nature of the noise in the image. This can be done by examining the noise distribution determined from a large number of image regions considered to be equivalent except for noise. A non-Gaussian distribution will lead to a modified criteria, but still based upon [7].

Structure determination

In order to determine the "unknown" structure, it is necessary to perform a comparison between calculated images, exit wave functions or diffraction patterns and experimentally obtained data. As described above the comparison can be done using different matching/mismatching criteria.

Ideally, the determination of the structure is done by modifying the structure until the mismatch between the experimental and calculated data is within the error in the experimental data. In principle the imaging parameters themselves can be allowed to vary together with the atomic coordinates. However, in practice the imaging parameters are optimized separately if possible. This reduces the complexity of the problem and reduces the number of steps involved in the search for a solution which optimizes the matching criteria. In cases involving unknown defects in the presence of a "known" structure, the imaging parameters and specimen thickness are first determined from the known structure.

Determination of an unknown set of input parameters requires the following

1) An image (in real or reciprocal space) obtained from the experimental data.

2) A computational method yielding an image to be compared to the 1).

3) A method for comparing 1) and 2)

4) A criteria based upon 3) for when 1) and 2) are statistically equivalent.

5) An initial set of adjustable input parameters which are to be optimized so that the final configuration results in satisfying 4)

6) A method for varying the adjustable parameters so that the final configuration is found within finite time.

There is an essential assumption being made above, which states that the computational method used in 2) will produce the image in 1) given the correct choice of input parameters. This is a separate issue which will not be addressed here. The validity of this assumption can be debated and it is acknowledged that computational methods are in need of further refinement. However, in what follows, the assumption is presumed to be valid.

Matching Images or Exit wave functions

In order to compare calculation with experiment, one can compare either images or diffraction patterns. For perfect structures it may be beneficial to compare diffraction patterns since the number of data points to compare are given by the possible Bragg reflections of the structure[11]. However, for defect structures, the information that describes the defect is located in the diffuse scattering between Bragg spots, and it is more efficient to compare images. The entire discussion relates to both real space and reciprocal space, although only real space images will be referred to.

Simulated Thermal Annealing

Simulated thermal annealing is a relatively new technique for finding the global minimum of a multivariate function[15]. The algorithm is based upon assigning an energy to the system which is a function of the parameters (x_1,x_2,x_3,....x_n) being varied, with the optimum configuration of the system being the minimum energy state, the ground state. A temperature is also assigned to the system and the temperature is slowly being reduced as the configuration is changed. From the initial configuration $E_0(x_1,x_2,x_3,....x_n)$, the parameters are varied in a random fashion and for each variation the new energy $E_j(x_1,x_2,x_3,....x_n)$ is calculated. The new configuration is always accepted if $\Delta E = E_j(x_1,x_2,x_3,....x_n) - E_{j-1}(x_1,x_2,x_3,....x_n) < 0$. Otherwise the new configuration has a probability P of being accepted, where

$$19) \qquad P = e^{-\Delta E / T}$$

E and T being dimensionless quantities.

For each temperature, the system undergoes a given number of variations, accepting or rejecting the new configuration based upon the above criteria.

When a specified number of successful transitions have taken place, the temperature is lowered by a certain amount, and the parameters are changed again. As a function of iterations, the energy of the system decreases towards what is hoped to be the minimum energy configuration and the process is terminated when either no more successful variations are made for a given number of attempts or the temperature reaches a lower limit.

When comparing calculated and experimental images, the energy of the system can be chosen to be c2 or any of the other quantities that measures image-mismatch. When basing the comparison on the cross-correlation coefficient, the energy can be taken as 1-CCF.

Simulated thermal annealing is a straight forward technique that has proven to be very powerful for finding global minimum without getting trapped in local minima. It is sensitive to the starting conditions and the choice of starting temperature and some experimentation may be required. Near the minimum, it tends to be less optimal than search techniques based upon gradient methods and switching to a different search algorithm may be an alternative once the simulated annealing algorithm has terminated.

