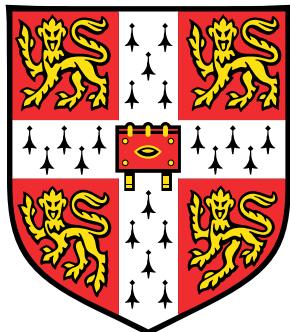


Exciton interactions in lead iodide perovskites



Wendy Niu

Department of Physics
University of Cambridge

This dissertation is submitted for the degree of

Doctor of Philosophy

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2014

I would like to dedicate this thesis to my loving parents ...

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 the result of my own work and includes nothing which is the outcome of work done in collaboration, except where specifically indicated in the text. This dissertation contains less than 65,000 words including appendices, bibliography, footnotes, tables and equations and has less than 150 figures.

Wendy Niu
2014

Acknowledgements

And I would like to acknowledge ...

Abstract

This is where you write your abstract ...

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Nomenclature

Acronyms / Abbreviations

AFM Atomic force microscope

BF Bright field

C12PI $(C_{12}H_{25}NH_3)_2PbI_4$

CHPI $(C_6H_9C_2H_4NH_3)_2PbI_4$

DF Dark field

LSP Localised surface plasmon

MQW Multiple quantum well

SEM Scanning electron microscope

SPP Surface plasmon polariton

Chapter 1

Introduction

Semiconducting behaviour was first reported in the 19th century. Michael Faraday noted in 1839 that the conductivity of silver selenide increased with temperature, opposite to what is expected in a metal [1]. In 1839 Alexandre-Edmond Becquerel reported photovoltaic behaviour in silver chloride coated platinum electrodes in an aqueous nitric acid electrolyte, where an increase in the voltage was observed when one electrode was illuminated with sunlight [2]. A photovoltaic cell was produced by Charles Fritts in 1883 using 30 μm thick selenium film with thin gold leaf contacts, with efficiency <1% [3]. Photoconductivity was first reported in selenium bars by Willoughby Smith in 1873, showing in the change in resistance as a result of exposure to sunlight [4], as well as William Grylls Adams and Richard Evans Day in 1876, who reported the production of electricity in selenium due to an exposure to sunlight [5]. Electroluminescent behaviour was first reported in SiC by Henry Joseph Round in 1907, who reported emission when a current was passed through crystals [6]. Simple principles such as these underpin the electronic devices we use today, yet it was not until Alan Wilson's work on band theory in 1931 that such behaviour could be understood and explained [7].

Throughout the 20th century research was focused on using semiconductors for devices, from the developments of rectifiers and diodes in the early part of the century to the first transistor built at Bell Laboratories in 1946 by William Shockley [Fig. 1.1(a)], John Bardeen and Walter Brattain using Ge. Much work has been undertaken to refine and improve the design of these devices to the more sophisticated designs used in our devices today, however the efficiency of such devices hinges on the quality of the semiconducting material used, and the ability to control impurities and dopants in the crystal. From the earliest devices using materials such as lead selenide, we now moved on to devices made using Si and Ge, which are still used today.

Si is the most widely used material in the semiconductor industry due to its abundance in the Earth's crust, low unit cost and simple and well-developed processing techniques. Its high band gap gives it thermal stability, allowing it to be used at high operating temperatures. Si is also highly mechanically stable with high electron mobility, and the native oxide insulating layer that spontaneously grows can be useful in electronics. Although Ge has higher conductivity than Si, its lower band gap gives devices more temperature sensitivity so it is less often used in devices. Alloys of Si and Ge may be used to combine the properties of both materials. One of the main disadvantages of Si is its indirect band gap, meaning it is a poor emitter for luminescence devices. However the GaAs has a direct band gap and can be used in luminescent devices, and its high electron mobility and band gap means it can be used in high speed devices. Other alloys, for example between Group III-V or II-VI atoms, can be used to tune the band gap and other optoelectronic properties. Alloys can also be combined to make lower-dimension structures, producing other desirable properties.

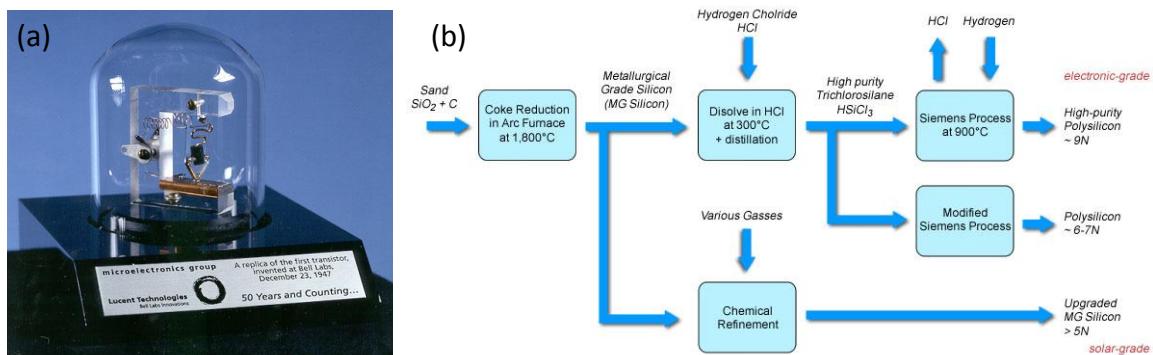


Fig. 1.1 (a) Replica of first transistor built in 1946. Reproduced from Ref. [8]. (b) Schematic of process required to produce device-grade silicon from starting materials. Reproduced from Ref. [9].

Despite the high stability and carrier mobility in inorganic semiconductors mentioned previously, one major disadvantage is with the processing of such materials. Although Si processing is well-developed, creation of device-grade material still requires many purification steps [Fig. 1.1(b)]. Alloys are often produced using vapour or electron-beam deposition, and as such the environmental parameters must be strictly controlled. Layer-by-layer growth can be used for the best quality material, however such processes are costly and time consuming. Current advances in technology require flexible, lightweight and more easily processable semiconductors.

Although conduction was noted in a mix of aniline and sulphuric acid by Henry Letherby in 1862, research on organic semiconductors began in earnest in the latter half of the 20th

century. Polycyclic aromatic compounds were found to form semiconducting charge transfer complexes with halogens in 1954 [10], and since then much work has been done on developing new molecules and polymers with desired optoelectronic properties, driven by the relative ease with which such molecules can be synthesised. Doping was developed in the 1970s to produce metallic and even superconducting complexes, for example highly conductive oxidised and iodine-doped polyacetylene in 1979 [11]. Organic semiconductors consist of conjugated molecules, whose overlapping π orbitals allow charge transport within the molecule. Given the low production cost, work has been undertaken to produce devices made of organic semiconductors, notably organic photovoltaics, thin film transistors, and light emitting diodes (OLED), probably the most mature organic electronic device as OLEDs have been used mobile phone displays and TVs, with better efficiency and brightness. However problems exist with the manufacturing of such devices, as mass production is not currently optimised for the organic electronic market. More fundamentally, organic semiconductors are less thermally, optically and electrically stable than their inorganic counterparts, leading to lower lifespan of devices. Charge mobility is also lower as hopping between adjacent molecules is required, and lower crystallinity of materials leads to scattering at grain boundaries.

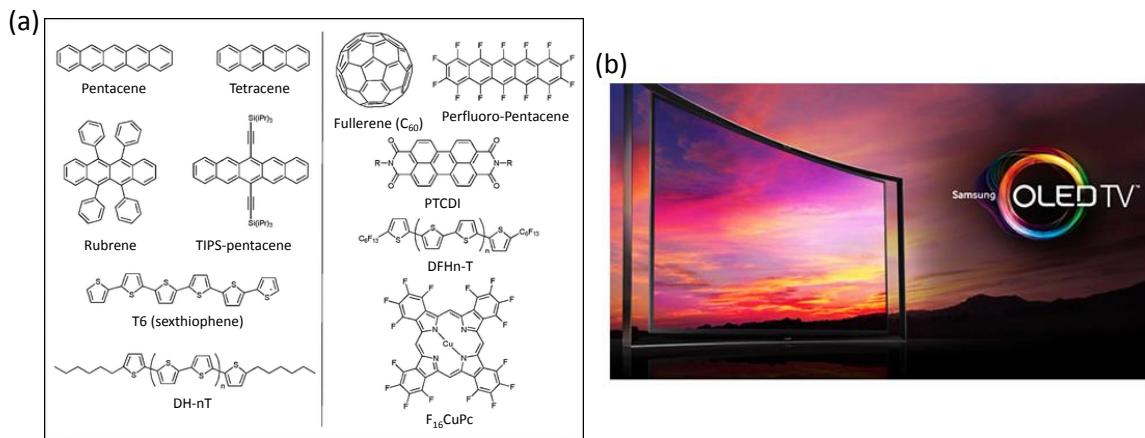


Fig. 1.2 (a) Common organic semiconductors, with p-type materials on the left and n-type on the right. Modified from Ref. [12]. (b) Samsung curved smart OLED TV. Reproduced from Ref. [13].

A new class of hybrid materials has emerged in the last 20 years. Metal halide based organic-inorganic perovskite semiconductors combine the stability of the inorganic semiconductors with the processability of organic semiconductors. A variety of inorganic frameworks can be formed, all of which are self-assembling. Charge carriers are excited in the

inorganic network, and thus share similar optical and electrical properties.

This thesis explores the optical properties of 2D lead iodide-based perovskites, particularly the interactions of perovskite excitons and collective electron oscillations in noble metal nanostructures called surface plasmons. We will first introduce the theory of excitons and review the research on lead iodide perovskites, before exploring the optical properties of surface plasmons. In Chapter 4 we will explore the fabrication of perovskite thin films via spin coating, then study the optical properties of ultrathin perovskite samples produced via exfoliation in Chapter 5. In Chapter 6 we will explore exciton-localised surface plasmon interactions in perovskite-coated metal island structures. Finally in Chapter 7 we look at the coupling between excitons and surface plasmon polaritons in coated plasmonic gratings.

Chapter 2

Excitons in lead iodide perovskites

2.1 Properties of excitons

In an ideal crystal, the orbitals of a periodic array of atoms $\sim\text{\AA}$ s apart overlap, and instead of discrete atomic energy levels, we instead find energy bands due to mixing between atomic levels. We can think of the resulting electron bands using two models: either by starting with a free conduction electron gas then adding the perturbation from the periodic potential of the ionic atomic cores (Bloch's theorem), or by adding weak interactions between the overlapping neutral atoms so electrons can hop between atomic sites (tight binding approximation, similar to linear combination of atomic orbital approach). In general this is complicated many-body problem, with both electron-electron and electron-core interactions to take into account in the Hamiltonian of the system, as well as the motions of carriers. Such a problem cannot be solved analytically, and requires computing power in order to calculate the band structure of even simple solids. The interactions mean the electron dispersion deviates from the free electron value of $E = \frac{\hbar^2 k^2}{2m_e}$ where E is the energy of the electron, k the wavevector of the electron wavefunction, related to its momentum p by $p = \hbar k$, and m_e is the rest mass of a free electron.

However we can still use some basic trends to describe the behaviour of charge carriers in solids. The most important energy bands are the valence band (VB), the highest band occupied by electrons (*c.f.* highest occupied molecular orbital in molecules), and the conduction band (CB), the lower non-occupied band (*c.f.* lowest unoccupied molecular orbital). The gap between the highest point of VB and lowest point of CB is known as the band gap E_g ; when this gap occurs at the same k point the material is said to have a direct band gap, and if not the band gap is indirect. The behaviour of free carriers depends on the relative energies of these bands with respect to the Fermi energy (E_F), defined as the energy of the

energy of the highest occupied state at 0 K [Fig. 2.1(a)]. Conduction depends on the availability of free electrons in the CB, where there are free states for electrons to move into. If E_F is within the VB (i. e. VB and CB overlap), then the solid is a metal. If E_F is in the gap between the VB and CB, then the material is a semiconductor if E_g is small ($\lesssim 1$ eV) so that electrons can be thermally excited to the CB, or an insulator if E_g is larger. Also as a result of electrons, the mass of an electron m_e is no longer given by the free electron rest mass value. Instead it depends on the curvature of the energy band, and the ‘effective electron mass’ of the electron m_e^* when responding to forces depends on the dispersion of the band it occupies as

$$m_e^* = \frac{\hbar^2}{\frac{d^2E}{dk^2}}. \quad (2.1)$$

For the rest of the Chapter we will drop the effective mass label * for brevity, however it should be noted that in general the free electron mass is only an approximation to the actual crystal value.

