# Diffraction Between the Spots: Scanning Electron Diffraction of Beamsensitive Disordered Materials



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# Declaration

I hereby declare that except where specific reference is made to the work of others, the contents of this dissertation are original and have not been submitted in whole or in part for consideration for any other degree or qualification in this, or any other university. This dissertation is my own work and contains nothing which is the outcome of work done in collaboration with others, except as specified in the text and Acknowledgements. This dissertation contains fewer than 60,000 words, including summary/abstract, tables, footnotes and appendices, but excluding table of contents, photographs, diagrams, figure captions, list of figures/diagrams, list of abbreviations/acronyms, bibliography and acknowledgements.

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# Abstract

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Crystallography loves order, but many organic materials are disordered and only partially, if at all, crystalline. Nonetheless, these disordered materials are functionally complex and require characterisation. Their lack of crystallinity poses not only fundamental questions about how to best describe such structures, but also blunts the typically precise tools of crystallography in describing atomic order. Yet, disordered structures not only have structural characteristics, but also complex micro- and nanostructures, defects, and phase distributions. Much of this can be gleaned from diffraction spots, but even more of the information lies in-between the spots in a diffraction pattern. The diffracted intensity outside of the diffracted spots contains the necessary information to not only obtain structural information, but to also characterise the disorder present. Recent developments in transmission electron microscopy (TEM) have enabled the collection of numerous spatially separated diffraction patterns across a specimen, and when combined with computational tools opened a new space for the crystallographic analysis of disordered materials.

In scanning electron diffraction (SED), a two-dimensional diffraction pattern is acquired at each probe position in a two-dimensional scan across a specimen. This four-dimensional (4D) data set can be extensively manipulated post-acquisition using computational tools, enabling the acquisition of multiple correlated conventional TEM experiments at once. Yet this is just the tip of the iceberg. Within such a 4D data set, any pixel can be correlated with another, even ones that may seem at first glance unphysical. In this work, such computational microscopy is applied to SED data to characterise the structure of disordered materials. In this work, the requisite computational methods are developed and applied to extract crystallographic information in metal-organic frameworks through pair distribution function analysis, in pharmaceutical cocrystals through nanoscale twist characterisation, and in polymers through semi-crystalline variance and correlation analyses. As all information is contained within a single scan, all of this analysis is done at doses low enough to avoid irradiation damage in the probed beam-sensitive samples.

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A schematic of a transmission electron microscope, as imagined by Midjourney, an artificial intelligence program that creates images from text prompts.

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# Nomenclature

The following nomenclature is used throughout this thesis, except where otherwise defined.

### **Physical Constants**

- c Speed of light
- *e* Elementary charge
- *h* Planck constant
- *h* Reduced Planck constant
- $m_0$  Electron rest mass

### Symbols

- $\alpha$  Convergence semi-angle
- $\chi(\mathbf{k})$  Aberration function
- $\lambda$  Electron wavelength
- $\varphi(s)$  Reduced intensity
- $\psi(r)$  Electron wavefunction
- $d_{\rm hkl}$  Interplanar spacing for lattice plane hkl
- *FT* Fourier transform
- g(r) Pair-distribution function
- G(r) Reduced pair-distribution function
- $I_g$  Intensity of diffracted beam g.
- **k** Reciprocal space vector, interchangeable with *s*
- k<sub>0</sub> Incident wave-vector
- **k**' Scattered wave-vector
- **Q** Reciprocal lattice wave-vector
- *r* Real space vector
- *s* Reciprocal space vector, interchangeable with **k**
- *s*<sub>g</sub> Excitation error

- *S*(*s*) Structure function
- V(r) Sample Coulomb potential
- Var(s) Normalised variance
- Z Atomic number

#### **Abbreviations/Acronyms**

- 2D Two-dimensional
- 3D Three-dimensional
- 4D-S(P)ED Four-dimensional scanning (precession) electron diffraction
- 4D-STEM Four-dimensional scanning transmission electron microscopy
- BF Bright-field
- BSS Blind-source separation
- CBED Convergent-beam electron diffraction
- DF Dark-field
- EDS Energy-dispersive X-ray Spectroscopy
- ePDF Electron reduced pair distribution function
- ICA Independent component analysis
- LRO Long range order
- MRO Medium range order
- NBED Nanobeam electron diffraction
- NMF Non-negative matrix factorisation
- PCA Principal component analysis
- PDF Pair distribution function, sometimes reduced pair distribution function
- RDF Reduced pair distribution function
- SED Scanning electron diffraction
- SNR Signal-to-noise ratio
- SPED Scanning precession electron diffraction
- SRO Short range order
- STEM Scanning transmission electron microscopy

- SVD Singular value decomposition
- TEM Transmission electron microscopy
- VDF Virtual dark-field
- XPDF X-ray reduced pair distribution function
- XRD X-ray diffraction

# 1 Structure of Organic Materials

The first organic compound was synthesised by Friedrich Wohler in only 1828, but now new organic structures discovered outnumber their inorganic brethren by nearly ten to one annually. Organic materials differ fundamentally from inorganic ones, such as metals or ceramics, by having basic packing units that are large and non-spherical. For inorganic materials, even though single atoms are not spherical due to the shape of atomic orbitals, they can nonetheless typically be approximated as a sphere in metals and inorganic compounds. Such spherical atoms usually pack in a cubic, tetragonal, orthorhombic, or hexagonal crystal systems. However, for organic molecules, a typical packing 'unit' contains multiple atoms, held together by strong covalent intramolecular bonds rather than the weak intermolecular bonds (dominated by van der Waals interactions) that govern molecular packing. These molecules are highly non-spherical, with regions of positive and negative charge pooling, and as a result pack in complex ways. It is therefore no surprise that molecular crystals are less symmetric, mainly monoclinic, triclinic, and orthorhombic. Furthermore, they typically have large unit cells with tens or hundreds of atoms and varied types of translational disorder. These types of disorder include partial crystallinity, twisting, defects, and high degrees of polymorphism. Vitally, they contribute to the remarkable functional complexity that organic structures have. This chapter describes some of the common types of structural and functional properties found within organic materials, as well as defects frequently found.

# 1.1 Order and Disorder

Order and periodicity are frequently interpreted synonymously in crystallography. However, they are not, and it is worth discussing the difference. Periodicity refers to a regularly repeating geometric element in a structure; the repeating nature of unit cells that form a macroscopic crystal is a case of periodicity. In comparison, order is less strict in its definition. It is present both in structures that are periodic and non-periodic. At the single atom level, it is clear a certain amount of order must be present in all atomic structures. After all, atoms cannot pack closer than a certain bonding distance. However, the extent to which such order extends in space is what separates crystals and amorphous materials. Crystals are both periodic and ordered, whereas amorphous materials have order only on the ångström scale and are not periodic. In between these two lies a wide variety of structural correlations, such as materials with medium-range order [1], correlated disorder [2], temporal disorder [3], and semicrystalline materials [4].

When discussing order on the length of one or two atomic separations, on the length scales of <0.5 nm, there is a tendency to refer to short-range order (SRO). Even amorphous materials exhibit short-range order, for example silica is composed of interlinked SiO<sub>4</sub> tetrahedra. A schematic example is shown in **Figure 1.1**. SRO can be probed using techniques such as pair distribution function (PDF) analysis. When this order is correlated over large distances, i.e. when the structure is also periodic, the structure is said to have long-range order (LRO), shown in **Figure 1.1a**. The benefit of LRO is that the structure can be probed easily using diffraction methods; the problem is heavily overconstrained [5]. Crystals are defined by their exhibiting of LRO. It is no surprise then, that most crystallography probes crystalline materials. In contrast, in non-crystalline materials, there is no LRO, and hence their structure remains exceptionally difficult to probe exactly. The absence of diffraction methods and lack of a periodic structure means that they are best described using distributions rather than exact structures. For a more thorough reading on order and disorder, the reader is guided to [6].

In between the two types of order sits medium range order (MRO). This refers to the structural correlations of atomic positions on length scales between 0.5 and 3 nm. It has been shown that many "amorphous" systems, such as clusters [7], [8], metallic glasses [9], and amorphous silicon [1] actually have significant MRO. However, probing MRO in materials has proved to be the most challenging of all these ranges of structural order, it is too irregular for bulk X-ray methods and too spacious for typical diffraction methods. MRO requires highly specific tools to probe, such as fluctuation electron microscopy [1], which is further discussed in Section 3.2.



**Figure 1.1**. Long-range and short-range order. (a-b) A 2D representation of long-range order in (a) a crystal formed of a single type of atom, and (b) of two types of atoms. (c-d) A 2D representation of an amorphous structure that only exhibits short-range order in (c) a crystal formed of a single type of atom, and (d) of two types of atoms. Adapted from [10].

# 1.1.1 Types of Disorder

Disorder refers to the absence of a certain type of order or periodicity, and is often directly correlated to the functional properties of a material [2]. That a material is disordered does not mean that it is without order. For example, a dislocation is a linear defect, a 1D line of disorder within an otherwise ordered lattice [11]. In contrast, amorphous glasses are disordered in all three dimensions, such that the structure is not periodic in any orientation. Types of disorder are defined both by their dimensionality, their size, and their frequency. Furthermore, disorder in a lattice may be correlated or uncorrelated [2]. In contrast, in random disorder, no such relationship exists. Examples are shown in **Figure** 

**1.2**. In practice, most disorder sits somewhere in the middle, with a tendency for random disorder when the density of defects is low, and a tendency for either correlated disorder or amorphous structure when the density is high.



**Figure 1.2**. Order, correlated disorder, and random disorder. (a)-(c) Atomic positions for water molecules in 2D ice, for ordered (a), correlated disordered (b) resulting in no hydrogen atoms being directly next to each other, and random disordered (c) structures. (d)-(f) The resultant simulated diffraction patterns. (g)-(i) The resultant reduced pair distribution functions g(r). Adapted from [2].

Notably, relatively little work has been done to describe types of correlated disorder. It tends to arise particularly in systems with either geometric frustration: an incompatibility between the ideal order and its geometric components, and in systems with highly underconstrained configurations, such as in ice where the number of possible molecular conformations is high compared to the number of ordered conformations. This geometric frustration, covered further in Section 1.4, is of particular interest in organic compounds due to the highly non-spherical molecules they are composed of. Different types of correlated disorder can be probed directly using diffraction methods, especially if

high-resolution 3D diffraction data of diffuse scattering can be acquired, such as with synchrotron Xrays [12], neutrons [13], or electrons [14]. Specialised methods such as 3D-ΔPDF [15] exist to directly display correlations within this disorder, and has been applied to solve the complex structures such as Prussian blue analogues [16]. In general, the pair distribution function (PDF) has been found to be a good general probe of both correlated and uncorrelated disorder [17]. Overall, disorder fundamentally changes material properties and presents a highly challenging extension to conventional crystallography, but one that can be probed with the correct characterisation methods.

# 1.1.2 Partially Crystalline Materials

Materials that exhibit both crystalline and non-crystalline states frequently also have both types present at once. Such materials are typically called some combination of semicrystalline, partially crystalline, and disordered, depending on the extent and type of non-crystallinity. Such materials also frequently exhibit local variation between more and less ordered states, such as in polymers, where regions of 'crystalline' polymer and 'amorphous' polymer typically exist, typically spatially separated [18]. Partial crystallinity can take many forms. Some structures have small nanocrystals in the midst of less ordered regions [19]–[21]; this is typical in many organic structures, and an example is shown in **Figure 1.3**. Others, such as polymers, permanently inhabit an intermediate ground where no region is highly crystalline or entirely amorphous. Typically, order is present only in a specific direction, such as chain-chain stacking for polymers, but the structure is otherwise disordered. Analysing partially crystalline materials requires a combination of crystalline and non-crystalline analysis tools. The latter pose a particular problem due to the nanostructure problem, detailed below in Section 1.2.



**Figure 1.3**. An amorphous-crystalline composite. (a) Three-dimensional tomography and the corresponding STEM-EDS mapping of a crystalline-amorphous MOF composite, highlighting the local presence of Al-rich regions. (b) A correlated scanning electron diffraction (SED) map showing crystallinity and energy dispersive X-ray spectroscopy (EDS) map showing compositional variation. The Al-rich regions are directly correlated with crystalline regions in SED. Adapted from [22].

# 1.2 The Nanostructure Problem

Characterising structure and understanding structure-property relationships is the fundamental goal of materials science. Simon Billinge coined the term 'Nanostructure Problem' [23] to refer to the challenge of characterising highly defective and non-crystalline materials. It is often forgotten that the reason one can determine the three-dimensional atomic coordinates of atoms in a crystal with a precision of 10<sup>-3</sup> Å is due to the periodicity of crystal structure over a much larger distance than a single unit cell. Solving the phase problem is only possible for highly symmetric, over-constrained structures, namely crystals [24]. Without long-range structural order, our ability to locate the positions of atoms in a material and correlate those to the macroscopic properties of the material is impossible as we cannot locate the atoms themselves. Furthermore, it is vital that the sample is resistant to probing radiation, which for organic materials is rarely the case [25]. Therefore, the characterisation of non-crystalline and semi-crystalline materials requires methods that are distinct from those used for crystalline structure solution [26]. Such methods, like the pair distribution function (PDF) are ensemble methods and do not aim to locate atoms, but rather describe the common relationships between atoms in the material.

Of course, if such non-crystalline materials were not scientifically and industrially relevant, the problem wouldn't be a problem. However, phases without long-range structural order are increasingly part of modern materials, from metallic [27] and inorganic [28] glasses, partially disordered materials [6], to protein-based materials [29], to nanomaterials dominated by interfaces [30]. Hence, the problem of solving disordered and non-crystalline structures is not only necessary, but extremely challenging. Nonetheless, advancements have been made in the last decade, particularly regarding detectors [31] and microscope functionality, and electron microscopy is now at the forefront of characterising these disordered structures at the nanoscale.

# 1.3 Organic Crystal Structure

Can you predict a crystal structure from just its constituent molecules? Generally, the answer turns out to be no. A molecule's crystal structure is not a simply predictable property [32], but rather dependent on the interplay between the types and positions of functional groups it contains. There are general themes, such as Kitaigorodskii's close packing principles [33], which dictate that in general the size, shape, and the minimisation of empty space within a molecular crystal dominate the expected crystal structure. However, such effects are modulated by the chemistry of the packing species [34].

**Table 1.1**. Frequency of different crystals systems for the 1,172,316 structures for which the space group is fullydefined in the CSD as of 1 Jan 2022 [35].

System	Structures	%CSD
Monoclinic	604,866	51.6
Triclinic	306,368	26.1
Orthorhombic	198,210	16.9
Tetragonal	25,601	2.2
Trigonal	23,584	2.0
Cubic	7,335	0.6
Hexagonal	6,352	0.5

In general, organic compounds are monoclinic, triclinic, or orthorhombic. Shown in **Table 1.1** are the most common crystal systems found in the Crystal Structure Database (CSD) [36]. 94.6% of documented structures fall into one of these three crystal systems, with monoclinic being the most common at 51.6%. The most common space groups are further shown in **Table 1.2**, with P2<sub>1</sub>/c and P1 making up 59.3% of the documented structures. This can be rationalised by Kitaigorodskii's principles. The highly non-spherical single molecules don't pack perfectly on top of each other, suffering from geometric frustration, detailed further below. As a result, the next layer is displaced laterally or at an angle with respect to the one below. Example unit cells of a monoclinic and orthorhombic organic crystal are shown in **Figure 1.4**. These molecules pack in a so-called herringbone structure, but each molecule is also at an angle, such that rotating the crystal reveals orientations down which single 'lines' of molecules align (**Figure 1.4**b,d). This herringbone stack is also frequently slightly tilted,

frequently resulting in a monoclinic structure if orientations are alternated, while in the orthorhombic structure in **Figure 1.4**c two of these monoclinic structures are effectively stacked side-to-side with opposite orientation (such that [100] and [ $\overline{1}$ 00] orientations are parallel), resulting in an overall unit cell that is orthorhombic.



**Figure 1.4.** A monoclinic paracetamol unit cell down the (a) [100] and (b) ca. [210] axes. [001] points to the right. An orthorhombic theophylline unit cell viewed down the (c) [001] and (d) [011] axes. [100] points to the left.

**Table 1.2.** The 10 most common space groups among the CSD structures for which the space group is fullydefined on 1 Jan 2022 [35]. No other group makes up more than 1% of structures. Space groups 1-2 are triclinic,3-15 are monoclinic, and 16-74 are orthorhombic.

Rank	Space Group No.	Space Group	Structures	%CSD
1	14	P21/c	400,548	34.2
2	2	ΡĪ	293,933	25.1
3	15	C2/c	97,014	8.3
4	19	P2 <sub>1</sub> 2 <sub>1</sub> 2 <sub>1</sub>	81,425	7.0
5	4	P21	60,123	5.1
6	61	Pbca	37,779	3.2
7	33	Pna2 <sub>1</sub>	15,983	1.4
8	9	Сс	12,150	1.0
9	62	Pnma	11,846	1.0
10	1	P1	11,496	1.0

## 1.3.1 Disorder in Organic Materials

Numerous types of disorder have been observed in organic crystals, such as point defects [37], [38], dislocations [39], [40], striations [41], crystalline grain boundaries [42], amorphous regions [43], voids [44], [45], and twisting [46], [47]. In addition, polymers are typically semicrystalline, as discussed further below in Section 1.5. Defects in organic materials have been correlated to functional properties, particularly in organic semiconductors, where the complete absence of crystallinity and semi-crystallinity has been correlated to reduced electronic conductance [48]. This is likely associated with defect-related trap states within the band gap [49], which have been shown to inhibit electronic conduction. However, a characterisation gap for such defect structures still exists, with little to no direct imaging of such defect structures. Instead, properties and structures are typically inferred from macroscopic and X-ray synchrotron studies due to the irradiation sensitivity of most organic structures.

# 1.3.2 Microstructure in Organic Materials

Organic materials have microstructures with a large variety of length scales, from large 100 µm crystals, to structures with nanodomains on the 10 nm scale. An example of the length scales of defect and microstructures typical in organic semiconductors are shown in **Figure 1.5**. These microstructures can be further complicated by the presence of multiple phases. Phase segregation and separation are common, as while some molecules form cocrystals [50], most are mutually immiscible. This is accentuated in polymer-molecule composites, as polymer strands do not naturally fit in molecular crystals. As a result, the sample microstructure is highly dependent on the processing method. For example, spin-coated films tend to show strong planar chain orientation anisotropy [51], while injection-moulded polymers have complex flow anisotropy [52]. Overall, the microstructure of a given sample is unique, but important, as microstructural features have been correlated with mechanical [53], functional [43], and degradation [54] properties .



**Figure 1.5**. Scale of morphological features in organic semiconductor devices. Top row: polymer and fullerene heterojunction. Bottom row: functional molecular crystal. Each dotted square is the area from which the zoomed in square to the left comes from. Adapted from [43].

# 1.4 Geometric Frustration

The complex shape of molecules results in an inevitable misfit in packing. This misfit is typically called geometric frustration<sup>1</sup> [55] and it occurs in any constituent materials where atoms cannot pack in an ideal organised manner, such as in molecular crystals [47], metallic glasses [56], proteins and other soft structures [57] or around defects [58]. Geometric frustration carries an energy cost, as it prevents atoms from sitting as close to each other as they would ideally like to and can be resolved through distortions to ideal packing, such as local strains. These distortions tend to fall into two categories, local and non-local distortions.

Local distortions refer to resolutions in which all atoms or molecules distort in an equivalent manner, resulting in a changed but uniform structure. For example, atomic rearrangements in bent crystals without defects exhibit such phenomena [59], but even tetragonal distortions in perovskites [60] could be interpreted as a resolution to a frustration in the cubic phase. Such local distortions are often referred to as non-cumulative, as the per volume energy of the distortion is independent of volume; the energy of the structure is not dependent on its size.

Non-local distortions are ones in which the packing units do not deform uniformly. Such non-local compromises often occur for small structures and due to the non-uniformity of the distortion, such structures do not have a repeating unit cell. Examples of non-local distortions are twisting in molecular crystals [46], icosahedral reordering in metallic glasses [56], and, in a way, the icosahedral nanocrystals without a periodic unit cell that are seen for many metals. Non-local distortions are often called cumulative, with the distortion increasing with the size of the structure, resulting in the per volume energy also scaling with size. Such distortions are therefore typically observed on the nanoscale.

Already in 1929, Ferdinand Bernauer discovered that a large fraction of small organic compounds could form twisted crystals [61], predicting that over 25% of such compounds could form twisting structures. Recent work suggests that the 25% might be an underestimate [62], [63]. Twisting on the nanoscale seems to be not only common, but akin to the non-periodic nanocrystals that form for small metallic assemblies, a standard method for nanoscale structures to re-order to reduce energy. Such structures twist more slowly as the structure gets larger [47], (**Figure 1.6**) tending to a zero twist rate for macroscale crystals.

<sup>&</sup>lt;sup>1</sup> As opposed to frustration with geometry, also common in crystallography.

### 1.4.1 Twisting Structures

Twisting has been observed in a wide variety of organic structures, such as cellulose [64], aspirin, urea, and many more [63]. Perhaps most commonly, the spherulite structures seen in polymers (see Section 1.5) also involves the twisting of the constituent polymer chains. These twists can take multiple forms depending on the origin of the geometric frustration [47], a few of which will be discussed below.

#### 1.4.1.1 Uniaxial Twist

The simplest twisting structures are ones with a uniaxial twist (**Figure 1.6**a). Such structures have a helical symmetry along the fibre axis and a twist rate t. In such cases, the elastic distortion increases towards the edges of the structure, and to ensure that the local twist rate at the edges doesn't exceed a structural limit, the fibre will twist more slowly as its diameter increases. The effect of this is, counterintuitively, that the greatest elastic energy density is in the centre of the fibre, rather than the edges [46]. Modelling of such structures gave a prediction for the twist rate t [46] of

$$t = \frac{6G_t}{G_s} \frac{t_0}{A}$$

(1)

where  $G_t$  and  $G_s$  are the twist and shear moduli of the one-molecule wide rod,  $t_0$  the twist rate of the one-molecule wide rod, and A the cross section of the fibre. t is inversely proportional to A, resulting in the pitch  $\rho$  (the 360° rotation period) being proportional to A. Notably, t should approach zero as A increases. This relation usually does not hold for very thin crystals, as those crystals have relatively low shear moduli, resulting in a strained crystal with near zero twist rate.



**Figure 1.6.** Twist geometry. (a) Uniaxial twist of a square cross section structure. Increasing the diameter of the structure decreases the twist rate due to elastic strain. The colour bar is elastic energy density. Adapted from [47]. (b) Twisting in a sheared monoclinic bilayer structure. Adapted from [63].

#### 1.4.1.2 Shear-layered Twist

Not all structures follow this t-A relationship predicted for uniaxial twisting. In fact, the first organic compound with discovered polymorphs [65], [66], benzamide, exhibits a twisting polymorph [67] (**Figure 1.7**) with a twist rate that is not inversely proportional to A. This can be explained through a strained bilayer structure, shown in **Figure 1.6**b. Such structures can arise from an unconventional monoclinic (010) twin (when  $\beta \neq 90^{\circ}$  is between the [100] and [001] directions) in the monoclinic unit cell, subsequently sheared to give an orthorhombic structure. This strain is resolved as a twist rather than as a shear along the typical monoclinic axis, giving rise to a pitch  $\rho$  of [47]

$$\rho = \frac{4\pi h}{3\sin\beta} \tag{2}$$

where h is the bilayer thickness, and  $\beta$  is the monoclinic angle, giving a twist rate of

$$t = \frac{2\pi}{\rho} = \frac{3\sin\beta}{2h}$$
(3)

which notably is inversely proportional to h rather than A. Some thicker structures did not follow this relationship, twisting faster than expected. This was attributed to a multilayer rather than a bilayer

structure. It is worth noting that both uniaxial and shear-induced twist rates are often much lower than expected for very thin crystals, as those crystals have relatively low shear moduli.

# 1.4.2 Diffraction from Twisted Structures

Twisting introduces significant correlated disorder to observed diffraction patterns. Such structures are typically analysed via fibre and helical diffraction [68]. A few examples are given in **Figure 1.7**. This is covered in more detail in Section 2.5.



**Figure 1.7.** Diffraction from a shear-induced twisting crystal of Benzamide. Adapted from [67]. (a) A plate-like Benzamide I crystal. (b),(d) Crystals with low twist rate, nearly straight crystals. (c) Diffraction from a fast twisting crystal.

# 1.5 Polymers

Polymers are formed from repeating a monomer unit along a backbone, forming a polymer chain. This chain is typically linear, although cyclic, branching, and networked alternatives also exist. A huge variety of polymers can be synthesised through complex polymerisation reactions, something that has raised significant scientific and commercial interest in creating tuneable polymer structures. Typically, polymers are first separated into three categories, thermoplastics, elastomers, and thermosets, of which thermoplastics are typically further separated into crystalline and non-crystalline thermoplastics. The polymers in this work are thermoplastics, and so the other two categories will not be covered here, instead the reader is directed to [69].

Thermoplastics are the most common type of polymer and are defined by their ability to melt upon applied heat. They tend to have a linear or branched chain structure. Their meltability means they can be injection moulded, extruded, or spin-coated into complex geometries. Thermoplastics tend to not crystallise perfectly, as their long and highly tangled macromolecules in the liquid state simply do not have enough time or driving force to form crystals. As a result, most 'crystalline' thermoplastics are in fact semi-crystalline, with both crystalline and amorphous regions throughout the specimen. However, many thermoplastics simply do not crystallise, often due to the steric hindrance of their macromolecules, and simply remain amorphous throughout.

Overall, relatively little is known about polymer microstructure; many polymers damage rapidly under the electron beam and their frequent lack of a crystalline unit cell makes their relative characterisation additionally challenging, as noted in Section 1.2.

# 1.5.1 Polymer Chemistry

Polymers are formed through polymerisation reactions of which there are numerous types. The two most common are step polymerisation and chain polymerisation. In step polymerisation, any monomer, dimer, or polymer unit can react with another, while in chain polymerisation an initiator molecule is used to ensure that growth is linear along a polymer chain. In both cases, these monomer units can be highly complex and tuneable, such as for some of the polymer semiconductors in Chapter 6.

# 1.5.2 Polymer Microstructure

Polymer microstructure and morphology have direct consequences for the final properties of polymer samples. However, such microstructure is relatively poorly known. Historically, polymer microstructures were not directly examined using electron microscopy due to beam-induced damage, and so were probed either through bulk methods or after processing such as via heavy metal staining [70]. While staining enabled the characterisation of some microstructural elements, it also caused significant structural damage, and is not suitable for all polymers. Non-electron microscopy methods, such as scanning probe microscopy [71] and light microscopy [72] have also been used to probe polymer microstructures although such methods remain insensitive to the underlying crystallography.

Typically, polymers are described either as amorphous, semi-crystalline, or spherulitic, as shown in **Figure 1.8**. The first two involve local partial chain ordering (or a lack of it) while spherulites are formed by the growth of crystalline lamellae from a melt around a single nucleation point and are somewhat analogous to dendrite growth in metals. These lamellae are typically bounded by amorphous regions. Spherulites showcase strong birefringence and they are relatively brittle compared to other polymers.



**Figure 1.8.** Polymer microstructure examples, (a) of a spherulite<sup>2</sup>, and (b) semicrystalline and amorphous regions.

However, methods that can directly probe the underlying crystallography of polymers with nanometre spatial resolution remain elusive. Recently, scanning electron diffraction (SED) has emerged as a low-dose characterisation technique capable of characterising polymer structure microstructure and crystallography with nm spatial resolution. For polyethylene, which has a 'shish-kebab' structure [73], [74], SED revealed a highly complex multi-twinned, multi-phase microstructure in one of the 'kebab's, far more complex than the simple crystalline lamellar structure previously assumed. Characterisation and orientation mapping of polymer chains can be achieved in SED [75], [76], which benefits from being a low-dose method due to its low convergence angle, and can be combined with cryogenic cooling where possible. In particular, polymer chain stacking, called  $\pi$ — $\pi$  stacking, tends to diffract strongly and can be mapped with relative ease.

<sup>&</sup>lt;sup>2</sup> From Wikimedia Commons (Materialscientist): https://commons.wikimedia.org/wiki/File:Spherulite2.PNG

#### 1.5.2.1 Polymer Unit Cells

Owing to their linear macromolecule structure, a typical polymer unit cell is oriented along the chain direction, with polymer chains packing side-by-side. Multiple chains tend to sit in each unit cell, with a single unit cell containing tens or even hundreds of atoms, depending on the chain and stacking complexity. These unit cells are typically monoclinic, triclinic, or orthorhombic. These spatial arrangements are held together by van der Waals and hydrogen bonding, resulting in highly anisotropic crystals. In general, polymer structures have been determined from either large single crystals (rare) or from semi-crystalline ensemble measurements, such as through XRD. Frequently, preferred orientations can be seen in such analyses, and polymers often display fibre-like diffraction [69].

#### 1.5.2.2 Disclinations

Disclinations are a type of topological defect commonly seen in liquid crystals [77], [78], and similarly expected in solid polymers. A disclination occurs at a location where there is an abrupt change in polymer chain orientation. Numerous types of such disclinations occur, characterised by their sign, their topological charge k and their shape (such as a point or a line). Examples of 2D (point) disclinations are shown in **Figure 1.9**. The topological charge k represents the strength of a disclination, with higher values of k being correlated to larger amounts of and more complex orientation changes at the disclinations can mutually annihilate. For example, k = +1/2 and k = -1/2 can annihilate to produce a defect-free structure. There has been limited direct imaging of disclinations in polymers, although some have been seen in SED [79]. Historically, some evidence has also been seen in stained samples [80], [81]. Overall, the study of disclinations has been limited by the rapid destruction of polymers under the electron beam. Low-dose SED provides an opportunity for further characterisation of such defect structures.


**Figure 1.9**. Examples of disclination structures in 2D nematic liquid crystals. The red arrow indicates defect orientation, and the value of k indicates the topological charge. Reproduced from [78].

# 1.5.3 Polymer Properties

#### 1.5.3.1 Molar Mass and Degree of Polymerisation

The molar mass M of a polymer refers to the expected mass of a single macromolecule and is typically given in g or kg mol<sup>-1</sup>. It is related linearly to the degree of polymerisation x, which is the number of monomer units in a polymer chain, by

$$M = xM_0 \tag{4}$$

where  $M_0$  is the mass of a monomer unit. It is worth noting that the molar mass and degree of polymerisation are distributions rather than single values, typically following an approximate Boltzmann distribution, with a sharp minimum and a long tail.

## 1.5.3.2 Kuhn Length

Polymer macromolecules have stiffnesses that are highly dependent on the monomer structure. To convey polymer backbone stiffness, it is convenient to use the definition of Kuhn length *b* [82]. When considering a polymer chain, it will tend to clump like a ball of yarn. The mean-square end-to-end distance  $< R^2 >_0$  can then be compared with the fully extended size  $R_{\text{max}}$  to give the Kuhn length

$$b = \frac{\langle R^2 \rangle_0}{R_{\max}}$$

(5)

The Kuhn length can be thought of as the length of polymer chain that takes 'steps' in a random walk model [69], shown in **Figure 1.10**. Effectively, the orientation of a segment  $b_1$  and the orientation of the next segment  $b_2$  are uncorrelated if both have a length equal to the Kuhn length. The Kuhn length is a proxy for measuring chain stiffness and can be directly extracted from dynamic light scattering experiments [83], [84].



**Figure 1.10**. A schematic of a random walk model for a polymer. Left: A polymer chain back bone with a monomer length l and a Kuhn length b. The orientation of each Kuhn length segment is independent of the previous. Right: A larger polymer chain. The mean-square end-to-end distance  $\langle R^2 \rangle_0$  is described by the number of Kuhn length segments  $N_b$  and the Kuhn length b.

# 2 Electron Diffraction Theory

In crystallography, most structures are solved using diffraction methods [85]. In diffraction measurements, probing radiation, such as electrons, electromagnetic radiation, or neutrons interact with the electronic and atomic structure of a material. The measured diffracted intensity in reciprocal space is proportional to the Fourier transform (FT) of the autocorrelation function  $FT[\rho(\mathbf{r} + \mathbf{r}')\rho(\mathbf{r})]$ of the three-dimensional electron density distribution  $\rho(\mathbf{r})$  [24]. The difficulty of diffraction analysis is in the so-called 'Phase Problem' [86]. As the measured kinematic intensity depends only on the autocorrelation of the electron density, the phase of the electron density is lost in the diffracted data. Hence predicting the electron density from a given diffracted intensity is a highly nontrivial problem [23]. This is in contrast to the relatively trivial direct problem of predicting the diffracted intensity given an electron density distribution. However, even the forward problem can be difficult in the dynamical scattering limit, although complex modelling methods, such as the multi-slice method, exist [87]. Structure solution methods solve the real-space structure from diffraction measurements. In the crystalline case, this is typically an over-constrained problem [5], often to the degree that a 3D scan can be reduced to a 1D scan and still return a unique solution, such as in the case of powder diffraction. However, with partially crystalline and non-crystalline materials, the opposite becomes true very quickly, these systems are under-constrained, due to the loss of information in the Bragg peaks due to finite size effects [23] and the increasing complexity of the repeat unit. Mathematically, there is no unique solution obtainable from the diffraction data; any solution is degenerate.

Due to their charge, electrons interact with both the atomic nuclei and the electron cloud, in comparison to X-rays that do not interact with the charged nuclei. As a result, the scattering cross-section of electrons is higher by a factor of ca.  $10^6$  [88]. The result of this is two-fold. Firstly, information can be gleaned from much smaller volumes. Electrons can obtain sufficient diffracted signal from Å<sup>3</sup> and nm<sup>3</sup> sized volumes, impossible for X-ray diffraction. Secondly, electron samples must be thin, as the mean free paths of fast electrons through a sample are on the order of 10-250 nm, depending on accelerating voltage and sample composition. All in all, X-ray diffraction is a bulk method, while electron diffraction is a nanoscale method. Furthermore, while spatially localised X-ray methods, such as X-ray nanobeam [89], have been developed at synchrotron beamlines with spatial resolutions of ca. 100 nm, X-rays will not replace electrons for nanoscale characterisation of beamsensitive materials. This is due to the relative elastic and inelastic cross-sections. For the same elastic (diffraction) signal, X-rays generate several hundred times more damage through inelastic scattering [90] than electrons. Therefore, electron scattering is the highest signal-to-noise ratio (SNR) crystallographic characterisation tool available.

It is worth noting that there are non-diffraction methods, such as atomic force microscopy (AFM) [91] and atom probe tomography (APT) [92] that exist for the purposes of solving atomic structures and properties. However, these methods are either restricted in the structural information they can extract (AFM) or their applicability to specific materials and sample geometries (APT).