Simulated Evolution

Simulated evolution is another technique for obtaining the global minima which is modeled after Darwin's principle of "survival of the fittest"[16] It starts with an initial configuration of all the variables to be fitted and produces a number of sets l from the initial set using a random generator (mutation generator). This set l represents the first generation of children. The algorithm proceeds in the following way:

i) Evaluate a quality function Q (goodness of fit) for all l children.

ii) Select a subset ($\mu < l$) of survivors which will be he parents

of the next generation.

iii) Create a new generation by applying the random generator, after selecting and mixing a part of the parent's parameter vectors.

iv) Loop back to i) until one of the following criteria are met: a) a maximum number of generations have been reached or b) a critical goodness-of-fit has been reached.

Other Techniques

There are other ways to do the refinement which is based upon changing the input parameters so that the system moves in a path where the gradient with respect to the fit is the largest[17]. Each method has its advantages. Simulated thermal annealing and simulated evolution are good techniques for getting close to the optimum fit. Once close to the minimum, gradient methods may be used for further refinement until the match is within the uncertainty of the measurement.

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Optimal and Near-Optimal Filters in High Resolution Electron Microscopy

1. Introduction

Most images recorded in High Resolution Transmission Electron Microscopy (HRTEM) show the presence of amorphous layers, such as native oxide, contamination or support film. The amorphous layer degrades the information which comes from the crystalline material of interest and lowers the signal to noise ratio. What is referred to as signal and noise in this case are actually two signals, one coming from the crystalline region of the specimen and the other coming from the amorphous layer(s).

The signal which is recorded on film, image plate or slow scan CCD camera is a function of the interaction of the electrons with the specimen and cannot in principle be separated into components from an amorphous region and a crystalline region. However, as a first approximation, the effect of the amorphous material is to add the image of the amorphous material to the image of the crystalline material. Mathematically, this can only be shown within the validity of the Weak Phase Object Approximation (WPOA). For thin specimens and using the WPOA (Cowley & Moodie, 1957), the electron wavefunction at the exit surface of the specimen can be written

real space (1)

$$\Psi(\mathbf{r}) \approx 1 + i\sigma(V_c(\mathbf{r})t_c + V_a(\mathbf{r})t_a)$$

or

$$\Phi(\mathbf{u}) \approx \delta(\mathbf{u}) + i\sigma(\bar{V}_c(\mathbf{u})t_c + \bar{V}_a(\mathbf{u})t_a)$$

reciprocal space (2)

where σ is the interaction constant, t_c and t_a are the thicknesses of the crystalline and the amorphous material, and Vc and Va are the two respective projected potentials.

The effect of the objective lens on the electron wavefunction is to add in reciprocal space a phase factor **X(u)** such that to the first approximation the Fourier transform of the recorded intensity is

$$F(\mathbf{u}) \approx 2\sigma(\bar{V}_c(\mathbf{u})t_c + \bar{V}_a(\mathbf{u})t_a)\sin\chi(\mathbf{u}) \equiv F_c(\mathbf{u}) + F_a(\mathbf{u}) \quad (3)$$

with the corresponding real space representation

$$I(\mathbf{r}) \approx I_c(\mathbf{r}) + I_a(\mathbf{r}) \quad (4)$$

The components F(u**)** are complex and because the image is real, F*(**u)** = F(-**u)**.

The expression above does not include the effects of partial coherence in which the recorded image is a sum of incoherent images from electrons with a spread of energies and incoming directions. However, including partial coherence does not change the decomposition of the recorded signal into the two components, one crystalline and one amorphous, provided that non-linear terms are neglected in the image formation (O'Keefe 1979). The weak phase object approximation is usually valid for the amorphous layer and also applies in many cases for the crystal for the thicknesses where the amorphous layer shows a large effect on the image quality.

The two signals have distinctly different characteristics in the frequency domain. The amorphous contribution is spread out in reciprocal space with amplitudes that depend on the magnitude of the reciprocal vector and can be considered independent of the azimuthal angle except for random variations. The signal from the crystalline material is peaked around specific spatial frequencies which correspond to spacings between atomic planes. Thus while the information in real space is inseparable, the two signals can to a large extent be separated in the frequency domain.