When electrons are excited from the VB they leave behind ‘holes’, quasiparticles used to describe an absence of electrons with a charge of $+e$, $k_h = -k_e$ and $m_h \approx m_e$ although this depends on the dispersion of the energy bands in question. The attraction between the electron and hole binds them together to form a hydrogen-like neutral particle in the crystal. We can therefore use results from the hydrogen atom to find the binding energy E_B and Bohr radius a_B :

$$E_B = \frac{\mu e^4}{32\pi^2 \epsilon^2 \epsilon_0^2 \hbar^2 n^2} = \frac{R_H}{n^2} \frac{\mu}{\epsilon^2 m_e} \quad (2.2a)$$

$$a_B = \frac{4\pi\epsilon\epsilon_0\hbar^2}{\mu e^4} = a_0 \frac{\epsilon m_e}{\mu}, \quad (2.2b)$$

where $\mu = (\frac{1}{m_e} + \frac{1}{m_h})^{-1}$ is the effective mass of the exciton, ϵ is the dielectric constant of the material, and n is the energy level of the exciton, with $n = 1, 2, 3, \dots$. The Rhydberg constant $R_H = 13.6$ eV, and most probably distance between a proton and electron in the ground state $a_0 = 0.5$ Å are defined with respect to the hydrogen atom. Due to E_B , exciton energies have energies below free electrons in the CB, and form a series of hydrogenic level below the lower edge of the CB [Fig. 2.1(b)]. The energy of an exciton E_{ex} is given by

$$E_{ex} = E_g - E_B + \frac{\hbar^2}{2M}(k_x^2 + k_y^2 + k_z^2), \quad (2.3)$$

where $M = m_e + m_h$ is the total mass of the exciton, and the k_i terms describe its motion in

3D. Eq. 2.3 gives the energy of free electrons that are able to move throughout the crystal, however excitons can be bound to impurities, further lowering their energy.

Mott-Wannier excitons typically exist in inorganic materials, where high ϵ gives rise to $E_B \sim 10$ eV and $a_B \sim 100$ Å. Therefore such excitons extend over many unit cells, and due to low E_B low temperature measurements are required to observe Mott-Wannier excitons optically, although we can often observe multiple n states [Fig. 2.1(c)]. Conversely Frenkel excitons general exist in organic materials with low ϵ , so $E_B \sim 1$ eV and $a_B \sim 10$ Å. Frenkel excitons are limited to a few unit cells, or in the case of molecular organic semiconductors, the same molecule. If the charge carriers can hop between molecular sites then such excitons are known as charge-transfer excitons.

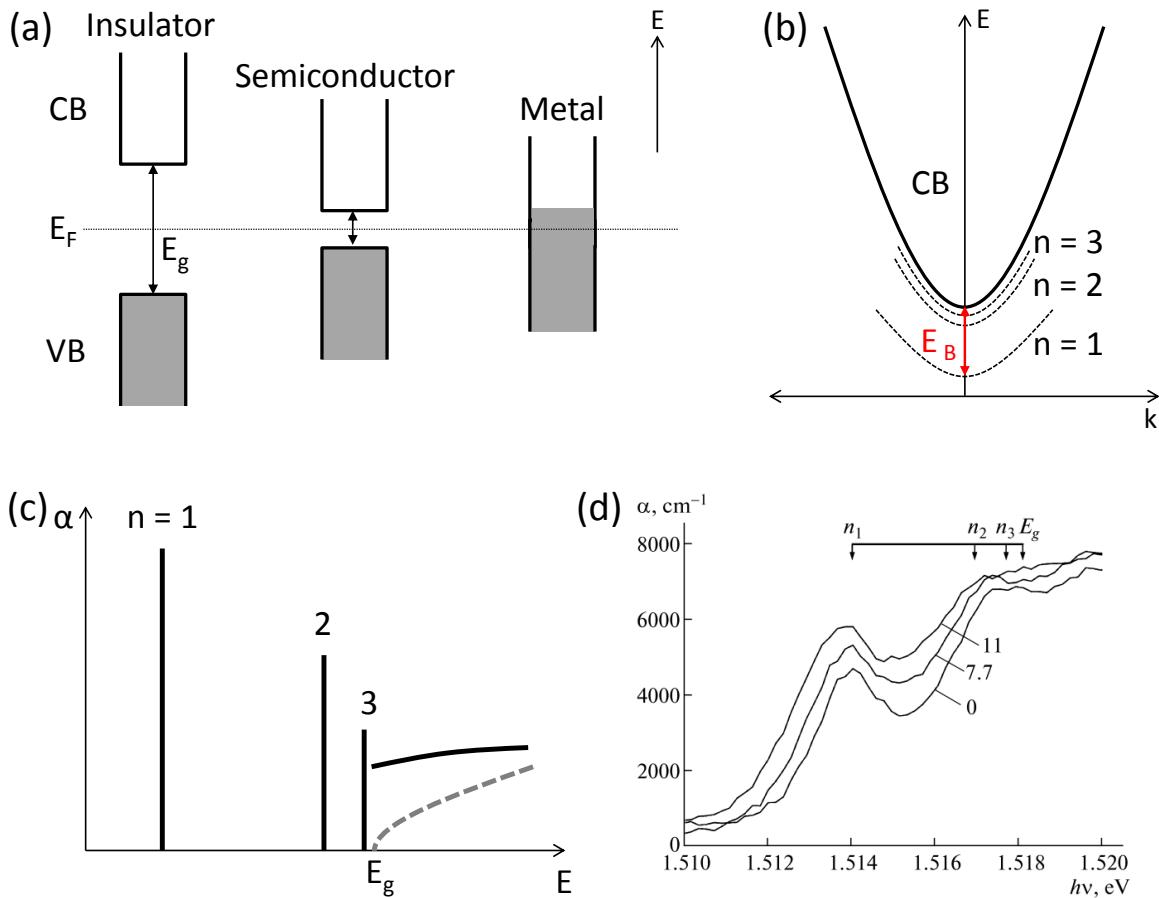


Fig. 2.1 (a) Schematic of band structures of insulators, semiconductors and metals. (b) Schematic of exciton bands below the conduction band. (c) Theoretical absorption spectrum of a 3D semiconductor according to Eq. 2.4 (not to scale). The dotted line shows the expected band edge absorption without the Sommerfeld factor (see text). (d) Experimental absorption spectrum of GaAs crystal at 1.7 K, with the labelled laser power (W/cm²). From Ref. [14].

As mentioned previously, excitons can be created by the absorption of a photon. In bulk semiconductors, the absorption coefficient α depends on the initial ground state $|i\rangle$ of a filled VB and empty CB and final state $|f\rangle$ of one promoted electron as

$$\alpha = |\langle f | i \rangle|^2 \rho(E), \quad (2.4)$$

where the matrix element $\langle f | i \rangle$ gives rise to selection rules of possible electronic transitions, and the joint density of states $\rho(E)$ is the number of states per unit energy range between the CB and VB at the photon energy E . In 3D $\rho(E)$ has a $\sqrt{E - E_g}$ dependence, however in experiment the Coulomb interaction leads to an enhancement in the absorption by the Sommerfeld factor, and α is not discontinuous at zero at E_g as expected from Eq. 2.4 [15]. This gives rise to the theoretical absorption spectrum shown in Fig. 2.1(c), with discrete exciton absorption lines decreasing in intensity as n^{-3} [15], and continuous absorption at higher energy due to interband transitions.

One difference between the theoretical and measured absorption spectrum is due to the overlap of the exciton lines and band edge absorption as a result of the finite width of exciton peaks (Γ_{ex}). The finite exciton width is due to a number of factors: firstly homogeneous variable that affect each exciton equally, such as the radiative lifetime of the exciton ($\frac{1}{\Gamma_{ex}(0)}$), and collisions with phonons. Secondly inhomogeneous factors such as impurities (Γ_{imp}), whose effects will be more localised. Overall this leads to a measured linewidth Γ_{ex} that varies with temperature T as

$$\Gamma_{ex}(T) = \Gamma_{ex}(0) + \Gamma_{imp}(T) + A_{ac}T + \frac{B_{op}}{\exp^{\frac{\hbar\omega_p}{k_B T}} - 1}, \quad (2.5)$$

where the contribution to phonon collision can be separated into coupling with acoustic phonons (A_{ac}), and interactions with polarised optical phonons of energy $\hbar\omega_p$ [16].

Once excitons are excited via absorption, they can also emit that energy in the form of a photon. Thus in emission spectra the strength of exciton peaks depends also on the oscillator strength f , given by the matrix element between the exciton and photon wavefunctions. The strong coupling between excitons and photons also gives rise to exciton-polaritons. The non-propagating exciton solution at $k = 0$ gives rise to the longitudinal exciton-polariton frequency ω_L , while the travelling plane-wave solution gives rise to the transverse exciton-polariton with frequency ω_T as $k \rightarrow \infty$. The absorption of a photon to form the charged constituents of an exciton causes deformation of the crystal, and as a result of the structural rearrangement the annihilation of the exciton causes emission of a photon at a lower energy.

This is known as the Stokes shift, which has a larger value for Frenkel excitons in organic semiconductors due to the more easily deformed molecules.

2.2 Excitons in 2D systems

Exciton motion can be confined to 2D in quantum well (QW) systems, where a material with smaller E_g (well) is sandwiched between layers of material with higher E_g (barriers). If the well and barrier layers are periodically arranged then a multiple quantum well (MQW) or superlattice structures is formed. Four types of band alignment can be achieved [Fig. 2.2(a)]. In type I structures potential steps appear in both the VB and CB between the well and barrier materials, thus both electrons and holes are confined to the well region. In type II QWs the band edges of the barrier are both shifted in one direction with respect to the well material, creating a staggered band alignment, where the electrons are confined in the well, but holes in the barrier region. In the most extreme case top of the barrier VB is above the well CB, creating a type II broken-gap arrangement. Type III QWs occur when using a semimetal (which has a small overlap between the VB and CB) as the well and another semiconductor as the barrier. For the rest of the Chapter we will consider the properties of type I QWs.

Type I QWs can be formed from many III-V materials, for example GaAs/AlGaAs, differing the group III compound in GaAs/AlAs, or differing the group V compound in GaAs/GaP. In general these inorganic QWs are grown atomic layer-by-layer, either using molecular beam epitaxy (MBE) or metalorganic chemical vapour deposition (MOCVD). In both these cases the relevant atoms are deposited onto the substrate, either in the gas phase or via a molecular beam, and the stoichiometric mix of relevant atoms must be carefully controlled. In designing a QW system, one must also consider the lattice constants of the materials in question. If there is a large lattice mismatch there will be strain in the layers, and the growth will not be epitaxial, i.e. there will not be a well-defined crystal structure throughout the layer. The AlAs/GaAs system has good lattice matching, alternatively a ternary alloy can be used, where the atomic proportions can be adjusted to reach the lattice constant needed. It is also possible to use materials that are highly strained but can adapt to the local lattice constant up to a critical thickness.

The layered structure of a QW confines carrier motion in the direction of layer growth, thus creating a 2D system. If the width of the well layer L is smaller than or on the order of the electron de Broglie wavelength (~ 10 nm), then we can observe wavelike effects of the carriers. The confinement then reduces to the problem of a particle in a finite well, and

produces discrete energy states in which the electron/hole can reside [Fig. 2.2(b)]. Using the results of a particle in an infinite well for simplicity, the allowed states have energy E_u

$$E_u = \frac{\hbar^2}{2m} \left(\frac{u\pi}{L} \right)^2, \quad (2.6)$$

where m is the mass of the carrier, and the index u labels the energy level. Instead of having one VB/CB, we instead have a series of minibands with difference energies [Fig. 2.2(b)]. In order to find the absorption spectrum of QW systems Eq. 2.4 still holds, however the due to the orthogonality of eigenstates of the well potential, transition from state $|u_1\rangle$ in the VB and $|u_2\rangle$ in the CB are only allowed if $u_2 - u_1 = 0$. The joint density of states $\rho(E)$ also changes in 2D, and becomes step-like, again with the Sommerfeld factor enhancing absorption [Fig. 2.2(c)].