# 2.1 Electron Diffraction Theory

As electrons in a TEM pass through a sample, they are partially scattered. This scattering can be elastic (without energy transfer) or inelastic (with energy transfer), as well as coherent (preserving phase relationships) or incoherent (not preserving phase relationships). Electron diffraction, the coherent, elastic, structured scattering of an electron beam by a sample, can be modelled as coherent elastic scattering by the static sample Coulomb potential  $V(\mathbf{r})$ .

The relativistic scattering of a high voltage electron beam can be treated as the scattering of electrons with a relativistic mass  $m = m_0(1 - \frac{v^2}{c^2})^{1/2}$ , where  $m_0$  is the electron rest mass, v the electron velocity, and c the speed of light [93]. The electron wavefunction  $\psi(\mathbf{r})$  is the solution to the timeindependent Schrödinger equation by the electrostatic sample Coulomb potential  $V(\mathbf{r})$ 

$$\nabla^2 \psi(\mathbf{r}) + 4\pi (k^2 + U(\mathbf{r})) \psi(\mathbf{r}) = 0$$

(6)

where  $k^2 = \frac{2m}{h^2}$ ,  $U(\mathbf{r}) = \frac{2meV(r)}{h^2}$ , E is the electron accelerating potential, e is the magnitude of the elementary charge. There are no general solutions to this equation, and while rigorous solutions exist for a few well-defined geometries, in general approximate numerical solutions are used to interpret electron diffraction from arbitrary orientations and potentials [94].

## 2.1.1 Born Approximation

The Schrödinger equation can be transformed for the scattering of an incident plane wave  $\psi_0(\mathbf{r}) = \exp(2\pi i \mathbf{k_0} \cdot \mathbf{r})$ , giving the integral form [94], [95]

$$\psi(\mathbf{r}) = \exp(2\pi i \mathbf{k_0} \cdot \mathbf{r}) + i \int \frac{\exp(2\pi i k |\mathbf{r} - \mathbf{r}'|)}{|\mathbf{r} - \mathbf{r}'|} U(\mathbf{r}') \psi(\mathbf{r}') d^3 \mathbf{r}'$$
(7)

where the first term is the incident plane wave  $\psi_0(\mathbf{r})$  and the integral represents the scattered wave  $\psi_g(\mathbf{r})$  by the potential  $U(\mathbf{r}')$ . In the first-order Born approximation, the scattering is assumed to be weak, such that wavefunction  $\psi(\mathbf{r}')$  inside the integral is approximately equal to the incident wave, i.e.  $\psi(\mathbf{r}') = \exp(2\pi i \mathbf{k_0} \cdot \mathbf{r})$ . This approximation is equivalent to saying that only the incident wave contributes to scattering and is often called the kinematical approximation.

# 2.1.2 Kinematical Electron Diffraction

Kinematical electron diffraction can be further refined in terms of the Fourier transforms of the Coulomb potential  $V(\mathbf{r})$  [94]. For an arbitrary arrangement of atoms, the total potential  $V(\mathbf{r})$  can be written as a convolution of the atomic potentials  $V^a(\mathbf{r})$  and delta functions describing their positions:

$$V(\mathbf{r}) = \sum_{i} V_{i}^{a}(\mathbf{r}) * \sum_{j} \delta\left(\mathbf{r} - \mathbf{R}_{i,j}\right)$$
(8)

where  $R_{i,j}$  is the atomic position of the *j*th atom of type *i*.

The scattered wave amplitude  $\phi(\mathbf{k}')$  for scattering vector  $\mathbf{k}' = \mathbf{k} - \mathbf{k_0}$  is recovered from the Fourier transform of this potential

$$\phi(\mathbf{k}') = \sum_{i} f_{i}(\mathbf{k}') \sum_{j} \exp(-2\pi i \mathbf{k}' \cdot \mathbf{R}_{i,j})$$
(9)

which is the sum of the atomic scattering factors  $f_i(\mathbf{k}')$  and their relative phases, encoded in the position-dependent phase factor  $\exp(-2\pi i \mathbf{k}' \cdot \mathbf{R}_{i,j})$ . The periodicity of  $V(\mathbf{r})$  is directly correlated to the periodicity in observed diffraction patterns. For semi-crystalline materials, such as the ones in this work,  $V(\mathbf{r})$  is partially periodic. No analytical solutions exist for such a potential, but the general features of semicrystalline scattering can be understood by considering each extreme, fully amorphous and perfectly crystalline structures, and interpolating.

#### 2.1.2.1 Diffraction from a Structureless Material

The scattering from a completely structureless material is equivalent to the sum of the scattering amplitudes from single atoms of each type present in the material. The scattering amplitudes are typically called *atomic scattering factors*  $f(\mathbf{k}')$ , and are found from the first-order Born approximation by substituting the electrostatic potential of a single atom  $V^{a}(\mathbf{r}')$  into equation (7).  $V^{a}(\mathbf{r}')$  is a function of the electron potential of an atom through Poisson's equations [93], giving the Mott-Bethe relation between the electron atomic scattering factor  $f^{e}(\mathbf{k}')$  and the X-ray atomic scattering factor  $f^{X}(\mathbf{k}')$ , where  $f^{X}(\mathbf{k}')$  is simply the Fourier transform of the electron charge density:

$$f^{e}(\mathbf{k}') = \frac{me^{2}}{2\pi\hbar^{2}{\mathbf{k}'}^{2}} \int_{-\infty}^{\infty} V^{a}(\mathbf{r}') \exp(2\pi i\mathbf{k}' \cdot \mathbf{r}') d^{3}\mathbf{r}'$$
$$\frac{2\pi me^{2}}{\hbar^{2}{\mathbf{k}'}^{2}} \int_{-\infty}^{\infty} 4\pi e[Z\delta(\mathbf{r}') - \rho_{e}(\mathbf{r}')] \exp(2\pi i\mathbf{k}' \cdot \mathbf{r}') d^{3}\mathbf{r}' = \frac{2me^{2}}{\hbar^{2}{\mathbf{k}'}^{2}} [Z - f^{X}(\mathbf{k}')]$$
(10)

where  $\rho_n(\mathbf{r}') = Z\delta(\mathbf{r}')$  is the nuclear charge density,  $\rho_e(\mathbf{r}')$  is the electron charge density, and Z is the atomic number.  $f^X(\mathbf{k}')$  tends to zero at high  $\mathbf{k}'$ , leading to

$$f(\mathbf{k}')_{\mathbf{k}'\to\infty} = \frac{2me^2}{h^2{\mathbf{k}'}^2}Z$$
(11)

which also tends to zero at high k' as  $1/k'^2$ . Note that the observed intensity  $I(k') = f(k')^2$ , and at high values of k', such as for high-angle annular dark field (HAADF) imaging, the atomic number dependence is roughly  $I_{HAADF} \propto Z^2$ . These atomic scattering parameters are parametrised in tables [96]. In general, electron scattering factors are higher by a factor of ca.  $10^4$  compared to their X-ray equivalents, leading to considerably stronger scattering, as well as larger amounts of multiple and inelastic scattering, especially for thick samples. Both scattering factors decrease with the magnitude of k' due to phase differences between waves from different spatial locations around an atom. [93] Example forms are shown in **Figure 2.1**. Notably, while both X-ray and electron scattering parameters tend towards contrast proportional to Z at high k', the scattering factors are considerably different at low scattering vector magnitudes, with electron factors not being proportional to Z below 3 Å<sup>-1</sup>. In addition, in general electron scattering factors fall faster as a function of k' than X-ray factors do.



**Figure 2.1**. Electron scattering from a structureless potential. (a) Electron scattering geometry. Adapted from [94]. (b) Relative comparison of X-ray, electron, and neutron scattering factors. Reproduced from [95]. (c-d) More detailed scattering factors as a function of scattering vector for (c) electrons and (d) x-rays for a few choice elements. The magnitude of the electron factors further depends on the relativistic electron mass, which is only a constant pre-factor.

### 2.1.2.2 Diffraction from a Perfect Crystal

Of course, in practice no material is structureless. At the opposite end of the spectrum, atoms within a periodic unit cell of a perfectly crystalline material have exactly defined positions defined in terms of a lattice vector  $\mathbf{r}_{\text{lat}}$  and a position vector within the unit cell  $\mathbf{r}_{i,j}$ , as

$$\mathbf{R}_{i,j} = u\mathbf{a} + v\mathbf{b} + w\mathbf{c} + \mathbf{r}_{i,j} = \mathbf{r}_{lat} + \mathbf{r}_{i,j}$$
(12)

where *a*, *b*, and *c* are the lattice basis vectors in real space.

The scattered wave amplitude of a finite crystal  $\phi_g^c(\mathbf{k}')$  near a particular Bragg condition g can then be written as a finite sum over the atoms in the repeating unit cell using the Fourier transform:

$$\phi_{g}^{c}(\mathbf{k}') = F(\mathbf{k}') \cdot FT\left[\sum_{u,v,w}^{N} \delta(\mathbf{r} - \mathbf{r}_{lat})\right]$$
$$= \sum_{i} f_{i}(\boldsymbol{g} + \boldsymbol{s}_{g}) \sum_{j} \exp(-2\pi i(\boldsymbol{g} + \boldsymbol{s}_{g}) \cdot \mathbf{r}_{i,j}) \sum_{u,v,w}^{N} \exp(-2\pi i(\boldsymbol{g} + \boldsymbol{s}_{g}) \cdot \mathbf{r}_{lat})$$
(13)

where  $\mathbf{k}' = \mathbf{g} + \mathbf{s}_g$ . Importantly, this can be simplified by noting that  $|\mathbf{s}_g| \ll |\mathbf{g}|$ , meaning that the  $\mathbf{s}_g \cdot \mathbf{r}_{i,j}$  can be neglected, and by noting that  $\mathbf{g} \cdot \mathbf{r}_{lat}$  evaluates to an integer, to give

$$\phi_g^{\rm c}(\mathbf{k}') = \sum_i f_i(\mathbf{g}) \sum_j \exp(-2\pi i \mathbf{g} \cdot \mathbf{r}_{i,j}) \sum_{u,v,w}^N \exp(-2\pi i \mathbf{s}_g \cdot \mathbf{r}_{\rm lat})$$
(14)

such that  $\phi_g^{nc} = F(g) \cdot S(s_g)$ , where F(g) is the structure factor of that particular Bragg condition gand  $S(s_g)$  is the shape factor, representing the deviation from an infinite crystal.  $S(s_g)$  is related to the Fourier transform of the crystal shape function. The structure factor F(g) is then given by

$$F(\boldsymbol{g}) = \sum_{i} f_{i}(\boldsymbol{g}) \sum_{j} \exp(-2\pi i \boldsymbol{g} \cdot \boldsymbol{r}_{i,j})$$
(15)

that gives the kinematical intensity of a diffracted beam. The second term accounts for lattice geometry and encodes the Bragg condition:

$$\boldsymbol{g} = h\mathbf{a}^* + k\mathbf{b}^* + l\mathbf{c}^*$$

(16)

where g is a reciprocal lattice vector for plane hkl and  $\mathbf{a}^*$ ,  $\mathbf{b}^*$ , and  $\mathbf{c}^*$  are the reciprocal lattice basis vectors.

## 2.1.2.3 Diffraction from an Imperfect Crystal

Semicrystalline materials can be described as exhibiting some amount of crystalline diffraction, although diffracted intensities are typically affected by imperfections in this semicrystalline 'lattice'. Hence, it is worth considering scattering from imperfect crystals, where atoms do not sit in their ideal atomic sites due to finite temperature effects, defects and other distortions. For displacements  $u(r_{lat})$  that varies between different unit cells  $\mathbf{r}_{lat}$ , the position vector  $\mathbf{R}_{i,j}$  is given by

$$\mathbf{R}_{i,j} = \mathbf{r}_{lat} + \mathbf{r}_{i,j} + \boldsymbol{u}(\mathbf{r}_{lat})$$
(17)

Following the same methodology as for Equation (14), this leads to the scattered wave amplitude from a finite imperfect crystal  $\phi_g^{ic}$  (neglecting  $s_g \cdot u(\mathbf{r}_{lat})$  for being small):

$$\phi_{g}^{\rm ic}(\mathbf{k}') = F(\mathbf{k}') \cdot FT\left[\sum_{u,v,w}^{N} \delta(\mathbf{r} - \mathbf{r}_{\rm lat} - \boldsymbol{u}(\mathbf{r}_{\rm lat}))\right]$$
$$= F(\boldsymbol{g}) \sum_{u,v,w}^{N} \exp(-2\pi i \boldsymbol{s}_{g} \cdot \mathbf{r}_{\rm lat}) \exp(-2\pi i \boldsymbol{g} \cdot \boldsymbol{u}(\mathbf{r}_{\rm lat}))$$
(18)

which includes an additional geometric phase term as part of  $S(\mathbf{s}_g)$  with phase  $\alpha = -2\pi i \mathbf{g} \cdot \mathbf{u}(\mathbf{r}_{lat})$ . This is nonzero when  $\mathbf{g} \cdot \mathbf{u} \neq 0$ , i.e. if the displacement  $\mathbf{u}$  is not perpendicular to  $\mathbf{g}$ . As a result imperfections, such as from static disorder or finite temperature effects, result in diffuse scattering away from the Bragg condition. This is particularly strong in amorphous and partially crystalline materials, where diffracted halos, rather than peaks, are frequently seen.

# 2.1.3 Diffraction Geometry

A common way to visualise the Bragg diffraction condition (Equation (16)) is by using the Ewald sphere construction (**Figure 2.2**). The Ewald sphere is a graphical representation of the Bragg condition in reciprocal space and is made with the incident wavevector  $\mathbf{k}_0$  drawn towards the origin of the reciprocal lattice. The surface of the resulting sphere with radius  $|\mathbf{k}_0|$  represents the Bragg condition, with diffraction occurring when a reciprocal lattice point intersects the sphere surface.



**Figure 2.2**. Ewald Sphere construction for a perfect crystal for scattering from the diffraction condition g with planar spacing with d. Reproduced from [97].

By noting that  $|\mathbf{k}_0| = \mathbf{k} = \frac{1}{\lambda}$  and  $|\mathbf{g}_{hkl}| = \frac{1}{d_{hkl}}$ , we can see that the Ewald sphere condition in **Figure 2.2** is equivalent to Bragg's law:

$$\lambda = 2d_{\rm hkl}\sin\theta$$

(19)

### where $\theta$ is the Bragg angle.

In a TEM, the electrons are accelerated such that the wavelength  $\lambda$  of the electron is of the order of picometres, and as such  $|\mathbf{k}_0| \gg |\mathbf{g}|$ . This effect, combined with the fact that samples are thin in the electron beam direction (< 200 nm) to enable transmission, results in the excitation of many diffraction conditions at once, especially when near a zone axis (**Figure 2.3**). The diffracted spots seen in the

central part of the diffraction pattern are known as the zero-order Laue zone, and additional diffracted spots can often be seen in higher-order Laue zones.



**Figure 2.3**. Zone-axis Ewald sphere construction for electron diffraction with equal excitement of +g and -g. Reproduced from [97].

#### 2.1.3.1 Excitation Error

In practice, some diffraction also occurs near, but not exactly at, a Bragg condition. The extent (in reciprocal space) of this near-Bragg condition diffraction is described by the deviation parameter  $s_g$ , often called the excitation error. The excitation error is a result of the relaxation of the Bragg condition in a finite-size sample. In a TEM, it is therefore primarily in the beam direction, as samples need to be thin for electron transparency. This deviation typically takes a  $\operatorname{sinc}^2$  functional form, as it is related to the Fourier transform of the sample profile (a top hat) in the beam direction. An example of a deviation from the diffraction condition is shown in **Figure 2.4**. In the two-beam condition, this feature can be explored further to give an expression for the dynamical diffracted intensity [94] as a function of deviation g:

$$I_g(s,t) = \left(\frac{1}{\xi_g s'}\right)^2 \sin^2 \pi t s' \qquad s' = \sqrt{s^2 + \frac{1}{\xi_g}}$$
(20)

where  $\xi_g$  is the extinction length for reflection g, t is the crystal thickness, and  $\sqrt{\frac{1}{\xi_g^2} + s^2}$  is sometimes referred to as the effective extinction error s'. It is notable that both  $\xi_g$  and t can be simultaneously estimated from the above equation if  $I_g$  can be measured as a function of s.



**Figure 2.4**. Definition of the excitation error (shown positive) for a beam at the Bragg condition (solid line) and near the Bragg condition (dashed line). Right:  $sinc^2$  intensity typically observed near the Bragg spot g. Adapted from [94].

#### 2.1.3.2 Bend Contours

When imaging a bent sample, it is sometimes possible to see oscillating dark and light fringes in both the bright field image and the diffracted intensity across the specimen. This can be interpreted through Equation (20), with a local change in orientation or thickness causing oscillations in the observed intensity. As the sample tilts, its angle to the incoming beam changes, and it passes through the deviation *s* of a given Bragg spot *g*. This variation is often called *Pendellosung*, and extinction occurs when Equation (20) goes to zero; which occurs when

$$\sin \pi t s' = 0$$
$$\pi t s' = n\pi$$
$$s = \pm \frac{1}{\xi_g} \sqrt{\left(\frac{n\xi_g}{t}\right)^2 - 1}$$

(21)

for integer n, such that intensity goes first to zero when  $t = \xi_g$ . This will occur for both +g and -g, but the contrast is often relatively weak for thick samples due to high angle scattering out of the detector. The real-space spacing between *Pendellosung* fringes is inversely proportional to the curvature.

# 2.1.4 Dynamical Electron Diffraction

The approaches used to describe crystalline diffraction above fail for all but the thinnest of specimens due to dynamical and multiple scattering. In comparison to X-rays, electrons interact much more strongly with Coulomb potentials [93], and as a result the First-order Born Approximation doesn't hold in practice. For crystalline materials, this results in dynamical diffraction, especially near zone-axis orientations. For the organic semicrystalline materials in this work, dynamical electron diffraction is generally not needed to interpret data. It has been briefly used to derive the dynamical excitation error (Equation (20)), but otherwise is not covered. The reader is instead guided to [85] for a rigorous treatment of the Bloch-wave description of dynamical scattering and to [98] for the computational approximation through the multislice method.

# 2.1.5 Diffuse Scattering

In practice, a simple combination of structureless and imperfect crystal scattering is a relatively poor description of the scattering that occurs in partially crystalline materials. For a more accurate description, it is worth briefly considering diffuse scattering in more detail. Diffuse scattering refers to the scattered intensity recorded in the diffraction patterns that is not from crystalline Bragg spots nor scattering from a structureless sample. It can result from inelastic phonon, plasmon, or electron excitations, as well as elastic scattering due to structured but imperfect packing. In the kinematical limit, diffuse scattering by electrons is equivalent to diffuse scattering by X-rays or neutrons [94]. For a complete review on the scattering from disorder, the reader is guided to [99].

# 2.1.5.1 Diffuse Scattering from Disorder

As shown for an imperfect crystal in Section 2.1.2.3, structural disorder leads to a local breaking of the Bragg condition, resulting in diffraction to regions in k-space that would be otherwise forbidden. Depending on the whether such disorder is correlated or uncorrelated, the diffraction can frequently manifest either as blurring out of the diffraction spots (uncorrelated) or as lines of intensity in the diffraction pattern (correlated). An example of both is given in **Figure 2.5**. This disordered scattering encodes a large amount of information, and has been used to solve complex structures, such as disordered forms of aspirin [100], defect structures in molecular crystals [101], dislocation effects in semiconductors [102], phonon modes in organic semiconductors [103], the structures of high temperature superconductors [104], and many more [99].

Typically, the dominant small-angle diffuse scattering seen in electron diffraction patterns is due to inelastic scattering by plasmons [105]. These can be reduced by energy-filtering, although this is not available on all microscopes. This source becomes more problematic with thicker samples, as the multiple scattering can cause Bragg spots to broaden due to the low-angle inelastic scattering.



**Figure 2.5**. Diffuse scattering in (a) TIPS-pentacene, a polymer semiconductor. Streaks of diffuse intensity are seen that were correlated to phonon modes. Adapted from [103]. (b-c) Diffraction patterns of structurally disordered Cu-Pt alloys, with (b) 35.0 at. % Pt and (c) 29.8 at. % Pt, showing notable intensity in between the primary Bragg spots. Adapted from [106].

## 2.1.5.2 Thermal Diffuse Scattering

Finite temperature has two primary effects on electron scattering. The first is to reduce the intensity of Bragg spots due to disorder by a factor  $\exp -2M$ , where  $M = B(\sin \theta/\lambda)^2$ , where  $\theta$  is the scattering angle,  $\lambda$  is the electron wavelength, and B is the Debye-Waller factor and related to the average atomic displacement magnitude  $\langle u \rangle^2$  by  $B = 8\pi^2 \langle u \rangle^2$ . The second effect is called thermal diffuse scattering (TDS) and is a complex interaction related to lattice vibrations. It can be shown that the dominant first-order divergence  $I_1$  from a Bragg scattered intensity  $I_0$  due to phonons is given by

$$I_{1} = \frac{f^{2}N^{2}}{4} \exp(-2M) \sum_{\mathbf{k}_{p},j} \langle \boldsymbol{a}_{\mathbf{k}_{p},j}^{2} \rangle \left(\boldsymbol{g} \cdot \boldsymbol{e}_{j}^{\rightarrow}\right)^{2} \left[\delta\left(\boldsymbol{g} + \mathbf{k}_{p} - \mathbf{k}\right) + \delta\left(\boldsymbol{g} - \mathbf{k}_{p} - \mathbf{k}\right)\right]$$
(22)

where  $a_{k_p,j}$  is the vibrational amplitude for phonon j with wavevector  $k_p$ ,  $e_j^{\rightarrow}$  is its direction, g the scattering vector,  $\mathbf{k}$  the electron wavevector, and N is the number of scattering atoms. In practice, equation (22) show that TDS arises from correlated atomic motion in the lattice and acts to disperse

scattered intensity from Bragg spots. From a single phonon mode, TDS would result in a pair of satellite peaks at  $g \pm k_p$ . However, as phonons occupy a continuous set of frequencies, instead smooth, often streaky contrast is seen instead, as shown in **Figure 2.5**. Work fitting full phonon dispersion relationships in X-ray diffraction has been made [107], with similar but more restricted works in electron diffraction [108], [109]. For thick samples, like with electron diffraction, multiple diffuse scattering can occur.

## 2.1.5.3 Multiple Diffuse Scattering

Multiple scattering complicates the analysis of diffuse scattering. Diffuse scattering, including inelastic scattering, carries some of the phase information from among the Bragg reflections [110], [111]. The interaction between elastic and inelastic scattering during multiple scattering leads to Kikuchi lines [112] as well as other complex interactions. Based on the multislice method, applying a distorted wave Born approximation (DWBA) [113], detailed descriptions of multiple inelastic scattering have been given [114], [115]. Primarily, multiple diffuse scattering includes one diffuse scattering event and one or more elastic scattering events due to the low cross-section of diffuse scattering, and results in progressively more smeared out Bragg spots with increasing thickness. A derivation of a general perturbation model is given in [94].

# 2.1.6 Inelastic Electron Scattering

Experimentally, the fit between the theoretical scattering profile for an amorphous material described in Section 2.1.2.1 and experimental scattering is poor for organic materials. This is primarily due to inelastic scattering. The elastic scattering cross section scales roughly as  $Z^2$  while the inelastic scattering cross section only scales as Z, where Z is the atomic number, meaning that as Z increases, the proportion of elastic scattering also increases proportionally to Z [116]. For example, the inelastic scattering cross section is approximately 4 times higher than the elastic scattering cross section for amorphous carbon [117]; approximately 80% of electron scattering is inelastic!

Inelastic scattering occurs when there is energy transfer between the incident electron and the specimen, resulting in energy deposition in the specimen. The biggest sources of inelastic scattering in TEM are core excitations, outer-shell excitations, plasmons, and phonons, and for a complete guide, the reader is guided to [118].

**Core excitation** scattering occurs when there is energy transfer from the incident electron to inner shell electrons in the solid. The electrons are promoted to near or above the Fermi level, resulting in energy transfers of a few keV and scattering typically to the same k-range as Bragg scattering [118]. The subsequent de-excitation is the source of characteristic X-rays and Auger electrons [119].

**Outer-shell excitation** scattering occurs when the energy transfer is to a (typically delocalised) outershell electron in the conduction or valence band of the material, which is excited to a higher energy level. The energy transfer is typically a few eV to tens of eV and is typically scattered to low angles (< 0.05 Å<sup>-1</sup>). The de-excitation results in secondary electrons, but for organic materials, these excited electrons frequently do not return to their original states, resulting in a permanent change to the bonding configuration [118]. This is termed ionisation damage and is discussed further in Section 2.8.

**Plasmon** scattering occurs due to the collective excitation of the valence-electron density. The energy transfer is typically 5-30 eV, and scattering occurs at low angles (< 0.05 Å<sup>-1</sup>). Plasmons are typically short-lived and deposit their energy as heat. Plasmons and outer-shell excitations can be described as alternate methods of excitation, and one tends to be favoured over another. In insulators, as many organic compounds are, the tendency is towards single excitation, while systems with large amounts of free electrons, such as conductors, tend towards plasmon scattering.

**Phonon** scattering occurs due to the scattering of the incoming electron by atomic lattice vibrations, and is often described as quasi-elastic, as the energy losses are below 0.1 eV. It occurs throughout the diffraction pattern, and can be reduced by cooling of the sample. Its effect for light elements (Z < 13) is less than that of electron excitations and plasmons [120]

# 2.1.6.1 Energy Filtering

Energy-filtering enables the removal of almost all inelastic scattering except for phonon scattering, for which the energy loss is less than the energy spread of the electron beam. This has been shown to greatly increase the quality of diffraction patterns acquired in the TEM, especially on organic materials [94], [121], [122], as shown for diamond in **Figure 2.6**. However, energy filtering is not always available on a TEM, especially in conjunction with scanning electron diffraction. In these cases, data acquisition is still possible, but will have lower SNR than filtered data.



**Figure 2.6**. An extreme case of a (110) zone-axis CBED pattern from a diamond sample approximately 1  $\mu$ m thick, without energy filtering (left) and with energy filtering (right). A dramatic improvement in contrast is observed. Reproduced from [122].

#### 2.1.6.2 Modelling Inelastic Scattering

For the purposes of PDF analysis on EM data collected without energy filtering, it is important to understand the approximate angular distribution of inelastically scattered electrons. Lenz [123] formulated the total inelastic scattering cross section  $\frac{d\sigma_i}{d\Omega}$ , rewritten [124] as

$$\frac{d\sigma_i}{d\Omega} = \frac{4\gamma^2 Z}{a_0^2 (2\pi k_i)^4} \left\{ 1 - \frac{1}{[1 + (2\pi k_i r_0)^2]^2} \right\}$$
(23)

where  $\gamma^2 = \left(1 - \frac{v^2}{c^2}\right)$  where v is the electron velocity and c is the speed of light, Z is the atomic number,  $a_0 = 0.529 \times 10^{-10}$  m is the Bohr radius,  $r_0$  is a screening radius that depends on the element, and  $k_i$  is the scattering vector magnitude for inelastic scattering that is related to the incident wavelength  $k_0$  by

$$k_{i}^{2} = k_{0}^{2}(\theta^{2} + \theta_{E}^{2})$$
(24)

where  $\theta$  is the scattering angle and  $\theta_E$  is a characteristic angle (both in radians) associated with the mean energy loss  $\overline{E}$  by  $\theta_E = \overline{E}/(\gamma m_0 v^2)$  where  $m_0$  is the electron mass.

Overall, this results in a scattered intensity  $I(\theta)$  that is linear below  $\theta_E$ , which for carbon is under 0.01 Å<sup>-1</sup>, before following a  $\theta^{-2}$  profile until a critical angle  $\theta_0$  based on  $r_0$ , ca. 0.8 Å<sup>-1</sup> (at 100 kV, changing slowly with accelerating voltage). After this, the profile follows  $\theta^{-4}$ , similarly to elastic scattering. However, the problem with interpreting total scattering profiles in electron scattering is further complicated by multiple inelastic scattering.

#### 2.1.6.3 Multiple Inelastic Scattering

The total scattering cross section of inelastic electrons in organic materials is acceleration voltagedependent, but is typically between 3 and 5 times higher than that of elastic scattering for low *Z* materials [117]. As a result, multiple inelastic-inelastic and inelastic-elastic scattering contribute significantly to the inelastic scattering profiles before multiple elastic scattering contributes. Multiple scattering can be predicted by a Poisson model if the thickness is known (**Figure 2.7**), but the elastic-inelastic multiple scattering is particularly poorly predicted by simulation due to the necessity to know both scattering cross-sections accurately. This, combined with the different dependencies of the elastic and inelastic scattering cross-sections on scattering vector magnitude, result in a background profile whose form is highly thickness dependent. Therefore, a significant divergence of the predicted and measured structureless scattering profiles occur for all but the thinnest of samples when inelastic scattering contributes. Overall, inelastic scattering contributes to the interpretation of diffraction data in organic materials in two key ways. Firstly, it results in a difficult-to-predict thickness-dependent scattering profile, and secondly it is a key contributor to electron beam damage [125]–[127], discussed further in Section 2.8.



**Figure 2.7**. Probabilities of no scattering (P<sub>0</sub>), one occurrence of scattering (P<sub>1</sub>) and so on as a function of thickness *t*. The thickness is divided by the mean free path. At  $t = \lambda$ , multiple scattering contributes 42% of the total scattered intensity, rising to 69% at  $t = 2\lambda$ . Adapted from [118].

# 2.2 Focused Probe Electron Diffraction

# 2.2.1 The Column Approximation

In a focused probe STEM experiment, it is important to be able to consider pixels independently of one another. Theory in diffraction contrast imaging has established the *column approximation* [85], where the sample can be considered as a series of narrow columns oriented in the electron beam direction (**Figure 2.8**). Importantly, such columns can be considered independent such that any change to the wavefunction can only be considered as a function of column length, i.e.  $\psi(z)$ . For this to be true, scattered beams must remain in the column. In an example SED experiment, such as shown in **Figure 2.8**, with t = 200 nm,  $\lambda_e = 2 \text{ pm}$ , and for scattering out to 0.5 Å<sup>-1</sup>, such a column has a width of ~ 2 nm, lower than typical scan step sizes in SED, and can be less than the diffraction-limited width of the beam. As such, pixels in SED can be treated independently.



Figure 2.8. The Column Approximation and its parameters. Reproduced from [97].

# 2.2.2 Diffraction from Focused Electron Probes

SED experiments are performed using a focused probe, typically with a relatively low convergence semi-angle (0.2-1 mrad) in the 'near-parallel' configuration. Such probes are formed by the condenser aperture, and the electron source is focused onto the specimen in the so-called 'critical illumination' condition [128]. The finite convergence angle results in a section of reciprocal space being sampled (**Figure 2.9**), and the image of the aperture being projected onto the diffraction plane as a circular disc at each diffracted spot. To ensure that spots do not overlap, the convergence angle must be kept low, typically < 3 mrad for inorganic samples and < 1 mrad ideally for organic samples with large unit cells. In such a condition, each diffracted spot can analysed independently.



**Figure 2.9**. Convergent Electron Beam Diffraction. Reproduced from [97]. (a) The critical illumination condition, where the source is imaged in the specimen plane and subtends a semi-angle  $\alpha$  to the optic axis. The probe function can be approximated by an airy disk. Adapted from [128]. (b) The Ewald sphere for a finite convergence angle. (c) Schematic CBED pattern showing the diffraction pattern from the optical condition in (b).

Data in this work is taken primarily in the nanobeam electron diffraction (NBED) configuration, with a ca. 0.5 mrad convergence semi-angle and using a small diameter (ca. 10 µm) condenser aperture  $d_A$ . Under such conditions, coherent illumination is obtained when the transverse coherence length of the electron beam  $X_a = \frac{D\lambda}{\pi d_s}$  is larger than the diameter of the aperture. See **Figure 2.9**a for the definitions of the variables. This is generally true when a field-emission gun and a small illumination aperture are used [128]. This results in a probe wave function

$$\psi_{p}(\mathbf{r}) = \int_{-\infty}^{\infty} A(\mathbf{k}_{t}) \exp(i\chi(\mathbf{k}_{t})) \exp(2\pi i \mathbf{k}_{t} \cdot \mathbf{r}) d\mathbf{k}_{t}$$
$$= FT [A(\mathbf{k}_{t}) \exp(i\chi(\mathbf{k}_{t}))]$$
(25)

where  $A(\mathbf{k}_t)$  represents the illumination aperture, typically a Heaviside step function and  $\chi(\mathbf{k}_t)$  is the wave aberration function of the transverse component  $\mathbf{k}_t$  of the electron probe. Notably, in NBED, the use of a small illumination aperture gives  $\chi(\mathbf{k}_t) \sim 0$  [128], and as a result the probe function is just the Fourier transform of the Heaviside function, which is the Airy function [129], as shown in **Figure 2.9**a. Therefore diffraction spots are expected to appear as Heaviside discs once again.

# 2.3 Scanning Electron Diffraction

Properties of both crystalline and non-crystalline materials are frequently determined by defects present in the structure. It is thus technically critically important to be able to determine the nature of such defects, often with nanoscale precision. Electron back-scatter diffraction (EBSD) is highly effective on the length scales of tens of nanometres up to hundreds of micrometres [130]. However, a gap exists in structure determination between the ten nanometres and sub-nanometre scale. Scanning electron diffraction (SED) in the scanning transmission electron microscope (STEM) has emerged as a contender to fill this gap. In SED, a nanometre-sized near-parallel electron probe is rastered over an area of interest, and a diffraction pattern is recorded at each probe position, forming a 4D data set which is then analysed. SED's rise has been driven by recent STEM developments in highly coherent high brightness electron sources [131], flexible probe-forming electron optics [132], fast and efficient pixelated detectors [133], [134], and the reducing cost of computational power to process the large obtained data sets. For the purposes of non-crystalline materials, it is easy to analyse the obtained diffraction patterns using methods more typically suited to them, such as pair distribution function (PDF) analysis [135]. No additional data has to be taken, and the only requirement is to have a code package that can analyse the obtained files, such as the in-house crystallographic data suite Pyxem [136]. A schematic of SED is shown in Figure 2.10.

It is worth noting that SED data sets can be acquired using scanning precession electron diffraction (SPED) [137], requiring a slightly more complex setup, where the electron beam is tilted away from the optic axis by the precession angle - and the beam then rotated about the optic axis both above and below the specimen. Precession has the advantages of having a large range of accessible scattering angles due to the rotation of the zero-order Laue zone, and that the intensities of the reflections in SPED patterns are closer to their kinematically predicted values [138]–[140]. SED and SPED have been successfully applied to phase mapping [141], strain mapping [142], and three-dimensional interphase crystallography [143]. These have primarily been crystalline applications, where analysis has been based on the analysis of Bragg peaks.