The aim of this paper is to outline automatic procedures for estimating the

signal $Fc_{(}u)$ from F(u) (obtained from the Fourier transform of the recorded signal I(r)) and thus obtaining an estimate of the image $Ic_{(}r)$ which would have been recorded in the absence of the amorphous material.

2. Mathematical derivation of the filters

2.1. Optimal Filter (Wiener)

The estimate for the signal $Fc_{(}u)$ is written as:

(5) $$F_c^{est}(\mathbf{u}) = F(\mathbf{u}) \cdot \Theta(\mathbf{u})$$

where we seek to optimize the solution such that the summed square differences between the true signal $F_C(u)$ and the estimate $F^{est}(u)$ is minimized. This is expressed as minimizing the quantity

(6)
$$\sum_{\mathbf{u}} | F_c^{est}(\mathbf{u}) - F_c(\mathbf{u}) |^2 =$$
$$\sum_{\mathbf{u}} | (F_c(\mathbf{u}) + F_a(\mathbf{u})) \cdot \Theta(\mathbf{u}) - F_c(\mathbf{u}) |^2$$

The expression above is minimized by minimizing each term. Setting the derivative with respect to equal to zero gives

(7) $$\Theta(\mathbf{u}) = \frac{| F_c(\mathbf{u}) |^2}{| F_c(\mathbf{u}) |^2 + | F_a(\mathbf{u}) |^2}$$

The filter above is well known and is referred to as the Wiener filter (Rabiner & Gold, 1975). In the derivation of the Wiener filter, it is important to note that the quantities $F_C(u)$ and $F_a(u)$ are treated as uncorrelated. Thus there is an underlying assumption that there is no correlation between the signals from the crystalline and the amorphous materials. The degree of correlation between the estimates of the two signals will vary depending on the thickness of the amorphous and crystalline part (Hÿtch &

Chevalier, 1994), but will be considered to a first approximation to be negligible. The Wiener filter and variants thereof, including the Cannon filter (Cannon, 1977), and their use in HREM have been discussed in detail by Marks (Marks, 1996).

In order to determine $\Theta(\mathbf{u})$ we need estimates for both quantities $|F_c(\mathbf{u})|^2$ and $|F_a(\mathbf{u})|^2$.

"Because the optimal filter results from a minimization problem, the quality of the results obtained by applying the optimal filter differs only from the true optimum by an amount that is second order in the precision to which the filter is determined" (Press et al., 1986). Thus excellent results are often obtained by fairly crude estimates of the filter.

Writing

$$|F_c(\mathbf{u})|^2 + |F_a(\mathbf{u})|^2 \approx |F(\mathbf{u})|^2 \quad (8)$$

gives as the estimate for the optimal filter

$$\Theta_{WF}(\mathbf{u}) \approx \frac{|F(\mathbf{u})|^2 - |F_a(\mathbf{u})|^2}{|F(\mathbf{u})|^2} \quad (9)$$

so that, together with the Fourier transform of the recorded signal, the only quantity that needs to be determined is $|F_a(\mathbf{u})|^2$. Section 3 will discuss how to determine $|F_a(\mathbf{u})|^2$ from $|F(\mathbf{u})|^2$.

2.2. Average Background Subtraction Filter (ABSF)

$$\Theta_{WF}(\mathbf{u}) \approx \frac{|F(\mathbf{u})|^2 - |F_a(\mathbf{u})|^2}{|F(\mathbf{u})|^2} = \frac{(|F(u)| - |F_a(u)|) \cdot (|F(u)| + |F_a(u)|)}{|F(\mathbf{u})|^2} \quad 10$$

By approximating the Wiener filter as

$$\Theta_{WF}(\mathbf{u}) \approx \frac{|F(\mathbf{u})| - |F_a(\mathbf{u})|}{|F(\mathbf{u})|} \cdot \frac{|F(\mathbf{u})| + |F_a(\mathbf{u})|}{|F(\mathbf{u})|} \equiv \Theta_{ABSF}(\mathbf{u}) \cdot \frac{|F(\mathbf{u})| + |F_a(\mathbf{u})|}{|F(\mathbf{u})|} \quad 11$$