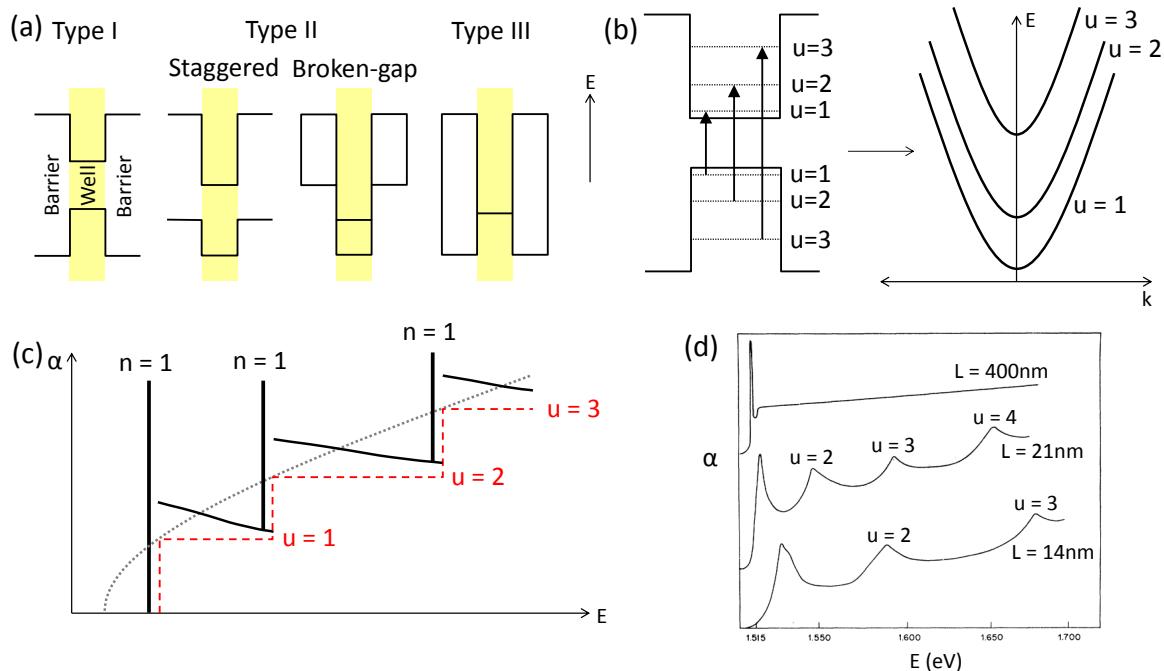


Fig. 2.2 (a) Schematic of band structures of quantum well classifications. (b) Schematic of allowed discrete energy states inside the quantum well, and the resultant miniband structure. (c) Theoretical absorption spectrum of a 2D quantum well according to Eq. 2.4 (black lines, not to scale). The dotted grey line shows the band edge absorption in 3D, the dashed red line the absorption in 2D, both without the Sommerfeld factor. (d) Experimental absorption spectrum of GaAs/Al_{0.2}Ga_{0.2}As quantum wells with the labelled well width L at 2 K. Only the $n = 1$ excitons are observed, but the minibands u are labelled. Adapted from Ref. [17].

Exciton bands exist below each of the minibands, and in order to find E_B we solve the

problem of a hydrogen atom in 2D, so that

$$E_B = \frac{\mu e^4}{32\pi^2 \epsilon^2 \epsilon_0^2 \hbar^2 (n + \frac{1}{2})^2} = \frac{R_H}{(n + \frac{1}{2})^2} \frac{\mu}{\epsilon^2 m_e}. \quad (2.7)$$

Therefore the energy of the exciton bands E_{ex} below the minibands can be described by

$$E_{ex} = \Delta E_u - E_B + \frac{\hbar^2}{2M} (k_x^2 + k_y^2), \quad (2.8)$$

where ΔE_u describes the transition between energy levels u in the CB and VB. The expected absorption spectrum of a QW structure is shown in Fig. 2.2(c), while an experimentally measured spectrum for GaAs/Al_{0.2}Ga_{0.8}As QWs are shown in Fig. 2.2(d). Note how the change in L affects the allowed ΔE_u transitions in the measured energy range, with $L = 400$ nm looking essentially like a 3D system.

We can see from Eq. 2.7 that the reduction in dimensionality leads to a factor of 4 increase in E_B for the $n = 1$ state, meaning exciton effects should be observable at higher temperatures in QWs. Similar a_B is reduced by a factor of 2 in QWs compared to the 3D value as a result of the confinement. For this reason it is often said that excitonic effects are stronger in QWs due to increased overlap between the electron and hole wavefunctions.

2.3 Properties of PbI perovskites

Organic-inorganic metal halide semiconductors can combine the distinct properties of its organic (e.g. ease of processing, structural diversity, plastic mechanical properties) and inorganic (e.g. band structure variability, electrical mobility, thermal and mechanical stability) constituents into one composite. Two dimensional organic-inorganic perovskites self-assemble to form a multiple quantum well (MQW) structure, with alternating sheets of high band gap (~ 6 eV) organic cations and lower band gap (~ 3 eV) semiconducting inorganic layers [18, 19]. The wide range of organic molecules that can be incorporated into the structure enables tunability of the system [20]. These hybrids produce strong optical responses, and exhibit many interesting optical properties, for example room temperature excitonic effects due to quantum and dielectric confinement [21], third order non-linear optical effects [22], and biexciton effects [?]. Due to the self-assembled nature of the structure, such perovskites are much easier to produce than conventional inorganic QWs, and can be fabricated using a variety of techniques. Thin films of these materials can also be processed using simple techniques such as spin coating [23], making them ideal for use in optoelectronic

devices.

2.3.1 Structure and bonding

Metal halide RMZ_3 organic-inorganic semiconductors are based on the ABX_3 perovskite crystal structure [Fig. 2.3(a)], consisting of a corner-sharing octahedra network of halogen atoms Z (most commonly I, Br or Cl), with a metal atom M in the centre of each octahedron (+2 valence metals such as Pb, Sn, Cd, Zn, Cu, Co), and organic molecules in the interstices between octahedra [Fig. 2.3(b)]. The organic molecules must fit into the interstitial space, and as a result very short molecules are used, the most common of which is CH_3NH_3 , which hydrogen bonds to the nearby halogen atoms. The band gaps of such semiconductors can be engineered by changing the metal and halogen composition, and recently lead halide-based semiconductors have been used as a light absorbing layer in solar cells, producing efficiencies of up to 16%, and there is currently a drive to find lead-free alternatives for wider use [24–27].

A change in the stoichiometric mix of organic and inorganic molecules results in the formation of lower dimension structures, for example 2D layered systems, or 1D inorganic wires [28–31]. From here we will focus on $<100>$ oriented 2D lead iodide (PbI) perovskites [32].

The structure of 2D hybrid PbI perovskites with formula $(\text{RNH}_3)_2\text{PbI}_4$ are shown in Fig. 2.3(c), and consist of alternating layers of corner-sharing PbI_6 octahedra and interdigitating RNH_3 molecules (where R is an organic moiety). The organic molecules have larger E_g ($\sim 6 \text{ eV}$) than the inorganic layers ($\sim 3 \text{ eV}$), thus forming a type I MQW structure. The bonding in inorganic layers is primarily ionic as Pb-I distances are more comparable to the sum of ionic radii [33]. Inorganic sheets are sandwiched between layers of organic molecules via hydrogen bonding between NH_3 groups and I atoms (see Fig. ??), where there are two environments for iodine atoms: bridging (shared by octahedra) or terminal (not shared by octahedra). Thus the three hydrogen atoms on NH_3 groups can adopt two hydrogen bond formations, with either two bridging iodines and one terminal iodine, or two terminal iodines and one bridging iodine. In the latter formation, the bonded iodines can either form an equilateral or right angled triangle [34]. Organic molecules are usually held together by van der Waals or aromatic-aromatic interactions that can (de)stabilise the perovskite structure [35] and as the weak force it is possible to cleave the structure and produce mono- or few-layer regions. In this way the structure of such PbI perovskites is comparable to other layered materials such as graphene or transition-metal dichalcogenides.

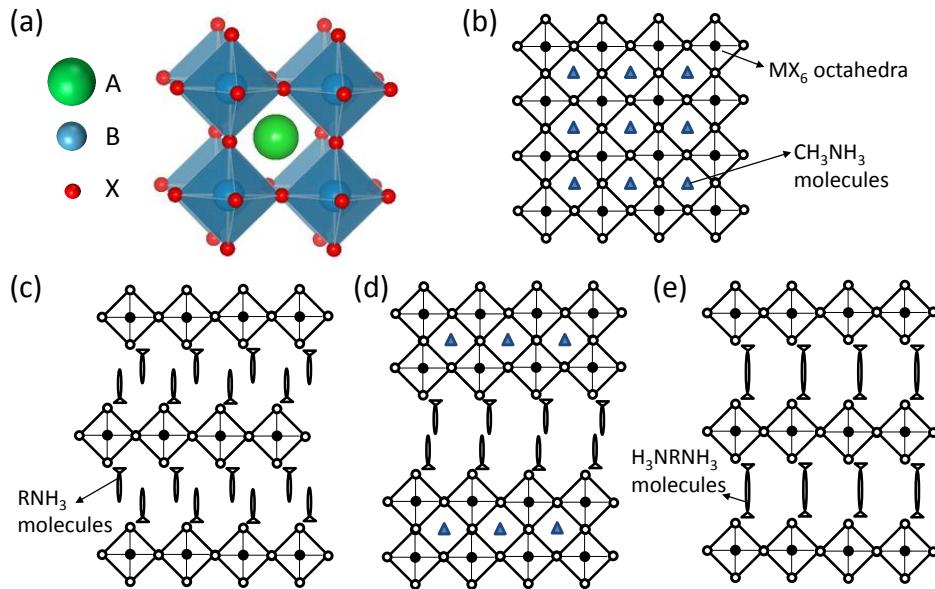


Fig. 2.3 (a) Schematic of ABX_3 perovskite. Adapted from Ref. [36]. Structure of (b) 3D $(\text{CH}_3\text{NH}_3)\text{PbI}_3$, (c) 2D $(\text{RNH}_3)_2\text{PbI}_4$, (d) 2D multilayered $(\text{RNH}_3)_2(\text{CH}_3\text{NH}_3)\text{Pb}_2\text{I}_7$, and (e) 2D diammonium $(\text{H}_3\text{NRNH}_3)\text{PbI}_4$ viewed along the crystallographic \vec{b} axis.

The width of the inorganic layer, defined as the width of the PbI octahedra, is typically around 6.5\AA [37], whereas the organic layer width is obviously molecule dependent. There are two main orientations for consecutive inorganic layers: eclipsed, giving a monoclinic structure, or staggered, giving an orthorhombic structure [38]. The orientation chosen mostly depends on the organic molecule used, and how it fits into the overall structure. Phase transitions from the orthorhombic phase to the monoclinic phase will lead to a halving of the c lattice parameter due to the increased symmetry [34].

The basic 2D layered structure can be varied in a series of ways. For instance the inorganic layers can be extended to contain multiple layers of $[\text{PbI}_6]$ octahedra [Fig. 2.3(d)], with general formula $(\text{RNH}_3)_2(\text{CH}_3\text{NH}_3)_{n-1}\text{Pb}_n\text{I}_{3n+1}$ [39], thus varying the width of the quantum wells. Another is with the use of diammonium organic molecules, which can hydrogen bond to two consecutive inorganic layers, therefore eliminating the need for van der Waals interactions between the interweaving molecules. The general formula for such systems is $\text{H}_3\text{NRNH}_3\text{PbI}_4$, and the structure is shown in Fig. 2.3(e).

There is little limit on the organic molecule R that can be used in the structure. The cross section of the organic molecule should be small enough so that it can fit into the interlayer space between four adjacent octahedra, and must have a cross-sectional area $\lesssim 40\text{\AA}^2$ [35]. However their lengths are not constrained so long as the intermolecular forces are strong enough to hold the structure together. Systems with aromatic molecules tend to be

better organised with more crystallinity, since such molecules allow for self-assembly using stronger aromatic-aromatic interactions [40]. Large organic R groups will also hinder self assembly and reduce crystallinity [40]. In general very simple organic molecules are used, for example those based on simple alkane chains ($(C_nH_{2n+1}NH_3)_2PbI_4$, C_nPI hereafter), ring structures ($(C_6H_9NH_3)_2PbI_4$, CHPI), or aromatic molecules ($(C_6H_5C_2H_4NH_3)_2PbI_4$, PAPI). However more complex organic molecules can be used, for example optically active ligands [38, 41] and fullerene derivatives [42, 43]. The formation of perovskite structure was particularly problematic due to the large size of the fullerene molecules, however band gap of the ligands gave rise to energy transfer between organic and inorganic layers.