**Figure 2.10**. Schematic of SED. Left to right. A beam is scanned in *x* and *y*, with a 2D-diffraction pattern acquired  $(k_x \text{ and } k_y)$  at each pixel. Subsequently, a number of experiments can be calculated in post-processing (right). Adapted from [97].

# 2.3.1 Electron Optics

The theory of electron optics and lenses in an SED configuration is now well established and can be read in [97], [144]. A few key findings are noted here.

#### **Lens Aberrations**

Electron lenses must obey Laplace's equations, and as a result they do not act as perfect lenses. Perfect lenses would focus an incoming wavefront to a coherent single focus point [94]. Typical circular lenses have well-known aberrations that can be represented by the aberration function  $\chi(\mathbf{k})$ , where

$$\chi(\mathbf{k}) = \frac{2\pi}{\lambda_e} W(\mathbf{k})$$

(26)

defined as a phase difference between the ideal spherical wavefront and the aberrated wavefront.  $W(\mathbf{k})$  represents the error in the wavefront measured as a distance. A single ray is aberrated by an amount  $\mathbf{r} \propto \nabla W(\mathbf{k})$ . This error is typically described as an expansion about the optic angle  $\rho$  and angle around the axis  $\theta$  [85].

Notably, for non-aberration corrected SED, after correcting for the defocus, two-fold astigmatism, and coma by aligning beam-tilt through the lens, the probe becomes dominated by the defocus and spherical aberration. In SED, aberrations are typically negligible due to the small condenser apertures used to obtain the near-parallel beam, except in the case of scanning precession electron diffraction (SPED).

## 2.3.2 Electron Probe

The electron probe for an SED experiment is typically formed in 'microprobe' mode with a small condenser aperture (C2 aperture in a three condenser lens system), which involves a beam crossover above the mini-condenser lens, resulting in a near-parallel beam with a small convergence semi-angle (0.2 - 1 mrad). Given a sufficiently small enough aperture, such alignments can also be done in 'nanoprobe' mode, where the mini-condenser lens is not used. The optical configuration is highly dependent on whether the system has two or three condenser lenses (CLs), as shown in **Figure 2.11**. Three condenser lens systems offer additional control of the convergence angle by the control of position of the CL2-CL3 crossover, which can be controlled by the relative CL2-CL3 excitation [94]. Aberration corrector optics may also need to be considered, and are not generally required in operation due to the small probe-forming apertures used in SED.



**Figure 2.11**. Optical configuration for a typical SED experiment in (a) two-condenser lens nanoprobe, (b) three condenser lens nanoprobe mode, and (c) three condenser lens microprobe mode. A probe corrector sits within the dashed box, but is not typically used in SED. CL: Condenser lens, MCL: Mini-condenser lens, UOL: Upper objective lens, LOL: Lower objective lens, CA: Condenser aperture, C2: Condenser 2 aperture The MCL-UOL-LOL vertical spacing is exaggerated on the diagram for clarity.

### 2.3.2.1 Probe size

The alignments of the electron microscope optics are performed to minimise the effect of the correctable lowest-order aberrations (defocus, two-fold astigmatism, and coma) on the electron probe. The subsequent probe size will be limited by a combination of spherical aberration, chromatic aberration, and the Rayleigh diffraction limit. Use of field emission guns (FEGs) reduces the chromatic effect, resulting a probe diameter d of [94], [145]

$$d^{2} = d_{d}^{2} + d_{s}^{2} = \left(\frac{1.22\lambda}{\alpha}\right)^{2} + (0.5C_{s}\alpha^{3})^{2}$$
(27)

where  $C_s$  is the coefficient of spherical aberration,  $\alpha$  is the convergence semi-angle, and  $\lambda$  is the electron wavelength. However, in typical SED with a near-parallel beam,  $\alpha$  is small, such that the diffraction term dominates, giving

$$d \approx \frac{1.22\lambda}{\alpha} \tag{28}$$

In practice the probe size is typically slightly larger. Some typical probe sizes are given below in **Table 2.1**.

**Table 2.1** Calculated diffraction-limited values of probe diameter d with accelerating voltage  $E_0$  and convergence semi-angle  $\alpha$ .

$\boldsymbol{E_0}$ (kV)	lpha (mrad)	<b>d</b> (nm)
80	0.5	10.2
200	0.2	15.3
200	0.5	6.1
200	1.0	3.1
300	0.2	12.0
300	0.5	4.8
300	1.0	2.4

# 2.3.3 Scanning Precession Electron Diffraction

For highly crystalline samples, it can be preferential to tilt away from the optic axis to reduce dynamical interactions in the sample. One such method is precession electron diffraction (PED) [146] in which a double-rocked electron beam is precessed around the optic axis at a precession angle  $\phi$  and then rocked back into focus below the optic axis to measure integrated intensities across a hollow cone of illumination. Diffraction patterns recorded in PED have been shown to have diffracted intensities that are closer to the kinematical prediction than without precession. Precession is generally required to solve crystal structures accurately through electron diffraction, such as with 3D-ED [147]. Extensive literature exists on PED, and scanning precession electron diffraction (SPED) [25], [137], [148]. In addition, detailed methods for implementations of PED [149] and SPED [97] exist. A schematic of PED optics is shown in **Figure 2.12**a.

PED has two primary effects on the observed diffraction patterns. Firstly, the intensities are closer to the predicted kinematical intensities [140] (**Figure 2.12**), and secondly, due to the hollow cone illumination, a larger range of reciprocal space is acquired. This second effect can be highly beneficial for orientation and phase mapping in crystalline organic structures. A side effect of precessing the electron probe is an increase in the probe size, with an additional precession contribution [146], assuming a non-corrected TEM, such that the coefficient of spherical aberration  $C_s$  is the dominant aberration, of

$$d = 4C_s \phi^2 \alpha \tag{29}$$

where  $\phi$  is the precession angle, although by optimising the alignment procedure and choice of convergence angle, this can be reduced [97], [103].



**Figure 2.12**. Precession electron diffraction. (a) An optical diagram of a typical PED alignment, showing rocking and de-rocking of the electron probe pre- and post-specimen. (b-e) Changes to the unprecessed (b) [001] zone axis diffraction pattern of Er<sub>2</sub>Ge<sub>2</sub>O<sub>7</sub> with 20 mrad precession (c), and 47 mrad precession (d). (e) is a kinematical simulation. Reproduced from [25].

# 2.3.4 Data Acquisition on Organic Samples

Significant optimisation of SED acquisition conditions on organic materials was done as part of this work. In a typical SED experiment on organic samples, a few parameters were found to be common:

- The current is low (2-10 pA) to minimise beam damage.
- Approximately 256x256 pixels are acquired in real space, typically limited by data size after acquisition, although this can be reduced.
- The convergence angle is between 0.2 and 0.5 mrad, smaller than in conventional SED due to the larger unit cells present.
- The step size is between 0.5*d* and *d*, where *d* is the beam diameter, estimated from the convergence angle of a diffraction limited probe as in Equation (28). Higher step size reduces damage from beam overlap, while lower improves spatial sampling.
- A 256x256 or 512x512 direct electron detector is used. Larger detectors will quickly increase file size.
- A short acquisition time (ca. 1 ms) is used to reduce electron dose and enable fast acquisition.

Subsequently, pre-processing steps, such as calibration are done in Pyxem [136]. Details of the preprocessing steps are available on the Pyxem GitHub, as well as in [97]. Furthermore, while exact conditions will be sample-dependent, a few general suggestions exist:

# **Crystalline Samples**

Organic crystals typically have short Bragg vectors ( $0.05 - 0.15 \text{ Å}^{-1}$ ), which can easily result in peak overlap on detectors with poor MTF performance. Therefore, it is recommended to use a relatively high camera length if this is a concern. In addition, precession should be used if orientation mapping is needed to maximise the number of visible reflections in each pattern.

# **Amorphous Samples**

In contrast to crystalline samples, a low camera length should be used, such that the  $|\mathbf{k}|_{max}$  is at least 2 Å<sup>-1</sup>. In order to probe the short (< 1.5 Å) bonds in organic samples, this high frequency information is needed. If dose is particularly important, it should be ensured that 2 Å<sup>-1</sup> is the distance to the flat edge (rather than the corner) of the detector, to ensure that all electrons at this k are detected. Amorphous samples typically do not have sudden changes in structure, so a lower convergence angle (< 0.2 mrad) with a larger beam diameter can also be used to improve pair distribution function (see Section 2.6) quality if available.

# **Polymer Samples**

Polymers typically have strong intensity around 0.25 – 0.4 Å<sup>-1</sup> due to the diffraction from  $\pi$ - $\pi$  and backbone stacking. In the case of more strongly ordered polymers, additional intensity will likely be seen at higher scattering vector, and in this case similar rules as with crystalline samples should be considered. However, in the case of polymers with low crystallinity one should go to a high camera length to maximise the angular resolution within the  $\pi$ - $\pi$  ring, such that  $|\mathbf{k}|_{max}$  is approximately 0.4 Å<sup>-1</sup>.

# 2.4 Diffraction Contrast Imaging

### 2.4.1.1 Dark Field Imaging

In TEM mode, it is possible to both tilt the direct beam and insert the objective aperture to form a dark field (DF) image. The aperture and beam tilt are chosen to align a specific diffraction disc onto the optic axis of the microscope, effectively forming an image of the source of the diffracted intensity. DF imaging has been used to image defects such as dislocations [150], and to identify inorganic regions in biological samples [151], and is a powerful tool for analysing nanoscale structure. However, the use of dark field imaging is limited by its impracticality for beam-sensitive samples. Diffraction patterns must be acquired, a suitable diffracted beam identified, the sample tilted onto the correct diffraction axis, and then the diffracted beam aligned with the objective aperture. A schematic is shown in **Figure 2.13**. As a result, it is impractical to measure more than a few dark-field images for a single sample. In addition, a significant amount of electron dose is spent aligning the setup. As a result, DF imaging has not been used extensively in TEM, particularly for the analysis of organic samples. However, in SED it is possible to reconstruct virtual dark field images in post-processing (below).



**Figure 2.13**. Optical configuration for dark field imaging. A beam is tilted to the diffraction condition, and a diffraction pattern is formed on the objective aperture, such that a select hkl is used subsequently to form an image on the detector. This is almost equivalent to placing an objective aperture around a diffraction spot.

## 2.4.1.2 Virtual Dark Field Imaging

Virtual dark field (VDF) imaging is a computational technique used in SED, where a virtual (computational) objective aperture is placed in the detector plane during data processing. A schematic is shown in **Figure 2.14**. Effectively, any number of DF experiments can be performed computationally. As such, dark field imaging in SED is a very powerful tool that can highlight orientation changes [152], separate nanoparticles [148], quantify irradiation damage [153], and much more. In addition, it is an extremely useful tool for qualitatively analysing data sets and is used extensively in the later chapters.



**Figure 2.14**. Virtual dark field imaging. In an SED scan (left), a diffraction pattern is acquired at each pixel. Subsequently, a virtual aperture (middle) is placed in the same position in diffraction space for each diffraction pattern. The intensity within the aperture is integrated to form a virtual dark field image (right). Any arbitrary shape or size aperture can be used to create a VDF image.
# 2.5 Fibre Diffraction

Fibres and twisted crystals can have high degrees of short- and long-range order while falling short of periodic three-dimensional crystallinity. Typically one axis, close to or parallel to the fibre axis, is constant, while the other directions are cylindrically averaged. This cylindrical averaging results in the phenomenon of fibre diffraction [154]. In fibre diffraction, a helical symmetry for the diffracting unit is assumed, resulting in the presence of so-called 'layer lines', where all seen diffracted vectors appear in parallel lines of constant k across the diffraction pattern (**Figure 2.15**) [155]. The central line perpendicular to the rod axis is called the equator, and the line perpendicular to the equator is called the meridian.

Helical diffraction is typically described in cylindrical polar coordinates, with  $(r, \varphi, z)$  in real space, and  $(R, \psi, Z)$  in reciprocal space [154]. Conventionally, z is the helix axis, and the line at z = 0,  $\varphi = 0$  is the x-axis. The repeat distance along the z-axis is c, and there are u repeating structural units in t turns in the space 0 < z < c. Cochran *et al*. [156] showed that the structure factor on layer line l of a repeating helix is

$$\boldsymbol{F}(R,\psi,Z) = \sum_{j} \sum_{n} f_{j} J_{n} \left(2\pi Rr_{j}\right) \exp\left[i\left(n\left(\psi+\frac{\pi}{2}\right)-n\varphi_{j}+\frac{2\pi lz_{j}}{c}\right)\right]$$
(30)

where diffraction occurs only for Z = l/c, where l is an integer.  $r_j$ ,  $\varphi_j$ , and  $z_j$  are the real-space coordinates of atom j in the unit cell,  $f_j$  is its atomic scattering factor, and  $J_n$  is the Bessel function of the first kind of order n. However, the summation over n should only be done over those values of nthat satisfy the helical selection rule

$$l = tn + um \tag{31}$$

where *m* is any integer and there are *u* repeating structural units in *t* turns in the space 0 < z < c as mentioned above. The usual  $I \propto |\mathbf{F}|^2$  holds for kinematical scattering. Equation (30) can be rewritten in the form [68]:

$$\boldsymbol{F}(R,\psi,l/c) = \sum_{n} \boldsymbol{G}_{n,l}(R) \exp\left[in\left(\psi + \frac{\pi}{2}\right)\right]$$
(32)

Where the Fourier-Bessel structure factor  $G_{n,l}(R)$  is independent of  $\psi$  and is given by

$$\boldsymbol{G}_{n,l}(R) = \sum_{j} f_{j} J_{n} \left( 2\pi R r_{j} \right) \exp \left[ i \left( -n\varphi_{j} + \frac{2\pi l z_{j}}{c} \right) \right]$$
(33)

Notably, in this form  $J_n(0)$  is 1 when n = 0 and 0 otherwise, resulting in the important property that structure factors on the meridian (R = 0) are nonzero only on the layer lines where l is an integer multiple of the number of repeat units in a helical unit cell u.

Fibre diffraction patterns typically fall into two categories, crystalline and non-crystalline. The crystalline patterns show clear diffraction spots, while non-crystalline patterns only show broad lines of continuous intensity. An example of each is shown in **Figure 2.15**. Nonetheless, intensities are still limited to layer lines in both structures.



Software exists to simulate fibre diffraction data, such as the program Helix [157].

**Figure 2.15**. X-ray fibre diffraction of (a) an amorphous oriented nucleic acid fibre, and (b) a polycrystalline and oriented polysaccharide fibre. Common terminology is overlaid on (a). Layer lines are also present in (b), but with much finer spacings, shown by the short red arrows. Adapted from [154].

# 2.6 Pair Distribution Function Analysis

## 2.6.1 Pair Distribution Function Theory

### 2.6.1.1 The Pair Distribution Function in Real Space

Describing non-crystalline structure in a material is a challenging task. Without long-range order, a perfect description of the structure would only be obtained by specifying the positions of every single atom in the structure. This is obviously untenable, so three-dimensional ensemble models must instead be used. It is possible to describe the structure in terms of models and packing of atoms, such as shown by Bernal [158] with packing of random spheres. An alternative, and often more powerful, method is using statistical distributions to describe the atomistic structure. The pair distribution function (PDF) is one such method and is powerful as it can be obtained from experiment through the transform of diffraction data. The notation below follows [159] but within the electron scattering vector framework, where *s* is the scattering vector magnitude, taken as  $s = \frac{2 \sin \theta}{\lambda}$  for scattering semi-angle  $\theta$  and de Broglie wavelength for the electron beam  $\lambda$ . This is in contrast to the typical X-ray community notation of wavevector magnitude  $Q = \frac{4\pi \sin \lambda}{\lambda}$ .

The PDF g(r) measures the normalised probability density of finding two atoms separated by a distance r and oscillates about unity. It has peaks at distances  $r_{ab}$  where r is the separation between two atoms a and b in the solid, as is shown in **Figure 2.16**. Mathematically, it is given by

$$g(r) = \frac{1}{4\pi\rho_0 N r^2} \sum_a \sum_b \delta(r - r_{ab})$$
(34)

where  $\rho_0$  is the average number density in a system of N atoms.  $\delta$  is the Dirac delta function. Peaks in the PDF represent deviations from the average number density in the material and can be used to characterise structure. In continuum terms, the equivalent equation is

$$g(r) = \frac{1}{\rho_0^2 V} \int \rho(r') (\rho(r+r') dr'$$
(35)

where V is the sample volume,  $\rho(r)$  is the single atom density function, and  $\rho_0$  its average. In short, the PDF quantifies the oscillations away from the average atomic density.



**Figure 2.16.** A schematic of the reduced PDF G(r) modelled from a sheet of graphene and an experimental X-ray PDF. In practice, peak broadening occurs due to thermal motion and missing information as a result of a finite *s* range. Adapted from [160].

#### 2.6.1.2 Projection of the PDF in TEM

In practice, this 3D sum is slightly altered in a TEM, as the electron beam only probes momentum transfer perpendicular to the beam path, as noted in the column approximation. As a result, the acquired PDF is in fact the 2D atomic column-column correlation

$$g_{\text{column}}(r) = \frac{1}{A} \int_{r' \in x, y} \rho_{1\text{D}}(r') \rho_{1\text{D}}(r+r') dr'$$
(36)

where the integral is over x and y, parallel to scattering vector s, A is the probed area of the sample perpendicular along the beam path (z) and  $\rho_{1D}(r)$  is the column-averaged single atom density function along z

$$\rho_{1D}(r) = \rho_{1D}(x, y) = \frac{1}{L} \int \rho(r) dz$$
(37)

where *L* probed length of the sample in the *z* direction. The measured directional PDF in a TEM probes the integrated column-to-column correlation in the x - y plane, where each column is integrated along *z* first, and then the column-column correlation is calculated, rather than the true 3D-PDF. However, given the relatively large probed volume at low convergence semi-angle, for non-crystalline materials, this probed correlation is typically assumed to be equivalent to the true 3D-PDF.

In addition, the PDF is in general a vector quantity that depends on the direction in *s*-space. However, provided that the probed sample is uniform and that all orientations are probed along the beam path, which is an approximation that holds for amorphous materials, the two are equivalent. In crystalline samples with multiple orientations, the PDF would therefore be expected to vary by real space position, as it depends on the local orientation.

### 2.6.1.3 Obtaining the Pair Distribution Function from Scattering Data

Like all structure-probing functions, the PDF is related to the scattered amplitude from a sample. The PDF is a total scattering technique; it is a measure of both the Bragg peaks and the diffuse scattering intensities that appear in diffracted data. Using diffuse scattering is key to the interpretation of scattering data of non-crystalline materials, as it carries information about the local structure [159]. This is necessary for the interpretation of any short- and medium-range order present. Classically, the PDF g(r) is extracted from the structure function S(s):

$$G(r) = 4\pi r \rho_0(g(r) - 1) = 8\pi \int_0^\infty s[S(s) - 1]\sin(2\pi sr) \, ds$$
(38)

where the reduced pair distribution function G(r) is commonly presented instead. The structure function S(s) is related to the magnitude I(s) of the sample scattering amplitude  $\psi(s)$ 

$$S(\mathbf{s}) = \frac{I(\mathbf{s})}{\langle f(s) \rangle^2}, \qquad I(\mathbf{s}) = |\psi(s)|^2$$
(39)

where  $\langle f(s) \rangle$  is the compositionally-averaged sample scattering factor, given by  $\langle f(s) \rangle = \sum_i c_i f_i(s)$ where  $c_i$  and  $f_i(s)$  are the atomic fraction and scattering amplitude of element i,  $\psi(s)$  (called the the scattering amplitude) is the coherent sum of the scattered waves from the atoms in the sample

$$\psi(s) = \sum_{v} f_{v}(s) e^{2\pi i \, \mathbf{s} \cdot \mathbf{R}_{v}} \tag{40}$$

where  $\mathbf{R}_{v}$  is the position of the vth atom,  $f_{v}(s)$  is the scattering amplitude of the atom v, and s is the diffraction vector defined by

$$\boldsymbol{s} = \boldsymbol{k}_{\rm f} - \boldsymbol{k}_{\rm i} \tag{41}$$

where  $k_i$  is the wavevector of the incoming beam and  $k_f$  is the wavevector of the scattered beam. For elastic scattering  $\lambda_i = \lambda_f$  and so the magnitude of s is given by

$$s = |\mathbf{s}| = \frac{2\sin\theta}{\lambda} \tag{42}$$

In an experiment, to obtain the PDF, one starts from the one-dimensional scattered intensity profile I(s). In a TEM, the profile I(s) is obtained by an azimuthal integration of the detector plane around the central beam. I(s) is normalised to obtain the reduced intensity  $\varphi(s)$ , defined for a monoatomic sample as

$$\varphi(s) = s[S(s) - 1]$$

$$\varphi(s) = s \frac{I(s) - Nf(s)^2}{Nf(s)^2}$$
(43)

where N is a fitted scattering parameter related to the number of scattering atoms. However, for multi-element systems, the definition of the reduced intensity is more complex, as the sample coherent scattering cross-section and structureless scattering differ, resulting in a modified sum [17], [159]

$$\varphi(s) = s \frac{I(s) - N < f_i(s)^2 >_i}{N < f_i(s) >_i^2}$$
(44)

over the elements *i* in the sample. The divergence of  $\langle f_i(s)^2 \rangle_i = \sum_i c_i f_i(s)^2$  and  $\langle f_i(s) \rangle_i^2 = [\sum_i c_i f_i(s)]^2$  in the numerator and denominator is due to the coherent scattering cross-section being proportional to  $\langle f_i(s)^2 \rangle_i$ , while the sample's structureless scattering factor, which is the correct normalisation to obtain the PDF, is proportional to  $\langle f_i(s) \rangle_i^2$ .

The practical details of this fit are covered in Section 4.3. Finally, the PDF is calculated from  $\varphi(s)$  according to Equation (38). An important point to make is that to extract the PDF, a value for  $\rho_0$  must be assumed, or the result will not oscillate around unity. Hence the reduced PDF G(r) presents a cleaner way to present the data as it does not explicitly depend on  $\rho_0$ , is the direct Fourier sine transform of the data, and conveniently oscillates around zero. Due to the reasons above, the data presented in this work are exclusively reduced PDFs. Although harder to physically interpret, the reduced PDF does have an additional advantage when considering variation in peak heights. In an ideal crystal, the peaks in the reduced PDF oscillate to infinity with no change in height as a function of r. Hence, any loss in peak height is a sign of loss of structural order or a function of acquisition conditions, and the reduced PDF can be used to interpret variance in crystallinity.

#### 2.6.1.4 Pair Distribution Function of Multi-component Systems

When probing systems with multiple atoms or components, it can often be desirable to extract data about the local coordination of a specific atomic species or specific component, or to understand the summation of PDFs from multiple partial components. It is possible to define a partial pair distribution function  $g_{\alpha\beta}(r)$  that gives the distribution of atom pairs in the material coming from atoms of type  $\beta$  around atoms of type  $\alpha$  [161], [162]. The partial distribution function is given by:

$$g'_{\alpha\beta}(r) = \frac{1}{4\pi\rho_0 N r^2} \sum_{\nu \in \{\alpha\}} \sum_{\nu \in \{\beta\}} \delta(r - r_{\nu\mu})$$

$$(45)$$

where  $v \in \{\alpha\}$  means 'for atoms that are type  $\alpha$ '. This then results in the straightforward summation to the full PDF:

$$g(r) = \sum_{\alpha} \sum_{\beta} g'_{\alpha\beta}(r)$$
(46)

This also translates to the structure function S(s), although the convention is to normalise the scattering factors by their relative scattering cross section  $\langle f_{\alpha\beta}(s) \rangle^2 / \langle f(s) \rangle^2$  [159] where  $\langle f_{\alpha\beta}(s) \rangle^2$  is the sum of the scattering cross section b for the partial structure function of  $\alpha$  and  $\beta$  over atoms of type  $\alpha$ . This gives the conventionally used normalised partial structure function

$$S_{\alpha\beta}(s) = \frac{\langle f(s) \rangle^2}{\langle f_{\alpha\beta}(s) \rangle^2} S'_{\alpha\beta}(s)$$
(47)

where  $S'_{\alpha\beta}(s)$  is the scattering occurring due to atoms of type  $\beta$  around atoms of type  $\alpha$ . This normalisation is imperative as it makes  $S_{\alpha\beta}(s)$  and  $g_{\alpha\beta}(r)$  tend to the same limits as S(s) and g(r)respectively. In practice, this means that we can analyse and quantify the partial PDF  $g_{\alpha\beta}(r)$ . This in turn allows us to define a partial structure factor  $\varphi_{\alpha\beta}(s)$ 

$$\varphi_{\alpha\beta}(s) = s [S_{\alpha\beta}(s) - 1]$$
(48)

from which can be defined a partial reduced pair distribution function  $G_{\alpha\beta}(r)$ 

$$G_{\alpha\beta}(r) = 8\pi \int_0^\infty \varphi_{\alpha\beta}(s) \sin(2\pi s r) \, ds \tag{49}$$

such that

$$G(r) = \sum_{\alpha\beta} \frac{\langle f_{\alpha\beta}(s) \rangle^2}{\langle f(s) \rangle^2} G_{\alpha\beta}(r).$$
(50)

However, complications arise from the weighting factor  $\langle f_{\alpha\beta}(s) \rangle^2 / \langle f(s) \rangle^2$  which is itself a function of the scattering vector *s*. As a result, decomposing a multi-component PDF systems is very challenging. Approximation such as the Morningstar-Warren approximation [163] do show that multiple PDFs add roughly in summation and analysis on binary systems has shown decomposition of a multi-signal system to be possible [164] but in general this scattering-vector dependence of the summation means that decomposing a multi-component PDF into its constituents is non-trivial. In electron-PDF, multiple scattering further complicates this as  $f_{\alpha}(s, t)$  is also a function of thickness *t*. It has been shown that separation from a compositional series is possible through unsupervised machine learning methods [165], although this only works for thicknesses where multiple scattering does not play a major part.

### 2.6.2 X-ray PDF

Most PDF analysis in the literature is performed using synchrotron X-rays. This is an effective way to characterise bulk samples but remains a spatially-averaged and depth-averaged method. State-of-theart PDFs, with high resolution structural information beyond 100 Å can be obtained routinely [159]. Xray-PDF can be done using both mono- and polychromatic X-rays [166], although the former is more common. Laboratory sources have maximum scattering vectors  $s_{max}$  of between 1.3 Å<sup>-1</sup> (Cu-sources) and 3.2 Å<sup>-1</sup> (Ag sources), while synchrotron sources have attainable  $s_{max}$  between 4 Å<sup>-1</sup> and 8 Å<sup>-1</sup>. Corrections need to be applied to X-ray scattering data, due to the large thickness of the sample and the high accuracy desired. The necessary corrections are well known for X-ray scattering [159] and correspond mainly to absorption, polarisation, and Compton scattering. Inelastic scattering at high *s* is often discriminated during acquisition, as is incoherent Compton scattering.

### 2.6.3 Electron PDF

The primary advantages of electron PDF (ePDF) over X-ray PDF is a far greater scattering cross section, allowing smaller volumes of material to be investigated, as well as a localisation of signal, allowing spatially resolved mapping of structure. However, the greater scattering cross section means electron diffraction is subject to multiple scattering in all but the thinnest samples. This is discussed further below in Section 2.6.5. A more thorough review by Karle of the factors that limit quantitative PDF analysis can be found in [167]. A disadvantage of ePDF is the difficulty of extracting partial PDFs, as well as the reduced *s* range. Well established methods exist to extract the partial PDFs in X-ray and neutron PDF data, but these methods require measuring at conditions with different relative scattering factors f(s), something not easily achievable in the electron microscope [135].

In STEM diffraction, even with a near-parallel beam as used in SED, there is a broadening of the diffracted signal in s-space due to the finite convergence angle of the beam. This can be easiest seen in the broadening of diffracted spots from crystalline material, although the same effect also occurs in non-spotty diffraction patterns. This results in the degradation of s-resolution, which in turn results in the loss of information at high values of r.

### 2.6.4 Neutron PDF

It is possible to do bulk PDF analysis using neutron scattering. Top end maximum scattering vectors  $s_{\text{max}}$  of greater than 15 Å<sup>-1</sup> can be achieved. However, there are relatively few neutron sources and intensities are comparatively low, resulting in X-rays being usually a preferable source for bulk analyses. Neutron PDF will not be covered in detail here; for more detail, the reader is guided to [159].

## 2.6.5 Effect of Sample Thickness on the PDF

Theoretically, the PDF is independent of thickness. However, in electron scattering, multiple and inelastic scattering introduce a thickness *z*-dependent change to the scattered intensity I(s), as noted in Section 2.1. As a result, any measured reduced PDF is affected by thickness. In crystalline samples, this is typically seen as "averaging out the intensities" where strong diffracted spots become weaker and vice versa [159]. In amorphous materials the effect is similar, with peaks in the diffracted profile slowly "broadening" as thickness increases [168]. As a result, the observed diffraction profile diverges from the expected scattering profile given in Equation (44) with increasing thickness, particularly at low *s*. As the profile also appears in the denominator for the reduced intensity  $\varphi(s)$ , this can result in a diverging  $\varphi(s)$  at high value of *s*. Methods of overcoming this problem are discussed in Section 4.3.

Anstis et al. [168] investigated this effect and found that peak positions in the PDF were relatively unaffected whilst  $t < 5\lambda_0$ , where  $\lambda_0$  is the elastic mean free path of the sample. The effect was primarily to broaden peaks in the PDF, although coordination numbers were also affected. This could be reduced by normalisation of the PDF to the intensity of the first peak [169]. This makes the PDF quite robust to analysis in regimes with thickness changes, as is common in nanoparticle samples, as identifying features of the PDF are unchanged. Methods to recover elastic single-scattering PDFs exist [169], [170] but either require accurate estimates of sample thickness or negligible inelastic scattering, both of which are not fulfilled for typical SED scans. In general, it was found that the PDF is also more robust to thickness change than many of the other intermediate structural signals, such as the reduced intensity.

## 2.6.6 Other Related Functions

It is worth noting that other correlation functions mathematically related to the PDF exist, such as the Patterson function. They will not be covered here, but Egami and Billinge provide an excellent summary in [159].

# 2.7 Electron Detection

In order to accurately measure diffraction data, is it vital to have appropriate electron detectors. For SED experiments, it is vital to have fast (1-10 ms) detectors, in order to acquire scans with a reasonable size (256x256) within a reasonable time considering microscope stability. When combining this with low-dose measurements, as done in this work, a high detective quantum efficiency (DQE) to avoid 'wasting' electrons is key. For this, the answer has been a shift to hybrid pixelated detectors and direct electron detectors [134], [171], where electrons impinge directly onto either a silicon sensor or the electronics, rather than via an intermediate scintillator, as used to be typical for slower scanning charge coupled devices. Such detectors showcase near-perfect DQEs, are fast (< 1 ms or less), have strong linearity, and have low dark noise levels. A hybrid Merlin Medipix3 has been used throughout this work [133].

## 2.7.1 Detective Quantum Efficiency

The detective quantum efficiency (DQE) is given by

$$DQE(\omega) = \frac{SNR_{output}^{2}(\omega)}{SNR_{input}^{2}(\omega)}$$

(51)

where  $\omega$  is spatial frequency, and SNR is the signal-to-noise ratio of the electron wavefunction (input) and the measure intensity (output) respectively. The DQE can have a maximum of 1 and can be very high for direct electron detectors [172], [173].

### 2.7.2 Modulation Transfer Function

An ideal detector has a small point spread function (PSF) to minimise the spread of measured intensity from one pixel to the next. The PSF describes the spread of an electron impinging on the detector. With an ideal point spread function, each pixel functions independently of one another, but a spread in the PSF reduces the available pixel count. A narrow PSF gives rise to the highest modulation transfer function (MTF) at the Nyquist sampling frequency [174]. The MTF describes the ratio of the input and output of spatial modulation on a detector as a function of spatial frequency  $\omega$ :

$$MTF(\omega) = \frac{C'(\omega)}{C(\omega)}$$
(52)

where  $C(\omega)$  is the contrast of the true signal as a function of spatial frequency, and  $C'(\omega)$  is the contrast in the detected signal. It is related to the PSF by

$$MTF(\omega) = f(\omega) * FT[PSF(x)]$$
(53)

where  $f(\omega)$  is the Fourier transform (*FT*) of the detector pixel shape.

While the MTF has been shown to be almost ideal for the Medipix3 detector at low kV, it is considerably worsened at 200 kV and 300 kV, typical high operating voltages to reduce radiolysis damage in organic structures. The degradation is due to travel of the primary and secondary electrons in the silicon sensor of the Medipix3, and the MTFs for 80 and 300 kV were measured and are shown below in **Figure 2.17**. The detector performance is considerably degraded at 300 kV. At 80 kV, the use of the charge-summing mode (CSM) [31] will give near perfect DQE and MTF, as it uses a hardware algorithm to integrate areas within a 2x2 grid. As the PSF contains virtually zero contribution at >1 pixel at 80 kV, this mode improves detector performance. In contrast, at 300 kV, the effect of this mode would simply be to give multi-pixel excitation with some spatial separation between the excited pixels. Instead, multiple pixels will be excited (up to 6 pixels away) by incident electrons at 300 kV. It is currently unclear whether the best SNR for low-dose imaging is at 300 kV or lower due to degraded detector performance at high accelerating voltage.



**Figure 2.17**. The MTF (a) and PSF (b) of the Medipix3 detector at 80 (blue) and 300 kV (orange) measured using a knife-edge, compared to the ideal case (red) in single-pixel mode. The detector thresholds, which is the amount of charge that has to be deposited in a single pixel for the detector to count an event, were 20 kV and 50 kV respectively. The detector behaves in a near-ideal way at 80 kV, but at 300 kV the pixel spread reaches up to six pixels away from original incidence. Note that in the PSF plot, the ideal PSF corresponds to 100% of pixels being classed at their correct pixel.