This gives

$$12 \quad \Theta_{ABSF}(\mathbf{u}) = \frac{|F(\mathbf{u})| - |F_a(\mathbf{u})|}{|F(\mathbf{u})|}$$

with the relationship between the two filters as:

$$(13) \quad \Theta_{ABSF}(\mathbf{u}) \approx \Theta_{WF}(\mathbf{u}) \cdot \frac{|F(\mathbf{u})|}{|F(\mathbf{u})| + |F_a(\mathbf{u})|}$$

The asymptotic behavior of the two filters are the same, with the filters being close to equal when $F(\mathbf{u}) >> F_a(\mathbf{u})$. The background subtraction filter results in a little stronger filtering in the intermediate regime, but as will be seen from the examples shown in Section 4, the two filters give comparable results.

In the above the filter $\Theta_{absf}(u)$ has been derived from the filter $\Theta_{wf}(u)$. In fact the historical background of the filter $\Theta_{absf}(\mathbf{u})$ is that it was introduced and used at the NCEM for a number of years before its relationship to the Wiener filter was understood. The background subtraction filter is incorporated into the NCEM image processing extension (Kilaas & Paciornik, 1995) for the software package Digital Micrograph (Gatan, Inc.). The ABSF filter should not be confused with another filter of similar name (Sattler & O'Keefe, 1987) which is a subjective filter, requiring the user to identify reflections associated with the crystalline material and to construct a background by replacing strong peaks with Fourier components whose phases are random and whose amplitudes are taken from regions near the peaks.

The average background subtraction filter has a simple geometric description from which it derives its name. If one writes the result of the signal as

$$(14)\qquad F_c^{est}(\mathbf{u}) = \Theta_{ABSF}(\mathbf{u}) \cdot F(\mathbf{u}) = \frac{|F(\mathbf{u})| - |F_a(\mathbf{u})|}{|F(\mathbf{u})|} \cdot F(\mathbf{u})$$
$$= (|F(\mathbf{u})| - |F_a(\mathbf{u})|) \cdot \hat{F}(\mathbf{u})$$

then the vectorial representation of the resulting Fourier component is obtained from the Fourier component of the recorded signal minus a vector in the same direction whose length is given by the estimated amplitude of the signal $F_a(\mathbf{u})$. This is illustrated in figure 1.

3. Estimating the background

The signal from the amorphous background is spread out in reciprocal space with an amplitude that depends on the radial frequency |u|, and is independent of direction, except for random variations. Thus in order to estimate the amplitude |Fa(u)| it is necessary to find the average Fourier amplitude for a given radial frequency |u|, excluding in principle any contributions from the signal Fc(|u|). However, the signal Fc(u) is localized in reciprocal space and for any given radius |u| more than a few pixels away from the center, Fc(u) only corresponds to a limited number of sampling points along the curve |u| = constant. In addition, the signal Fc(u) usually is much larger than Fa(u) and can often be distinguished from Fa(u) based upon its magnitude. It turns out that the results obtained by applying the filters are not very sensitive with respect to how the average of the background signal is computed. Estimating the background signal by averaging over the entire circle |u|=constant, often give good results even though the estimated magnitude of the background will be higher than the actual value when crystalline peaks are present along the circle. However, the background can be better estimated from the histogram of the amplitudes. The histogram will separate the background signal from any large crystalline contribution, with the majority of the pixel values coming from the background signal due to the amorphous material.

Thus the average amplitude $|F_a(|\mathbf{u}|=\text{constant})|$ can be estimated by

determining the histogram of amplitudes along the circle $|\mathbf{u}|$ = constant and computing the mean of the histogram, excluding outliers which will reduce any contribution from the signal $F_a(\mathbf{u})$. A typical histogram of Fourier amplitudes for |u|=constant when crystalline peaks are present is shown in figure 2. In this case there are only a relatively small number of pixels that correspond to the crystalline material and the histogram distribution becomes bi-modal with the majority of the pixels contributing to the lower cluster of peaks.