Phase transitions in C_nPI

Four main phases of C_nPI perovskites have been identified: at low temperatures, $\lesssim -30^\circ C$ (phase I), the crystal exhibits twinning so atomic positions cannot be determined. Above this, between ~ -30 and $15^\circ C$ (phase II), organic chains are ordered. At room temperature, between ~ 15 and $65^\circ C$ (phase III), organic chains have become disordered, and at high temperatures, above $65^\circ C$ (phase IV), organic chains appear to be "melting" [21, 23, 37]. For C_{12-} , C_{14-} , C_{16-} , and $C_{18}PI$ phase III is orthorhombic with staggered inorganic sheets, and phase IV is monoclinic with eclipsed inorganic sheets [34]. The changes in structure are shown in Fig. ???. The transition between phases II and III is known as "pre-melting", and causes dynamic rotational disorder of $[NH_3]$ groups in C_{12-} , C_{16-} , and $C_{18}PI$, actually leading to a small decrease in the c lattice parameter [44]. Above the transition, there is a sharp increase in conformational disorder as the alkylammonium chains become tilted at different angles, leading to a large increase in c , and thus an increase in volume since no lateral motion occurs [44]. Simulations show that after the melting transition alkyl chains are no longer all-trans, and the introduction of gauche defects leads to a shortening of the chain and an increase in its effective cross-sectional area. Conflicting demands of close packing that optimise dispersive interactions and the larger area required by conformationally disordered chains can no longer be met by a uniformly tilting arrangement, and the non-uniform tilt allows for increased space for individual chains [45].

Changes in conformation during phase transitions also cause a spatial shift in the coupling between NH_3 groups and inorganic octahedra [46]. In order to accommodate the changes in alkyl chains, PbI_6 octahedra can undergo two types of structural change: they can tilt perpendicular or parallel to the inorganic sheets. During perpendicular tilting octahedra are tilted with respect to each other, whereas parallel tilting leads to an overall corrugation of inorganic layers [34].

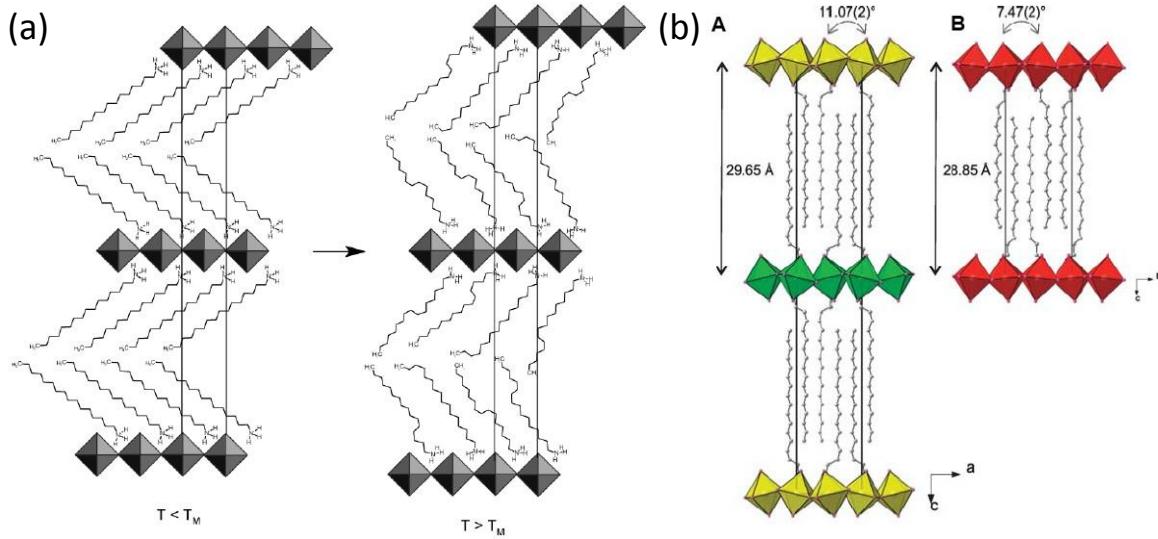


Fig. 2.4 (a) Structure of C_nPI ($n=12, 14, 16, 18$) before (A) and after (B) the pre-melt transition. A is orthorhombic, and B is monoclinic so the lattice parameter c is halved. (b) Schematic of the melting transition in $C_{10}PI$. Below the melting temperature T_M , the chains adopt an all-trans conformation (no gauche defects [45]) with a well defined tilt angle of around 55° [47, 48]. Above T_M , alkylammonium chains are conformationally disordered with different tilt angles. (a) Reproduced from Ref. [34], (b) from [44].

The switching between orthorhombic and monoclinic phases does not only occur because of temperature changes. Pradeesh *et al.* showed that the phases present in spin coated films of $C_{12}PI$ depended on the thickness of the sample, as well as aging effects. They postulated that this was due to strain effects, where high strain in thicker samples favours the flatter orthorhombic phase. Aging effects, where a sample in the monoclinic phase gradually shifted to the orthorhombic phase, were stopped by annealing the samples or capping them with PMMA. Hysteresis was seen in the thin films during heating-cooling cycles, and the high temperature phase found to be metastable below the transition temperature as sample strain impeded the transition [46].

2.3.2 Fabrication and processing techniques

Silica-gel method

An aqueous solution of $Pb(CH_3COO)_2$, $C_nH_{2n+1}NH_2$, CH_3COOH , and Na_2SiO_3 is prepared in a test tube, and becomes a gel after approximately one week. At this point an aqueous solution of KI is poured into the gel and the I^- ions diffuse slowly into the gel to form C_nPI single crystals. About a month after the introduction of KI, platelet-like crystals

form, approximated $2 \times 2 \times 0.1$ mm in size. However crystals for perovskites where $n=4$ or 6 cannot be produced using this method [21].

The silica-gel method allows multiple components to be mixed in solution on a nanometer scale, which can produce very homogeneous materials. The technique can also be used to make thin films by dissolving the raw ingredients in a suitable solvent (e.g. an alcohol) that is fast gelling and drying. The necessary condensation reaction and solvent evaporation will take place after the solution has been applied to a substrate, leaving behind a gel film. The wet film can then be heated to produce a dry film, but in order to avoid cracking the wet film should generally be less than 1 μm thick. The thin gel films are usually amorphous, but surfactants on the substrate can help with assembly, and annealed films are generally crystalline [49].

Solution crystal growth

Although single crystals can be produced using the silica-gel method, it is rather time intensive. More commonly crystals of a similar or larger size can be produced from solution. PbI_2 and the organic ammonium iodide (e.g. RNH_3I for $(\text{RNH}_3)_2\text{PbI}_4$, produced by reacting RNH_2 and HI) are usually mixed in stoichiometric amounts (e.g. [50, 51]) in solution and left for the solvent to evaporate [18]. After about a week single crystals form, although the rate of evaporation can be controlled to change the morphology of crystals [52]. The difficulty lies in finding solvents that will dissolve both the inorganic and organic parts of the perovskites, and examples include acetone [53], DMF [50], and HI [44].

Layered solution method A layered solution method can be used to create $(\text{C}_6\text{H}_5\text{C}_2\text{H}_4\text{NH}_3)_2\text{PbCl}_4$ (PAPC) at room temperature. PbCl_2 in concentrated aqueous HCl is placed in a tube, followed by a less dense layer of methanol, carefully placed using a syringe. Finally a stoichiometric quantity of $\text{C}_6\text{H}_5\text{C}_2\text{H}_4\text{NH}_2$ (relative to PbCl_2) is added at the top of the column as shown in Fig. 2.5, which then disperse into the methanol. As the layers slowly diffuse together, crystals of PAPC form at the interface between the PbCl_2/HCl and methanol layers. X-ray diffraction (XRD) studies confirm that crystals had the structure expected [54].

Spray pyrolysis/drying

Spray pyrolysis produces particles when misted streams of precursor solution are introduced to a furnace. The solvent evaporates, followed by diffusion and separating out of solute, drying, precipitation, and a reaction between components in the tubular furnace reactor. Multicomponent particles can be prepared via spray pyrolysis due to microscale reactions

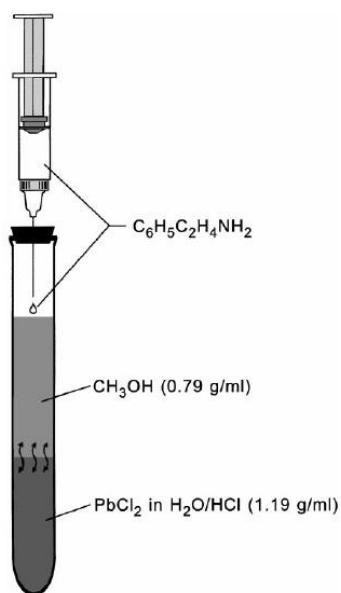


Fig. 2.5 Schematic of experimental set up used for the layered solution method of crystal growth. Crystals were formed at the PbCl_2/HCl and methanol interface. Densities of the pure solvents are shown in brackets. Reproduced from Ref. [54].

happening inside the micro-sized droplets, and new phases with narrow size distributions and non-agglomeration can be obtained rapidly due to the high temperature and inert gas stream inside the furnace reactor [55].

The precursor solution used to prepare a PAPI powder consists of stoichiometric quantities of PbI_2 and $\text{C}_6\text{H}_5\text{C}_2\text{H}_4\text{NH}_3 \cdot \text{HI}$ in THF. XRD data indicate the structure is less organised than PAPI prepared by other methods, e.g. spin coating, and from FE-SEM images the powder particles appear to be $\sim 1\mu\text{m}$ in size. The photoluminescence (PL) spectrum of PAPI powder showed a strong exciton peak, indicating the formation of the required layered structure. However the wavelength is shifted by around 5 nm, which is probably due to distortion and a different linkage of PbI_6 octahedra [55].

A similar technique of spray drying can be used to produce perovskite nanoparticles. The precursor solution is prepared by bubbling a flow of dry HI into a dry ethereal solution of organic amine. Drying droplets (initial mean diameter $\sim 35\ \mu\text{m}$) are carried from the aerosol generator by dry air into an evaporation chamber settled in an oven heated at 250°C . Dried particles are collected as made and stored at room temperature. Nanoparticles created using this method are mostly spherical, but with a large size distribution (50 – 500 nm diameter; average $60 \pm 10\ \text{nm}$). The perovskite actually crystallises at the edge of the particle while the centre is depleted, therefore larger nanoparticles are hollow. Smaller particles dry

too fast for the above process to take effect, and are therefore denser. XRD data indicates the nanoparticles are crystalline, with a small red-shift of a few nm in PL exciton energy. The nanoparticles are much larger than the Bohr radius of excitons, so the shift is likely due to strain effects caused by spray drying, not quantum confinement. The nanoparticles are also fairly photo-stable, with a drop in PL intensity of around 30-50% after illumination of 1hr [?].

Langmuir-Blodgett technique

The Langmuir-Blodgett (LB) technique uses a movable barrier to apply pressure to a monolayer of molecules at a liquid-gas interface. The applied pressure causes molecules to be close enough so that van der Waals forces can achieve close packing, thus forming a thin film (see Fig. 2.6) [49]. Era *et al.* used LB to create thin films of $(C_{22}H_{45}NH_3)_2PbBr_4$. The long chain ammonium bromide ($C_{22}H_{45}NH_3Br$) is spread on an aqueous subphase containing $PbBr_2$ and CH_3NH_3Br from a chloroform and DMF solution. The monolayer is then pressed to a surface pressure of $30mNm^{-1}$, then deposited on hydrophobidised fused quartz substrates. Two layers of the monolayer are required to create the MQW structure needed. The just deposited film shows a strong exciton absorption peak, indicating the formation of the quantum well structure, and remain unchanged for more than 12 hours [56]. Fig. 2.6 shows the absorption spectra after multiple depositions.

Intercalation

Although there has been no direct investigation, it is believed that the driving forces for self assembly in PbI perovskites include a preference for the inorganic octahedra network, the hydrogen bonding between NH_3 groups and inorganic sheets, and the construction of interdigitated organic groups by van der Waals/aromatic-aromatic bonds [52]. Thus perovskite films can be formed by the intercalation of organic molecules into the inorganic framework. PbI_2 films are deposited onto substrates by vacuum deposition or spin coating, and a solution of organic ammonium iodide prepared. The PbI_2 covered substrates are dipped in the iodide solution, then the iodide solution solvent to remove excess organic salts, and finally pumped in the loading dock of the drybox used in order to remove all remaining solvents. The resulting films have the same properties as films created using other methods [57]. The film thickness is controlled by the initial PbI_2 film thickness, and in the case of $(C_6H_9C_2H_4NH_3)_2PbI_4$ (CHPI) only 10s in the iodide solution was needed in order to complete intercalation [58].