# 2.8 Electron Beam Damage

The imaging of soft materials in the electron microscope is one of dose-limited microscopy. The recordable signal is not limited by electron optics, but by the signal that can be extracted prior to structural destruction and subsequent loss of meaningful signal. Electron-induced damage is a well-studied phenomenon in the TEM [125]. Several types of damage can occur leading to the amorphisation of the probed structure, typically classified into two primary categories: displacement damage and radiolysis damage. Understanding beam damage is particularly significant here since the materials probed are beam-sensitive; they have low damage thresholds. As a result, the current used is low, practically controlled by the use of small condenser apertures and either low monochromator current or small spot size, depending on the microscope. The monochromator affects the current by demagnifying the gun image. The downside of using a low beam current to limit exposure is a loss of SNR in the data. This problem is unavoidable, and hence machine learning based data reduction techniques are on occasion used to extract higher SNR data, as detailed in Chapter 3.

### 2.8.1 Types of Electron Beam Damage

The most common types of damage in the TEM are displacement damage and radiolysis. Sputtering, specimen heating and hydrocarbon contamination are also potential problems. The first two will be discussed here, with more thorough analysis in Egerton *et al.* [125]. In general, for low *Z* materials, bond breaking caused by radiolysis damage is the limiting factor for signal-to-noise ratio (SNR) in acquired data. Typically damage resistance is measured as a critical dose, defined as the complete destruction of the electron diffraction pattern of a crystalline or partly crystalline sample. This critical dose is temperature-dependent, with cryogenic cooling increasing the critical dose [126].

### 2.8.1.1 Displacement Damage

Displacement damage, also called knock-on damage, occurs due to the transfer of energy to atomic nuclei during elastic scattering, and is significant at high accelerating voltage. The energy transfer in this 'pseudo-elastic' scattering is typically less than 0.1 eV, although can be several eV at high angles [175]. It occurs for all specimens, typically above an element-specific accelerating voltage, although generally requires a large current to be significant. At 200 kV, carbon- and silicon-based structures will be damaged. Since the majority of the probed species in this work are carbon based (such as the probed metal-organic frameworks), displacement damage is expected to occur. However, the more significant damage mechanism is that of radiolysis damage.

#### 2.8.1.2 Radiolysis Damage

Radiolysis, also called ionisation, is the result of inelastic scattering of the electron beam, either by single electron excitation or sometimes through plasmons. This single electron, which may be an excited core electron or a valence band electron typically has between a few eV and a few keV of energy. In highly conductive materials, electrons from above the Fermi level will typically fill these gaps in < 1 fs, resulting in little change to the material. However, in the case of insulators or semiconductors, this time scale is much slower, usually > 1 ps, but occasionally up to µs, and as a result, de-excitation will also occur both through phonon scattering and additional secondary electron emission [176]. As a result, permanent change to the electron potential may occur; electrons do not return to the ground state, bonds break and the crystal structure becomes amorphized with increasing dose. The radiolysis effect is small in good conductors, but unfortunately most organic compounds are not such.

Radiolysis is the more significant damage mechanism in insulating organic materials [125] in contrast to inorganic compounds where mass loss of volatile species is typically dominant. In general it is favourable to operate at high accelerating voltage to reduce the inelastic scattering cross-section and subsequent amount of radiolysis damage. This delays the loss of crystallinity and medium range order (MRO). Among organic materials, a range of critical doses  $D_C$  comprising orders of magnitude are observed. In general, aromatic compounds have higher resistances, attributed to the stability of benzene and phenol rings. A few values at different accelerating voltages  $E_0$  are provided in **Table 2.2** for comparison [177]. Both radiolysis damage and mass loss can be reduced by cooling the specimen [178], [179]. Other options include reducing the beam current and reducing the convergence angle to spread the beam over a larger area. Curious results have been observed regarding damage in TEM versus STEM mode. For example, the amorphisation dose increased by several orders of magnitude on polymers by reducing beam diameter from 1 µm to 1 nm, suspected to be partially due to damage in radiolysis not being localised to the probed area. The general trends are not yet known, but there is suggestion that a dose exceeding the critical dose can be used, provided the scan is fast enough.

Material	$\boldsymbol{E_0}$ (kV)	<b>D<sub>C</sub></b> (e⁻/Ų)
Glycine	60	0.9
Polyethylene	100	6
Phthalocyanine (Pc)	60	60
Cu-Phthalocyanine (Cu-Pc)	60	940

Table 2.2. Critical Dose D	c for a few selected	materials at selected	accelerating	voltages $E_0$ .
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# 3 Statistical Methods

# 3.1 Unsupervised Machine Learning

Unsupervised machine learning methods are well suited for signal separation analysis [180], particularly here since supervised machine learning methods are not applicable due to a lack of training sets. Since the number of local measurements is in general much greater than the number of distinct microstructural elements, there is significant redundancy in the measured microscopy data. The goal of using machine learning methods is to uncover these underlying physical signals, and unsupervised machine learning methods exploit this redundancy automatically when finding common signals in the data [58]. Such methods have been applied in electron microscopy to separate signals corresponding distinct microstructural elements, such as crystal phases or strained regions [143], [181]–[185]. In crystalline SED, such methods can also be used to unmix signals from regions where multiple microstructural elements have been sampled. This unmixing is typically referred to as blind source separation (BSS) and is of interest in the mapping of structural domains in data. BSS is the process of separating source signals from mixed signals without knowledge of the underlying source signals. BSS problems are generally highly underdetermined, but useful solutions can frequently be extracted. Such methods have also been applied to non-crystalline SED analysis [186].

Most of the methods used rely on linear matrix decomposition methods such as principal component analysis (PCA), independent component analysis (ICA), and non-negative matrix factorisation (NMF). These linear methods describe data by a linear combination of latent variables that are learned from data. Since the number of physically distinct signals in the data is comparatively low, the features in the data should be well approximated using a small number of basis vectors. These basis vectors can be learned from the data, and with appropriate constraints, interpreted as physical signals. The basic form of these linear matrix decomposition methods is that the data matrix *X* can be represented as a linear sum of a basis vector matrix *W* and a corresponding coefficient matrix *H*, showcased in **Figure 3.1**. Mathematically *X* is given by:

$$X = WH$$

(54)

What differentiates PCA, ICA, and NMF are the restrictions placed on the basis vectors  $w_i$ . It is worth briefly mentioning the concept of reconstruction error. Unless the number of basis vectors is equal to the number of distinct data points, the reconstruction of X from the basis vectors is not perfect. Then the reconstruction error can be defined as:

$$\|\boldsymbol{X} - \boldsymbol{W}\boldsymbol{H}\|_F^2 \tag{55}$$

where  $||A||_F$  is the Frobenius norm of matrix A, which is defined as the square root of the sum of the absolute squares of its elements:

$$\|\boldsymbol{A}\|_{F} = \sqrt{\sum_{i} \sum_{j} |a_{ij}|^{2}}$$
(56)

The reduced decomposition factors are calculated by minimising the reconstruction error, although not necessarily directly by a least-squares minimisation. It is worth noting that NMF has not been used on the PDFs, despite being a generally efficient algorithm when analysing diffraction data [187], as the machine learning methods were generally applied to the reduced PDF, where the non-negative assumption of NMF is not appropriate.



**Figure 3.1**. Unfolding of a multidimensional data set **X** with **Y** real-space pixels into a weights matrix W and a components matrix H through PCA.

## 3.1.1 Principal Component Analysis

A more mathematically thorough definition of PCA and ICA is provided in [58]. The important parts for SED data are summarised below.

In PCA, the covariance matrix of the data is decomposed into a series of eigenvectors called "principal directions", where the data projected along these directions are called the principal components. These principal directions are required to be orthonormal and are the directions along which the covariance of the data shows maximal variance. It is hence key to standardise the data before

decomposition, or the data will tend to be biased towards direction with high measurement scale. Most importantly, in PCA the principal directions can be ranked using their eigenvalues, and often a truncation is intentionally applied to reduce signals that likely corresponds to noise or measurement error. The error in the reconstruction is then equal to the sum of the eigenvalues discarded and should be chosen such as to be small.

PCA can also be used as a denoising tool [188] by performing a PCA decomposition on the data, followed by a reconstruction using only components with high explained variance ratio. Effectively, components which explain only small parts of the variation in the signal are discarded. This typically corresponds to noise, as the noise varies randomly at each real space pixel. In this sense, PCA effectively averages the noise over the entire recorded data set, exploiting the redundancy as mentioned before. The reconstructed data set formed of only the `non-noise' signal components hence has much lower noise

PCA does not find physically relevant signals, as it maximises the explained variance of each component. For example, in a sample composed of two different signals with mixed regions a two-signal PCA decomposition would return principal components approximately corresponding to a summed and a difference signal. This contrasts with ICA (detailed below) which returns statistically independent signals.

## 3.1.2 Independent Component Analysis

ICA is a blind source separation (BSS) technique, related to PCA. In ICA, there is no ranking of components, and instead of maximising the variance of each component, instead the non-Gaussianity [180] is maximised. In practice, this results in signals that are statistically independent from each other in the data set. Hence, ICA can return physically relevant signals, although care should be taken as this is not necessarily true, particularly if distinct signals in the data are very similar. It is nonetheless a powerful tool.

A problem with ICA is that noise covariance in the data is only partially identifiable [189]. In effect, noisy data cannot be unmixed effectively. This is not a concern if each pixel has a high SNR, but at the low SNR associated with low-dose SED this quickly becomes a problem. To circumvent this problem, a PCA decomposition can first be applied to the data, to remove noise, and ICA then applied to the reconstructed data in order to extract the independent signals present in the reconstructed data. This combined PCA-ICA method is powerful as a low-noise BSS algorithm.

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# 3.2 Fluctuation Electron Microscopy

Fluctuation electron microscopy (FEM) is a probe of medium range order in a material [190]. In FEM, spatial fluctuations in non-crystalline materials are observed, which appear as `speckle' fluctuations in diffraction patterns. An example of such variation is shown in Figure 3.2. Mathematically, the 'speckliness' of an image can be estimated from the mean-squared normalised variance of the intensity (between spatial locations). This works well provided there is no change in sample thickness, composition, or illuminated intensity across the illuminated area. Treacy and Gibson showed that additional maxima are seen in the variance pattern when compared to the diffraction pattern, corresponding to higher order reflections due to medium range order [191]–[193]. Medium range order is notoriously difficult to probe in materials, as it is on too small a scale for bulk methods, such as X-ray PDF, but on too large a scale to be probed by a standard electron beam in STEM mode. Furthermore, as shown by Krivanek et al [194], standard TEM imaging of amorphous samples is not readily interpretable. They randomised the phases of the Fourier components in a TEM image of an amorphous material, recovering an image that is qualitatively similar. The "fringiness" of the image is primarily an artifact of the contrast transfer function of the microscope. In fact, a random array of pixels passed through the contrast function of a TEM setup again qualitatively resembles an actual image of amorphous material [190]. FEM, like the PDF, remains one of the only effective methods of probing medium range order in non-crystalline materials.

Treacy and Borisenko showed that medium range order is pervasive in amorphous silicon using FEM, as it encodes information not seen in a bulk PDF scan [1], suggesting that the structure of amorphous silicon is best described using a paracrystalline structural model [195]. Such analysis has also been applied to other amorphous materials, such as Al-RE (RE: rare earth element) films [196], [197] and phase-change alloy  $Ge_2Sb_2Te_5$  [198]. Some materials do show essentially no speckliness, such as amorphous silica and some amorphous carbon films [7], consistent with a continuous random network structure, but speckliness is more the norm than the exception.

The exact theory of scattering fluctuations in the kinematical approximation is relatively complex and will not be covered in detail. The reader is instead pointed to [190]. The fundamental quantity is that of the normalised variance Var(s) is given by

$$Var(\mathbf{s}) = \frac{\langle I^2(\mathbf{s}) \rangle}{\langle I(\mathbf{s}) \rangle^2} - 1$$
(57)

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where  $\langle a \rangle$  refers to the mean of a. The variance is mean-squared normalised to eliminate the dependence of the speckle on the sample thickness. Through theory, it is possible to show that the variance is dependent on 4-body correlations in the structure. The variance is usually unaffected by aberration terms, as FEM studies tend to use a broad beam and operate relatively far from the resolution limit. An example of a calculated variance pattern is shown in **Figure 3.2**.

In practice, the variance in an FEM image can be acquired in two equivalent ways. Either in TEM mode, where a series of images is taken with varying diffraction vector s, or in STEM mode, where a series of diffraction patterns are acquired at varying locations r. The latter is used in this work, as its acquisition is functionally equivalent to an SED scan. The two are equivalent if the TEM mode scans all values of s, and the STEM mode all values of r.



**Figure 3.2**. FEM Data Acquisition: (a-c) Data "speckle", the pattern varies as successive pixels are scanned. Data from Sample A1 in Section 7.1. (d),(g) Mean pattern and radial profile from sample A1 in Section 7.1 (e),(h) Normalised variance pattern from the same data as (d) and the corresponding radial profile below. Note that the total variance grows without bound as the mean decreases. (f),(i) "Corrected" mean variance pattern acquired by subtracting a term equal to 1/mean and corresponding radial profile. Note the high spottiness. All the patterns have the shadow of the HAADF detector on the detector plane, hence the intensity goes to zero suddenly at high s.

# 4 Scanning Electron Pair Distribution Function Analysis

# 4.1 Background and Motivation

Metal-organic frameworks (MOFs) are highly porous crystalline structures consisting of metal centres and organic ligands. Most exhibit pores on the nanoscale and the resulting extremely high internal surface area means MOFs are of particular interest in gas storage, such as for H<sub>2</sub>, CO<sub>2</sub>, CH<sub>4</sub> [199]–[202], and more toxic gases such as PH<sub>3</sub> and BH<sub>3</sub> in the semiconductor industry [203]. MOFs have multitudes of other applications, such as in selective gas separation [204]–[207], heterogenous catalysis [208]– [210], water harvesting [211], and drug delivery [212]. The microporosity responsible for the wide variety of applications can be designed through chemical and structural variation controllable in turn by coordination chemistry between ligand and metal sites [213]. This tunability is a particular strength of MOFs; over 80,000 MOF structures are reported in the Cambridge Structural Database [214].

The search for MOFs with ever greater accessible surface area has produced new challenges. Producing bulk-scale MOFs in volumes beyond a few mm<sup>3</sup> that retain porosity without undergoing mechanical or thermal collapse, and ideally with little to no large-scale defects such as high-angle grain boundaries has proved to be a particular challenge. This causes further problems, as the manufacturing of less stable MOFs by traditional ball milling and sintering techniques can destroy the porous crystalline structure. To answer this, there has been significant development of MOF glasses [215] and methods to produce them, such as blending [216], flux melting [217], and crystal-glass composite formation [22]. These blending methods have enabled the production of shapable bulk glasses with intermediate mechanical properties [218], improved gas uptake [219], and chemical stability [220]. Manually engineering defects and disorder [221] has introduced further tuning, such as improving accessible surface area in Fe-BTC [222] or by creating coordinatively unsaturated metal sites for catalytic applications [223]. These structural properties are direct results of the phase distributions and interfaces within these materials.

However, characterisation of these non-crystalline structures has proved challenging due to the "Nanostructure Problem", as noted in Section 1.2. X-ray and neutron PDF has played an integral part in characterizing both amorphous MOFs and varying degrees of disorder in MOF crystals [218], [223]–[225]. However, this has been limited to single-phase PDF studies, as the unmixing of multiple PDF signals from a single scattering signal is extremely difficult, although has been shown to be possible with compositional series [226]. These multi-component MOF glass-derived composite structures,

however, pose a significant challenge to unambiguous structural characterization due to PDF signals containing unresolved and overlapping features from the constituent phases. To showcase the structural separation of the glassy phases in a MOF-inorganic glass composite, electron PDF analysis was employed. Important questions on the length scale of interface structures can be answered without needing to solve the unmixing problem by obtaining spatially resolved-PDFs in SED. Such analysis has been successful previously at high electron doses [186], [227], [228], but not for beam-sensitive materials.

Developing such methods further enables the application to a wide variety of amorphous composites. Amorphous composites with microscopic to nanoscale variation encompass bulk metallic glasses [229], [230], ceramic composites such as those prepared by sintering [231]–[233], phase separation processes in silicate glasses [28], [234], nanoscale dispersions [235], and polymers and polymer blends [236].

## 4.1.1 Zeolitic Imidazolate Frameworks

Zeolitic Imidazolate Frameworks (ZIFs) are the dominant category of glass forming MOF structures [223]. These 3D-networks are composed of charged transition metal centres linked by imidazolate ( $Im^-$ : [ $C_3H_3N_2$ ]<sup>-</sup>)-based ligands (see **Figure 4.1**). In the crystalline state, they adopt structures similar to zeolites due to the similarity in coordination geometry and bridging ligands. Intriguingly, the melting point of Zn based ZIFs, such as ZIF-4 [215] can be controlled by substitution of the  $Im^-$  by more strongly bonding ligands, such as benzimidazolate ( $bIm^-$ : [ $C_7H_5N_2$ ]<sup>-</sup>). ZIFs can thus be controllably melt processed and quenched to form glassy states. However, these glassy states suffer from the same characterisation difficulties as all glasses do as detailed in Section 1.2, and most characterisation is done using bulk synchrotron XRD methods [223]. In this report, we use a nanocomposite containing ZIF-62 [237], a ZIF with ultrahigh glass-forming ability [238] and a chemical formula of Zn( $Im_2$ -xbIm\_x).



Figure 4.1 An imidazolate molecule linked to two Zn ions.

# 4.2 Sample Preparation

The samples for the following section were prepared by Dr Louis Longley, with the help of Dr Thomas Bennett, Dr Courtney Calahoo, and Dr Lothar Wondraczek.

Samples of a ZIF-62-glass —  $0.67([Na_2O]_{0.9}[P_2O_5])-0.33([AlO_{3/2}][AlF_3]_{1.5}-glass composite were made, through a complex thermal and pressure treatment described in detail below. The two phases were heavily intermixed during processing and subsequently ground up in a pestle and mortar to prepare them for viewing in a TEM. ZIF-62 consists of <math>Zn^{2+}$  ions connected by imidazolate (Im,  $C_3H_3N_2^-$ ) and benzimidazolate (bIm,  $C_7H_5N_2^-$ ) linkers in a 1:1.75:0.25 ratio  $[Zn(Im)_{1.75}(bIm)_{0.25}]$  with a *cag* topology in crystalline form.  $a_gZIF-62$  is used to refer to the amorphous ZIF-62 phase. The inorganic 0.67( $[Na_2O]_{0.9}[P_2O_5]$ )-0.33( $[AlO_{3/2}][AlF_3]_{1.5}$  is a complex amorphous material with no found crystallinity. This glass-glass mixture was then examined in the TEM. The X-ray and electron samples presented below were made from two separate filtrations of the  $a_gZIF-62$ . The more detailed procedure is given below.

**ZIF-62:** Crystalline ZIF-62 was synthesized by adding zinc nitrate hexahydrate (1.65 g, 5.55 mmol), imidazole (8.91 g, 131 mmol) and benzimidazole (1.55 g, 13.12 mmol) to DMF (75 ml). The mixture was then heated at 130°C and stirred for 48 hours. The resultant product was washed with DMF (2 x 20ml) under vacuum to obtain a white crystalline powder (yield 42.6% from moles zinc). To increase the yield the filtered reaction mixture was placed back into the oven at 130 °C for a further 48 hours and then more product obtained through washing under vacuum twice with DMF (20 ml). Two filtrations of ZIF-62 were used, one for the X-ray and one for the electron microscopy samples.

**0.67([Na<sub>2</sub>O]**<sub>0.9</sub>[**P**<sub>2</sub>**O**<sub>5</sub>])-**0.33([AIO**<sub>3/2</sub>][**AIF**<sub>3</sub>]<sub>1.5</sub>): High purity reagents (optical grade) of NaPO<sub>3</sub> and AIF<sub>3</sub> were melted in a Pt crucible in an electric muffle furnace. Due to the known volatility of fluoride, care was taken to initially melt the mixture at 800°C for one hour to allow NaPO<sub>3</sub> to melt and dissolve the AIF<sub>3</sub> before heating to a temperature of 1000°C for half an hour to achieve complete homogeneous dissolution. To prevent crystallization, the inorganic glass required quenching between two copper metal plates, and then was annealed at Tg + 50°C (460°C) for one hour before returning to room temperature. The small pieces were then pulverized by ball-milling for 30 minutes in a Retsch PM 100 grinder at 350 rpm with 1 min intervals using Si<sub>3</sub>N<sub>4</sub> balls (with roughly equal sample and ball volume).

**0.67([Na<sub>2</sub>O]<sub>0.9</sub>[P<sub>2</sub>O<sub>5</sub>])-0.33([AlO<sub>3/2</sub>][AlF<sub>3</sub>]<sub>1.5</sub>): ZIF-62 50:50 wt% Composites:** Approximately 300 mg crystalline ZIF-62 and 300 mg  $0.67([Na<sub>2</sub>O]_{0.9}[P<sub>2</sub>O<sub>5</sub>])-0.33([AlO<sub>3/2</sub>][AlF<sub>3</sub>]_{1.5})$  powders were mixed together by ball-milling in a Retsch MM400 mixer mill with a 15 ml stainless steel jar and one 5 mm

stainless steel ball bearing for 5 minutes at 25 Hz. A 200 mg sample of the ball milled powder mixture was placed in a 13 mm die and compacted into a pellet using 10 tonnes of pressure applied for one minute. This pellet was placed in a tube furnace (Carbolite 12/65/550) which was left to equilibrate under argon for one hour before heating to 410°C at 10°C/min and holding for 1 minute. All heat treatment was done under constant argon flow. The heat-treated pellet was left to cool under argon at the natural rate of the tube furnace; the sample was removed from the tube furnace at temperatures equal to or below 200°C.

# 4.3 Practical Acquisition of Electron PDF

## 4.3.1 Pre-processing

To acquire electron PDFs (ePDFs) from an SED scan, first a calibration of the scan step size and diffraction pattern pixel size was performed using a standard 500 nm gold diffraction grating replica with latex spheres (Ted Pella). The gold diffraction data was also used to determine residual elliptical distortions in the diffraction plane due to post-specimen optics.

The SED data was then pre-processed in Pyxem [136] as follows, partially using code written as part of this project:

- Centring the direct beam in each diffraction pattern using a cross-correlation routine to remove the effects of the beam-tilt impurity.
- 2. Applying an affine transformation to correct for elliptical distortion.
- 3. Integrating each diffraction pattern azimuthally around the centre of the direct beam to acquire a 1D scattering profile using PyFAI [239].

### 4.3.1.1 Choice of Camera Length

ePDF analysis has a slightly different camera length requirement to typical SED data. The two key parameters for  $s = 2 \sin \theta / \lambda$  are the reciprocal space resolution  $\Delta s$  and the reciprocal space limit  $s_{\text{max}}$ , which control the extent of the PDF  $r_{\text{max}}$  and the real space resolution  $\Delta r$  respectively [159]. The ideal  $s_{\text{max}}$  is higher than in typical SED data, as the high-frequency real space information is coded at high s. Ideally, an  $s_{\text{max}}$  of 2.5 Å<sup>-1</sup> or more is used. Note that the corners of the detector rarely extend to  $s_{\text{max}}$  in practice due to the lower signal-to-noise ratio (SNR). This inevitably results in a poorer reciprocal space resolution  $\Delta s$ , particularly as the observed signal is convolved by the probe function of the electron beam. However, a bigger loss of reciprocal space resolution o.ccurs due to the reduced MTF performance of the Medipix3 detector at > 80 kV, detailed further in Section 2.7.2 and in [240]. This spread is independent of the camera length and results in a degradation of the camera resolution, which is exacerbated at low camera length.

### 4.3.1.2 X-ray Diffraction Acquisition

X-ray data presented below were collected at the I15-1 beamline at Diamond Light Source, UK ( $\lambda$  = 0.161669 Å, 76.7 keV) by collaborator Louis Longley. Samples were loaded in borosilicate capillaries of 1.17 mm inner diameter. Data were collected in the region of ~0.4 < Q < ~22 Å<sup>-1</sup>. Correction for the background, multiple scattering, container scattering, Compton scattering, fluorescence and absorption were performed using the GudrunX program [241].

### 4.3.2 Background Fitting

In an ideal world, the subsequent acquisition of an electron PDF is done by fitting an accurate background to the scattered intensity I(s) as described in Section 2.6. I(s) is ideally normalised to obtain the reduced intensity  $\varphi(s)$ , defined as

$$\varphi(s) = s \frac{I(s) - N < f_i(s)^2 >_i}{N < f_i(s) >_i^2}$$
(58)

where  $f_i(s)$  is the average single atomic scattering factor of the constituent element i [96],  $\langle \cdots \rangle_i$ denotes the sum over  $f_i(s)$  (or  $f_i(s)^2$ ) weighted by the atomic fraction of element i, and N is a fitted scattering parameter related to the number of scattering atoms.  $N < f_i(s) >_i^2$  represents the sample's "structureless" or mean scattering factor, different for X-rays and electrons (see Section 2.1.2.1) [159].

However, in practice, there is some detector noise in the acquired data, meaning that the experimental reduced intensity  $\varphi_{\exp}(s)$  is instead fitted as

$$\varphi_{\exp}(s) = s \frac{I(s) - (N < f_i(s)^2 >_i + C)}{N < f_i(s) >_i^2}$$
(59)

where the additional parameter *C* is an additive constant that accounts for detector noise (close to zero for Medipix3).

For the Medipix3, fitting *C* is straightforward. However, in practice the theoretical scattering factor  $N < f_i(s)^2 >_i$  does not fit the measured intensity at both low and high *s* simultaneously. This is seen clearly in **Figure 4.2** and is a result of thickness-dependent multiple and inelastic scattering, particularly multiple inelastic scattering, a discussion of which is given in Section 2.1.6. The form of the divergence does not follow an analytical equation, and as a result, one must choose whether to fit the profile to low or high *s* even in thin samples (ca.  $t < 0.5\lambda_i \approx 0.15\lambda_e$ , where  $\lambda$  is the mean free path of inelastic of elastic scattering respectively). Fitting the high *s* region is generally accepted as causing less error in the PDF [159], [168], [231] due to a division by the fitted profile accentuating misfit at high *s*. This choice results in a low-frequency artifact in reciprocal space (**Figure 4.3**d) that in turn results in small changes to the observed peak positions in the PDF.

This error is unavoidable if the fitting is to be done by a physically-appropriate scattering factor, and has been previously noted [168], [227]. As a solution to this problem, it has been suggested to use a 4th order polynomial function that goes to zero at s = 0 and  $s = s_{max}$  to fit the reduced intensity after fitting the appropriate atomic scattering factor [242]–[244]. Using too high an order of polynomial would fit the structured scattering rather than the baseline, while too low an order of polynomial does not accurately reflect the baseline. A 4<sup>th</sup>-order polynomial has been found to be a good intermediate. This polynomial shifts the baseline to get a reduced intensity that does, on average, oscillate around zero as it physically should. An example is shown below in Section 4.3.3. This results in more accurately reproduced reduced intensities, as shown in **Figure 4.3**. However, the peak heights in the reduced intensity do vary with thickness, and hence it has been proposed to normalise the height of the resulting ePDF to the height of the first peak for more accurate comparison [168]. Note that this is an error rarely encountered in X-ray PDF analysis due to the much lower overall scattering cross-section resulting in virtually no multiple scattering.



**Figure 4.2** (a) Azimuthally-integrated profiles (solid) and their best fit (dashed) from thick (dark) and thin (light) regions of the  $a_g$ ZIF-62. The profiles were fitted to ensure correspondence at high s as explained. The profile diverges considerably more in the thick sample. (b) The natural logarithms of the same profiles. The divergence from the fit is clearer, particularly occurring at low s as expected. The thick sample does not follow the expected profile shape.



**Figure 4.3** A more detailed analysis of the thickness dependence of the reduced intensity across a variety of  $a_g$ ZIF-62 thickness (a). (b) The slope in the logarithm of the experimental profile shape changes with thickness, while (c) the fitted profile is unchanged. (d) The acquired reduced intensity as a function of thickness, with a noticeable undershoot at low *s* due to the poor fit. (e) The same reduced intensities when corrected by a 4<sup>th</sup>- order polynomial. A previous shift becomes an approximately multiplicative error.

The variation of the acquired ePDFs as a function of thickness was also measured for both the  $a_g$ ZIF-62 and the inorganic glass and are given in **Figure 4.4**. A 4<sup>th</sup>-order correction was applied. The average peak height decreased with increasing thickness, but the peak positions remained consistent as expected [60]. Minor peak shifts (< 0.2 Å) were observed for the inorganic glass, likely due to its greater density and inelastic scattering cross-section, resulting in faster changes with increased thickness relative to  $a_g$ ZIF-62. The consistency in peak position and profile shape is important for the PCA/ICA decomposition as it should primarily affect the total weight of each phase assigned to each pixel, but not result in pixels being misclassified as one phase over the other. At large thicknesses, particularly in the inorganic glass, this assumption is no longer fully satisfied, but the primary peak positions are still well defined, and so even thick regions can be separated using PCA/ICA (Section 4.4 below).



**Figure 4.4** A series of ePDFs acquired at different thicknesses of the inorganic NaPO<sub>3</sub>-AlF<sub>3</sub> glass (b) and the  $a_g$ ZIF-62 (c) from the areas shown in (a), overlaid with the area-averaged ('full') ePDF from the results below. Color coding follows (a). With increasing thickness (more yellow), the peaks become weaker and more blurred, with notable slight shifting and blurring of the peaks around 3.8 Å and 5 Å in the inorganic glass. The PDF from the whole area (dashed) has an intermediate result as it is acquired from a variety of thicknesses.

### 4.3.3 Damping the Reduced Intensity

To obtain a PDF, the Fourier transform of the reduced intensity should be calculated. However, due to the finite data acquisition range and low SNR at high s, directly taking a Fourier transform of the obtained reduced intensity will cause a highly oscillating PDF. This high noise level is due to a large Poisson uncertainty in the recorded signal at high s, such that the Poisson uncertainty is much higher than the mean. Subsequent division to obtain the reduced intensity causes large oscillations in  $\varphi_{exp}$ . It is thus necessary to damp the reduced intensity. Several methods can be used but all of them can cause a loss of information in the PDF, requiring a fine balance between reducing noise and losing signal. A decaying exponential term  $exp(-bs^2)$  [135] with a variable term b or a Lorch function [159] is typically used to force the PDF to near zero at  $s_{max}$  and the data is extrapolated from the edge of the direct beam to zero. This parameter is called s<sub>min</sub>. By forcing the function to zero at both s<sub>min</sub> and s<sub>max</sub>, large termination ripples that arise when taking a Fourier transform of the reduced intensity with a finite s-range are avoided [159]. The exponential was used for SED data, and the b term in the exponential  $exp(-bs^2)$  was typically of the order of 0.5-1.0. This term varied between the areaaveraged PDFs (low) and the per-pixel PDFs (high) due to differences in their signal-to-noise ratios. The value was adjusted until the spurious high r oscillations in the PDF were removed (typically identified by eye). A figure of the full procedure is given in **Figure 4.5**.

The need for damping depends on the data quality. Numerous methods were attempted, and our general recommended procedure for SED data is:

- 1. Extrapolation of the fitted profile to zero from the  $s_{min}$  chosen in 1 to remove effects of the direct beam.
- 2. Fit the 4<sup>th</sup>-order scattering correction with an  $s_{\min}$  just above the *s*-range affected by the direct beam and with an  $s_{\max}$  close to the maximum scattering vector. If the data is extremely noisy at high *s*, the spurious oscillations at high *s* can first be reduced by an exponential damping with a low coefficient ( $b \sim 0.1$ ).
- 3. Damp the high *s* oscillations using the exponential  $\exp(-bs^2)$  term [159] with 0.5 < b < 1.0. *b* should be chosen such as to damp 'noisy' oscillations without removing peaks in the data itself. Typically, I would recommend varying *b* and calculating the ePDF. The *b* at which high-frequency oscillations in the ePDF disappear is recommended.

Other damping functions such as the Lorch function exist [159], particularly in X-ray literature, many of which are incorporated in Pyxem.

### Choosing b and $s_{\max}$

Varying the strength of the damping parameter b and the value of  $s_{max}$  will affect the resultant PDF, as shown in **Figure 4.6** and **Figure 4.7**. Increasing b effectively reduces  $s_{max}$  by damping the information at high s. Ideally, they should be chosen such as to:

- 1. Ensure that the PDF approximates to a straight line with a negative gradient as it approaches s = 0. This gradient is linked to sample density.
- 2. Ensure that there are at most small oscillations at high *r* in the PDF.
- 3. Given the above constraints, the damping term should be as small as possible and  $s_{max}$  as large as possible.

One should iterate over the damping parameter and observe the resultant PDF in order to determine these parameters. Examples are given in the notebooks stored on Apollo (see Appendix B).


**Figure 4.5** Data processing and damping for the PDF. (a) Background fit (dashed) of the radial profile (full). (b) Extrapolation to zero (dashed) of the reduced intensity (full). The extrapolation is performed for values of *s* lower than the intersection of the two lines. (c) 4<sup>th</sup>-order polynomial fit (dashed) subtracted from the profile. (d) Damping (dashed) by either an exponential (red, b=0.6 in this case) or the Lorch function (blue), resulting in the final reduced intensity (e-f). (g) PDF of the two reduced intensities, acquired by a Fourier sine transform of (e) (red) and (f) (blue).



**Figure 4.6**  $a_g$ ZIF-62 PDF as a function of damping factor b, rescaled to the height of the first peak. I: Inset showing G(r) at low r without rescaling to the height of the first peak, the gradient observed is constant as is theoretically expected. II: Inset showing the 4 < r < 7 Å in more detail, where some inflection is observed. The feature is present until a damping factor of 3.



**Figure 4.7**  $a_g$ ZIF-62 ePDF variation as a function of  $s_{\min}$  and  $s_{\max}$ . Blue: reducing  $s_{\max}$ , Red: Increasing  $s_{\min}$ . For the data where  $s_{\min}$  was increased,  $s_{\max}$  is 2.0.