The average background signal is computed for every radius in order to derive the estimate for $|F_a(\mathbf{u})|^2$. Subsequently $|F_C(\mathbf{u})|^2$ is estimated from

$$(15) \quad |F_c(u)|^2 \approx |F(\mathbf{u})|^2 - |F_a(\mathbf{u})|^2 \approx |F(\mathbf{u})|^2 - \langle |F_a(|u|)|^2 \rangle$$

in order to compute the filters $\Theta_{wf}(\mathbf{u})$ and $\Theta_{absf}(\mathbf{u})$

4. Examples of application

In order to illustrate and to compare the effect of the two filters, they have been applied to a HRTEM image of Zeolite. Zeolite damages very rapidly under the electron beam and it is difficult to obtain images from the crystalline phase before the material becomes amorphous. Figure 3 shows the experimental HRTEM image and figure 4 shows the corresponding Fourier transform. The crystalline signal is clearly visible as sharp peaks against a background that varies with radial frequency |u|. The computer automatically analyzes the variation of intensities in the frequency domain in order to determine the estimated background. Figure 5 shows the estimated background intensity as a function of radial frequency |u|. At low frequencies the background becomes very sensitive to random variations as it is estimated from very few data

points. Although the crystalline signal can be seen to contribute slightly to the estimated background at some frequencies, the corresponding error in the estimated background has very little effect on the functionality of the filters. The actual peak intensities due to the crystalline material is typically 1-2 orders of magnitude greater than the background intensity.

After the background has been estimated, the computer automatically constructs either the filter $\Theta_{wf}(\mathbf{u})$ or $\Theta_{absf}(\mathbf{u})$. By applying the filters $\Theta_{wf}(\mathbf{u})$ and $\Theta_{absf}(\mathbf{u})$ on the Fourier transform of the original data, figure 4, the component of the signal due to the amorphous material is reduced. Figure 6 and figure 7 show the effect of the filters $\Theta_{wf}(\mathbf{u})$ and $\Theta_{absf}(\mathbf{u})$ respectively on the Fourier transform of the original data. The corresponding images which are obtained by the inverse Fourier transform are shown in figure 8 and 9. The images shown in figure 8 and figure 9 are only estimates of the signal produced by the crystalline part and should not be interpreted as the true signal. The crystalline region of the image looks less noisy and it is easier to identify the image produced by the individual unit cells. The filtered images also appear to preserve the shape of the crystalline material and careful examination of the images shows that there are no features in the filtered images that cannot also be identified in the unprocessed image. Finding regions that show the Zeolite structural unit is much easier starting from the filtered images rather than the original data. By extracting a region containing a structural unit one can calculate the cross-correlation coefficient between the template and the filtered image for every position of the template with respect to the image (Paciornik et al., 1996). This gives a new image of values in the range of -1 to +1 with values close to 1 implying regions that are similar to the template. Figure 10 shows the cross-correlation image with the template inserted. These positions can be used to extract regions from the original data to obtain an averaged image of the unit cell. The average template is shown in figure 11a. For comparison the average template obtained from the Wiener filtered image is shown in figure 11b.

5. Discussion

Many will argue that the only processing that safely can be performed is averaging successive images of the same region, ensuring that the only variant is noise, since even averaging similar regions from the same micrograph may introduce artifacts caused by systematic structural variations and systematic errors in imaging parameters.

With any type of filtering, there is a danger that the filtered image no longer represents the raw data in any meaningful way. However, there are many times when one makes intelligent guesses with respect to the nature of the recorded signal and the imaging conditions and proceeds accordingly. The underlying assumptions in this work is that the Fourier transform of the recorded image can be treated as a sum of two uncorrelated signals. Neither statement will strictly be true, but as a starting point, they are presumed to hold to a first approximation. The results obtained by applying the two filters need to be compared to the actual data to ensure that they are free of obvious artifacts. Obvious artifacts such as extension of crystalline features into non-crystalline regions indicate that the filter is inappropriate and often this is caused by too much filtering resulting from an over-estimation of the noise. Since the signal ("crystalline image") is unknown for an experimental image, it is difficult to assess quantitatively the effects of the filters on experimental data. However, for simulated data with known crystalline material and known noise, the results can be compared quantitatively. Tests show that depending on the size of the image and how the estimate of the noise relates to the true noise, the improvements in signal to noise lie in the range 3-7 (Marks, 1996).