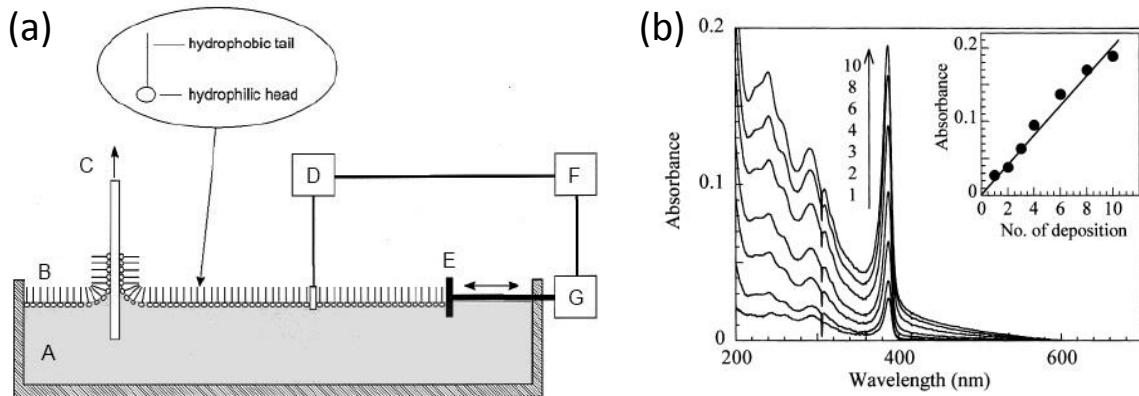


Fig. 2.6 (a) Schematic of Langmuir-Blodgett setup, with (A) a bath containing (B) the monolayer of amphiphilic molecules at the gas-fluid interface, (C) the substrates being withdrawn from the bath, and (D) a balance that measures the surface pressure of the monolayer, generated by the movable barrier. (E) The moving barrier is controlled by a feedback device (F), which gets information from the pressure sensor and adjusts the barrier position using the (G) motor. (b) Absorption spectra of LB film obtained using the labelled number of depositions. Inset shows exciton absorbance intensity against the number of depositions, and it can be seen that the absorbance is increased as the films become thicker with an increasing number of depositions. (a) reproduced from Ref. [49], and (b) reproduced from Ref. [56].

A similar dipping method has been used to make thin films of $\text{C}'_{12}\text{PI} \cdot \text{INH}_3(\text{CH}_2)_{12}\text{NH}_3\text{I}$ is dissolved in dioxane, then diluted with water to create a 5mM solution. A PbI_2 /dry oxane solution is also created, and quartz substrates cleaned in a solution of H_2SO_4 and H_2O_2 , then rinsed with distilled water and methanol. The hydrophilic quartz substrates are dipped in the iodide solution for 20mins, then immersed in a dioxane/water solution for 5mins to remove excess organic salts. Substrates are then dipped in the PbI_2 solution for 15mins, and finally washed with dioxane. The procedure can be repeated as needed in order to achieve multilayers of self assembled quantum wells and control the films thickness [59].

Gaseous intercalation has also been demonstrated by Era *et al.*. A 20 nm thick film of PbI_2 is vacuum deposited on a quartz substrate, then exposed to vaporised organic ammonium iodide ($\text{C}_6\text{H}_5\text{C}_2\text{H}_4\text{NH}_3\text{I}$). Both XRD and absorption confirm the formation of PAPI, although in XRD signatures of PbI_2 were still seen [60].

Recently in-situ measurements on liquid-phase intercalation has shown that intercalation occurs from the top surface of the PbI_2 film, and proceeds in a direction perpendicular to the inorganic layers. Organic molecules attaches to the top surface of the PbI_2 layer has terminal groups, therefore transforming the face-sharing PbI octahedra into a corner-sharing network, opening interstices for organic molecules to diffuse through and interact with the bottom surface of the same layer, thus converting the PbI_2 layer into full 2D monolayer. Further

diffusion continues intercalation for layers further from the interface [Fig. 2.7]. This process requires a non-polar solvent that will not compete with the hydrogen bonding between the inorganic and organic constituents, while steric hindrance between molecules also plays a role in the dynamics, thereby producing an optimum organic iodide concentration for the intercalation solution [61].

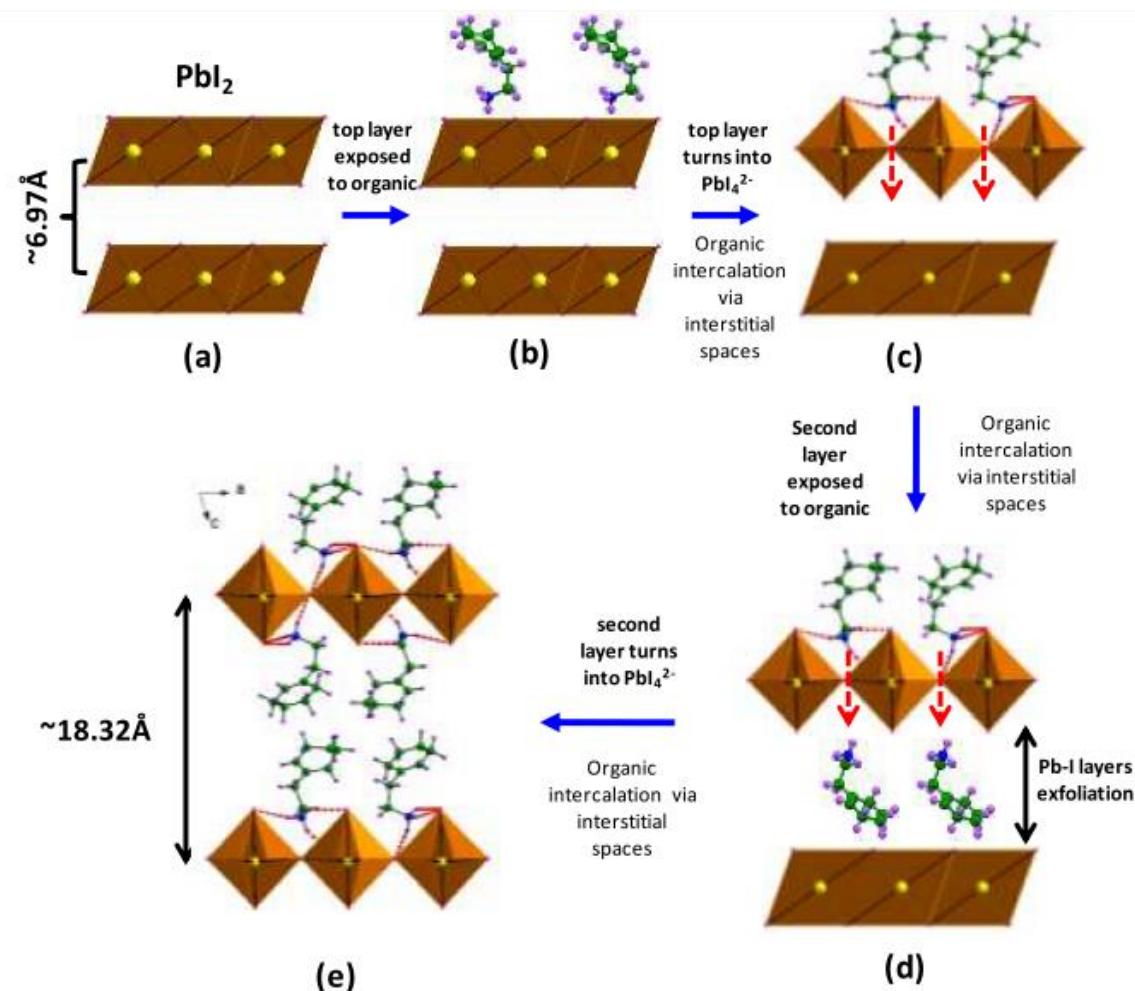


Fig. 2.7 Schematic of liquid-phase intercalation dynamics of organic ammonium iodide molecules into PbI_2 film. Reproduced from Ref. [61].

Dual-source vacuum deposition

PAPI films were also created using a dual-source vacuum deposition method, where both PbI_2 and $(\text{C}_6\text{H}_9\text{C}_2\text{H}_4\text{NH}_3)\text{I}$ were vacuum deposited simultaneously on a quartz substrate with rates of 7.1 and $21\text{ng cm}^{-2}\text{s}^{-1}$ respectively. The created film showed the same sharp

exciton absorbance peak as films created by other methods, however the XRD pattern of the film showed no peaks at high θ , so the films were disordered and possibly defective. The growth of the layered perovskite structure appears to occur in the solid phase on the substrate [62]. In addition the organic ammonium halide may dissociate into an amine and HI, so care needs to be taken with choice of evaporation rates, or else films can be multiphasic, disordered or defective [63].

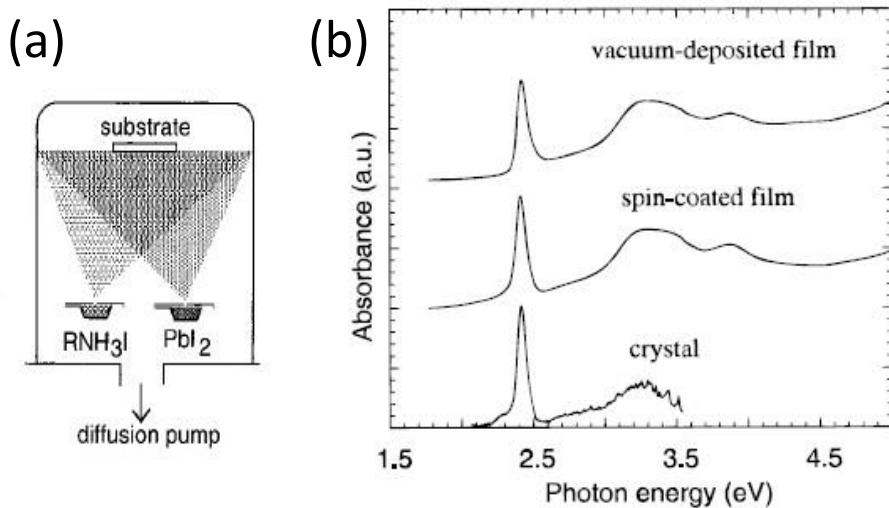


Fig. 2.8 (a) Schematic of dual-source vacuum deposition process. Deposition rates of the two sources needs to be controlled carefully in order to create a non-defective film. (b) Comparison of absorbance spectra of a vacuum deposited film, a spin coated film, and a single crystal sample of PAPI, where there appears to be very little difference between the three different methods. Reproduced from Ref. [62].

Single source thermal ablation

Single source thermal ablation involves the vaporisation of an initial material onto a substrate in order to form a film (see Fig. ??). The initial material can be placed on the tantalum heater as a powder (insoluble powders ideally as a suspension for better thermal and physical contact with the heater, and more even dispersion), crystal, or a concentrated solution that has been allowed to dry. The chamber is then pumped to vacuum, and a current passed through the heater. The starting material is vaporised from the heater surface and the perovskite is reassembled on substrates. The important control variable is the rate at which the heater reaches its final temperature, as a low rate may lead to multiphasic or defective films. Substrates can undergo multiple ablations, and the amount of initial material can also be used to control film thickness [63]. AETHPI ($\text{AETH} = \text{C}_{18}\text{H}_{28}\text{N}_2\text{S}_2$) films are created using

this method. Luminescence spectra show that as-formed films had only traces of a small exciton peak, so there is probably some short range order, but the MQW structure is not fully formed. Sharp exciton peaks are seen after the film is annealed, and the peak increased in intensity with annealing time as grain sizes increase. The Stokes shift also decrease with annealing time, showing an improvement in quantum well quality [64].