# 4.4 Results and Discussion

#### 4.4.1 Data Acquisition and Processing

4D-STEM data of a beam-sensitive nanocomposite of amorphous ZIF-62 and amorphous 0.67([Na<sub>2</sub>O]<sub>0.9</sub>[P<sub>2</sub>O<sub>5</sub>])-0.33([AlO<sub>3/2</sub>][AlF<sub>3</sub>]<sub>1.5</sub>) was acquired on a JEOL ARM300F STEM operated at 200 kV with a beam diameter of ca. 5 nm and a convergence semi-angle of ca. 0.6 mrad. The beam current was 3.4 pA and the frame rate was 1 ms per pattern, with slight overlap between the beam at adjacent pixels, leading to an electron dose of ca. 10  $e/Å^2$ , much lower than typically used in STEM imaging [244]. The diffraction patterns were acquired using a quad-chip Medipix3 direct electron detector with 512x512 pixels and a  $\Delta s$  of 0.011 Å<sup>-1</sup>. The four constituent read-out electronics are separated by a gap of 3 pixels, resulting in a total detector area equivalent to 515x515 pixels and leading to a maximum scattering vector magnitude  $s_{\rm max} = 2 \sin \theta / \lambda$  of 2.8 Å<sup>-1</sup> (corresponding to a maximum  $Q_{\rm max}$  =  $4\pi \sin \theta / \lambda$  of 17.6 Å<sup>-1</sup>). Two STEM-ePDF scans were acquired to gauge potential electron beam damage. Subsequently, correlative X-ray energy-dispersive spectroscopy (EDS) was performed to provide an independent method of confirmation of structural domains, as they correspond to compositional differences. For sufficient counts to be recorded in EDS, the current at the sample was significantly increased by using a larger condenser aperture. This was accompanied by beam damage and sample amorphisation. However, as shown in previous SED-EDS data on crystalline MOFs [22], this should not cause drift of the metal centres that are used for phase identification. These elements were Zn ( $a_q$ ZIF-62) and Al (inorganic glass).

Additionally, 256x256 patterns were acquired on the same microscope operated at 300 kV using a single Medipix3 direct electron detector with a beam diameter of ca. 5 nm and a convergence semiangle of ca. 0.6 mrad during a different session. The beam current was 4.6 pA and the frame rate was 1 ms per pattern, leading to an electron dose of ca. 13 e/Å<sup>2</sup>. The single Medipix3 detector had 256×256 pixels and a  $\Delta s$  of 0.016 Å<sup>-1</sup> leading to a maximum scattering vector  $s_{max} = 2 \sin \theta / \lambda$  of 2.1 Å<sup>-1</sup>. (corresponding to a maximum  $Q_{max} = 4\pi \sin \theta / \lambda$  of 13.4 Å<sup>-1</sup>).

The SED data was analysed as described above. The EDS data was analysed using Pyxem and HyperSpy [245]. A PCA and ICA decomposition as detailed in Section 3.1 was applied to the spatially resolved EDS data, due to a relatively low signal-to-noise ratio during acquisition. The SNR was limited by destruction of the inorganic glass under the electron beam. The decomposition has been shown to be effective at nanoparticle EDS analysis [246]. The ICA factors of the EDS signal from the particles were

then analysed to confirm the elements present, which matched with the compositions determined from X-ray analysis, and with the observed structural information.

### 4.4.2 ePDF-XPDF correspondence

For the most accurate comparison to X-ray data, area-averaged STEM-ePDFs were calculated through summing multiple diffraction patterns from each phase. The areas for summing were determined using PCA on the PDFs obtained at each pixel, detailed below, and were chosen to be thin where possible. The diffraction patterns were summed, with each ePDF contained approximately 3000 diffraction patterns, and then processed into ePDFs. These ePDFs were scaled and compared to PDFs acquired through X-ray total scattering [218] on separate samples, shown in **Figure 4.8**. The scaling was necessary due to the applied 4<sup>th</sup>-order correction. For both phases, the reduced intensities show agreement in peak positions, though relative intensities exhibit some variation, particularly in the inorganic glass. Some variation in intensity is expected, as the electron and X-ray scattering factors differ (**Figure 2.1**), meaning the two signals encode structural information slightly differently. An additional difference comes from a different  $s_{max}$  between the two acquisitions.

For the inorganic glass, the peak positions at 0.3, 0.7, 0.8, 1.3, and 1.5 Å<sup>-1</sup> are all accurately reproduced, although the relative intensities of the peaks at 0.7 and 0.8 Å<sup>-1</sup> are reversed with respect to each other (Figure 4.8a). Similarly, peaks at 2.6, 4.9, and 6.4 Å are present in both the ePDF and xPDF (Figure 4.8c) of the  $a_q$ ZIF-62. The peaks around 3.5 Å appear broadened in the ePDF relative to the X-ray PDF. For  $a_q$ ZIF-62, the peak positions in the reduced intensity at 0.6, 0.8, and 1.0 Å<sup>-1</sup> appear in both electron and X-ray scattering profiles (**Figure 4.8**b). However, the intensities of the 0.6 and 0.8  $Å^{-1}$  peaks, and the high s peaks (> 1.2 Å<sup>-1</sup>) are significantly reduced in the electron data. The high s attenuation is attributable to the acquisition conditions. Peaks in the PDF of  $a_q$ ZIF-62 (Figure 4.8d) can be assigned to N1-Zn (ca. 2 Å), C-Zn (N-C-N) (ca. 3 Å), N2-Zn (ca. 4 Å) and Zn-Zn (ca. 6 Å) (N1 and N2 are referenced to a selected Zn atom, with N1 directly coordinated to this Zn and N2 referring to the next nearest N atom to the same Zn) [22]. Additional contributions around 3 Å may be attributable to N-N distances [247], expected to be relatively stronger in ePDF. The remaining peaks are also in the correct positions, although the intensity around 6 Å is reduced in the electron data. A further peak is also present around 5 Å in the ePDF data, which has been seen in other XPDF work [248]. These notable intensity differences around 5 and 6 Å are attributed to relative differences in X-ray and electron scattering, with relatively stronger scattering of the associated Zn-H peaks around 5 Å [247] and relatively weaker Zn-Zn scattering around 6 Å expected in the electron PDF. Such differences between X-ray and electron PDF have not been previously reported but are also clearly visible in simulated electron and X-ray PDF intensities, which diverged notably around 5-6 Å for a ZIF-62 crystal (see Section 4.4.2.1). In addition, these 5-6 Å features changed very slowly with changing damping factor *b* and  $s_{max}$  (see **Figure 4.7**), all but ruling them out as an artifact of the data processing. The divergence arises from different atomic pairs that contribute to the PDF having different partial scattering factors as a function of *s*, shown in **Figure 4.9**. The X-ray scattering factors are dominated by Zn-Zn and C-Zn partial scattering factors, while the electron data is dominated by C-C and C-N scattering factors. X-H (where X is any elements) partial scattering factors are much higher in electrons than X-ray data, but still low in the electron data. The ratios of the electron and X-ray scattering factors are shown in **Figure 4.10**, and further highlight the reduced Zn and increased H contributions in electron scattering. Differences in PDFs due to different partial scattering factors are well known between X-ray and neutron PDFs [216], [249], but less studied between X-ray and electron PDF. This difference in scattering could be further exploited to characterize structures composed of multiple elements, subject to reduced inelastic and multiple scattering contributions, particularly as the largest differences in scattering among the elements in the sample are seen for features including hydrogen.



**Figure 4.8** Comparison of reduced intensities  $\varphi(s)$  and reduced PDFs G(r) from X-ray and electron scattering for (a), (c) the 0.67([Na<sub>2</sub>O]<sub>0.9</sub>[P<sub>2</sub>O<sub>5</sub>])-0.33([AlO<sub>3/2</sub>][AlF<sub>3</sub>]<sub>1.5</sub>) inorganic glass and (b), (d)  $a_g$ ZIF-62. The electron-derived  $\varphi(s)$  and G(r) are acquired by summing over pixels in a region of known composition. The  $a_g$ ZIF-62  $\varphi(s)$  and G(r) have both been rescaled due to thickness variation. The grey horizontal line marks zero, around which both the structure factor and the reduced PDF oscillate in all plots, with the structure factor going to zero at high *s*. The peaks for Zn-Zn and N-Zn have been labelled in (d).



**Figure 4.9** The relative partial pair scattering factors for the elements in the  $a_g$ ZIF-62, electrons (top) and X-ray (bottom). The sum of all scattering factors at each value in s is 1. Each scattering factor does decrease as a function of s, but at different rates.



**Figure 4.10** The natural logarithm of the ratio of the electron and X-ray partial scattering factors as a function of scattering vector *s* for  $a_g$ ZIF-62. (a) Full scale for *s* up to 3 Å<sup>-1</sup>. (b) Zoomed in for *s* up to 3 Å<sup>-1</sup>. Note that all intensities are ratios.

Overall, the greatest discrepancies are clearly seen at low s in the reduced intensity  $\varphi(s)$  (Figure 4.8b). This is to be expected, as the scattering profiles are fitted to minimize error at high s with its associated higher real-space frequency structural information. One of the effects of this data processing is to overshoot the scattering profile at low s, leading to a stronger damping of the signal there. In addition, some information is lost as a result of removing the direct beam from the electron data, which is larger than in X-ray scattering due to the finite convergence angle in SED, significant low-angle (mostly plasmon) scattering, and the MTF performance of the Medipix3 detector at 200 kV. The inelastic, multiple, and multiple inelastic scattering that cause the scattering factor misfit are unfortunately largely unavoidable, especially for typical sample thicknesses in the types of composite materials examined here. More quantifiable PDFs will be acquirable on thinner samples with consistent thickness, but micro- and nanoparticle samples often exhibit variable thickness, and additionally it is possible for very thin specimen PDFs to differ from the bulk due to incomplete orientational averaging [159]. Energy-filtered SED may offer further reductions in inelastic contributions, improving quantitative comparison with X-ray PDFs, but was unavailable on the used microscope. Because significant multiple and inelastic scattering contributions preclude quantification of intensities in the PDFs, no modelling of the ePDFs was done. However, the general correspondence between X-ray PDFs and STEM-ePDFs demonstrates that identifiable ePDFs can be obtained in total electron scattering. The consistency of the bulk X-ray PDF description with nanometer-resolved STEM-ePDF demonstrates a high degree of homogeneity in  $a_q$ ZIF-62, which was previously assumed but unverified. These results are particularly important for understanding  $a_g$ ZIF-62 and its composite formation processes. Such verification of a consistent bulk and nanoscale structure in composites is typically defined by diffraction in crystalline materials but has not been readily possible in amorphous analogues for MOF glass materials before.

#### 4.4.2.1 Simulating X-ray and electron PDFs

As noted in **Figure 4.9** and **Figure 4.10**, X-ray and electron PDFs are not expected to agree as different partial pair distributions are weighted differently at different values of *s*. For the  $a_g$ ZIF-62, scattering from Zn by electrons is particularly reduced, and scattering from H is increased from near zero to observable. To measure this effect, simulations were done on the  $a_g$ ZIF-62.

A simplified ZIF-62 crystal structure was used to calculate partial atomic pair distribution functions, which were then turned into partial reduced intensities and multiplied by the partial scattering factor distributions given above in **Figure 4.9**. The ZIF-62 crystal structure retrieved from the Cambridge Structural Database (CSD, CCDC number 671070) incorporates disorder in the benzimidazolate (blm) position as well as solvent molecules [237]. For ZIF-62 PDF calculations, a lowered symmetry structure was adapted from the same lattice parameters and derived from the fractional coordinates of the reported ZIF-62 structure, with partial occupancies and solvent molecules removed. This approach is similar to that used by Thorne *et al.* [248] though here we retain all hydrogen atoms in the structure.

The same data processing pipeline as for the respective X-ray and electron data was followed, and is presented in **Figure 4.11**a-b. The differences in the simulated electron and X-ray PDFs are very consistent with the observed differences, with the relative intensities of the 5 and 6 Å peaks accurately represented. The inaccuracies in peak intensities between simulated and observed PDFs are likely due to the lack of blm<sup>-</sup> linkers in the unit cell.

In addition, **Figure 4.11**c-f contains simulated electron and X-ray PDFs for a range of  $s_{max}$  values. The same processing was applied to each PDF, to ensure any differences are only due to different scattering factors. The partial structure factors were obtained, weighted, and damped by a Lorch function before being transformed into reduced PDF. Relative peak intensities differ throughout, but in a consistent manner. Peaks around 1.0, 1.5 Å and 5 Å are stronger in ePDF, while peaks are 2.6, 3.1, 4.9, and 6.4 Å are stronger in XPDF. Many of the distances stronger in the ePDF are associated with partial PDFs involving H, such as the C-H and Zn-H partials [247]. This suggests the difference in the electron and X-ray PDFs observed in **Figure 4.8** are real features that are differentially weighted in electron and X-ray scattering.

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**Figure 4.11** (a) Simulated and observed XPDFs of the described ZIF-62 unit cell, with the same processing conditions for each (b) The simulated and observed ePDFs, again following the same processing conditions. (c)-(f) Simulated X-ray (red) and electron (blue) PDFs for a given  $s_{max}$  using a Lorch function for damping. The  $s_{max}$  ranges from 2 Å<sup>-1</sup> (c) to 14 Å<sup>-1</sup> (f) (13 Å<sup>-1</sup> to 88 Å<sup>-1</sup> equivalent  $Q_{max}$ ).

### 4.4.3 ePDF Mapping via Unsupervised Machine Learning

The ability to construct area-averaged ePDFs from user-defined areas post-acquisition is a key benefit of acquiring ePDFs in STEM mode over TEM mode. However, any averaged method is still limited in spatial resolution by the need to spread out dose over multiple pixels to improve signal-to-noise ratio (SNR), in this case ca. 200 nm (Figure 4.12a). Diffraction patterns from individual probe positions in the STEM scan, in turn, have low SNR, such that single PDFs are too noisy to interpret directly. To overcome this problem, a linear matrix decomposition using Principal Component Analysis (PCA) was performed [250], [251]. This extracts common signals within the data that account for a disproportionately large fraction of the variance in the data. This was followed by an independent component analysis (ICA) blind source separation on the extracted PCA components to convert them into independent additive signals, rather than sums and differences recovered by PCA. The areas outside the particles were masked out, as was the data in the PDF below 2 Å. Both of these masks improved the signal separation in ICA. The matrix decomposition was performed on the PDFs as they are less susceptible to thickness-variation than the reduced intensities, as is discussed further below in Section 4.4.7. PCA, ICA, and related matrix factorisation approaches have shown success in unmixing signals in ePDF of bulk metallic glasses [252], though at much higher electron doses than used in the present study, as well as in X-ray PDFs of compositional series with known endmembers [226]. The PCA was applied to the 3D data cube containing the PDFs, although with pixels rebinned 2×2 from the original data (such that 4 pixels were summed together to improve SNR).

The applied PCA-ICA approach revealed two factors with variance above the noise floor. These ICAePDFs are shown in **Figure 4.12**e-f. The ICA-ePDFs were calculated using an average composition of the constituent phases, as their spatial distribution was unknown. Importantly, using the average composition did not change peak positions (Section 4.4.6). The ICA-ePDF signals show clear similarity with the area-averaged ePDFs, with the same peak positions reproduced in the ICA-ePDF as in the area averaged data. Additionally to recovering the ePDF profiles, ICA also recovers a corresponding map of the linear weights of each component. These spatial distributions recovered by ICA applied to the STEM-ePDF data clearly agree with EDS, showing consistent phase separation of the atomic structure and the corresponding compositional differences for the *a*<sub>g</sub>ZIF-62 and the 0.67([Na<sub>2</sub>O]<sub>0.9</sub>[P<sub>2</sub>O<sub>5</sub>])-0.33([AlO<sub>3/2</sub>][AlF<sub>3</sub>]<sub>1.5</sub>) glass. In the ICA-ePDFs, minor variations in peak height are seen around 2-4 Å, likely due to slightly imperfect unmixing of the two signals in the thicker sample regions. ICA assumes a linear summation of the separated signals, which is known to be approximate but not entirely precise for PDF analysis [159]. In addition, multiple scattering leads to thicknessdependent behaviour in the ePDF [168], discussed further below. Nevertheless, the maps of



**Figure 4.12** (a) A virtual annular dark field STEM image, constructed from the SED dataset by summing the pixels with 1.5 Å<sup>-1</sup> < s < 2.5 Å<sup>-1</sup> for each real-space pixel. The red and blue rectangles mark the regions used for area-averaged reduced ePDFs in **Figure 4.8** for the 0.67([Na<sub>2</sub>O]<sub>0.9</sub>[P<sub>2</sub>O<sub>5</sub>])-0.33([AlO<sub>3/2</sub>][AlF<sub>3</sub>]<sub>1.5</sub>) inorganic glass (red) and  $a_g$ ZIF-62 (blue). (b) EDS map of Al (red) and Zn (blue) from the data set, taken after the STEM-ePDF data. Scale bar 250 nm. An NMF-PCA algorithm was applied for denoising the EDS spectrum [253]. (c), (d) ICA loading maps for (c) the 0.67([Na<sub>2</sub>O]<sub>0.9</sub>[P<sub>2</sub>O<sub>5</sub>])-0.33([AlO<sub>3/2</sub>][AlF<sub>3</sub>]<sub>1.5</sub>) inorganic glass and (d)  $a_g$ ZIF-62. Scale bars 250 nm. (e), (f) ICA component profiles and corresponding area-averaged reduced ePDFs for (e) the inorganic glass and (f)  $a_g$ ZIF-62. Associated scree plots are given in **Figure 4.15**.

the two signals show agreement with the acquired EDS data as shown in **Figure 4.12**b, yielding a structural map from STEM-ePDF with a 5.5 nm pixel size (**Figure 4.12**c-d).

A second example of the same material in **Figure 4.13** is shown to showcase the repeatability of the STEM-ePDF method. The data was acquired on a single Medipix3 detector that was limited to only 256×256 pixels, but was chosen due to a through-thickness phase overlap. As the scattered data is convolved with the probe function with a non-zero convergence angle as discussed in Section 4.3.1.1, the resolution in *s* cannot simply be halved (as would be done by keeping  $s_{max}$  constant) and so the chosen  $s_{max}$  was lower than in the 512x512 data set. Nonetheless, the acquired ePDFs show clear structural correlation to the ones observed in **Figure 4.8**, although with slightly broader peaks. Peak positions are faithfully reproduced, and the ICA loading maps agree excellently with correlated EDS data. Furthermore, a small inclusion of the ZIF-62 in the inorganic glass was observed in both the EDS



**Figure 4.13** STEM-ePDF analysis of a second example of the  $a_g$ ZIF-62 and 0.67([Na<sub>2</sub>O]<sub>0.9</sub>[P<sub>2</sub>O<sub>5</sub>])-0.33([AlO<sub>3/2</sub>][AlF<sub>3</sub>]<sub>1.5</sub>) inorganic glass composite material. (a), (b) Comparisons of area-averaged reduced ePDF profiles obtained using a 512×512 pixel detector, for particles shown in **Figure 4.8** and using a 256×256 pixel detector, for particles shown in (c)-(d), for (a) the inorganic glass and (b)  $a_g$ ZIF-62. (c) EDS map of Al (red) and Zn (blue) from the data set, taken after the STEM-ePDF data. An NMF-PCA algorithm was applied for denoising the EDS spectra [253]. (d) The ICA maps corresponding to reduced ePDF components for the inorganic glass (red) and  $a_g$ ZIF-62 (blue). (e), (f) contrast-adjusted images of the small inclusion of  $a_g$ ZIF-62 on the lower particle for (e) EDS and (f) ePDF analyses. The PCA explained variance ratio for the map in (d) is given in **Figure 4.15**.

and the ICA decompositions, noted in **Figure 4.13**e and **Figure 4.13**f. While this may be the result of the overlap of two particles, the two-phase structure appears most likely to be the result of an interfacial contact between the  $a_g$ ZIF-62 and the inorganic glass. Such features have been seen in bulk structures of a closely related set of ZIF-62 inorganic glass composite samples [218]. The ICA decomposition of the ePDF signals showed presence of both components, showing that through-

thickness variation is also encoded in the ePDF signal. The ePDFs depicted in **Figure 4.13** illustrates that STEM-ePDF can offer spatial mapping capabilities with materials-specific signatures in the form of interpretable, though lower quality, PDFs even for sub-optimal (reduced  $s_{max}$ ) acquisition conditions. The ICA-ePDFs also showed clear correspondence with the area-averaged ePDFs, as shown in **Figure 4.14**, although the mixing was poorer and resulted in some deviation in peak positions. This was also reflected in the explained variance ratios of the PCA decompositions (**Figure 4.15**).



**Figure 4.14** (a),(b) ICA component profiles and area corresponding area-averaged ePDFs for (a) the inorganic glass and (b)  $a_g$ ZIF-62 in the 256x256 pixel data set used in **Figure 4.13**.



**Figure 4.15** Fraction of the variance explained by the most significant 25 principal components for the (a) 512x512 data set in **Figure 4.12** and (b) 256x256 data set in **Figure 4.13**. The PCA decomposition was performed on the pixel-by-pixel ePDFs. (a) Two components are strongly above the baseline and were used for the subsequent ICA decomposition. (b) Three components were above the baseline, but the largest two that were significantly higher were used for the subsequent ICA decomposition. (c-e) The loadings of the three-component ICA decomposition for the 256x256 data. One of the three (d) was found to be correlated with thickness, rather than crystallographic variation and reduced the accuracy of the two structural PDF loadings.

### 4.4.4 Interface analysis

A PDF was acquired from the area at the interface in the 512x512 dataset outlined in purple in **Figure 4.16**. The acquired ePDF was compared to ones acquired from single-domain regions. These PDFs are show in **Figure 4.16**b. The interface PDF shows small differences to a best fit of a sum of the component PDFs. Small differences are present in relative peak intensities, but it was not possible to distinguish signal from noise in this interface signal. The lack of a clear interface signal both in these area-averaged data and in the ICA suggests that any interfaces in the short-range order are abrupt to within a pixel size of  $\sim$ 6 nm. This observation is consistent with the expected viscosity and limited mixing of the two phases in previous studies [218]. Interfacial structures are fundamental to the porosity of MOF composite materials as well as their mechanical adhesion and properties [224].

However, due to only making up a small fraction of the total probed volume [254], interfaces have remained a challenge to probe using bulk methods such as X-ray PDF. Here, by constraining interface abruptness with STEM-ePDF mapping and separately confirming the retention of the determined bulk atomic structure near interfaces, strong, direct evidence is provided that amorphous MOF composites can be described by their bulk structures within 5-10 nm of structural interfaces.



**Figure 4.16** (a) Areas from which single domain ePDFs were calculated for the inorganic glass (red),  $a_g$ ZIF-62 (blue), and the interface (purple). (b) The single-domain and interface region ePDFs. The diffraction profiles were summed before an ePDF was extracted for each of the signals. A weighted average scattering factor was used for fitting at the interface. (c) The best fit of a sum of the red and blue ePDFs in (b), compared to the ePDF in the overlap region. Peak positions are reproduced, but relative height differ slightly. A difference is plotted at the bottom.

#### 4.4.5 Electron Beam Damage

The electron dose (ca. 10 e<sup>-</sup>/Å<sup>2</sup>) is similar to observed critical doses in crystalline ZIF structures [255]. In crystalline ZIFs, complete loss of crystalline Bragg diffraction, and structural collapse are typically recorded first (<100 e<sup>-</sup>/Å<sup>-2</sup>) and with increasing dose, both radiolysis [256], [257] and electron beaminduced mass loss are expected to occur. Such effects are also expected for the inorganic glass [258]. To gauge the loss of structural signal, the same area was scanned a second time. **Figure 4.17** presents area-averaged ePDFs from the first and second STEM scans, showing minimal changes with a total accumulated dose of ca. 20 e<sup>-</sup>/Å<sup>2</sup>. Almost all the observed peaks were unchanged, with very subtle differences observed in the inorganic glass around 7 Å. This suggests the work was done in a regime before significant damage was caused to the amorphous sample structure, considering the good correlation with the X-ray PDF in **Figure 4.8**. Direct comparisons between critical doses from crystalline samples with Bragg diffraction and amorphous samples with ePDF may not be possible, but ePDF presents a further tool for assessing damage as a function of dose in amorphous MOFs. The capability for recording multiple sequential scans without detectable changes also suggests it should be possible to do time-based studies in beam sensitive organic samples using scanning ePDF analysis.



**Figure 4.17** Area-averaged STEM-ePDF profiles for two scans of the same area for (a) the  $0.67([Na_2O]_{0.9}[P_2O_5])$ -0.33([AIO<sub>3/2</sub>][AIF<sub>3</sub>]<sub>1.5</sub>) inorganic glass and (b)  $a_g$ ZIF-62. The grey horizontal lines mark zero on the vertical axes.

## 4.4.6 Background Composition-dependence of the ePDF

The area-averaged PDFs from the  $a_g$ ZIF-62 and inorganic glass were fitted with varying background compositions. The fits are shown in **Figure 4.18**. The effect of an incorrect composition was primarily to increase peak heights in the  $a_g$ ZIF-62 and decrease peak heights in the inorganic glass. Any more complicated changes in the measured background scattering factors is effectively removed by the 4<sup>th</sup>- order correction. The increase and reduction in peak heights are due to the inorganic glass's theoretical scattering profile increasing at a faster rate at low *s*, resulting in a greater denominator in Equation (10) due to the misfit described in Section 4.4.2. Peak positions were virtually unchanged, meaning incorrect background fitting should have little to no effect on the ICA decomposition being able to distinguish the signals from one another.



**Figure 4.18** (a) Area-averaged ePDF as a function of background composition for the  $a_g$ ZIF-62, from 100%  $a_g$ ZIF-62 and 0% inorganic glass to 0%  $a_g$ ZIF-62 and 100% inorganic glass. (b) The opposite. The area-averaged ePDF as a function of background composition for the inorganic AIF<sub>3</sub>-NaPO<sub>3</sub> glass, from 100% inorganic glass and 0%  $a_g$ ZIF-62 and to 0% inorganic glass and 100%  $a_g$ ZIF-62.

## 4.4.7 Applying PCA-ICA to Other Signals

The PCA-ICA decomposition was also applied to the reduced intensities in the data. This decomposition revealed three components above the noise floor. The first strongest two components, were structural, but were significantly affected by thickness, particularly the inorganic glass, as shown in **Figure 4.19**. A third component was observed above the noise floor, but not found to correspond to a physical signal and was instead partially correlated to thickness. Hence, two components were used for the subsequent ICA decomposition. This thickness-dependent variation is a slow oscillation that manifests itself differently in the PDF, in a way that ICA was able to resolve, and as such ICA was directly applied to the PDF signals.

In general, what makes a specific signal differentiable in PCA and ICA is how well it adheres to the assumed variance condition. In ICA, non-Gaussianity is maximised. Seemingly, the thickness variation manifests in the PDF signals in a 'relatively non-Gaussian' manner, such that ICA is able to separate it. This is relatively less in comparison to the reduced intensities. Hence, the PDF was preferred as the signal for unsupervised learning decomposition, over the reduced intensity.



**Figure 4.19** (a) ADF of the particles used in the main text. All scale bars 250 nm. (b) Explained variance of the PCA decomposition of the spatially-resolved reduced intensity  $\phi(s)$ . Two components were used for the subsequent ICA decomposition. (c), (d) the ICA loading maps of the two components. (e), (f) The ICA components of the two phases, compared to the area-averaged reduced intensities shown in **Figure 4.12**.

# 4.4.8 Effect of the carbon background on the ePDF

The carbon background is not simply subtractable from the ePDF signal due to the complex behaviour of the PDF with multiple components described in Section 2.6.1.4, as well as the varied thickness across the sample resulting in dynamical scattering. Nonetheless, the structure factor and reduced PDF were acquired from the carbon to check whether the positions of its peaks corresponded to poor performance in the ICA decomposition and is shown in **Figure 4.20**. The ePDF agrees well with previous work [86], with peak positions at 1.4 Å and 2.5 Å accurately reflected. In general, the carbon signal is very weak in comparison to the nanoparticles. Notably, the ICA in **Figure 4.12** showed incomplete separation of the ePDFs at approximately 2.5 Å which may be related to the amorphous carbon contributions from the support film.



Figure 4.20 (a) Reduced intensity measured in the carbon film, averaged from a 50×50-pixel real space region.(b) STEM-ePDF of the amorphous carbon film, from the reduced intensity in (a).

# 4.5 Application of the Method to other MOF Data

The area-averaged ePDF method was also applied to Fe-BTC, a MOF sample with crystalline and amorphous regions [259], [260]. The structure of Fe-BTC, which is topologically disordered, is not fully known, but is a close analogue to crystalline MIL-100 [260]. The goal was to characterise amorphous regions within the structure, particularly whether the amorphous structure was homogeneous, for use in X-ray subsequent PDF modelling. Data was taken on the same JEOL-ARM300F at 200 kV as the data for the ZIF-62-inorganic glass composite. The total dose was ca. 10 e<sup>-</sup>/Å<sup>2</sup>. The samples had crystalline and amorphous regions and the PDF pipeline was applied to amorphous regions within the data, and the results are shown below in **Figure 4.21**. Overall, the amorphous regions were homogeneous, with some changes in the peak heights at 2.2 Å and 5 Å. This is likely due to the varied thickness in the data, as the amorphous nanoparticle regions were not a consistent thickness. Regardless, it showcases the repeatability of ePDF analysis when applied to MOF data.



**Figure 4.21**.PDF analysis of amorphous Fe-BTC. (a) Example SED scan, with 5 by 5 pixel summed diffraction patterns in (b), (c), and (d). (d) is crystalline, while (b), and (c) are amorphous. (e) PDFs from four different amorphous regions across three scans, normalised to the height of the first peak. A crystalline PDF is not reproduced due to being orientation dependent. The PDFs were processed in exactly the same way, although the number of pixels summed varied slightly.

# 4.6 Conclusion and Outlook

STEM-ePDF analysis was used for the phase determination and mapping of a composite glass comprised of  $a_g$ ZIF-62 and 0.67([Na<sub>2</sub>O]<sub>0.9</sub>[P<sub>2</sub>O<sub>5</sub>])-0.33([AlO<sub>3/2</sub>][AlF<sub>3</sub>]<sub>1.5</sub>) at ca. 6 nm spatial resolution at a low dose of ca. 10 e<sup>-</sup>/Å<sup>2</sup>. Benchmarked against X-ray PDF, area-averaged STEM-ePDF and ICA blind source separation revealed two primary structural signals independently recovering the distribution of bulk short-range order characteristics across domains of each glass component. The ePDF structural analyses were further verified using correlated EDS chemical imaging. Cumulatively, these findings provide critical direct evidence of the local, nanoscale retention of atomic structure described well by the average bulk PDF to within a pixel size of <10 nm of interfaces in this MOF/inorganic glass composite. The method shows high sensitivity, identifying and separating inclusions as well as side-by-side phases at low electron dose, establishing exciting possibilities for the use of STEM-ePDF for quantifying the atomic structure within numerous nanostructured non-crystalline materials systems. Especially, it was shown that it is possible to identify and characterise structural variation tool that encodes both structural information and has nanoscale spatial resolution for amorphous materials, enabling the characterisation of many non-crystalline materials systems.

Interestingly, the differences in the X-ray PDF and electron PDF of the ZIF structure highlight possible further research into combined X-ray-electron PDF analysis. Conceivably, both could be used simultaneously to independently constrain the structure of a probed amorphous material, possibly yielding more accurate structural models than either method independently. Such further work may be contingent on better acquisition conditions, particularly the use of energy-filtered diffraction to reduce inelastic contributions in the ePDF data.

# 5 Diffraction from Twisted Cocrystals

# 5.1 Background and Motivation

Pharmaceutical cocrystals [261] are emerging as a potential solution to poor solubility that plagues many pharmaceutical compounds and prevents their access to market [262], [263]. Aqueous solubility, in particular, remains an all-too-frequent hurdle for new pharmaceutical compounds, with as many as 90% of new chemical entities suffering from poor aqueous solubility [264]. In a cocrystal, a pair of molecules in a stoichiometric ratio are combined to form a unique crystal lattice [265]. The ability to design specific cocrystals to enable nanoscale solubility, hydration properties, stability [266] and dispersion of active compounds has become possible on an industrial scale in the last decade [267], and the design space is near infinite. However, such cocrystals are often difficult to characterise, as their crystal structure and microstructure are both dependent on the processing conditions [268], [269], and phase polymorphism is common. To address the polymorphism problem, 3D-electron diffraction (3DED) has emerged as a tool to probe the crystal structure of small volumes on the nanoscale [147], but significant questions remain about the microstructures of such compounds. Scanning electron diffraction (SED) is the perfect tool for probing such nanostructures due to its combination of spatial resolution and diffraction resolution.

In general, cocrystal structures have shown complex nano- and microstructure, primarily attributable to their multiple types of (mostly) planar molecules preferring to pack in layers. Perylene-TCNQ was found to form either nanowires or micron-scale blocks depending on the stoichiometric ratio [270]. The structure was also dependent on the solvent used, and even could change over time [271]. In addition, an orientation 'wiggle' was reported. Twisting has also been observed in cocrystals [272], [273]. However, few direct measurements of cocrystal microstructure have been observed, as the structures are generally highly beam-sensitive, such that very few tools other than SED can identify microstructure in such materials at the necessarily short length scales. The general solution of such structures using bulk methods such as XRD carry the possibility of missing out nanoscale structural phases and size-dependent phenomena that can occur in such compounds [274].

Paracetamol is a common painkiller with a global market of roughly \$500 million. It has two primary polymorphs, forms 1 and 2 [275], as well as an additional highly unstable form 3 that can be stabilised under certain chemical conditions [276]. For tablet processing, form 2 would be preferred due to better compressibility attributed a layered hydrogen-bonded structure [277], but its lower thermodynamical stability means that in practice form 1 is used in combination with a binder. To

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enable better processing, the paracetamol-theophylline cocrystal system has been explored due to improved mechanochemical compressibility in the cocrystal over either pure substance [269], primarily attributed to strong hydrogen-bonded layers in many cocrystal structures. Theophylline itself is a phosphodiesterase inhibiting drug used in respiratory therapy. A structure for the bulk cocrystal has been solved [269], [278] through XRD, but polymorphism of the system remains unexplored.

Here we show diffraction from a novel twisting polymorph of a paracetamol-theophylline cocrystal. While it is possible to characterise twisting crystals in 3DED [279], the twisting rate here was too high, such that each TEM diffraction pattern is in practice a sum of a large orientation range, resulting in fibre diffraction-like patterns. However, we show that using SED and subsequent characterisation of the twist rate present in the material, an orientation map can be recovered, enabling the construction of a slice of 3D reciprocal space from a single sample crystal. As a result, both a unit cell of the polymorph and the space group can be recovered. The recovered unit cell has highly anisotropic lattice parameters, likely due to the twisting nature of the crystal. Overall, we demonstrate that SED can solve the crystal structure of twisting molecular crystals by the combination of sufficient spatial resolution and correlation with local orientation.

# 5.2 Sample Preparation

Pure paracetamol, pure theophylline, and a 1:1 mixture of paracetamol and theophylline was ballmilled to induce mechanochemical transformations and solid-state processes analogous to mechanochemical production methods in the pharmaceutical industry, such as milling and blending. All pure compounds were purchased off-the-shelf. Slides were prepared for SED by transferring the powders directly from the ball mill to holey carbon STEM support grids. Using the same method, other cocrystal systems were also explored, but only the paracetamol-theophylline system showed the twisting crystals analysed below.

The samples were made and prepared by collaborator Duncan N. Johnstone.