Both the Wiener filter and the average background subtraction filter are created automatically from the data by estimating the power spectrum of the signal due to the amorphous background, assuming that, apart from random variations, the signal only varies with radial frequency |u|. The Wiener filter gives the optimal estimator for the signal $Fc_{(}u)$ in a least squares sense. The

background subtraction filter results in a slightly heavier damping of frequencies where the "signal" (F_c(**u**)) and "noise"(F_a(**u**)) are comparable in magnitude, but otherwise gives results very similar to the Wiener filter. Given the availability of both filters, it is clear that the Wiener filter should be chosen over the background subtraction filter. However, because the background subtraction filter has been in use by many scientists for a number of years, it is important to provide information on its applicability and its relationship to the Wiener filter. Both filters operate automatically. No choice is offered to the user, and no further assumptions are made, apart from those built into the algorithm. If properly applied, they can assist in extracting useful information from HRTEM images when the image consists of crystalline material in conjunction with an amorphous layer.

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Figures

Fig. 1 The Fourier component F(u) of the recorded signal depicted as the sum of the components $Fa_{(}u)$ and $Fc_{(}u)$. Using the filter $\Theta A_{BSF(}u)$ to estimate the desired signal $Fc_{(}u)$ results in $Fc_{e}^{st(}u)$

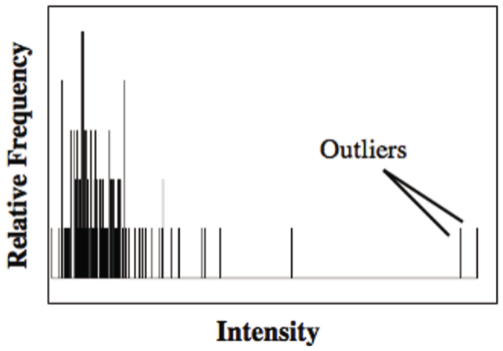


Fig. 2. Histogram of amplitudes along a circle in frequency space where |u| = constant. Most pixel values are associated with the background signal and contribute to the left cluster of peaks in the histogram. The peaks lying to the far right are associated with a signal coming from the crystalline material and can easily be separated from the background when estimating the average background amplitude.

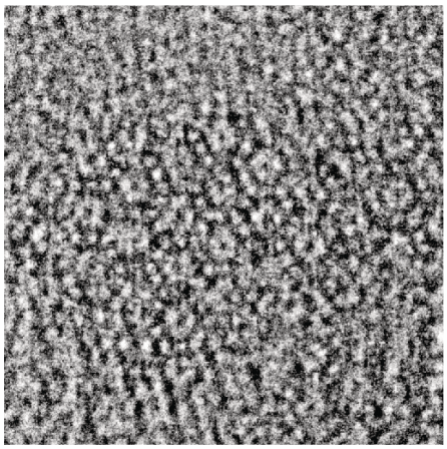


Fig. 3. Experimental image of Zeolite which is rapidly being damaged by the electron beam. Central portion shows a part that is still crystalline. Because of the large amount of amorphous "noise" it is difficult to obtain meaningful data from the crystalline area.

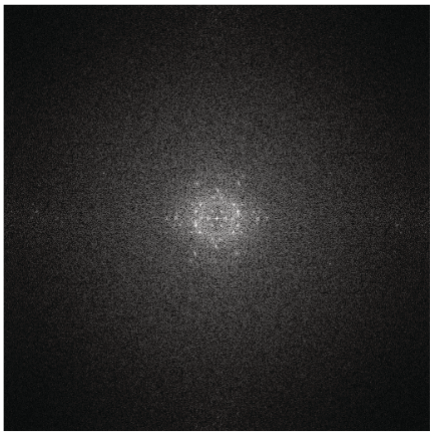


Fig. 4. Fourier transform of the image shown in figure 3. The crystalline region of the image contributes to the localized sharp peaks, while the amorphous material contributes to the smeared out background.