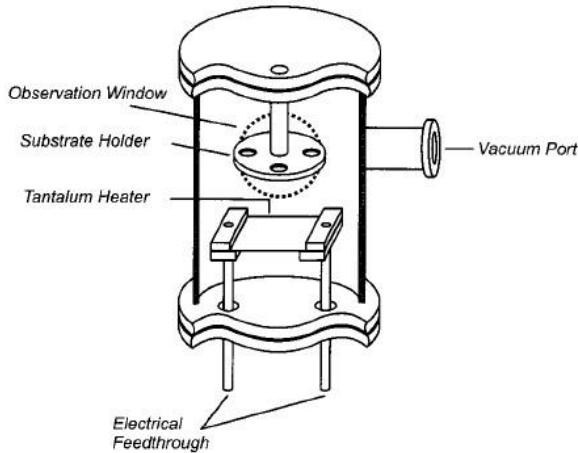


Fig. 2.9 Cross section of single source thermal ablation apparatus. The starting material is placed on the tantalum heater, then as a charge is passed through the solid sublimates and forms a thin film on the substrate above. Reproduced from Ref. [64].

Spin coating

Thin films can easily be produced by spin coating a solution of PbI perovskites on a substrate (suitable solvents include DMF [42], THF [65], and acetonitrile [?]). Drops of solution are added to spinning substrates, and as the solvent evaporates a polycrystalline film is left behind. The films produced have the same optical or electronic properties as single crystals, and the crystallographic *c* axis perpendicular to the surface of the films [66]. The films are generally smooth with a roughness of \sim 1-2nm, although the solvent, substrate, perovskite solution concentration, substrate temperature, and spin speed may all affect the thickness and morphology of films. Pre-treating substrates may also improve wetting properties and affect film morphology [49]. In thicker films, strain and uneven crystal planes lead to stacking imperfections, which may give quantum wells of differing widths. The organic molecules may also become misaligned, and inorganic layers distorted. All these effects lead to a decrease in the intensity of exciton absorption/luminescence [?].

PbI perovskite samples tend to degrade over time due to moisture in the air, so a PMMA matrix doped with nanocrystalline PAPI can be created in order to suppress degradation

[67]. PPMA, PbI₂, and (C₆H₉C₂H₄NH₃)I are dissolved in DMF then spin coated onto a glass substrate, and the resulting film annealed (thickness \approx 200nm). It can be determined through XRD that the *c* axis in PAPI crystals are perpendicular to the surface of the film. A strong exciton absorption at 2.4eV is seen as expected, as well as a step-like feature at 2.7eV due to interband transitions (see Fig. ??). After the PMMA doped sample is left in a humidity controlled box for two months, the spectrum appeared unchanged, but the PAPI film is degraded and no exciton peak can be seen (see Fig. ??). The binding energy of excitons in PMMA doped films is around 300 meV, larger than in pure PAPI samples (around 250meV) due to dielectric confinement of PMMA.

Adjusting the temperature and duration of annealing changes the size of the nanocrystals, but the absorption energy is not affected by the size of the crystals. On the other hand, annealing conditions do change the magnitude of the exciton absorption. At a temperature of 100°C, an increase in annealing time (up to 2 hours) increases the absorption intensity as the nanocrystals become larger, although the relative absorption is not changed much [Fig. 2.10(a)]. However at 125°C an increase in annealing time actually reduces absorption (both absolute and relative) as PAPI crystals decompose and PbI₂ signatures are seen in the spectra. Upon annealing for longer times, even at 100°C the relative absorption decreases, presumably also due to decomposition [68]. As seen in Figs. 2.10(c,d), both annealed films are more thermally and photo-stable than pure PAPI films. Oxygen(1s) signatures are seen in the photoelectron spectrum of photo-irradiated PAPI films, so photo-induced oxidation is a likely mechanism for photo-degradation [68].

Patterning

Patterned PAPI films have been produced using a micromoulding in capillaries method (MIMIC). Moulds are created by casting PDMS on silicon master moulds, then peeling away the PDMS layer. In order to make patterned films, PDMS moulds are placed in conformal contact with pre-cleaned silicon substrates so the channels of the mould formed capillaries with the substrate. A solution of PAPI dissolved in DMF is dropped on one end, and the channels are spontaneously filled by capillary force [Fig. 2.11(a)]. The mould and solution are then cured for 2hrs at 65° to evaporate the solvent. In general the film stripes in Fig. 2.11(b) are defect free, however some edge defects are seen in C (channel width 0.8 μ m) since the channels are more difficult to fill when the width decreases. The width of film stripes also tend to be a little smaller than the width of mould channels, and a shrinkage of around 25% was seen after evaporation of solvents. The patterned films have same the optical properties as unpatterned films [69].

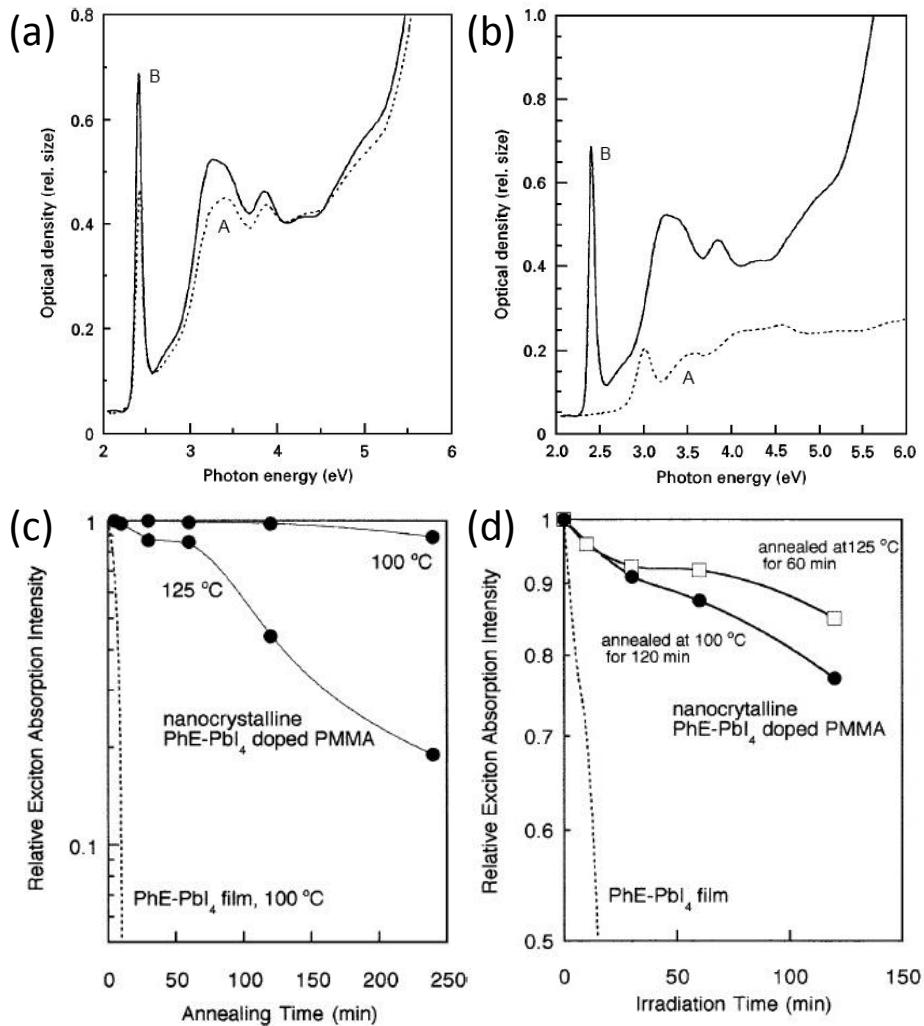


Fig. 2.10 (a) Absorption spectra of A) PAPI film spin coated using acetonitrile solution (dotted line), and B) PAPI doped PMMA annealed at 125°C for 10 minutes (solid line). The spectrum for both films have the same features, although the exciton absorption intensity of A is lower. (b) Absorption spectra of above films after being in a humidity controlled box for two months. Note the exciton peak has disappeared in A as the PAPI film degraded. (c) Changes in relatively exciton absorption intensity (I/I_0) as function of annealing time at different annealing temperatures. (d) Changes in relative exciton absorption intensity as function of photo-irradiation time. In both (c) and (d) the doped PMMA sample at 125°C decomposes with annealing time, although it still absorbs more than pure PMMA film. (a) and (b) reproduced from Ref.[67], (c) and (d) from [68].

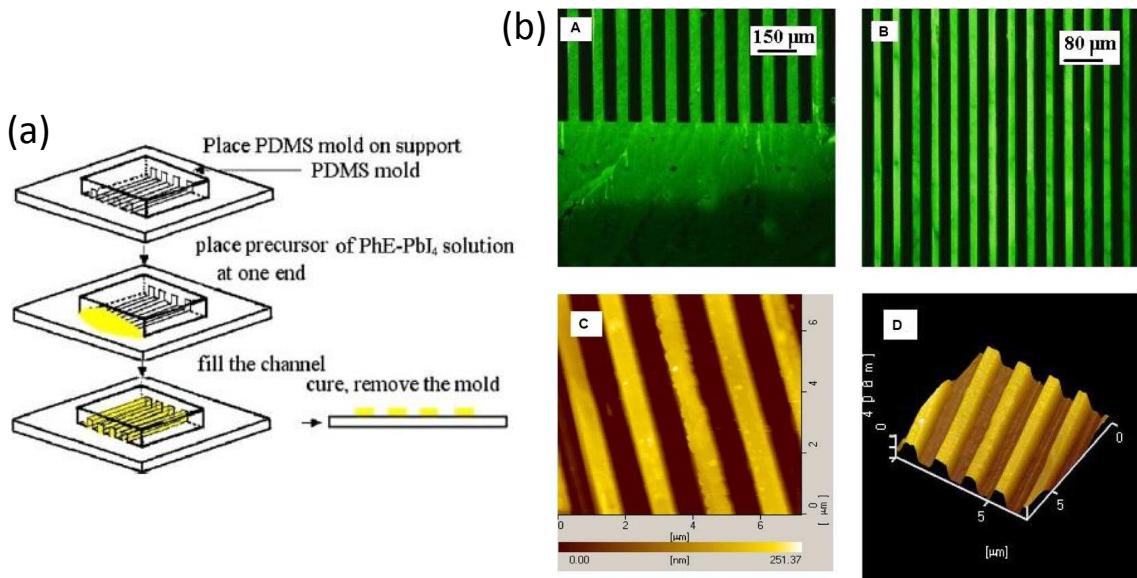


Fig. 2.11 (a) Schematic of MIMIC process for patterning PAPI films. (b) Fluorescent optical micrographs (A,B) and AFM images (C: planar, D: stereo) of patterned PAPI films. In all cases the black stripes represent bare substrate without PAPI. Film stripes widths are A) 50 μm wide, B) 15 μm , C) and D) 0.8 μm . Note in A an unpatterned film created from left solution that did not flow into the channels. In general film stripes are defect free, however in C some edge defects are seen due to the small channel width. D shows that film stripes are organised together like trapeziform ridges. Reproduced from Ref. [69].

2.3.3 Electronic structure

Umebayashi *et al.* calculated the electronic structure of C₄PI and its 3D extension CH₃NH₃PbI₃ using linear combination of atomic orbitals (LCAO) within density functional theory (DFT) [70]; the calculated band structures are shown in Fig. 2.12(a). As expected, the 2D compound has a higher band gap and narrower bandwidths due to a decrease in dimensionality. The 2D band structure also has flatter dispersions at the top of the valence band (TVB) and the bottom of the conduction band (BCB), leading to a larger effective mass of carriers, and thus larger binding energy for excitons.

Fig. 2.12 also shows the bonding diagrams for a single [PbI₆]⁴⁻ octahedron, as well as the 3D and 2D compounds above. In the 2D crystal the TVB consists of Pb(6s) and I(5p) σ -antibonding orbitals, whereas the BCB consists of Pb(6p) and I(5s) σ -antibonding orbitals and Pb(6p) and I(5p) π -antibonding orbitals (not labelled on figure). The crystal field also lifts the degeneracy between different iodine atoms, so the conduction band with bridging iodine atoms (I₁) is wider than that with terminal iodine atoms (I₂). In Ref. [71] the BCB was labelled non-bonding from first principles pseudopotential total-energy calculations with the

local density approximation, although it still consisted of Pb(6p) and I(5p) orbitals.