# 5.3 Results and Discussion

SED data of the ball milled pharmaceutical crystals was acquired on a JEOL ARM300F STEM operated at 200 kV with a beam diameter of ca. 5 nm and a convergence semi-angle of ca. 0.6 mrad. The beam current was ca. 1.5 pA and the frame rate was 1 ms per pattern, with slight overlap between the beam at adjacent pixels, leading to an electron dose of ca. 5  $e/Å^2$ .

The data showed significant, varied, crystal morphology. In the paracetamol-theophylline cocrystal data, a number of elongated plate-like crystals were observed, with peculiar VDF contrast. An example of such a crystal is shown in **Figure 5.1**, where it was noted that the diffraction patterns changed considerably across the observed structure, but with certain constraints. All reflections were observed in layer lines (see Section 2.5), indicative of fibre diffraction. A summed diffraction pattern of the sample also had clear layer lines (**Figure 5.2**). Overall, the diffraction patterns included many reflections at high scattering angles (> 0.8 Å<sup>-1</sup>), although such reflections were frequently weak (**Figure 5.1** inset 3 and **Figure 5.2**). These reflections can be attributed to the mosaicity of the crystal orientation. Mosaicity refers to the subtle changes in local orientation, such as through twisting or bending, that is common in molecular crystals. In effect, a range of orientations is sampled, resulting in more diffraction spots appearing than expected.

The observed diffraction patterns were fitted to a known 50:50 paracetamol-theophylline cocrystal structure (KIGLUI), but once "unravelled", the single diffraction patterns were not found to match any orientation, suggesting the twisting crystal is a previously unknown polymorph. As a result, classifying the orientation of a given diffraction pattern was extremely challenging.



**Figure 5.1**. Central Figure: Annular Dark-Field (ADF) image of a crystalline elongated paracetamol-theophylline cocrystal, with a range of  $0.5 < k < 1.0 \text{ Å}^{-1}$ . Insets 1-8: Diffraction patterns from selected pixels in the data set (indicated by coloured squares on the central figure). The scale bar is 0.3 Å<sup>-1</sup>.



**Figure 5.2**. Selected VDF contrast from the paracetamol-theophylline cocrystal shown in **Figure 5.1**. Top: Summed diffraction pattern from the whole crystal overlaid with 16 virtual apertures (numbered). Each line of reflections is made up of overlapping spots. Insets (1-16) Observed VDF contrast from the numbered virtual apertures, showing banded contrast. Field of view is 365 nm by 290 nm.

# 5.3.1 Observed VDF Contrast

Some paracetamol-theophylline cocrystals, showed complex VDF contrast, shown in **Figure 5.2**, with each reflection giving rise to bands of intensity. Reflections along the fibre equator showed intensity bands perpendicular to the fibre axis, while reflections whose g-vector is nearly parallel to the fibre axis showed contrast nearly parallel to the fibre axis. The remaining reflections showed VDF contrast that was at approximately the same angle to the fibre axis as the diffracted spot, but on the opposite side; diffracted spots clockwise of the fibre axis showed VDF contrast anti-clockwise of the fibre axis. This contrast was consistent between samples, a second example is shown in **Figure 5.3**, with notably less blurred diffraction spots. This was correlated to a slower twist rate (detailed calculation below), which in turn was correlated to a lower thickness.

Multiple reflections with very similar g vectors are seen in each VDF (see **Figure 5.4**). These families of reflections showcase diffracted bands with periodicity based on the layer line in which they lie. In **Figure 5.2**, reflections in the first layer line (insets 1, 2, 7, 8, and 15) have no overlap between bands in the VDF, with each band starting on one side of the fibre when the previous one ends on the opposite side, while reflections in the second layer line (insets 3, 4, 5, 6, and 16) have exactly one overlapping band at all times. In effect, their band spacing is halved. Reflections in the third layer line were generally too weak to get proper VDFs, but some extremely noisy VDFs suggested a continuation of this trend, with the spacing between bands reduced compared the second layer line. However, they were not conclusive.



**Figure 5.3**. Selected VDF contrast from a second thinner paracetamol-theophylline cocrystal. Top: Summed diffraction pattern from the whole crystal overlaid with 16 virtual apertures (numbered). Each line of reflections is made up of overlapping spots. Insets (1-16) Observed VDF contrast from the numbered virtual apertures, showing banded contrast. A small movement of the particle occurred during the crystal scan, visible most clearly in inset 9. Inset 15 showcases double diffraction. Field of view is 220 nm by 300 nm.

The VDF band frequency arises naturally from the structure, as multiple closely related reflections diffract from the same area. The band for a reflection hkl and 2h 2k 2l are in the same place. As (h + l) l k 2k 2l are in the same place. As (h + l) l k 2k 2l are in the same place. 1)kl and 2(h + 1) 2k 2l must also be in the same place, the (2h + 1) 2k 2l reflection that lies in between in reciprocal space has an intermediate VDF band. That the frequency doubles from the first to the second layer line suggests that the first layer line is a lowest-order reflection, which is consistent with all the twisted crystals both showing the same layer line spacing (0.26 Å<sup>-1</sup>) and showing no reflections closer to the central spot than this, except along the equator. Importantly, each VDF repeat period (the distance over which one band exists) does not correspond to a 360° rotation, as the period between different VDFs is not the same. This was confirmed by analysis of the +g/-g displacement (below) which suggests that the total twist across the crystal in Figure 5.2 is approximately 20°. The VDF bands are also notably spotty, oscillating between high and lower intensity. The +g/-g reflections are complementary in their positions of increased intensity, with the intensity oscillating between +g and -g. An example is shown in Figure 5.4. This suggests the spottiness may be due to either a local bending effect, such as the 'wiggles' in orientation seen in other cocrystal structures, or a moiré effect creating a 'beating' intensity. Curiously, no spottiness is observed for the equatorial reflections, suggesting that the origin of this effect leaves the equatorial planes unchanged. We think it is likely a feature related to the plane orientation rather than physical orientation of the VDF, as the VDFs of other reflections change perpendicular to the fibre axis (Figure 5.4a).



**Figure 5.4**. VDF variation for closely-related reflections. (a) Summed diffraction pattern showing the VDF apertures used for (b) and (c). (b) The spatial separation between a reflection g (magenta) and -g (cyan). Note the complementary 'spottiness' of the reflections. Such a separation was seen for all reflections. (c) VDF contrast across the bright first layer line reflection second nearest to the fibre axis, with red (inner) blue (middle) and green (outer) parts of the reflection all corresponding to different VDF bands. Hence the reflection is not one spread out reflection, but three distinct reflections at slightly different k values that each come into the diffraction condition during rotation.

### 5.3.2 The Case for Twisting

The VDF contrast combined with the real-space displacement of the +g/-g VDF positions being present for all reflections suggests a complex 3D orientation relationship within the crystal. To unentangle this relationship, it is simpler to break down this rotation as a series of rotations about different axes. First, we consider the rotation around the equatorial reflections. These reflections appear in distinct bands perpendicular to the long axis of the plate. Along this VDF, an orientation change does occur, as is visible both in the changing diffraction patterns (Figure 5.1) and more clearly in other VDF contrast changing across this band (Figure 5.2), which implies that those non-equatorial reflections leave the diffraction condition at some point across the equatorial VDF band. However, the presence of the equatorial band implies that across the short axis of the particle, the rotation is about the plane normal of the VDF reflection. A visual representation is given in **Figure 5.5**. This rotation is about the direction of the VDF band. An equivalent analysis for reflection as close to the meridian as possible implies that the rotation along the long axis is about the long axis. This rotation is a typical cylindrical 1D rotation, but importantly, it is not the only rotation present in the sample! If only one axis of rotation was present, each VDF would show unchanged contrast along that axis. However, this 3D combination of rotations is present in twisting, as shown in **Figure 5.6**, making it possible to map the orientation across the structure.



**Figure 5.5**. The rotations present implied by the orientation of the VDF contrast. (a) An equatorial VDF. The diffracting planes must remain perpendicular to the diffracted spot across the VDF. Therefore the only allowed rotation as one moves along the blue axis is around the blue axis. (b) A VDF as close to the meridian as possible. The diffracting planes must remain perpendicular to the diffracted spot across the VDF. Therefore the only allowed rotation as one moves along the red axis is around the red axis. (c) Both rotation drawn on the sample dark field image.



**Figure 5.6**. Twist geometry and projected views along axes. Left: Twist geometry for a cuboidal particle with long (z), medium (y) and short (x) in the direction of the given axes. (x,y,z) Projected views along those axes. Each axis has a projected rotation as the particle needs to be continuous in space. The amount of rotation is roughly proportional to the size of the object along that axis (i.e. rotation rate is approximately constant)

#### **Rotations Present / Reciprocal Space Probed**

Twisting probes a relatively complicated subset of reciprocal space orientations. It is not simple equivalent to a rotation, as shown in **Figure 5.7**. Instead, the combined effects of multiple rotations result in a larger portion of reciprocal space being probed than would be for an equivalent simple rotation.



**Figure 5.7**. Cubes showing the change in orientation across the specimen. The rotation is exaggerated by 5 times for clarity. The yellow and purple rectangles highlight regions where the two axes of rotation can be seen.
### 5.3.3 Measuring Twist Rate

#### Using the +g/-g Displacement

The consistent spatial separation between the +g and -g reflections in space enables the calculation of a twist rate for a given crystal. The angular rotation between the +g and -g intensity maxima is the scattering angle  $2\theta$  (see **Figure 5.8**d). An example of such a calculation is shown in **Figure 5.8**. The measured rotation rate seemed to be independent of the reflection, which is consistent with a relaxed twisting structure. An average twist rate of 0.049° nm<sup>-1</sup> was calculated for the crystal in **Figure 5.8**. This rate is approximately 0.014° per layer of stacked molecules, so locally a unit cell with almost no distortion would be expected to be observed. This rate was not consistent between data sets. The crystal shown in **Figure 5.3**, which is slightly smaller and thinner, had a twist rate of ca. 0.04° nm<sup>-1</sup>. In general, thinner crystals (which also tended to be narrower and shorter) showed even lower twist rate, with a small crystal shown in **Figure 5.9** having a twist rate of only 0.02° nm<sup>-1</sup>. With its slower twist rate, **Figure 5.9** highlights the displacement of the +g and -g reflections more strongly. The following insets are +g/-g pairs: 1&2, 3&4, and 11&12.



**Figure 5.8**. Measuring the rotation rate of the crystal twist. (a) Overlaid +g (red) and -g (blue) VDF contrast on the sample ADF. (b) Integrals of the red and blue VDFs along the arrow direction shown in (a). The peaks of the two are displaced by a distance  $\delta x$ . (c)  $\delta x$  for 5 different reflections at different angle, plotted against the magnitude of g. Regardless of orientation, the twist rate is similar. (d) A schematic highlighting that the rotation  $\gamma$  from the diffraction condition +g (red) to -g (blue) is equal to  $2\theta$ , as the red and blue corners are both right angles. This gives a twist rate of 0.049° nm<sup>-1</sup> in the crystal. The rate measured from only the equatorial reflections was very slightly higher at a twist rate of 0.052° nm<sup>-1</sup>.



**Figure 5.9**. Selected VDF contrast from a third small paracetamol-theophylline cocrystal. Top: Summed diffraction pattern from the whole crystal overlaid with 16 virtual apertures (numbered). Spots are seen outside the layer lines due to partial overlap with a second non-twisting particle. Insets (1-16) Observed VDF contrast from the numbered virtual apertures, showing banded contrast. Insets 1&2, are a +g/-g pair, as are 3&4, and 11&12. Inset 14 highlights the overlapping particle. Field of view is 170 nm by 190 nm.

#### Using the VDF Positions of the Equatorial Reflections

It is also possible to obtain an independent estimate of the twist rate using the positions of the VDFs of the equatorial reflections. These reflections, such as insets 10, 11, and 12 in **Figure 5.9**, have bands of VDF contrast across the long axis of the particle. The rotation from one of these bands to the next should be a simple rotation (although each band has a number of orientations within it). In effect, this means that all of the equatorial reflections lie in the same plane in reciprocal space. While this orientation), it can be reconstructed in post-processing by considering the positions along the particle of each equatorial reflection. Since the diffracted spots are from the same slice of reciprocal space, they should be coplanar and form a lattice of points within this coplanar slice. By calculating the position of each band (corresponding to a specific reflection) along the particle, shown in **Figure 5.10**, this slice can be reconstructed, as long as a rotation rate along the crystal is assumed.

Importantly, this rotation rate is in turn a variable that can be fitted. The reflections in this reconstructed reciprocal plane should be most 'lattice-like', in that the spot-to-spot vectors should be most consistent if the rotation rate is chosen correctly. This is shown in **Figure 5.11**, where an initial guess based on the rotation rate from the +g/-g displacement above is used (**Figure 5.11**a). These spots are already very grid-like, highlighting the validity of the coplanar assumption. To find the most 'lattice-like' rotation rate, the chi-square error  $\chi^2$  was calculated for each type of vector (in x and y) and the total sum  $\chi^2$  was compared. By plotting the  $\chi^2$  as a function of rotation rate, shown in **Figure 5.11**, it is found that 0.049° nm<sup>-1</sup> has a low error, but is not the minima, which is at 0.058° nm<sup>-1</sup>. This rotation rate also gives a visually improved lattice in the equatorial plane as well. The reciprocal **a**\* and 2**b**\* vectors are given in **Figure 5.11**e, further investigated below. The vector is 2**b**\* rather than **b**\* as systematic absences are seen in the equatorial plane, detailed below as well.



**Figure 5.10**. All the present equatorial VDFs (dark, left) and an example diffraction present with the reflection present (light, right), as well as their rough positions along the twisting crystal. All of these reflections were used to calculate the twist rate via fitting to a lattice.



**Figure 5.11**. (a) The equatorial reflections from **Figure 5.10**, with the reflections rotated by the rotation rate predicted by +g / -g displacement (b) Vectors between each nearby set of points, with the optimised rotation rate (from (d)). (c)  $\chi^2$  of the vectors as a function of rotation rate with (d) an inset in the low rotation rate region. The lower the metric, the more similar each vector. A minima is found at 0.058° nm<sup>-1</sup>. The  $\chi^2$  metric reduces at zero rotation rate, as the vectors converge to a line, and some points overlay. This gives a better 'grid' of points (b). (e) The reciprocal **a**\* and 2**b**\* vectors (as systematic absences are present in the equatorial plane).

#### 5.3.4 3D Reciprocal Space Reconstruction

As the twist rate is known, it is possible to reconstruct the expected orientation at each pixel within the crystal. This is a challenging process due to complex orientation relationships, as shown in **Figure 5.6**. Rather than try to predict this orientation analytically, we constructed an approximate model by constructing a 3D cuboid of positions in x, y, and z. These points can then each be rotated by a simple rotation. As a result, the cube defined by the positions  $(x_i, x_{i+1})$ ,  $(y_i, y_{i+1})$ , and  $(z_i, z_{i+1})$  is not only rotated about the rod axis, but about multiple axes, as each point moves independently. Subsequently, the orientation (and strain) at each real space position can be estimated by the strain and misorientation of the cuboid formed by the surrounding points, resulting in an orientation map of the sample. Note that we averaged the orientations through the sample thickness to reduce the calculation time, which scaled with the number of points  $N^3$ . Hence, this method ignored any throughthickness changes in orientation. There will be a rotation around the axis of the beam down the beam direction. However, this change should be small, as the particles are thinnest in the beam direction. An estimate was able to be acquired for the data set shown in **Figure 5.3** (see Section 5.3.6) and it was approximately 20-25 nm in thickness, equivalent to approximately 1° rotation when assuming a constant rotation rate. Furthermore, such rotations may be reduced at surfaces.

After a rotation was assigned to each pixel, each pattern can then be rotated back to see if this results in a consistent unit cell. To reduce the amount of data to rotate (there are thousands of frames), diffracting peaks were found by peak finding methods in Pyxem. The intensities at these peaks were also calculated. These vectors were then placed in a 3D diffracting cell. Both twist rates calculated in Section 5.3.3 were used. The results are shown in Figure 5.12 (0.049° nm<sup>-1</sup>) and Figure 5.13 (0.058° nm<sup>-1</sup>), with the twist axis along the vertical. (a)-(c) showcase the diffraction data down the three perpendicular rod axes, as shown in Figure 5.14. Not much information can be interpreted from these, although (a) strongly resembles the summed diffraction pattern from the data set. By rotating to the appropriate orientation, the spots that are not overlaid in (a)-(c) overlay, giving the diffraction data down the unit cell **a**\*, **b**\*, and **c**\* axes. **c**\* was chosen to be along the twist direction. It is also the unique axis. The shorter of the two reciprocal vectors was chosen as  $\mathbf{b}^*$  as  $2\mathbf{b}^*$  is very close in length to the  $\mathbf{b}^*$  axis of the previously classified polymorph. Systematic absences are seen along the 0k0 row, with reflections where k is odd missing, and the diffraction sampling along  $c^*$  is poor due to the rod orientation. Importantly, in the 3D diffraction pattern, the VDF intensity for each observed band condensed to a single point. This was only the case when the rotation axis was chosen appropriately, leaning at about 12° angle to the beam direction. With the rotation rate, both 0.049° nm<sup>-1</sup> and 0.058° nm<sup>-1</sup> resulted in converged points, but significant changes to the rate either resulted in non-periodic cells (for lower than 0.04° nm<sup>-1</sup>) or points spread into lines (higher than 0.07° nm<sup>-1</sup>). This confirms both twist rates are close to the correct value. We consider the twist rate measured from the equatorial reflections (0.058° nm<sup>-1</sup>) to be more accurate, as it is based on a larger number of reflections.



**Figure 5.12**. 3D Reciprocal space after orientation assignment of each pixel in the data in **Figure 5.2**, with a twist rate of 0.049° nm<sup>-1</sup>. Each panel is a projection of the found reciprocal vectors and their respective intensities. The orientation relationships between the different projections are given in **Figure 5.14**. (a-c) Down the sample axes. (a) Is the view along the beam path (along *x*). This is equivalent to the observed summed diffraction pattern in **Figure 5.2**. (b) View down the short axis and (c) down the long axis, at 90 degrees to (a). (d-f) Down the crystal axes. (d) Down the (chosen) **b**<sup>\*</sup> axis. (e) Down the **a**<sup>\*</sup> axis. (f) Down the **c**<sup>\*</sup> axis. All vectors along the respective axes align. The rotation from (d) to (e) is 74°. The **c**<sup>\*</sup> axis lies at 87° to the **a**<sup>\*</sup> – **b**<sup>\*</sup> plane.

The two twist rates gave slightly different length vectors and inter-axis angles, given below in **Table 5.1**. Note that the  $\alpha$  angle is not 90°, suggesting that the cell should be triclinic. However systematic absences in the 0k0 reflections (**Figure 5.12**e and **Figure 5.13**e) show that the unit cell must be monoclinic. Hence this change in angle is attributed to strain in the lattice, and when measured for the data set in **Figure 5.3**, was found to be 89°, further supporting this. The differences between the two rotation rates are small except in the measurement of the monoclinic angle  $\gamma$ , where the two are 98° and 106°. The basis vectors are about 10% shorter with the rotation rate of 0.058° nm<sup>-1</sup>. This, combined with the strain, makes the exact dimensions of the measured unit cell slightly uncertain.



**Figure 5.13**. 3D Reciprocal space after orientation assignment of each pixel in the data in **Figure 5.2**, with a twist rate of 0.058° nm<sup>-1</sup>. Each panel is a projection of the found reciprocal vectors and their respective intensities. The orientation relationships between the different projections are given in **Figure 5.14**. (a-c) Down the sample axes. (a) Is the view along the beam path (along *x*). This is equivalent to the observed summed diffraction pattern in **Figure 5.2**. (b) View down the short axis and (c) down the long axis, at 90 degrees to (a). (d-f) Down the crystal axes. (d) Down the (chosen) **b**\* axis. (e) Down the **a**\* axis. (f) Down the **c**\* axis. All vectors along the respective axes align. The rotation from (d) to (e) is 83°. The **c**\* axis lies at 87° to the **a**\*- **b**\* plane.



**Figure 5.14**. Schematic orientation relationship between the projected directions given in **Figure 5.12** and **Figure 5.13**. *x*, *y*, and *z* are in the scan orientation, with *x* aligned to the beam. The directions are based on the central pixel, since each unit cell has been 'detwisted'.  $\mathbf{a}^*$ ,  $\mathbf{b}^*$ , and  $\mathbf{c}^*$  are oriented to the crystal, with  $\mathbf{c}^*$  approximately 12° into the page.

 Table 5.1. Reciprocal lattice basis vectors and angles, and the real space lattice vectors of the calculated cell as well as the KIGLUI cell.

	Rotation Rate 0.049° nm <sup>-1</sup>	Rotation Rate 0.058° nm <sup>-1</sup>	KIGLUI
<i>a</i> * (Å⁻¹)	0.101	0.109	0.114
<b>b</b> * (Å⁻¹)	0.0340	0.0374	0.065
<i>c</i> * (Å⁻¹)	0.269	0.269	0.087
α (°)	87 (90)	87 (90)	90
β (°)	90	90	99.5
γ (°)	106 (74)	98 (82)	90
<b>a</b> (Å)	10.1	9.19	8.73
<b>b</b> (Å)	29.5	26.7	15.4
<i>c</i> (Å)	3.72	3.72	11.5
Volume (ų)	1045	904	1528

#### 5.3.5 Best-fit Unit Cell

#### Orientation

The *c*-axis was chosen as the unique axis for the structure, as it aligns roughly with the rod axis. The *b*-axis was chosen to align with the *b*-axis of the previously characterised cocrystal structure, as the length of the current *b*-axis is roughly twice the one in the known cocrystal structure. These two choices give a unit cell of 10.1 Å by 29.5 Å by 3.72 Å, with  $\alpha \approx \beta \approx 90^{\circ}$ , and  $\gamma \approx 106^{\circ}$  for a rotation rate of 0.049° nm<sup>-1</sup> and a unit cell of 9.2 Å by 26.7 Å by 3.72 Å, with  $\alpha \approx \beta \approx 90^{\circ}$ , and  $\gamma \approx 98^{\circ}$  for a rotation rate of 0.058° nm<sup>-1</sup>. The latter is used from here onwards.  $\alpha$  and  $\beta$  are not exactly 90° in the 3D reconstruction. However, the systematic absences present (**Figure 5.15**) in the diffraction data can only occur in a monoclinic unit cell, implying that both angles must be 90°. Any deviation from this measured angle is likely a combination of misassigned rotation rate and strain in the material, which is expected to be high in a twisting crystal.

#### Systematic Absences

With the chosen unit cell orientation, all of the hk0 reflections where  $k \neq 2n$  are absent (i.e 010, 030, 110 etc.), shown in **Figure 5.15**. However, such reflections are not absent in the hk1 plane. The 00*l* reflection is not seen due to the orientation of the crystal with respect to the beam. However, in the orientation of the data in **Figure 5.3**, 001 should be present (inset 15). It however, is not, with the very small amount of diffracted intensity due to intense double diffraction in pixels with both 121 and  $\overline{120}$  peaks. This implies that the structure has the monoclinic P2<sub>1</sub>/b (14) space group [280] (equivalent to P2<sub>1</sub>/c for a unique *b*-axis), and the divergence of  $\alpha$  from 90° is due to strain. P2<sub>1</sub>/b has systematic absences for: hk0:  $k \neq 2n$ , 0k0:  $k \neq 2n$ , and 00l:  $l \neq 2n$ , where *n* is an integer.

A small amount of intensity was seen in the hk0 reflections where  $k \neq 2n$ . However, this was attributed to double diffraction via the intense 121 and  $\overline{13}1$  reflections, as shown in **Figure 5.16**. These intensities did not show VDF bands, unlike other reflections, and they were localised to the crossover of the intense VDF bands from the first layer line. This indicates their source is double diffraction through the appropriate diffraction vectors.



**Figure 5.15**. Equatorial (a) and first layer line (b) reflections from the data reconstructed in **Figure 5.12**. All hk0 reflections where k is odd are systematically absent in (a), but all hk1 reflections are present in (b). The closest equatorial vectors are 120 and  $\overline{120}$  due to sampling.



**Figure 5.16**. Double diffraction among the hk0 reflections. (a) A reconstruction of the diffracted intensity in the hk0 plane. Inset: A zoom in on the 2k0 reflections. All reflections with  $k \neq 2n$  are forbidden in hk0 plane but do appear in the hk1 plane. Some intensity is measured at these spots. (b) Virtual dark field image of spots shown in (a), overlain on the VDFs of appropriate hk1 (yellow) and  $\bar{h}(\bar{k} + 1)1$  (magenta) reflections, showing that the forbidden reflection only appear at the double diffraction condition. Two  $\{1k1\}$  reflections make up each pink and yellow stripe.

## 5.3.6 Thickness Measurement

The cocrystal shown in **Figure 5.3** showed oscillating VDF contrast akin to a rotation or a bend in a two-beam diffraction condition for the closest {120} equatorial reflections, shown in **Figure 5.17**. The beating effect was not observed for other reflections. This type of beating is expected to occur with changing orientation in a two-beam condition, suggesting that no other reflection was strong enough to cause a two-beam condition. **Figure 5.17**b shows the ratio of the 120 and  $\overline{120}$  reflections. Both showed intensity oscillation, although they are slightly displaced from one another as with all observed reflections. The zeros in the observed profile (**Figure 5.17**c) are a function of the sample thickness and the extinction length of the reflection. That the profile is straight except at the edges (**Figure 5.17**b) implies that the sample has a consistent thickness, which is consistent with a plate-shaped particle. This also implies the primary analysed particle likely has a consistent thickness. The edges are likely tapered, or a rounded taper. Such a crystal morphology would be expected from a monoclinic crystal.



**Figure 5.17**. (a) VDF of the shortest equatorial reflection from **Figure 5.3** (inset 9). The contrast is akin to twobeam diffraction. Scale bar 50 nm. (b) +g divided by -g, displayed on a logarithmic scale (+g: red, -g: blue), highlighting the positions of the bands of intensity beyond the first peak. (c) The perpendicular profile down the lateral centre of the band. (d-e) The dynamical positions of the minima of (c) fitted such that the y-intercept is the thickness for (d) n=2 and (e) n=3.

To get an estimate of the thickness, the profile in **Figure 5.17**c was plotted in terms of scattering vector s, estimated from the calculated rotation rate. Only the bottom side (of the particle) was used to measure the positions of the zeroes due to the slight movement of the particle under scanning. Under a dynamical approximation, these positions occur at [94]

$$\sin^2 \pi t \sqrt{\left(s^2 + \frac{1}{\xi_g^2}\right)} = 0;$$

$$\pi t \sqrt{\left(s^2 + \frac{1}{\xi_g^2}\right)} = n\pi \tag{60}$$

where t is the thickness of the sample and  $\xi_g$  is the extinction length of a particular reflection. n takes a series of values for each minima, starting at n = 1 for  $0 < t < \xi_g$ , from n = 2 for  $\xi_g < t < 2\xi_g$ , from n = 3 for  $2\xi_g < t < 3\xi_g$  and so on. In practice, both can be determined by plotting  $s^2/n^2$  against  $1/n^2$ , such that  $1/\xi_g^2$  becomes the gradient and  $1/t^2$  the intercept. The two closest fits are given in **Figure 5.17**, for  $\xi_g < t < 2\xi_g$  and  $2\xi_g < t < 3\xi_g$ . The  $0 < t < \xi_g$  resulted in a line with a positive gradient. The fit is not perfectly linear, but did not improve if starting at n = 4 or n = 5. Which result is correct is unclear, but both give the same approximate thickness of 20-25 nm, which is consistent with a thin plate structure.

### 5.3.7 Twist origins

Twist originates from geometric frustration which can occur due to a wide variety of reasons (see Section 1.4). We suspect that the geometric frustration is a result of an unconventional (010)-style twin of the known KIGLUI structure due to two primary reasons. Firstly, such a twin would be expected to cause twisting with a twist rate that increases with particle size, in contrast to more conventional twisting where the twist rate is slower for large particles. This was observed in the twisting particles as described above. Secondly, the calculated unit cell has a lattice parameter that is close to twice the size of the KIGLUI cell along *b* (26.7 Å vs 15.4 Å) and a similar lattice parameter along *a* (9.2 Å vs 8.7 Å). In addition, the length of the 120 reciprocal lattice vector in the proposed structure (0.140 Å<sup>-1</sup>) is very similar to the 110 reciprocal lattice vector (0.133 Å<sup>-1</sup>) in the KIGLUI cell. The slight differences are likely a result of the strain that would have to occur for such a unit cell to form. Whether such twinning would occur on the scale of a single unit cell or over the whole fibre could not be determined.

### 5.3.8 VDF Band Spottiness

We were not able to conclusively determine the origin of the "spottiness" of the VDF bands, as shown in **Figure 5.2** and **Figure 5.3**. All non-equatorial reflections show this spottiness, with the +g/-g reflections being complementary in their spottiness (when one is intense the other is not), as shown in **Figure 5.4**. We suspect the spottiness is an interference effect, with a few possible origins. There is considerable strain in the structure due to twisting, and this strain is expected along the electron beam path. This strain could result in local orientation changes that manifest as spottiness in the diffraction pattern. In addition, it could be a moiré effect due to twinning. If the structure is formed due to a twin along the beam path, a moiré effect would be expected to be observed. Unfortunately, separating these two effects and exactly pinpointing the origin of the spottiness was not possible.

## 5.4 Conclusion and Outlook

Here, a novel twisting polymorph for the paracetamol-theophylline cocrystal system was identified. A scanning electron diffraction scan of these plate-shaped particles contained significant orientation variation due to the twisting. It was possible to characterise the twist rate in the structure, and subsequently reconstruct a slice of 3D reciprocal space. This slice made it possible to reconstruct a monoclinic unit cell with space group P2<sub>1</sub>/b with a unique *c* axis and enabled the characterisation of the crystal structure from a single 2D scan of the beam-sensitive material. In addition, SED revealed the morphology and complex nanoscale orientation relationships present in a way that could not have been done with conventional methods, due to the relatively fast twist rate of the identified crystal. Overall, SED can be used to characterise correlated disorder even in beam-sensitive materials.

To enable this characterisation, a number of new analysis tools were developed. A computational method for computing twisting in the plate-like crystals observed was developed. This enabled 'SED tomography', a 3D reconstruction of a slice of the reciprocal lattice from only a single SED scan. Such analysis has not been done before, and the method could be generalised for other twisting structures and be applied generally in organic specimens where twisting is present. Overall, it showcases the power of understanding the sample interactions and geometry in SED scans. Such understanding is vital to the analysis of correlated disorder, as shown here.

# 6 Polymer Semiconductors

## 6.1 Background and Motivation

Polymers are a class of materials formed of long macromolecules consisting of linked organic monomers. From natural polymers like DNA to synthetic ones like polyethylene, polymers are omnipresent. However, surprisingly little is known about their "microstructure", as measuring it is complicated by the heterogeneity in polymer chain environments. Bond energies and structures can be measured through spectroscopic methods, such as infrared spectroscopy or UV-vis spectroscopy [69] but the microstructural elements and features have been difficult to characterise due to the significant amorphous fractions of polymer present. Typically, X-ray diffraction and electron diffraction studies have found some order in amorphous polymers, such as polystyrene [281], but generally polymers are either described through the X-ray diffraction of large crystals or amorphous 'solidified liquid' models.

Polymers crystallise. However, in contrast to crystalline solids, or even molecular solids, polymers are normally only *semi-crystalline*, in that they have mixtures of crystalline and amorphous regions. This is easily deduced from polymer samples typically having densities in between their amorphous and fully crystallised structures (where such structures are possible to synthesise). The theory is that the chain-like nature of polymer 'packing units' generally prevents them from rearranging into perfect crystals. Therefore, polymers with high chain mobility and high thermodynamic driving factors are typically the most crystalline. However, orientation relationships and sizes of such domains are relatively poorly known. This is primarily because most measurements of amorphous and crystalline segments have come from a mixture of X-ray diffraction and large solution-grown single crystals that are not representative of actual cast samples [69].

Typically, polymer crystals are either characterised as spherulitic [282], twisting or some kind of semicrystalline [4] structure, with regions of greater and lesser ordering of polymer chains (**Figure 6.1**a-c). SED of polyethylene [283] revealed complex twinned, twisting, and multi-phase microstructures, and there is reason to believe that most polymers exhibit complex structure. However, as with many organic compounds the characterisation of polymers is typically limited by rapid damage under probing radiation and further complicated by a significant non-crystalline fraction of material. Furthermore, the structure can depend on processing conditions [282], [284] (**Figure 6.1**d). Low-dose SED can fill this characterisation gap, especially as it is capable of probing both crystalline and amorphous components of the microstructure at doses before significant structural damage.

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**Figure 6.1.** Polymer microstructure. (a) Polarised light image of a spherulite, adapted from [69]. (b) Schematic of the structure of such a spherulite adapted from<sup>3</sup>. (c) Semicrystalline and amorphous microstructures. (d) Flate plate X-ray diffraction patterns from polypropylene samples of an isotropic crystal (left) and an oriented fibre (right). Adapted from [69].

<sup>&</sup>lt;sup>3</sup> From Wikimedia Commons (Materialscientist): https://commons.wikimedia.org/wiki/File:Spherulite2.PNG

## 6.1.1 Organic Mixed Ion-electron Conductors

In this work, several known organic mixed ionic-electronic conductors (OMIECs) are probed. Simultaneous ionic and electronic conductivity is a fundamental functional property of solid-state electrochemical devices used for energy storage [285], desalination [286], electrocatalysis [287], electrochromics [288], and bioelectronic devices [289]. Despite widespread use, the dynamics and details of the coupling of ionic and electronic motion in OMIECs are not fully understood, particularly in polymer semiconductors [290]. This is partly because the general micro- and nanostructure of these polymer semiconductors is also poorly understood, with a classic view generally involving 'crystalline' and 'amorphous' regions [291]. Structural models are generally restricted to approximate models due to a lack of a repeating unit cell and a lack of experimental data on polymeric microstructures [292]. SED provides a route to probing these rich semi-crystalline structures, with recent work enabling large-scale mapping of chain orientation through careful analysis of the variation in the scattering in the  $\pi$ - $\pi$  stacking of polymer chains [144], [293].

# 6.2 Sample Preparation and Characterisation

## 6.2.1 Sample Preparation

BBL

A commercial sample of poly(benzimidazobenzophenanthroline) (BBL) was dissolved in methanesulfonic acid to create a solution for spin coating.

## p(g1T2-g5T2) and p(g0T2-g6T2)

The samples were made by dissolving 5 mg mL<sup>-1</sup> of the glycolated polythiophene polymer (p(g1T2-g5T2) or p(g0T2-g6T2)) in chloroform.