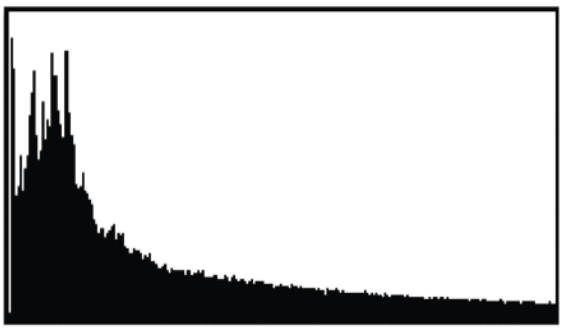


Fig. 5. The estimated background intensity as a function of radial frequency |u| obtained from the Fourier transform shown in figure 4.

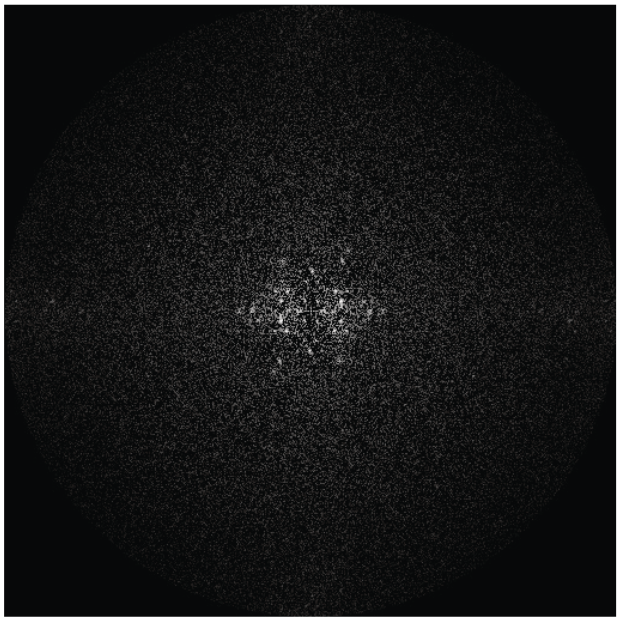


Fig. 6. The result of applying the Wiener filter on the Fourier transform of the original data. The background is reduced compared to the unprocessed data.

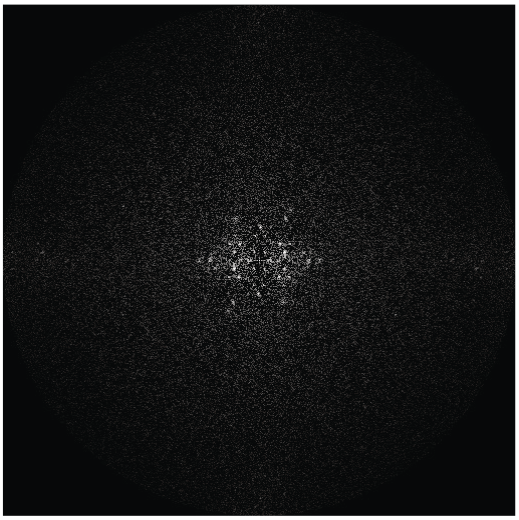


Fig. 7. The result of applying the background subtraction filter on the Fourier transform of the original data.