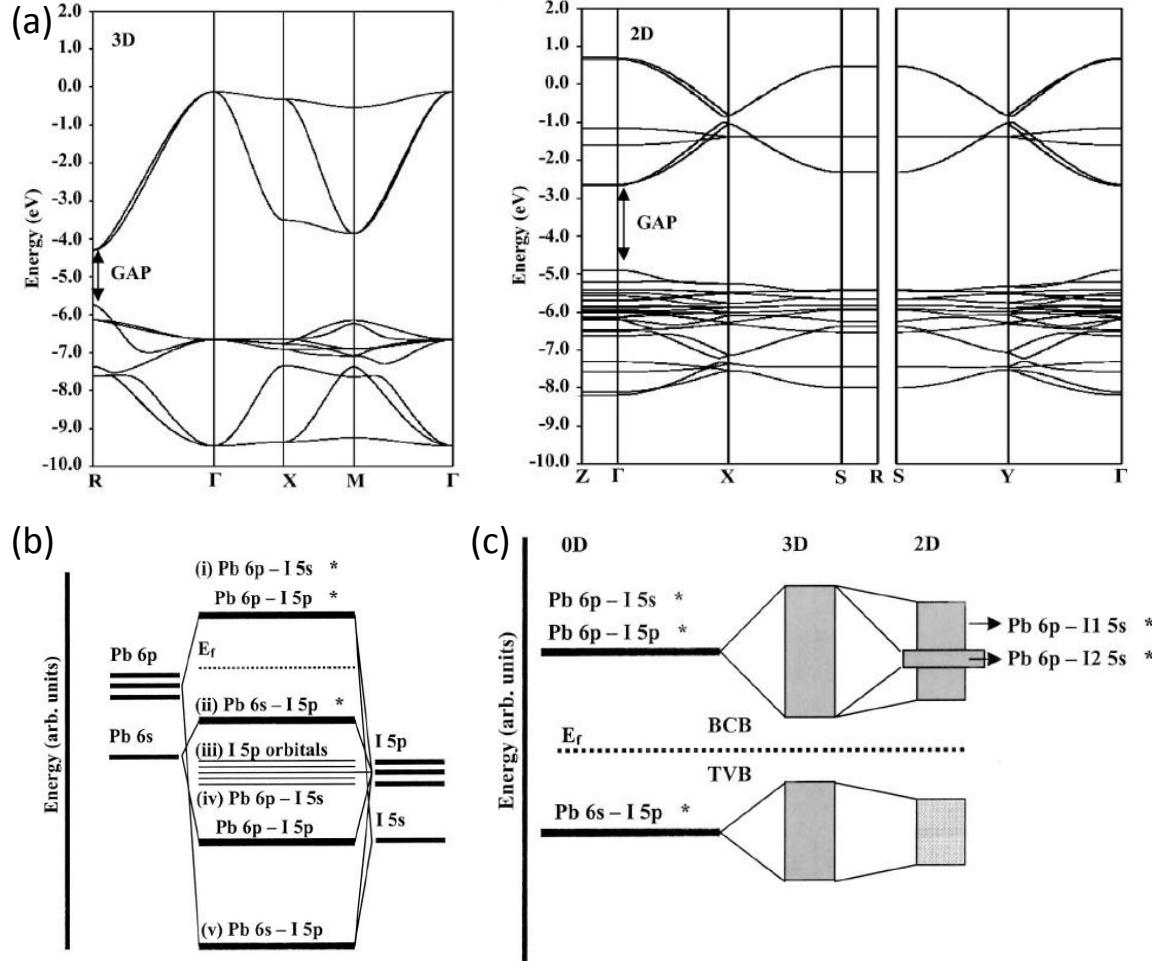


Fig. 2.12 (a) Calculated band structures of CH₃NH₃PbI₃ (3D) and C₄PI (2D) along high symmetry lines in the first Brillouin zone. The band gap is labelled. Bonding diagrams of (b) individual [PbI₆]⁴⁻ octahedra, and (c) extension to CH₃NH₃PbI₃ and C₄PI. The bottom of the conduction band (BCB), top of the valence band (TVB), and Fermi energy level (E_f) are labelled. Reproduced from Ref. [70].

2.3.4 Optical properties

Excitons

Excitons are formed due to transitions between the TVB and BCB in the inorganic layers of PbI perovskites. Excitons in C₁₀PI have an energy of around 2.4 eV above 275K, and the value changes to around 2.55eV at below 268K due to a phase transition of the alkylammonium molecules [Fig. 2.13]. The binding energy E_b was calculated from the difference

between the exciton peak and the step-like structure (interband transitions) in absorption spectra, and a large value of 320 ± 10 meV explains the presence of excitons even at room temperature [21]. The exciton also has a large oscillator strength of around 0.7 ± 0.1 [21]. The above values were for the lowest energy ($n = 1$) free exciton, but bound excitons are also seen, for example in Fig. 2.13(b) the peak at 2.55 eV is assigned to the lowest free exciton in the inorganic layers, whereas the peak at 2.53 eV decreases in intensity with temperature, and is thus assigned to a shallow bound exciton. The peaks seen at low temperature at 2.45 eV is sample dependent, and thus assigned to deeply impurity-bound excitons. Peaks at lower energies were sample dependent, and assigned to excitons deeply bound to impurities [21]. As seen from Fig. 2.13(c) the Stokes shift for C_{10}PI is small (less than 5 eV), and this is generally true for PbI-based perovskites (e.g. for C_{12}PI in Ref.[46]). In Ref.s [21, 37], both longitudinal and transverse exciton-polaritons were observed in reflection spectra of C_{10}PI at 1.6 K, with a splitting of around 60meV [21]. The exciton radiative lifetime was found to be ≈ 7 ps at 8 K [72].

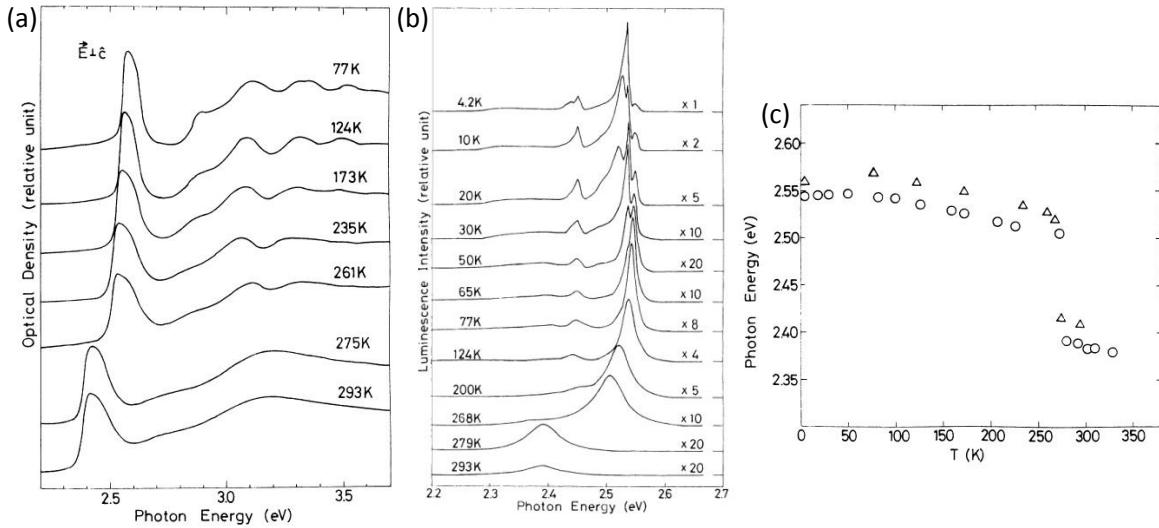


Fig. 2.13 (a) Optical density of C_{10}PI crystal at labelled temperatures. The peak furthest left is assigned to the lowest energy free exciton, and the step-like feature next to it (seen most clearly at 77 K) is due to interband transitions. Thus the binding energy has been calculated to be 320meV. (b) Photoluminescence spectra spectra of C_{10}PI . For assignment of peaks see main text. (c) Energies of absorption (\triangle) and photoluminescence (\circ) peaks as a function of temperature, showing small Stokes shifts of around 5meV. Figures reproduced from Ref. [21].

Polarisation-dependent excitons were seen in the absorption spectra of $\text{C}_6'\text{PI}$, as calculated from Kramers-Kronig analysis of polarisation-dependent reflectivity spectra [Fig. 2.14].

When the incoming light is polarised parallel to the crystallographic b axis, the lowest exciton polarised parallel to b is at 2.5272 eV. Two shoulders seen at 2.566 and 2.718 eV (indicated by small arrows) are due to vibronic bands. The very small peak at 2.819 eV is attributed to $n=2$ excitons, as determined using exciton activation energies calculated from PL spectra. Light polarised parallel to c can generate excitons polarised to both the \vec{a} and \vec{c} axes since the crystal is monoclinic, so the narrow peak at 2.559 eV is due to excitons polarised parallel to \vec{c} . As the \vec{a} and \vec{b} directions are very nearly isotropic, the peak at 2.5272 eV, also seen in Fig. 2.14(a), is attributed to excitons polarised parallel to \vec{a} . However the peak due to the exciton parallel to \vec{c} only appears at very low temperatures, as the peak width normally makes the two exciton peaks indistinguishable. The $n = 2$ peak is seen at the same position as in Fig. 2.14(a). The binding energy for the material is calculated to be 330meV, and the Bohr radius 8.2Å[73].

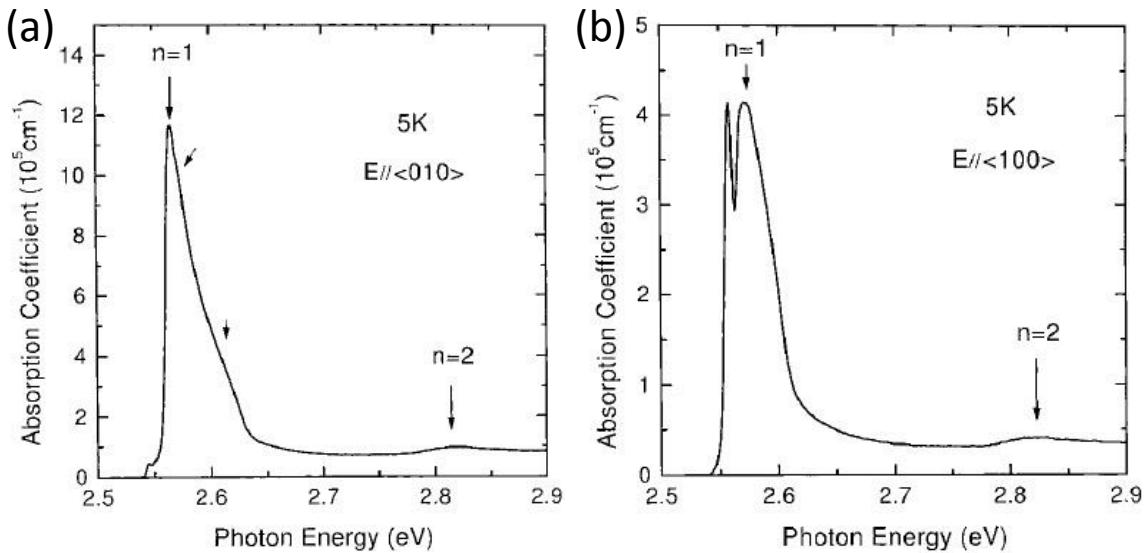


Fig. 2.14 Absorbance spectra for $\mathbf{E} \parallel$ crystallographic (a) \vec{b} , and (b) \vec{c} axes for C₆PI, calculated using Kramers-Kronig relations from reflection spectra. In (a) the $n = 1$ peak is due to excitons polarised parallel to \vec{b} , and the shoulders are seen due to vibronic bands. In (b) the sharp $n = 1$ peak is attributed to excitons polarised parallel to \vec{c} , whereas the broad $n = 1$ peak is due to excitons polarised parallel to \vec{a} . Reproduced from Ref. [73].

There has been some discussion regarding whether the excitons are Wannier- or Frenkel-like in nature. Xu *et al.* used magneto-optical measurements on polycrystalline C₁₀PI thin films and found that exciton peaks were not shifted when a magnetic field B was applied in the plane of the film, but when the field was applied perpendicular to the plane of the film

the energy of the exciton changed according to

$$E = E_0 \pm \frac{1}{2}g_{\perp}\mu_BB + c_0B^2, \quad (2.9)$$

where E is the energy of the exciton, E_0 is the energy of the exciton at zero field, g_{\perp} is the Landé g factor perpendicular to the plane of the film, μ_B is the Bohr magneton, and c_0 is the diamagnetic constant. The energy shift depends on the polarisation of the magnetic field (plus sign for σ^+ and minus sign for σ^-), and both g_{\perp} and c_0 depend on exciton wavefunction. Calculated value showed g_{\perp} were ~ 1 , and $c_0 \sim 10^{-7}$ eV/T². From the magneto-absorption measurements the Bohr radius a_0 of C₁₀PI was estimated to be 12Å. The peak shifts indicate C₁₀PI excitons are Wannier-like, since Frenkel excitons would have no extended motion and thus show no energy shift at all. However as the size of each [PbI₆] octahedron is around 6Å [21], so excitonic motion only extend over a few octahedra [74]. Further magneto-optical measurements revealed a small anisotropy in the 2D $a - b$ plane, as well as evidence of strong interaction between the exciton and phonon modes (phonon sidebands), suggesting a small extension of the internal wavefunction of the exciton [75].