## **TEM Samples**

The TEM samples were all made by first spin coating a layer of poly(3,4-ethylenedioxythiophene) polystyrene sulfonate (PEDOT:PSS) at 3000 RPM onto a polished silicon substrate. The PEDOT:PSS was dried at 120°C for 10 minutes, before a second layer (BBL, p(g0T2-g6T2), or p(g1T2-g5T2)) was spin coated at 3000 RPM. The layers were scored, after which the film was placed in deionised water, dissolving the PEDOT:PSS. A loop tool was used to pick up the samples from the water and they were subsequently placed on a TEM grid before being dried with filter paper. The p(g0T2-g6T2) and p(g1T2-g5T2) films were subsequently annealed at 120 °C for 20 minutes to ensure similarity between TEM samples and in-device films.

## 6.2.2 Film Thickness

The film thicknesses were measured using a DekTak XT Profilometer with a scan rate of 17  $\mu$ m s<sup>-1</sup> and a stylus force of 1 mg. The p(g0T2-g6T2) and p(g1T2-g5T2) films were measured to be approximately 80-100 nm in thickness, while the BBL film was measured to be between 35 and 50 nm.

## 6.3 Analysis Methods

## 6.3.1 2D Azimuthal Integration

2D-azimuthal integration, often also called polar reprojection, is not so much an integration as a change-of-basis operation, and is performed as an intermediate step in data processing. A diffraction pattern is 'unravelled' from  $k_x - k_y$  space into  $k - \varphi$  space, shown below in **Figure 6.2**. A  $\Delta k$  and  $\Delta \varphi$  are specified, and all data within a  $\Delta k - \Delta \varphi$  window of the original diffraction pattern are integrated to calculate the 2D azimuthal integral at a given k and  $\varphi$ . The resultant 2D-azimuthal integral is often further processed, such as to produce  $k - \varphi$  correlation maps, and it is much easier to post process than the direct diffraction patterns. The `dp.get\_azimuthal\_integral2d()` method can be used to do this in Pyxem. Virtual dark field images (VDFs) can subsequently be acquired from these data much in the same way as from the diffraction pattern, but the 2D-azimuthal integral can be used to separate  $\varphi$ -space more cleanly than a circular or rectangular aperture on the diffraction plane.



**Figure 6.2** A schematic of 2D-azimuthal integration of a diffraction pattern (left) into the 2D-azimuthal integral (right). Each angular slice  $\Delta \varphi$  is integrated in  $\varphi$  at each value of k.

## 6.3.2 2D Angular Correlation

The 2D-azimuthal integrals (above) can further be processed to measure angular correlations in  $\varphi$  at each value of k. Pyxem can do this for a whole 2D-azimuthal integral using the `ai2d.get\_angular\_correlation()` method. An autocorrelation is performed, giving the 2D angular correlation  $C(\phi, \mathbf{k})$ 

$$C(\phi, \mathbf{k}) = \frac{\langle I(\varphi, \mathbf{k}) * I(\varphi + \phi, \mathbf{k}) \rangle_{\varphi} - \langle I(\varphi, \mathbf{k}) \rangle_{\varphi}^{2}}{\langle I(\varphi, \mathbf{k}) \rangle_{\varphi}^{2}}$$
(61)

where the mean  $< \cdots >$  is computed over  $0 < \varphi < 2\pi$ . This 2D-map quantifies the likelihood of finding intensity at both angles  $\varphi$  and  $\varphi + \phi$  at each value of k. A negative value implies anticorrelation. An example is shown in **Figure 6.3**. For SED data, it is important to measure this correlation at each pixel, and then subsequently sum over the pixels. This can then unravel symmetry at values of k that could not be obtained from TEM diffraction, where the 'summing' is done before obtaining correlations.



**Figure 6.3** An example 2D angular correlation for the diffraction pattern in **Figure 6.2**. Black is high intensity. The observed intensity is stronger and comparatively less noisy at low k.

### 6.3.3 Flow Maps

After obtaining a 2D-azimuthal integral of a dataset, it is straightforward to cut this dataset into in k or in  $\varphi$ . In polymers, where the chain orientation is primarily encoded in the first major ring (usually around 0.3 Å<sup>-1</sup>), it can be useful to try and map the dominant orientation in that ring [293], [294]. An example is shown in **Figure 6.4**. By integrating the appropriate k-range, a profile of the intensity around the diffracted ring responsible for the chain orientation can be acquired. This k-range is typically chosen to be equal to the size of the diffracted spots. After integration, this ring can still often be very noisy, and to more accurately find the centre of the strongest peak, a Savitzky-Golay filter may be applied to the data [295]. This has the effect of smoothing the data in the case of high noise. Subsequently, the orientation of the highest intensity is picked at each pixel to form a flow map.

It is worth noting that in the case of multiple through-thickness orientations, as is the case in **Figure 6.4**, only the most intense orientation will be picked for the flow map. If multiple through-thickness domains are expected, a series of VDFs is more suited to display the data.



**Figure 6.4** Schematic acquisition of a polymer flow map. The 2D azimuthal integral (a) is integrated in k across the box selected, giving a 1D profile (b). The box was selected as it contained the intensity from the  $\pi$ - $\pi$  stacking of the polymer. This profile may then be convoluted with a Savitzky-Golay filter (c) followed by picking of the highest intensity peak for a dominant local orientation (d), with colours representing the g-vector direction (top right inset).

#### 6.3.3.1 Visualising Polymer flow maps.

There are multiple methods to visualise flow maps described. Several methods can be used directly within Python, such as Matplotlib's streamplot() method. The method works by moving 'points' through an image, where each point is described by a vector direction. In effect, it tries to draw continuous lines across oriented areas. Colour wheels are frequently used to describe orientation in

both magnetic imaging and EBSD, and such methods have also been employed here. However, note that only a 180° degree symmetry needs to be used, as only a chain orientation, rather than vector direction, is plotted. A few different methods of visualising are shown in **Figure 6.5**.



**Figure 6.5** Flow maps and misorientation maps of p(g0T2-g6T2) using streamplot. With (a) orientation of grains plotted according to the colour wheel in (e), with black lines showcasing polymer orientation. (b) The average 3x3 kernel misorientation, plotted as the square root of the reduced variance from Eq. (63). Bright areas have low kernel misorientation. Sharper boundaries with larger angles appear darker. (c) The orientation map from (a) plotted on the misorientation map from (b). (d) The same data as in (c), but with flow lines orientation plotted by colour and the misorientation plotted as in (b) and with inverted contrast.

#### 6.3.4 Kernel Misorientation

In addition, once a map of local orientations has been acquired, kernel misorientation [296] can be used to identify regions of 'messy' orientations. Different misorientation metrics can be used, in this case, we use the complex variance  $\operatorname{Var}_{x,y}$  of the assigned local orientation  $\phi_a$  in the 3x3 kernel at position at position x, y:

$$\operatorname{Var}_{x,y} = \sum_{x_i=x-1}^{x+1} \sum_{y_i=y-1}^{y+1} \frac{\left| (\cos 2\phi_a - \langle \cos 2\phi_a \rangle) + i [\sin 2\phi_a - \langle \sin 2\phi_a \rangle] \right|^2}{N-1}$$
(62)

where  $\langle ... \rangle$  denotes the mean,  $i = \sqrt{-1}$ , and N is the number of pixels in the kernel. Use of  $2\phi_a$  rather than  $\phi_a$  is to incorporate 180° symmetry in chain orientation. In practice, this can be implemented using Numpy's var() method, and passing in a signal of  $\cos 2\phi_a + i \sin 2\phi_a$ . This method is susceptible to single pixel misalignment in non-crystalline samples. Then, it is frequently better to use a more conservative estimate that ignores the pixel contributing most to the variance, using a reduced variance  $\operatorname{Var}_{x,y}^r$ 

$$\operatorname{Var}_{x,y}^{r} = \min_{x_{j},y_{j}} \left( \operatorname{Var}_{x,y} \left( x_{i} \neq x_{j} \cap y_{i} \neq y_{j} \right) \right)$$
(63)

which is measuring the variance for a reduced kernel without element  $x_j$ ,  $y_j$  for all elements  $x_j$ ,  $y_j$  in the kernel and picking the minimum. In effect, the pixel contributing most to the variance is removed. Such maps are particularly useful for materials with low crystallinity. An example misorientation map is shown above in **Figure 6.5**. Using  $\cos 2\phi_a + i \sin 2\phi_a$  as the measure for variance means that boundaries with high angles of misorientation appear dark, while those with low angles appear brighter. Maximum intensity is seen when all orientations with a kernel align.

#### 6.3.5 Variance Imaging

As noted in Section 3.2, an increase in crystallinity at a defined scattering vector magnitude k is associated with increased variance in the intensity distribution at k. This can be a powerful tool in the analysis of semicrystalline materials, where the presence of crystalline spots is not guaranteed, and they further vary in intensity. After obtaining a 2D Azimuthal Integral (see Section 6.3.1), one can extract a 1D intensity distribution  $I(\phi)$  at a given value of k, typically a summation of a range of k values as the diffracted spots have a finite size in the detector plane. The variance Var(k) in this intensity distribution is then correlated to the 'speckliness' of the pattern at k, defined as

$$Var(k) = \frac{\sum_{\phi} [I(k,\phi) - \langle I(k) \rangle]^2}{N-1}$$

where  $\langle ... \rangle$  denotes the mean and N is the number of pixels in  $I(\mathbf{k}, \phi)$ . Real-space regions with intense peaks, which are generally correlated with crystallinity, will have high variance, giving a proxy for a crystallinity map. Numpy's var() method is used for the practical implementation. A schematic acquisition example is shown below in **Figure 6.6**.



**Figure 6.6** Schematic acquisition of a variance map. (a) The 2D azimuthal integral from a single pixel (c) is chosen. The 2D azimuthal integral (a) is integrated in k across the box selected, giving a 1D profile (b). The variance of this profile gives a value per pixel, allowing construction of a variance map (d).

## 6.4 Results and Discussion

#### 6.4.1 BBL

BBL is a highly-stable [297] semiconducting ladder polymer that is often used as an *n*-type transport layer in polymer-based devices [298], including in organic field effect transistors in monolayer form [299]. Its monomer unit is shown in **Figure 6.7**, and it has the monstrous full chemical name of poly[(7oxo-7H,12H-benz[de]-imidazo[4',5':5,6]benzimidazo[2,1-a]isoquinoline-3,4:II,12-tetranyI)-12carbonyI] (BBL) [300]. It is an archetypal polymer semiconductor; unusually stiff with a Kuhn Length of 154 nm [301], beam-resistant for an organic compound, and showcasing both ionic and electronic conductivity [302]. However, relatively little is known about its structure, with structural studies primarily giving out a bare-bones unit cell [303] and no microstructural maps.



Figure 6.7. Monomer repeat unit of BBL.

BBL was probed in this work due to its ionic conduction properties, which were investigated by collaborator Scott T. Keene. It was chosen as the first sample due to its significant resistance to probing radiation. This helped optimise acquisition conditions in a regime where signal-to-noise ratio was not a concern, which was immensely helpful on the microscope. Its unique and bizarre VDF contrast, shown below, highlighted the need for further study, which is shown below.

#### 6.4.1.1 Data Acquisition

S(P)ED data of a spin-cast BBL film was acquired on a TS Spectra 300 operated at 300 kV with a beam diameter of ca. 5 nm and a convergence semi-angle of ca. 0.6 mrad and a precession angle of 1°. Data were also acquired without precessing. The beam current was varied between 2 and 80 pA and the frame rate was 1 ms per pattern, with slight overlap between the beam at adjacent pixels, leading to an electron dose of between ca. 8 e<sup>-</sup>/Å<sup>2</sup> and ca. 320 e<sup>-</sup>/Å<sup>2</sup>. The critical dose for BBL is not known but is expected to be higher than most polymer semiconductors. PCBM, with a critical dose of 300-400 e<sup>-</sup>/Å<sup>2</sup>, is a good comparison [304]. The diffraction patterns were acquired using a single-chip Medipix3 direct electron detector with 256x256 pixels and a  $\Delta s$  of 0.011 Å<sup>-1</sup>, leading to a maximum scattering vector of 1.4 Å<sup>-1</sup>. The data was acquired in TEM microprobe mode.

In addition, flat-field TEM diffraction data were acquired on a FEI Tecnai F20 FEG TEM operated at 200 kV. A selected area aperture was used to acquire data over a field of ca. 1  $\mu$ m<sup>2</sup> and was acquired on a Gatan OneView with 4096x4096 pixels and a  $\Delta s$  of 5.6×10<sup>-4</sup> Å<sup>-1</sup>. The diffraction pattern was slightly off-centre, leading to a maximum scattering vector of roughly 1.2 Å<sup>-1</sup>. A later tilt series was acquired under the same conditions bar a  $\Delta s$  of 9.1×10<sup>-4</sup> Å<sup>-1</sup>.

#### 6.4.1.2 Results

The films investigated had a thickness of ca. 30-50 nm, as noted above in Section 6.2.2. The TEM diffraction data in **Figure 6.8** showed a clear first ring at 0.29 Å<sup>-1</sup> associated with the  $\pi$ — $\pi$  bond stacking of the BBL monomers [305]. An additional triple ring was seen at ca. 0.5 Å<sup>-1</sup>, with exact positions of 0.41 Å<sup>-1</sup>, 0.50 Å<sup>-1</sup>, and 0.58 Å<sup>-1</sup> respectively. These are likely associated with the lateral packing of the chains, discussed further below. An additional broad ring, possibly consisting of multiple rings, was seen between 0.78 Å<sup>-1</sup> and 0.91 Å<sup>-1</sup>. A reduced intensity and reduced PDF was acquired on the same data, shown in **Figure 6.8**d-e. Notably, no peak was seen at 3.4 Å, the perpendicular distance of the  $\pi$ - $\pi$  stacking. As most of the chains are aligned in-plane with the  $\pi$ - $\pi$  stacking also in-plane (molecules lying on their sides), described further below, the probed atom-to-atom distances will be primarily along the polymer chain and parallel to the  $\pi$ - $\pi$  stacking (see Section 2.6.1). The chains must be laterally displaced, with a chain-chain separation of roughly 4.1-4.6 Å. This is consistent with previous studies produced a staggered unit cell with a chain-chain distance of roughly 4.3 Å [303]. The PDF has fairly broad peaks at low r, which is consistent with the range of bond length present in the monomer units, although finer resolution is limited by  $s_{max}$ .



**Figure 6.8** (a) Flat-field TEM scattering from BBL. (b) (c), the same figure at different brightness and contrast, highlighting the triple ring around 0.6 Å<sup>-1</sup> And a subsequent broad ring around 0.9 Å<sup>-1</sup>. (d) Reduced intensity of the signal from (a), clearly showing the aforementioned features. (e) The reduced PDF of (d) with the notable peaks labelled between 1 Å and 10 Å.

The SED data showed 'spottiness', as noted in **Figure 6.9**, with notable intensity variation in the first ring. Intensity variations were also observed in the triplet of rings further out, although the difference there was weak. The sample also showed thickness variation in SED, likely as a result of the lift-off process, showcased in **Figure 6.9**a-b. The variation in ADF intensity was 35% of the maximum observed intensity and will likely contain some structural contrast due to 0.8 Å<sup>-1</sup> < s < 1.2 Å<sup>-1</sup>. The thickness

variation had a 'bulbous' texture. The texture might also be associated with areas of relatively low density, which is likely to occur between more crystalline regions of the polymer due to the high Kuhn length of BBL of 154 nm [301]. These two effects cannot be easily ruled out, so it was not interpreted further.



**Figure 6.9** (a) Annular dark field image of a scanning electron diffraction (SED) scan with 0.8 Å<sup>-1</sup> < s < 1.2 Å<sup>-1</sup>. (b) Inset with changed contrast showing up to 35% (from maximum) thickness variation. (c-d) Typical single pixel diffraction patterns, showing multiple diffraction 'spots' at 0.29 Å<sup>-1</sup>.

Virtual dark field (VDF) imaging of the 1<sup>st</sup> ring of the same data shown in **Figure 6.10** revealed strong needle-like contrast with the orientation of the needle long axis perpendicular to the diffracted spot. These needles had widths between 5 and 30 nm, and lengths between 10 and 100 nm, with occasional larger domains having lengths of over 250 nm. This is consistent with needle-like domains of ordering in the BBL, with order primarily sustained in the direction of the polymer chains, likely due to the high Kuhn length of the BBL. Most domains were less than 20 nm in length, but the occasional larger domains dominate the total VDF contrast due to their much larger volume. The orientation of the strongest intensity at each pixel was plotted to try to create a polymer orientation map, but the map shows no 'domains' of orientation, in contrast to the VDF analysis above. This is a result of multiple ordered domains through-thickness with different orientations, resulting in the most intense orientation varying rapidly. The multiple orientations can be seen in single patterns shown in **Figure 6.9**c-d.

Overall, BBL shows a clear 'needle'-domain structure, with the needles in the orientation of the stiff BBL chains. The overlap of multiple needles is expected, as with the assuming 5-30 nm domain thickness range and a film thickness 30-50 nm, multiple domains would be expected to be overlaid to cover the thickness of the film. Such a 'pot of needles' further poses questions of intra-film porosity and density variation. The 35% (from maximum) ADF intensity variation suggests the film density may vary, although this could be a result of surface roughness from the processing method which involved an acid bath. Such surface roughness was visible in TEM images after tilting (**Figure 6.11**), where a knobbly texture was visible during tilting of the sample, on the scale of 10s of nm. Further combined SED and HAADF-STEM tomography is planned to separate these two effects.



**Figure 6.10** Virtual dark field images of the BBL. (a,b,e,f) VDF apertures placed on the first ring of the SED BBL scattering data. (c,d,g,h) The corresponding needle-like nanodomain VDF contrast.



**Figure 6.11** TEM image of BBL under (a) 0° and (b) 45°  $\alpha$ -tilt. The tilt axis is close to vertical.

Due to the Kuhn Length (154 nm) being several times the film thickness (50 nm), there are unlikely to be needles oriented out of the film plane. This was investigated by TEM diffraction on the Tecnai while tilting the sample, to see if the diffraction pattern would become a non-circular 'fibre pattern' under rotation (as only the tilt-axis remains unrotated). The results are shown in **Figure 6.12**. Under a 45°  $\alpha$ tilt, a clear change in the diffraction pattern was observed as expected, indicating a strong planar preferred orientation in the film. The  $\pi$ - $\pi$  ring faded to near zero perpendicular to the rotation axis (**Figure 6.12**c), while the triplet around 0.6  $Å^{-1}$  faded to zero at around 45° (in the diffraction plane) to the rotation axis, before being replaced with a single broader peak at around 60-90° to the rotation axis (Figure 6.12d). The broad peak around 0.85 Å<sup>-1</sup> became more intense, reaching a maximum around 40° to the rotation axis (Figure 6.12e), before dropping back down at 90° to the axis. Interestingly, these indicate that not only are the chains aligned in-plane, but that the  $\pi$ - $\pi$  ordering is roughly perpendicular to the film surface. If the  $\pi$ - $\pi$  order was random, for each domain rotating out of the diffracting condition, another would be expected to rotate in, resulting in a diffracted intensity that was unchanged across the  $\pi$ - $\pi$  ring under  $\alpha$ -tilt. Its absence indicates that polymer units are aligned in-plane with the  $\pi$ - $\pi$  stacking orientation also in-plane, and is consistent with a 'nanobelt' domain structure previously reported by Briseno et al [305]. Their nanobelts (prepared in a stirred chloroform-methane mixture) were measured to have a thickness of 10-50 nm, and a width of 200-1000 nm, with the  $\pi$ - $\pi$ -stacking perpendicular to the width (Figure 6.13). Since the spin coated film here had a thickness of 30-50 nm, it is easy to rationalise that these belts are in-plane, resulting in strong  $\pi$ - $\pi$ -orientation, and a 'pot of needles' microstructure (Figure 6.13b). However, not all the needle domains are in-plane, as some intensity in the  $\pi$ - $\pi$  ring is observed up to approximately 40° rotation from in-plane (projection of the 80° in-plane and 45° tilt). This is consistent with the same data plotted for 30° tilt (Figure 6.14), where the intensity of the ring does not go to zero at any angle to the rotation axis. In general, all observed effects were reduced at 30° tilt, with the exception that no intensity peak at was seen around 40° at k=0.85 Å<sup>-1</sup>. Possible sources of this tilting are bending, orientation variation, and twisting, but distinguishing the three is not possible without a diffraction tilt series.



**Figure 6.12** BBL Diffraction under 45°  $\alpha$ -tilt. The horizontal axis is the tilt axis and remains unchanged (blue), moving to 45 degrees from the tilt axis (red) to 90° to the tilt axis (green). (a) Diffraction pattern, with fibre-pattern halo. (b) Diffraction trace as a function of angle from the horizontal. The sudden drop off in the blue trace is due to the beam stop. (c)-(e) Zoom into the dashed line areas in (b), at around 0.28, 0.5, and 0.8 Å<sup>-1</sup>.



**Figure 6.13** BBL (a) Nanobelt structure suggested by Briseno *et al.*, adapted from [305], with the chain orientation perpendicular to the medium length axis b. (b) A schematic for the 'pot of needles' structure observed in the BBL films.



**Figure 6.14** BBL Diffraction under 30°  $\alpha$ -tilt. The horizontal axis is the tilt axis and remains unchanged (blue), moving to 45 degrees from the tilt axis (red) to 90° to the tilt axis (green). (a) Diffraction pattern, with fibre-pattern halo. (b) Diffraction trace as a function of angle from the horizontal. The sudden drop off in the blue trace is due to the beam stop. (c)-(e) Zoom into the dashed line areas in (b), at around 0.28, 0.5, and 0.8 Å<sup>-1</sup>.

#### **Diffraction Correlation**

The SED dataset was integrated using 2D-azimuthal integration, giving a map in  $k - \phi$  space where  $\phi$  is the angle around the diffraction plane, rather than  $k_x - k_y$  space. Subsequently, an angular correlation in  $\phi$  was acquired at each pixel by measuring the convolution  $\int I(\phi')I(\phi - \phi')d\phi'$  as a function of  $\phi$  and k. The results are shown in **Figure 6.15**. A strong 0° correlation was observed around 0.29 Å<sup>-1</sup>, but correlations at no other angles were observed. Additional 0° correlations were observed in the 0.4-0.6 Å<sup>-1</sup>, and an additional 72° and 108° correlation was observed at 0.54 Å<sup>-1</sup>, and a 48° and 132° correlation was observed around 0.61 Å<sup>-1</sup>. The non-0° correlations had some range of k over which they occurred, and they were assigned to the second and third triplet ('2.2' and '2.3') rings.

The 0° correlations reflect the intensity of the 'spottiness' at that *k*-vector, while non-0° correlations imply 'symmetry' present in the patterns at a given *k*. The lack of non-0° correlation around 0.29 Å<sup>-1</sup> implies that the orientations of multiple diffraction spots in the first ring are not correlated. This is consistent with the 'pot of needles' structure. The 48° and 72° correlations are unexpected, as the *k*-

spacings correspond to intra-molecular distances, but the expected angle for such a correlation would be 60° from the six-fold rings. Therefore, the 48° and 72° likely reflect the relative shear of chain-chain packing within the material, something that has not been previously studied in BBL. These correlations are not visible in single diffraction patterns as they are relatively incredibly weak, under 5% of the intensity of the  $\pi$ - $\pi$  stacking correlation.



**Figure 6.15** Angular correlation as a function of scattering vector magnitude. The scattering vector ranges from the edge of the direct beam (red) to 1.3 Å<sup>-1</sup> (blue). Ring 1 corresponds to the primary ring at 0.29 Å<sup>-1</sup>, while rings 2.1, 2.2, and 2.3 refer to the three components of the triplet ring seen in **Figure 6.8**. The calculation was done at each pixel and summed afterwards. Higher peaks imply higher intensity of diffracted spots.

To further probe the origin of this angular correlation, the self-correlation data from **Figure 6.15** was transformed back into  $k_x - k_y$  space, shown in **Figure 6.16**. The resulting image, while looking somewhat like a diffraction pattern, should not be interpreted as such. It has positive and negative intensities due to being a correlation rather than an intensity, and all intensities may not appear at once in a single pattern, especially as the correlation at each value of k is calculated independently of one another, such that at each value of k only the self-correlation is measured. Correlations for each value of k with the  $\pi$ - $\pi$  stacking ring were also calculated, shown in **Figure 6.16**c. As the intensity of the diffraction pattern is dominated by the  $\pi$ - $\pi$  stacking ring, this should result in a more diffraction-pattern like correlation. Similar features were observed as in the self-correlation, with slightly changed intensities. This is likely fortuitous, as the observed correlations are near the relatively strong diffracted spot at half the  $\pi$ - $\pi$  spacing observed at k=0.58 Å<sup>-1</sup>. This strong spot results in a strong measured correlation. The data are plotted after a background subtraction to highlight the correlations.


**Figure 6.16**  $k_x - k_y$  reprojection of the measured angular correlation in the BBL. (a) The data from **Figure 6.15**, projected into 2D. 0° correlation is horizontal. A missing wedge of data occurs due to the artificially high overlap at low correlation angle, that was subsequently removed. Black is high intensity. (b) The data in (a), but with the median correlated intensity at each value in k subtracted. As a result, correlation at high k are easier to visualise. The 48° and 72° correlation angles are highlighted, as is a potential 'unit cell'. (c) A median reduced map, as in (b), but for correlation between data in the strong diffracting ring. An additional broad (horizontal) band near 90° correlation is observed for the 2.1 ring.

In both cases, the 48° and 72° correlations both appear in the same horizontal line in the 2D scan, forming a line of 'reflections' as shown clearly in **Figure 6.16**b-c. This horizontal line is at a distance of 0.45 Å<sup>-1</sup> (2.2 Å). The relative displacement of the two correlations in the horizontal direction is ca. 0.6 Å<sup>-1</sup>, which is roughly twice the  $\pi$ - $\pi$  stacking distance. A potential monoclinic unit cell is drawn overlaid in **Figure 6.16**b. Note that, as the correlation is symmetric, both the unit cell and its mirror should show reflections, as demonstrated in **Figure 6.16**c. This monoclinic cell has an internal angle of 104°, and may reflect a longitudinal chain-chain displacement between polymer chains. The monomer unit is ca. 12 Å in length. The 104° angle implies a chain-chain longitudinal displacement of ca. 0.9 Å, which would subtend an angle of ca. 22° between equivalent positions, shown in **Figure 6.17**. Intensity is seen for the 2.1 ring in **Figure 6.16**c. This was not visible in **Figure 6.16**b (the self-correlation), suggesting it is directly correlated to the  $\pi$ - $\pi$  stacking peak being present. This band of intensity occurs between -22° and +22° from the vertical. As this reflection is perpendicular to the  $\pi$ - $\pi$  peak, it likely is from planes along the polymer chain alignment. The wide band, not seen for other reflections suggests that chains may have small random displacements with respect to each other. A visual example is given in **Figure 6.17**.



**Figure 6.17**. A schematic structural model for a BBL film with random small lateral displacement between chains. The three columns (left, centre, and right) describe different cases. The solid black lines represent the BBL chains. The same position along the chain is displaced slightly between each layer. The monomer units (top right) are aligned as shown between the two changes in axes on the left and the right. Other distances seen in the diffraction data are marked.

The observed correlations, when combined with the PDF data (**Figure 6.8**), suggest that the polymer chains do not simply stack on top of each other. Interestingly, a strong peak at 4.1 Å would be expected in the PDF if the chains were laterally displaced by 2.2 Å, the same distance as the horizontal line in **Figure 6.16**c. Such a distance is drawn on the monomer in **Figure 6.17**, and is approximately the distance between the centres of the carbon rings in the monomer structure. A 2.2 Å seems likely, and would likely be accompanied by a small displacement perpendicular to the chain (along *z* in **Figure 6.17**). Although the SED here is not sensitive to that displacement due to film texture, it does corroborate with previous work by Briseno *et al.* In addition, the broad lines of intensity (in the non  $\pi$ - $\pi$  stacking intensities) of the diffraction pattern suggest that some disorder is present, dominated by displacements between chains. These displacements are typically average 0.9 Å or less, resulting in a line of intensity between -22° and +22° from the vertical. This disorder may be dynamic or static. Overall, a significant amount of structural information can be found by correlating diffraction patterns in SED that cannot be found when examining larger areas of view in TEM mode, as each diffraction pattern is from a local area.

#### Measuring the Average Domain Size

To measure an average domain size in the BBL, which exhibited strong domain overlap, a VDF autocorrelation method was employed. Initially, VDFs were extracted for the  $\pi$ - $\pi$  stacking at a 6°angular interval; the integrated intensity from each 6° wedge was used for the same VDF. (Figure 6.18a-b). Subsequently, autocorrelations of these VDFs were then obtained (Figure 6.18c-d). The autocorrelations showed clear directionality along the domain orientation. A linear autocorrelation along the domain and perpendicular to it were then extracted at each angle, and the average was calculated. These autocorrelations were plotted and their average was used to measure an estimate of the average domain size. Autocorrelations should reveal a central 'peak', with its edge marking the average size of the domains. The edge, rather than the FWHM, is used as it was found to be consistent with the domain size in simulations. Along the domain direction, this edge was found to be at a length of 185 nm, which is very similar to the Kuhn length of the polymer (149 nm), suggesting that the chain stiffness is the primary driving factor in the unusual microstructure of the BBL films. In the perpendicular direction, two kinks were seen, at 65 nm, and at 105 nm. It is unclear which of these better represents the average domain width. Notably, both are greater than the thickness of the film (30-50 nm), explaining why these 'nanobelts' are aligned in plane, and hence why the primary  $\pi$ - $\pi$ stacking orientation of the film is also in the plane of the film.

#### Conclusion

Overall, the SED of BBL revealed a rich and complex micro- and nanostructure, with nanobelt-shaped domains. These domains stack on top of each other and are strongly oriented in-plane. Through the use of autocorrelation in both images and diffraction, additional information can be gleaned that could not be done using bulk methods. These give estimations of the crystallite domain size and the longitudinal chain-chain displacement respectively. When combined with bulk PDF analysis, an even more complete image of BBL's local packing environment can be calculated. Overall, SED enables us to understand the nanoscale variance and local structure of conducting polymers like BBL.



**Figure 6.18** Estimating the domain size of the BBL film using autocorrelation. (a-b) VDFs at different orientations around the central beam. (c-d) Their respective autocorrelations. The widths of the directional integration windows are exaggerated for clarity. The domain direction has a consistently larger correlation length. (e) All 60 autocorrelations along the domain direction for 6° angular VDFs (red dots) and their average (black line). Inset: Zoom into the edge of the central peak. The first kink (a local minimum) is seen at 185 nm. (f) All 60 perpendicular autocorrelations (cyan dots) and their average (black line). Inset: Zoom into the edge of the central peak. Two possible first kinks are seen, at 65 nm, and 105 nm.

### 6.4.2 p(g1T2-g5T2)

Thiophene-backbone ethylene glycol (EG)-conjugated polymers have shown great promise for use in organic electrochemical transistors due to their chemical synthesis and high ionic conductivity [306]. They are primarily formed of a thiophene backbone and are conjugated with different amounts of ethylene glycol sidechains attached to the backbone (**Figure 6.19**). The structure for p(g1T2-g5T2) is also shown in **Figure 6.19**, and other ethylene glycol (EG)-conjugated polymer structures are shown in **Figure 6.20**. These polymers have shown complex interlinked ionic and electronic conduction properties related to the local arrangement of their carbon chains as well as the local electronic doping [307]. p(g1T2-g5T2) is the polymer with the highest figure of merit for power amplification,  $\mu C^*$  where  $\mu$  is the charge mobility, and  $C^*$  the volumetric capacitance, known to date. Collaborator Scott Keene had found unusual ion charge transport at low dopant levels, that seemed to be related to polymer film microstructure, highlighted by small angle X-ray scattering experiments. SED was used to further probe such microstructural elements, as the microstructure of the films is not known.

These compounds are synthesised by functionalising partial monomer units with side groups that will only react with each other, in the presence of the catalyst tris(dibenzylideneacetone)dipalladium(0),  $(Pd_2(dba)_3)$  resulting in a complex monomer unit. The polymer chains had a length of between 16 and 36 monomer units. The p(g1T2-g5T2) solution was prepared by dissolving p(g1T2-g5T2) in chloroform (5 mg mL<sup>-1</sup>) and stirring at room temperature overnight. The films were made by spin-coating it onto a PEDOT-PSS film that was subsequently dissolved in water.



Figure 6.19. Structure of (a) p(g1T2-g5T2) and (b) p(g0T2-g6T2).

#### 6.4.2.1 Data Acquisition

Scanning electron diffraction data from the p(g1T2-g5T2) and p(g0T2-g6T2) films were acquired on a TF Spectra 300 operated at 300 kV with a beam diameter of ca. 3 nm and a convergence semi-angle of ca. 0.8 mrad. The beam current was 20 pA and the frame rate was 1 ms per pattern, with a step size of 5.6 nm, leading to an electron dose of ca. 40 e/Å<sup>2</sup>. The expected damage threshold for p(g1T2g5T2) is not precisely known, but the critical dose has been measured for similar semiconducting polymer P3HT and found to be ca. 18 e/Å<sup>2</sup> [308]. Data were acquired at doses above this value as previous data taken on polymers suggested that the critical dose in STEM may be considerably higher. Additional data were acquired at lower dose, with no discernible change in diffraction patterns. The diffraction patterns were acquired using a single-chip Medipix3 direct electron detector with 256x256 pixels, a camera length of 580 mm and a  $\Delta s$  of 2.8×10<sup>-3</sup> Å<sup>-1</sup>, leading to a maximum scattering vector of 0.5 Å<sup>-1</sup>. Additional data were acquired at a lower camera length of 185 mm and a  $\Delta s$  of 8.7×10<sup>-3</sup> Å<sup>-1</sup>, leading to a maximum scattering vector of 1.1 Å<sup>-1</sup>. In addition, flat-field TEM diffraction data were acquired on a FEI Tecnai F20 FEG TEM operated at 200 kV. A selected area aperture was used to acquire data over a field of ca. 1  $\mu$ m<sup>2</sup> on a Gatan OneView with 4096x4096 pixels and a  $\Delta s$  of 9.1×10<sup>-4</sup> Å<sup>-1</sup>, leading to a usable maximum scattering vector of roughly 1.8 Å<sup>-1</sup>. Additional data were acquired at 30° and 45°  $\alpha$ -tilt.