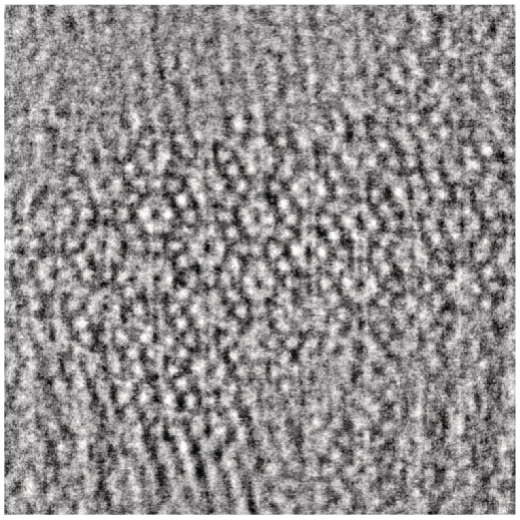


Fig. 8. The resulting image obtained from performing the inverse Fourier transform of the data obtained by the use of the Wiener filter

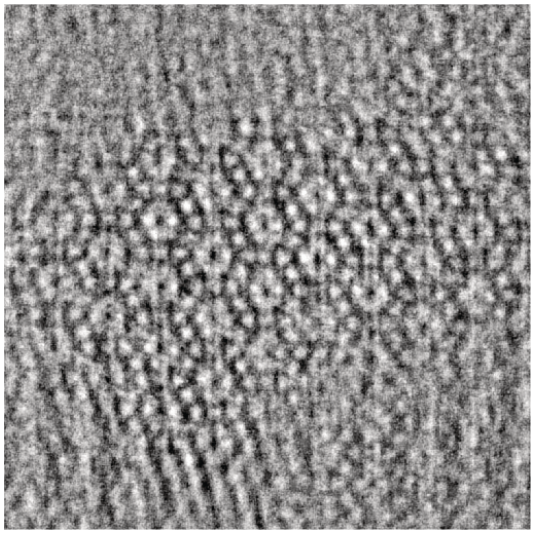


Fig. 9. The resulting image obtained from performing the inverse Fourier transform of the data obtained by the use of the background subtraction filter.

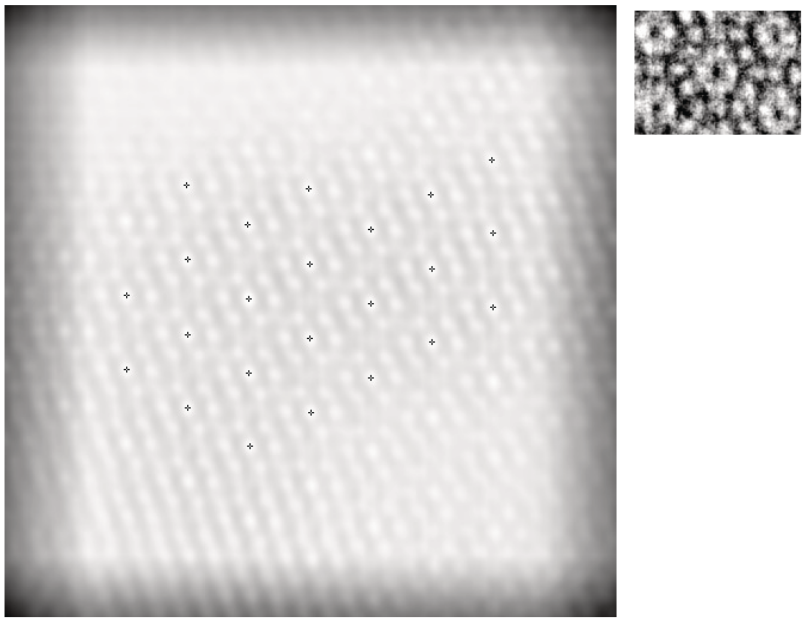
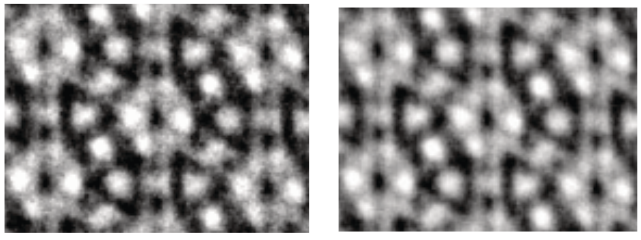


Fig. 10. Image showing the cross-correlation coefficients between a template containing the image of a Zeolite unit cell and the Wiener



filtered image. The template has been extracted from the filtered image. The crosses mark the positions of regions that are similar to the template image.

the filtered image (Fig. 10).

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