**Figure 6.20** Common thiophene-backbone ethylene glycol-conjugated organic semiconductors, including the family (F) to which p(g0T2-g6T2) and p(g1T2-g5T2) belong to. Figure adapted from [306].

#### 6.4.2.2 Results

#### **Palladium Inclusions**

HAADF-STEM data acquired simultaneously with SED data (k > 1 Å<sup>-1</sup>) of the p(g1T2-g5T2) showed very sharp nanometre-scale contrast (**Figure 6.21**). This was attributed to Pd inclusions from the catalyst, as reflections in the SED diffraction patterns of these inclusions were consistent with fcc-Pd (**Figure 6.22**). The HAADF-scan had a bimodal distribution of intensities, shown in **Figure 6.21**b-c. Most pixels followed a Gaussian distribution, while about 20% did not. By attributing the gaussian distribution to film thickness variation, and the additional distribution to Pd, a measure of the fraction of the scattered intensity that belongs to Pd could be estimated. These pixels are at nearly exactly the same positions as the intense regions in the HAADF image (**Figure 6.21**a and **Figure 6.22**a). A Gaussian was fitted to the low end of the data, and it was assumed that these 'intense' pixels had an average amount of scattering from the Gaussian as well as an additional Pd contribution. This is reasonable as the fraction of Pd is small, such that the intensity from the rest of the thickness would be expected to be almost the same. By integrating these profiles (**Figure 6.21**) Pd accounted for approximately 1.5-1.7% of the scattered intensity. To calculate the number fraction of Pd, the data was normalised by the relative scattering factor (assuming  $I \propto Z^2$  for HAADF scattering [85], which is not exactly true at this scattering range, but likely close) of Pd and p(g1T2-g5T2):

$$\frac{N_{\rm Pd}}{N_{\rm pol}} = \frac{I_{\rm Pd}Z_{\rm pol}^2}{I_{\rm pol}Z_{\rm Pd}^2}$$

(64)

giving an estimate of 0.02-0.03 at% Pd (200-300 ppm). Interestingly, following this method the maximum of the Pd-related intensity is equivalent to roughly 70 atoms of Pd within the probed beam volume at a single pixel. While this may have an effect, it is unlikely to shape the functional properties of the polymer.



**Figure 6.21** (a) HAADF-STEM image of p(g1T2-g5T2), showing strong local contrast attributed to Pd. (b) The intensity histogram of (a) in black dots, and the best fit (red). (c) The difference, plotted as a function of intensity subtracted by the centre of the gaussian in (b), the peak is at ca. 70 atoms Pd within the probed volume.



**Figure 6.22** Pd inclusion diffraction. (a) A map of the regions with intensities outside the range of the fitted Gaussian in **Figure 6.21**. (b) and (c): Example indexed diffraction patterns, consistent with fcc-Pd. (d) The summed 1D intensity from the pixels in (a) after background removal via a median kernel filter.

#### **Domain Structure**

The p(g1T2-g5T2) is nearly amorphous, showcasing relatively little chain alignment. Individual patterns appear 'nearly' amorphous, with little to no spottiness (**Figure 6.23**a). However, flow map analysis of the polymer reveals small, 20-100 nm size domains of preferential chain orientation (**Figure 6.23**b). There is a small preferential alignment in the film, unsurprising as it has been spin coated. The observed domains have varying degrees of internal misalignment, often showcasing high misorientation even within a domain (**Figure 6.23**d), with occasional stronger alignment. Domain and boundary positions were uncorrelated to locations of Pd inclusions (**Figure 6.23**c). It is likely that there are multiple through-thickness domains in the collected data. Given the low crystallinity of the sample such overlap could not be imaged, in contrast to the BBL.

Overall, the polymer is highly amorphous, with chains seemingly showing near-minimal alignment to their surrounding regions. A flow map was obtained and is presented in Figure 6.23c. Most of the observed domains are small and significant regions have no dominant orientation. More ordered regions are highly isolated with amorphous regions in between, as seen in the kernel misorientation map in Figure 6.23d. This isolation was correlated with decreased ionic mobility at low dopant doses, especially in comparison to typical OMIECs like PEDOT:PSS, and even compared to p(g0T2-g6T2). The degree of ordering was further analysed using variance measurements as a function of  $\phi$  across the  $\pi$ - $\pi$  stacking ring. The results are shown in **Figure 6.24**. Regions of stronger ordering are more clearly visible in the variance measures (Figure 6.24c), in contrast to a VDF (Figure 6.24b). These areas are small and isolated, and correlate to areas of low local misorientation (Figure 6.23d). The misorientation is quantified in Section 6.4.3.3. To check the variance is correlated to diffraction contrast, a variance map was acquired at a lower k (0.12-0.16 Å<sup>-1</sup>). No contrast was observed (Figure **6.24**d), highlighting the variance contrast is a result of variation in the intensity of  $\pi$ - $\pi$  stacking. These low misorientation regions have lengths up to a few 100 nm, but most are smaller and not well connected to each other. The relatively weak crystalline order in p(g1T2-g5T2) can be contrasted to the stronger variance seen in p(g0T2-g6T2) (Figure 6.27 in Section 6.4.3). No 2D angular correlations except at 0° were observed in the data.

Additional data sets are provided in **Figure 6.25**. **Figure 6.25**a highlights the separation of these slightly more ordered domains, while **Figure 6.25**b-c, at higher magnification, give better estimates of domain size, with 10s to 100 nm sized domains in general. Some highly ordered domains are seen in all three additional data sets, but once again these domains are isolated in the observed variance maps.

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**Figure 6.23** SED data from p(g1T2-g5T2). (a) A typical diffraction pattern. (b) Polymer flow map with a colour wheel inset in the top right for chain orientation. (c) Concurrent ADF image taken with the SED scan, with an angular range of roughly 0.5-1.5 Å<sup>-1</sup>. (d) Square root of the reduced kernel misorientation map.



**Figure 6.24** Variance imaging of p(g1T2-g5T2). (a) Concurrent ADF image taken with the SED scan, with an angular range of roughly 0.5-1.5 Å<sup>-1</sup>. (b) VDF of the intensity of the  $\pi$ - $\pi$  ring in the range 0.26-0.32 Å<sup>-1</sup>. (c) Angular variance of the  $\pi$ - $\pi$  ring, showing areas of locally high ordering. (d) Angular variance for 0.12-0.16 Å<sup>-1</sup>, showing a variance baseline.



**Figure 6.25** Other data sets from p(g1T2-g5T2), each with an ADF image, variance map, and flow map overlaid on a misorientation map. Data sets are at a lower magnification (a) (scale bar 500 nm) and higher magnification ((b) and (c)) (scale bar 100 nm). In (c) no HAADF was taken due to detector saturation at high current. Instead a VDF of the diffracted ring is provided.



**Figure 6.26** p(g1T2-g5T2) Diffraction under 45°  $\alpha$ -tilt. The horizontal axis is the tilt axis and remains unchanged (blue), moving to 45 degrees from the tilt axis (red) to 90° to the tilt axis (green). (a) Diffraction pattern, with fibre-pattern halo. (b) Diffraction trace as a function of angle from the horizontal in (a). The sudden drop off in the blue trace is due to the beam stop. (c)-(e) Zoom into the dashed areas (b), at around 0.28, 0.45, and 0.8 Å<sup>-1</sup>.

#### **TEM Diffraction**

The effects of the 45° tilt on the observed TEM diffraction pattern are shown in **Figure 6.26**. The weak  $\pi$ - $\pi$  peak at ca. 0.27 Å<sup>-1</sup> reduced in intensity from the unrotated axis, reaching a minimum perpendicular to the tilt axis. The peak did not go to zero. A second order-reflection seen around 0.45 Å<sup>-1</sup>, which was reduced in intensity while a reflection 0.5 Å<sup>-1</sup> increased in intensity. This region is where the Pd 111 and 200 peaks are seen. As a result it is unclear whether the change is to do with polymer ordering or with specific orientation of Pd inclusions. The small expected size of the Pd nanoclusters (< 5 nm in diameter) suggests they are unlikely to be oriented, and therefore that it is most likely something to do with the polymer ordering. This, however, is not conclusive. No further changes were observed. Overall, the effect of tilting is fairly small, especially when compared to BBL (see Section 6.4.1), but some in-plane ordering is observed.

### 6.4.3 p(g0T2-g6T2)

p(g0T2-g6T2) is a variant of p(g1T2-g5T2) with a different side chain arrangement (**Figure 6.19**). The lack of a glycolated tail on the shorter side chain pair tends to planarise with the polymer backbone, resulting in a relatively stiffer polymer chain than p(g1T2-g5T2). Its power amplification properties are not quite as high as p(g1T2-g5T2) but are nonetheless good. It was probed due to the chemical similarity to p(g1T2-g5T2), with small angle X-ray scattering implying microstructural differences. The sample was synthesised in the same way as the p(g1T2-g5T2). The polymer chains had a length of between 6 and 11 monomer units.

#### 6.4.3.1 Data Acquisition

The p(g0T2-g6T2) data was acquired under the same conditions as the p(g1T2-g5T2) film, described in Section 6.4.2.1.

#### 6.4.3.2 Results

#### **Palladium Inclusions**

As with p(g1T2-g5T2), HAADF-STEM (k > 1 Å<sup>-1</sup>) data of p(g0T2-g6T2) showed very sharp nanometrescale contrast attributed to Pd inclusions from the catalyst (**Figure 6.27**c). Interestingly, these inclusions are larger and fewer in number than in p(g1T2-g5T2), but the atomic fraction was calculated to be roughly equal at ca. 0.027% (270 ppm) using the same method as for p(g1T2-g5T2). The sparser distribution was also present in as-cast films, suggesting differences in the distribution between p(g1T2-g5T2) and p(g0T2-g6T2) are a result of the domain structure of each polymer. As a catalyst, the Pd likely plays a part in the domain growth, although it is possible it is simply shunted out of the domains during growth without playing a chemical part. While it may mediate growth, it is not the source of the different structural domains seen in the two samples, as roughly an equal amount of Pd was present in both samples.

#### **Domain Structure**

Flow map analysis of p(g0T2-g6T2) revealed large (100 nm to > 1  $\mu$ m) sized domains (**Figure 6.27**b) of strongly aligned orientation. The ordered domains contain clear diffraction spots (**Figure 6.27**a), in contrast to p(g1T2-g5T2), which showed little spottiness. In contrast to BBL, the domains in general are longer perpendicular to the chain orientation. This can be rationalised through the short length of the p(g0T2-g6T2) polymer units (6-11 monomers), with packing more akin to molecular crystals than BBL, which is common on many polymers [69]. The ordered domains were bordered by less ordered regions, as visible in the misorientation map (**Figure 6.27**d). Notably, not all changes in orientation had

boundaries, some areas showcased a change in orientation with a smooth boundary (red dashed region, **Figure 6.27**d), while others showed a sharp boundary (blue dashed region, **Figure 6.27**d). Further quantification of these misorientations is done below in Section 6.4.3.3.



**Figure 6.27** SED data from p(g0T2-g6T2). (a) A typical diffraction pattern. (b) Polymer flow map with a colour wheel inset in the top right for chain orientation. Dashed lines mark two regions of interest. (c) Concurrent ADF image taken with the SED scan, with an angular range of roughly 0.5-1.5 Å<sup>-1</sup>. (d) Square root of the reduced kernel misorientation map. Dashed lines mark the same two regions of interest.

Within regions, diffraction patterns were generally similar. An example of summed patterns from different regions for the data in **Figure 6.27** is shown in **Figure 6.28**. The average diffraction patterns matched well with the calculated chain orientations, with diffracted intensity perpendicular to the assigned chain orientation. The patterns showed arcing, consistent with slight local variation in orientation. The extent of this arcing varied, with the highly oriented region in **Figure 6.28**c showing less arcing than a similarly oriented but more varied region in **Figure 6.28**b. Little to no diffraction was seen elsewhere in the ring, suggesting that generally there is little to no through-thickness domain overlap.



**Figure 6.28** Regionally-averaged diffraction patterns from different parts of the scan shown in **Figure 6.27**. (a) Polymer flow map reproduced from **Figure 6.27**b, with dashed lines showing the approximate regions from which average diffraction patterns were obtained. (b-d) Square roots of the average diffraction patterns from the areas shown in (a).

Variance analysis of the p(g0T2-g6T2) (**Figure 6.29**) confirms much of the same results seen in the flow maps. Regions of high variance, corresponding to stronger crystalline ordering, are seen in the regions of low misorientation. The average value of the measured variance was 5.5 times higher than in p(g1T2-g5T2). The two are plotted side-by side with the same contrast in **Figure 6.30**, highlighting that p(g0T2-g6T2) has considerably higher diffraction variance. These regions are also uncorrelated to both the VDF of the diffracting ring (**Figure 6.29**b) from which the variance is measured, and to variance measured at a lower k range (**Figure 6.29**d). Hence, the bright regions are explicitly correlated to regions of high structural ordering in the sample, which were not present in p(g1T2-g5T2). These ordered domains are important for electronic transport at low dopant levels. Additional data sets at higher and lower magnification are shown in **Figure 6.31**. High variance regions are consistently seen

in the middles of domains, while ranges of smooth and misoriented boundaries are seen (Figure 6.27). Several interesting micro- and nanostructural features are annotated on Figure 6.31, such as a k = +1/2 disclination that resembles a fold in the chain orientation (Figure 6.31a), and k = +1 disclination with radiating chain orientation (bottom left, Figure 6.31b). A boundary structure with a line of defects is also seen [309] (top of Figure 6.31b), with the line consisting of 'spots' of misorientation across an otherwise smooth boundary, visible in both the variance and local orientation maps. At a smaller scale, small localised misoriented 'grains' are seen in the middle of oriented regions (Figure 6.31c). Their boundaries seem to correspond to spatially delocalised pairs of k = +1/2 and k = -1/2 disclination. Such a pair should mutually annihilate, but the presence of the second (orientation-wise inverted) pair likely prevents this from occurring. Disclinations are known to interact with each other [78]. Overall, all of these complex domains are not only highly ordered, but also well connected to each other, with sizes of 100s nm.



**Figure 6.29** Variance imaging of p(g0T2-g6T2). (a) Concurrent ADF image taken with the SED scan, with an angular range of roughly 0.5-1.5 Å<sup>-1</sup>. (b) VDF of the intensity of the  $\pi$ - $\pi$  ring in the range 0.26-0.32 Å<sup>-1</sup>. (c) Angular variance of the  $\pi$ - $\pi$  ring, showing areas of locally high ordering. (d) Angular variance for 0.12-0.16 Å<sup>-1</sup>, showing a variance baseline.



**Figure 6.30** Square roots of the variance images from p(g0T2-g6T2) (a,c,e) and p(g1T2-g5T2) (b,d,f) plotted with the same contrast levels. (a-b) with a maximum of 2.5, (c-d) with a maximum of 3.5, (e-f) with a maximum of 8.0.

#### **TEM Diffraction**

TEM diffraction under 45°  $\alpha$ -tilting of the p(g0T2-g6T2) film is shown in **Figure 6.32**. The results are very similar to p(g1T2-g5T2), with the  $\pi$ - $\pi$  peak at ca. 0.27 Å<sup>-1</sup> reduced in intensity from the unrotated axis, reaching a minimum perpendicular to the tilt axis. The peak did not go to zero but is closer to zero than in p(g1T2-g5T2), suggesting slightly stronger in-plane ordering. This is consistent with the larger domain size observed. A similar second order-reflection movement was seen around 0.45, with a decrease in intensity away from the unrotated axis and a neighbouring increase in intensity at 0.5 Å<sup>-1</sup>. This change could not be attributed to Pd or to the polymer. Overall, the effects of tilting were again small, especially when compared to BBL (see Section 6.4.1), but some in-plane ordering is observed.



**Figure 6.31** Other data sets from p(g0T2-g6T2), each with a ADF image, variance map, and flow map overlaid on a misorientation map. Some observed micro- and nanostructural features are annotated on top. Data sets are at a lower magnification ((a) and (b)) and higher magnification (c). (c) The black arrows represent k = -1/2 disclinations, and the red arrows represent k = +1/2 disclinations.



**Figure 6.32** p(g0T2-g6T2) Diffraction under 45°  $\alpha$ -tilt. The horizontal axis is the tilt axis and remains unchanged (blue), moving to 45 degrees from the tilt axis (red) to 90° to the tilt axis (green). (a) Diffraction pattern, with fibre-pattern halo. (b) Diffraction trace as a function of angle from the horizontal. The sudden drop off in the blue trace is due to the beam stop. (c)-(e) Zoom into the dashed areas in (b), at around 0.28, 0.45, and 0.8 Å<sup>-1</sup>.

#### Summary

Overall, it is possible to trace winding pathways that do not cross areas of high misorientation throughout p(g0T2-g6T2). This was correlated to higher mobilities of holes at low doping levels in contrast to p(g1T2-g5T2), in which no such pathways are possible. This enables the lowest energy doped states, that correspond to states in the more ordered regions, to contribute to conduction even before states in amorphous regions are able to contribute (which occurs at higher dopant levels), a key difference in functional properties.

#### 6.4.3.3 Misorientation Analysis

The two samples showcase notably different domain scales and structures. To quantify this, the misorientations present were plotted as a histogram (**Figure 6.33**). Overall, p(g0T2-g6T2) exhibits considerably lower misorientation, with most real-space positions showcasing near zero kernel variance. In contrast, p(g1T2-g5T2) has no such peak, highlighting its more amorphous nature. Interestingly, p(g0T2-g6T2) showcases small peaks in the misorientation. These were distributed within the sharp domain boundaries. The values of the peaks correspond to a roughly 90° boundary orientation, with 0.45, 0.75, and 0.9 corresponding to 1, 2, and 3 of the 8 orientations sampled (as one is discarded, see Equation ( 63 )) being at roughly 90° to the 7, 6, or 5 at a sharp boundary. No peak is seen for 4 of the 8 being one orientation as it would end up getting discarded to the reduced variance condition, which had to be used to reduce random noise in the kernel data. This suggests that a near 90° boundary may be favoured in the p(g0T2-g6T2), such a boundary would likely have some type of herringbone structure.



**Figure 6.33** Square root of the kernel misorientation of (a) p(g1T2-g5T2) and (b) p(g0T2-g6T2), reproduced from **Figure 6.23** and **Figure 6.27**. Both scale bars are 200 nm. (c) Histograms of the misorientation plotted as a function of variance, p(g1T2-g5T2) in red and p(g0T2-g6T2) in blue. The maximum for p(g0T2-g6T2) was approximately 12000.

## 6.5 Conclusion and Outlook

The three different polymers analysed here showcase complex microstructure present despite partial crystallinity. SED provides a key combination of sufficient spatial resolution and reciprocal space resolution to characterise local structural order, particularly in the  $\pi$ - $\pi$  stacking of the polymers. While some analysis of polymers in SED has been done before, here the boundaries of information that can be extracted from polymer samples is pushed to the limit. By carefully analysing the  $\pi$ - $\pi$  stacking orientation and VDF images, varied microstructures are revealed, with near amorphous structure in p(g1T2-g5T2), to semi-crystalline domain structures in p(g0T2-g6T2) to highly-ordered 'pot of needles' microstructures in BBL. The ability to characterise microstructural elements, domain boundaries, and even unit cell structures in these polymers highlight the power of SED in characterising semi-crystalline samples. Using a combination of amorphous and crystalline analysis tools, several visualisation methods are developed here, such as the kernel misorientation and variance methods. These are highlights of computational microscopy; the equivalent images could not be observed directly in the TEM, as they require computational analysis of correlated intensities within diffraction patterns. In addition, interacting disclination structures are seen for the first time in polymer thin films, and highlight the complex microstructure present. Low-dose SED is likely a key development in characterising polymer micro- and nanostructures. There is a wealth of information in their diffraction data, but it is frequently hard to observe due to the weak scattering from low atomic number elements. Yet, that information can be extracted by thorough analysis of intensities, such as in SED. SED promises exciting possibilities for the characterisation of polymers.

# 7 Other Projects

## 7.1 Metallic Glasses

### 7.1.1 Background and Motivation

Bulk metallic glasses (BMGs) remain a structurally and scientifically intriguing group of materials even as we near six decades since their discovery [310]. From a practical perspective, BMGs are of interest due to their unique structural and functional properties. They exhibit impressive mechanical properties [311], very high strengths (>5 GPa [312]), high hardness, high specific strengths, and excellent elastic limits. In addition, they present an intriguing case study for methods of plastic deformation in non-crystalline material [313]. Structurally, they retain the disordered atomic structure of high temperature melts and provide an intriguing analogue to the liquid state in terms of structure. Although disordered, BMGs exhibit strong medium range order, with typically frustrated icosahedral structural units on the order of 1-5 nm packing in an ordered but aperiodic structure [56]. However, the exact structure of both the icosahedral units and their long-range correlations remain generally elusive, although direct observations of single icosahedra have been made [56].

Probing the medium range structure is a common application of fluctuation electron microscopy (FEM) [314]–[316]. The greatest practical difficulty in FEM lies in typically needing to use a varying probe size to determine the scale of the MRO in the material. FEM was investigated here due to the high DQE, high speed direct electron detectors available. Kinematical models of FEM suggest that the observed fluctuations should be much higher than they are typically observed to be [1]. This may be due to electron-beam induced change or fluctuation of the sample under observation, which means that when recording a single diffraction pattern for a long period of time, a "smeared out" signal is observed. This would reduce the observed speckle in the data, and the subsequent observed variance. Hence, by using a faster and more efficient detector, and thus a considerably lower dose, much larger variance should be observed in the FEM data.

#### 7.1.2 FEM Results

5 samples with varying types of heat treatments were probed. These samples are termed A0, A1, A2, B0, and B1. A and B refer to different processing conditions, resulting in local packing that resembles fcc and bcc respectively, calculated using X-ray PDF analysis, while the number refers to the length of heat treatment. Hence, A2 is nearly entirely fcc-nanocrystalline, while A0 is almost entirely amorphous. The goal was to confirm this structural trend and see if one can extract further information about the local structure of the samples using local FEM analysis.

A SED data set of a series of BMGs with a composition of  $(Fe_{0.25}Co_{0.25}Ni_{0.25}Cr_{0.125}Mo_{0.125})_{0.89-0.86}B_{0.11-0.14}$ was acquired using a JEOL ARM300F STEM operated at 200 kV with a beam diameter of ca. 8 nm and a convergence semi-angle of 0.65 mrad. The beam current was 2 pA and the frame rate was 1 ms, leading to an electron dose of ca. 5 e<sup>-</sup>/Å<sup>2</sup>. Electron-induced damage is not a concern at these doses. The diffraction patterns were acquired using a Medipix3 direct electron detector with 256x256 pixels and a maximum scattering vector *s* of 0.7 Å<sup>-1</sup> (corresponding to a maximum  $q = 4\pi \sin \theta / \lambda$  of 9 Å<sup>-1</sup>). The data set was then used for FEM analysis as outlined in Section 3.2.



**Figure 7.1** Extracted FEM spectra from samples A0, A1, A2, B0, and B1 (each labelled top right) as well as a radial profile from A2 (top right). Colours in the FEM spectra constitute bcc-like peaks(red), fcc like peaks (blue), and something different (green), perhaps related to icosahedral packing. Clear structural differences are apparent. A2 is nanocrystalline fcc, but A1 and B1 both have mixtures of both fcc and bcc-like MRO, in different proportions. The peaks are considerably higher above the baseline in FEM than in a radial profile (top left).

The measured corrected normalised variance spectra for the five BMG samples are presented in Figure 7.1, along with a radial profile from sample A2 for comparison. Higher angle reflections are greatly strengthened in the variance spectrum. Each variance spectrum shown is calculated from 10,000 diffraction patterns. For the A2 sample, which is fcc-nanocrystalline, the peaks correspond to a lattice parameter 3.65 Å using the 220 reflection. A0 and B0 both appear almost amorphous, with slightly larger peak heights in A0, while both A1 and B1 show an in-between result.

It must be noted that the corrected variance background converges to a value below zero between the peaks in the data. The data should converge to one and shows that the acquired data set has subpoissonian variance, as the 1/mean correction effectively overcorrects for the poissonian variance. The potential source of this error is discussed further in Section 7.1.3 below.

As is shown by the presence of peaks in the corrected variance patterns, there is significant MRO in the metallic glass samples, particularly in A1, B1, and A2. The intensity of these variance peaks is very high compared to the baseline, orders of magnitude higher than those reported in literature [69, 118, 119], which typically have intensities of approximately 10<sup>-2</sup> to 10<sup>-3</sup>. This seems to suggest that electron dose indeed reduces the speckle observed. However, a problem is that the baseline itself is below zero. This problem was identified in all of the data sets measured.

Two groups of peaks are seen in the data, highlighted in **Figure 7.1**. In a powder case, these could be seen as corresponding to a potential fcc and a bcc structure. This is supported by data from collaborator Iurii Ivanov, who had noted a similar trend. The bcc and fcc refer to the trend of local packing in the glass, packing of the icosahedral-like units. They are not representative of the full crystal structure (except in the A2 case, which is mostly fcc-nanocrystalline as shown by X-ray PDF analysis).

Regardless of the baseline level, FEM has shown itself to an excellent alternate method for calibration, similar in accuracy to peak finding on AuPd data. The peaks are easy to fit, with a flat baseline and a gaussian shape. Hence, they can be used to effectively identify the positions of peaks in the data for calibration.

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### 7.1.3 Statistical Issues with the Merlin Medipix3

Several attempts have been made at identifying the source of error that causes the baseline in the corrected variance. Theoretically, the peaks should appear above the baseline, which should have a value of 1 [69]. This is clearly not the case as shown below in **Figure 7.2**, as the baseline is significantly suppressed to even below zero. This is a sign of below-poissonian variance in data. It was initially thought that this might be a counting-related problem, where multiple pixels were being excited by a single electron, leading to an artificial suppression of the poissonian variance in the data. Experiments were performed on gold with detector settings that prevent multiple excitation (using a threshold voltage greater than half the accelerating voltage), and the baseline was found to be further suppressed, as shown in **Figure 7.2**. This problem was persistent throughout detector conditions, but seemed to be accentuated by low signal-to-noise ratio. We think it is likely caused by either the noise related to the dark-current characteristics or readout noise on the Medipix3. Such noise is extremely low on the Medipix3 but is non-zero [133], and with such low dose per pixel in the acquired data, we end up in a regime where such low intensity noise-characteristics contribute a non-zero component to the overall noise, which in turn leads to the suppression of the baseline observed above.



**Figure 7.2** Gold FEM Spectra. (a) at 50 keV detector threshold TH\_0 energy (with a 300 kV beam) (b) at 200k keV detector energy. (b) has both higher peaks and a lower in-between signal. (c) Variation of spectra with beam current, 1.8 pA (blue) and 3.9 pA (red). Again, lower total counts (due to lower current) suppresses the variance in between the peaks further below zero, while increasing the height of the observed peaks. Green inset, more highly oscillating variance observed due to a reduction in integrated detector angle.

## 8 Conclusion and Future Prospects

The scanning electron diffraction (SED) of non-crystalline and semi-crystalline materials has been developed in multiple ways and its utility for understanding the properties of such materials has been demonstrated throughout the work in this thesis. All of this has been achieved at low electron fluence, a vital parameter that extends the range of materials that can be probed by SED to most organic materials. Organic materials are complex, more so than their inorganic counterparts, and their interpretation requires further development and understanding of their packing, defects, and microstructure. Yet, they are ever more important, with organic-inorganic hybrid materials, such as metal-organic frameworks and lead-halide perovskites some of the highlights from the last decade. The lower ordering and different nature of defects challenges crystallography. This work takes key steps in addressing gaps within the current suite of crystallographic characterisation tools. These tools can be applied depending on the crystallinity of the underlying specimen. Three key groups of analysis methods are developed, which range from completely amorphous to crystalline materials with correlated disorder.

## 8.1 Amorphous Characterisation

The development of low-dose scanning electron pair distribution function (SEPDF) enabled the characterisation and structural mapping of metal-organic frameworks, polymers, inorganic glasses, as shown in Chapters 4 and 6. These structures showcase structured elastic scattering, but relating this scattering to real-space structural information is exceedingly difficult, as the scattering does not contain crystalline periodicity. In this work, the SEPDF analysis, of which a few previous works exist, was further developed, using a mixed amorphous-amorphous glass. By using a combination of diffraction and compositional information, the SEPDF method was shown to work even at low electron fluences, in the presence of high poissonian noise. This is key, as the samples that most require it, organic samples, are typically prone to radiation damage. SEPDF analysis enabled further structural interpretation, especially regarding interfaces. Furthermore, it was highly resistant to thickness variation within the sample and could be acquired concurrently with crystalline scattering.

In addition, the work revealed fascinating further avenues for research. Particularly, the difference in X-ray and electron scattering factors could be exploited for combined X-ray-electron PDF analysis to better determine the underlying crystallography of multi-element organic structures. The sensitivity of electron scattering to hydrogen looks especially promising. In addition, SEPDF analysis is rapidly

gaining popularity in the analysis of amorphous samples using TEM, and I suspect will become a standard characterisation method in the years to come.

## 8.2 Polymer Characterisation

The polymer samples analysed in Chapter 6 highlight a different challenge. Polymeric samples are partially amorphous, and while methods like the SEPDF method developed in Chapter 4 can be used, more interesting information is recovered by analysing the subtle changes within the diffraction patterns. Analysis methods based on diffraction variance, kernel misorientation, and orientation mapping in polymers were developed in this work for use in SED in Chapter 6. The variance methods are novel for polymers, and highlight the unique brand of semi-crystallinity present. The difference to the amorphous samples in Chapter 4 highlights a key facet of analysing non-crystalline materials. Namely, it is vital to identify the type of order and disorder present, to subsequently be able to characterise the material most effectively. The three examples in Chapter 6 showcase three very different nanostructures, from near amorphous to domain-structured to nearly multi-crystalline. Such variety of crystallographic order has not been previously shown in polymers using direct imaging. SED is the key, as it enables the design of the experiment post-acquisition. As a result, different types of analysis can be attempted, and the most effective characterisation route identified in post-processing. Overall, this work also shows polymers contain an incredible amount of structural information. Lowdose SED will certainly be used for further characterisation of polymers. However, it is a difficult technique to apply, much more so than typical crystalline analysis, but I see great hope in its progress.

### 8.3 Organic Defect Structures

Organic structures have a wealth of unique defects, such as twisting and disclinations. These structures had not been studied in detail in SED before this work. Chapter 5 shows a method of structure solution in SED through unravelling a twist geometry in a rapidly twisting nanocrystal. By once again utilising computational power and computing the relative orientations of different pixels in SED, it was possible to characterise the underlying crystallography of the molecular cocrystal. Novel methods were developed to identify twist rates in the sample. Furthermore, in Chapter 6, several types of disclinations and disclination networks are seen. Both of these are likely just the tip of the iceberg. The defect structures of organic materials demand further characterisation, and SED remains a uniquely poised tool capable of both sufficient spatial resolution and obtaining crystallographic information, all while operating at a sufficiently low electron dose.

### 8.4 Outlook

Electron diffraction is changing. A decade ago, scanning electron diffraction was at its infancy, and most of transmission electron microscopy (TEM) experiment design was done before data acquisition. That has changed with SED (and other novel methods such as electron ptychography), where now most work is done post-acquisition. This change has dramatically increased the computing requirements of doing TEM experiments, with data sets of 10-30 GB now routine (and acquirable in a few minutes). In conjunction with improved direct and hybrid detectors, this has changed the landscape of materials that can be probed with a TEM. Now, it is not just inorganic and stained specimens, but organic and inorganic-organic hybrid specimens that demand characterisation, and with that, new problems have arisen. Materials probed are not only less crystalline, but they exhibit different types of defects than those typical of their inorganic brethren.

This work takes steps in addressing gaps in characterisation methods for these organic samples. In the next decade, I suspect more and more work will be done in characterising non-crystalline materials. However, some hurdles remain. In an ideal world, all of the data within this thesis would have been taken with an energy filter, to remove inorganic contributions that obscure the crystallographic signal. That was not possible since a microscope with sufficient detector performance at the end of an energy filter is not currently available. Such setups should ideally incorporate precession as well, but do not do so yet. However, such problems should be resolved within the next decade, enabling even higher quality data acquisition.

Overall, the shift within the TEM community is towards computational microscopy. The ability to arbitrarily relate any one pixel on the detector to another is incredibly powerful, and to do so with spatial resolution, as in SED, even more so. However, a new challenge has appeared in the last five years. We now take so much data, we don't have time to analyse it all! I theorise that the greatest developments in the near future of electron microscopy will be in data pre-processing. Identifying methods, perhaps supervised machine learning methods, that are able to let the scientists focus on the interesting structural elements within huge data sets is vital, or the details will be lost to the bigger picture. For this type of analysis, SED is promising. By acquiring large amounts of data at each probe position, a far larger number of pre-processing methods can be applied, in turn enabling further identification of areas of interest within samples. It is an exciting time in electron microscopy, but especially so if you like computation.

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## 9 References

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## Appendix A: List of Publications

Joonatan E. M. Laulainen, Duncan N. Johnstone, Ivan Bogachev, Sean M. Collins, Louis Longley, Thoms D. Bennett, and Paul A. Midgley "Mapping Non-Crystalline Nanostructure in Beam Sensitive Systems With Low-dose Scanning Electron Pair Distribution Function Analysis." Microscopy *and Microanalysis* 25.S2, p. 1636-1637, 2019.

Duncan N Johnstone, Phillip Crout, Joonatan Laulainen, Simon Høgås, Ben Martineau, Tina Bergh, Stef Smeets, Sean Collins, Jędrzej Morzy, Håkon Ånes, Eric Prestat, Tiarnan Doherty, Tomas Ostasevicius, Mohsen Danaie, Rob Tovey. "Pyxem/Pyxem: Pyxem 0.10. 0." (2019).

In addition, the following publications are currently in peer review:

Joonatan E. M. Laulainen, Duncan N. Johnstone, Ivan Bogachev, Louis Longley, Courtney Calahoo, Lothar Wondraczek, David A. Keen, Thomas D. Bennett, Sean M. Collins, and Paul A. Midgley, "Mapping Short-Range Order at the Nanoscale in Metal–Organic Framework and Inorganic Glass Composites" (2022). *Accepted to Nanoscale*.

Scott T. Keene, Joonatan E. M. Laulainen, Raj Pandya, Maximilian Moser, Christoph Schnedermann, Paul A. Midgley, Iain McCulloch, Akshay Rao, and George G. Malliaras, "Hole-limited electrochemical doping in conjugated polymers" (2022)

## Appendix B: Data Access

All data used in the production of this thesis, as well as example workflows and Jupyter notebooks are on Apollo at the following link <u>https://doi.org/10.17863/CAM.87958</u>. Additional information can be requested from the author.