C₆PI crystals were used in magneto-optical measurements, and the samples were of good enough quality that a single peak in Ishihara's measurements could be resolved into an exciton peak and its phonon sidebands. The calculated g_{\perp} and c_0 values were similar to those in C₁₀PI, however the authors believed that as the exciton Bohr radius is on the order of Pb-Pb distances in the crystal, the exciton may be better described by the Frenkel model (exciton wavefunctions made of s and p orbitals of the Pb²⁺ cation, and involve transitions from the $(ns)^2$ to $(ns)(np)$ molecular orbitals). The orbital wavefunctions of Pb²⁺ were then used to calculate g_{\perp} and c_0 , and the results agreed well with experiment [66]. On the other hand Tanaka *et al.* used electroabsorption and two-photon absorption studies and showed that excitons in thin film C₆PI samples were Wannier-like in nature, with 1s, 2s, 2p, and 3p energies at 2.34, 2.60, 2.61, and 2.64 eV respectively. The Wannier excitons exhibited strong 2D behaviour, with a large binding energy of the 1s exciton (310meV) [76].

The band gap of the material can be engineered by altering the halogen and metal atoms in the perovskite structure, thus tuning the exciton energy. The exciton energy ranges from 2.5 eV for PbI perovskites, to 3.1 eV for PbBr and 2.6 eV for PbCl. Indeed mixed-halide perovskites of the form (RNH₃)₂PbA_xB_{1-x} allows to further variation within this range [50, 77]. It has been shown that while the change in absorption wavelength generally varies linearly with the halogen concentration x as the process is averaged over excitons at all possible sites, the excitons formed then preferentially diffuse to lower energy halogen

atomic sites, thus emission wavelength does not show the same linear variation with x [78].

Biexcitons and triexcitons

An increase in excitation power can lead to the formation of bi- or tri-exciton complexes, where two or three free excitons are bound together. On the other hand, induced photo-carriers screen Coulomb interactions and make complexes less stable, so electron-hole plasmas usually form before triexcitons. In order to see triexcitons there needs to be a strong Coulomb interaction between carriers to withstand screening, and a low dimensionality also helps as screening by carriers only has a limited effect [79]. The radiative decay of a biexciton to a transverse exciton has been observed in $C_{10}PI$, with biexciton binding energy of $\approx 50\text{meV}$ [80]. By creating a waveguide configuration (200nm C_6PI film spin coated on Ti-containing SiO_2 with Al-mirrors formed on the optically flat output faces of the substrate) with transverse pumping, biexciton lasing was observed in C_6PI . The lasing threshold is 20 kW/cm^2 at 16 K and increases sharply with temperature. Fig.2.15(b) shows emission spectra above and below the lasing threshold: a broad biexciton band is seen below the threshold, but a sharp peak at 2.281eV is seen above the threshold. The biexciton band is probably isotropic as the emission was not polarised [81].

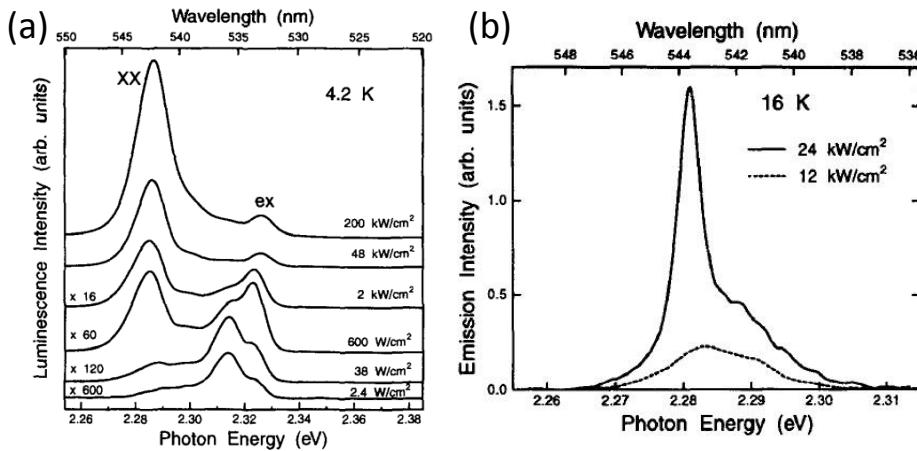


Fig. 2.15 (a) Photoluminescence spectra of C_6PI film excited by 337 nm nitrogen laser at 4.2K with labelled excitation intensities. Free exciton PL (labelled ex) is seen at 2.324 eV with some bound exciton PL at 2.315 eV, but the ex band dominates at low excitation. The biexciton band (labelled XX), due to the radiative recombination of a biexciton leaving behind an exciton, is located 40meV below the free exciton band. (b) Emission spectra of C_6PI waveguide above (24 kW/cm^2) and below (12 kW/cm^2) the lasing threshold at 4.2 K. The lasing wavelength is 543.6 nm. Reproduced from Ref. [81].

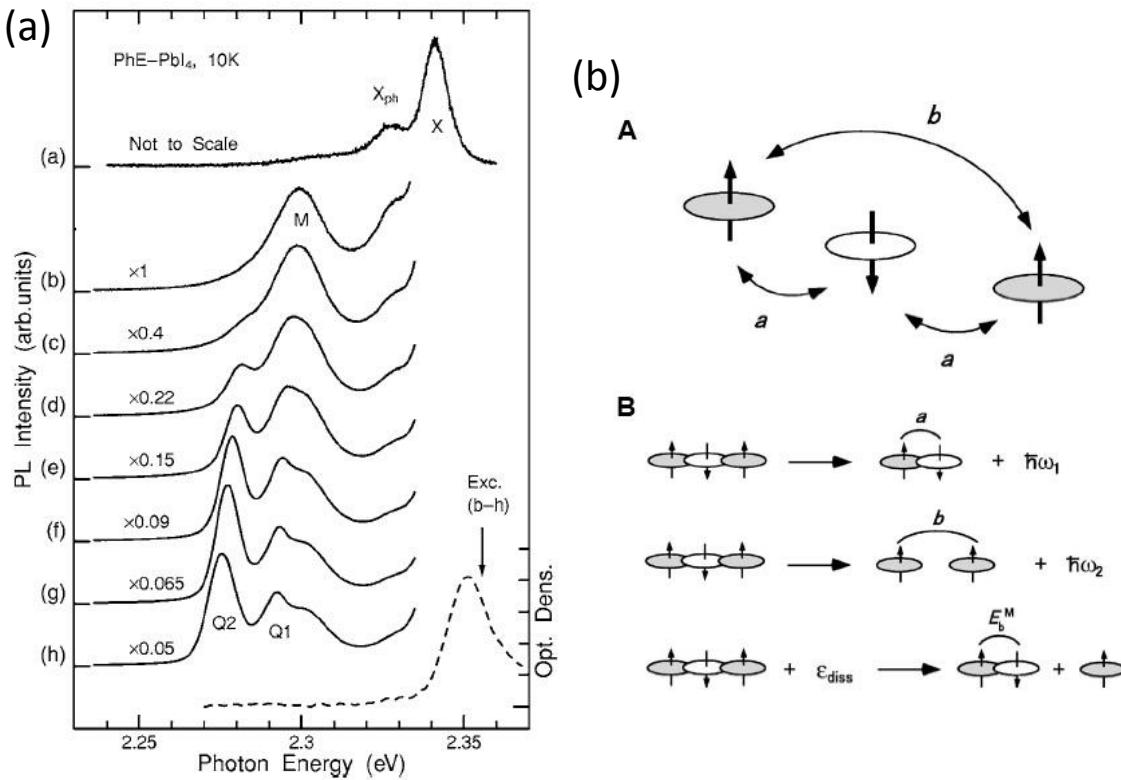


Fig. 2.16 (a) PL spectra of PAPI at different excitation intensities and energies. (a) was excited at 2.58eV, and (b)-(h) at 2.355eV. The excitation intensities are (a) 4.6×10^{10} , (b) 4.6×10^{12} , (c) 1.4×10^{13} , (d) 2.8×10^{13} , (e) 4.6×10^{13} , (f) 9.2×10^{13} , (g) 1.6×10^{14} , and (h) 3.2×10^{14} photons cm⁻². The dotted line shows the absorption spectrum. X is the free exciton band, X_{ph} phonon sidebands, M the biexciton band, Q₁ amplified spontaneous recombination of biexcitons, and Q₂ the $\hbar\omega_2$ triexciton process in Fig. ???. (b) A) shows the triexciton model, and B) shows some likely radiative and dissociation mechanisms. For the meaning of symbols see the main text. Reproduced from Ref.[79].

Shimizu *et al.* observed triexciton formation in the PL spectra of PAPI. In Fig. 2.16(a), the free exciton band is labelled X, phonon sidebands X_{ph}, and M is the biexciton band. At higher excitation intensities, two other bands can be seen, labelled Q₁ and Q₂. Q₁ was assigned to the amplified spontaneous emission due biexciton recombination, and Q₂ to a triexciton process. Fig. 2.16(b) shows two of the likely radiative processes of the triexciton state, which is considered to consist of a bound state of three spin singlet excitons. *a* denotes the interaction energy between opposite spin excitons (< 0), *b* denotes the interaction between same spin excitons (> 0), ε_{diss} is the dissociation energy of a triexciton decaying to a biexciton and a single exciton, and E_b^M is the binding energy of a biexciton. As Q₂ is at a lower energy than M, it can only be due to the $\hbar\omega_2$ process, although it is unclear why the $\hbar\omega_1$ process was not observed. From the data collected, it was calculated that $a=-37.5$ meV,

$b=11\text{meV}$, $\epsilon_{\text{diss}}=14\text{meV}$, and $E_b^M=50\text{meV}$ [79].

Dielectric confinement and the image charge effect

Both the binding energy and oscillator strength of (transverse) excitons in 2D PbI perovskites are much larger than those in the inorganic 3D equivalent, PbI₂ [82]. In materials where a sheet is sandwiched between barrier layers with lower dielectric constant ϵ_b and higher band gap, three types of confinement affect excitons. Firstly “dielectric confinement”, where the lower barrier ϵ_b reduces the effective dielectric constant of the entire structure, thus providing less shielding and giving a higher binding energy E_b . Secondly “quantum confinement”, where the reduction in dimensionality to 2D gives a binding energy four times larger than expected in bulk 3D material [83]. Thirdly “mass confinement”, where carrier wavefunctions extending into the barrier region lead to a larger effective mass, thus increasing the binding energy. Mass confinement depends on quantum confinement in order to determine how much of the carrier wavefunction is leaked into the barrier region, and generally only has a small effect [84]. The interfaces between layers also act as mirrors which create an infinite series of image charges (see Fig. 2.17). Using carrier wavefunctions which fit the boundary conditions of the well, as well as self and image-charge Hamiltonians, exciton properties can be calculated. The results showed that the binding energy of excitons increase if the barrier height decreases, the excitons have larger effective mass in carrier region, or the barrier regions have a smaller dielectric constant [84]. Muljarov *et al.* found that the potentials created due to the image charge effect causes charges in the inorganic layers to be repelled from the interface, whereas charges in the organic layers are attracted to the interface [85].

Pressure induced changes

Fig. 2.18 shows how the absorption and PL spectra of C₈PI changes when pressure is applied to the material in a diamond cell. Both sets of spectra show that the exciton energy decreases as the pressure increases, since a reduction in Pb-I distance leads to an increase in the energy of the antibonding valence band, with the non-bonding conduction band less affected [71]. After a phase transition at 12GPa, absorption spectra show much broader “tails” that can be described by Urbach’s rule, and correlates with the disappearance of exciton signatures in PL spectra. Fig. 2.18(c) shows changes in PL intensity on the up-stroke (pressure increasing from ambient) and down-stroke (pressure decreasing down to ambient). A large hysteresis is observed, with PL intensities significantly smaller on the down-stroke than on

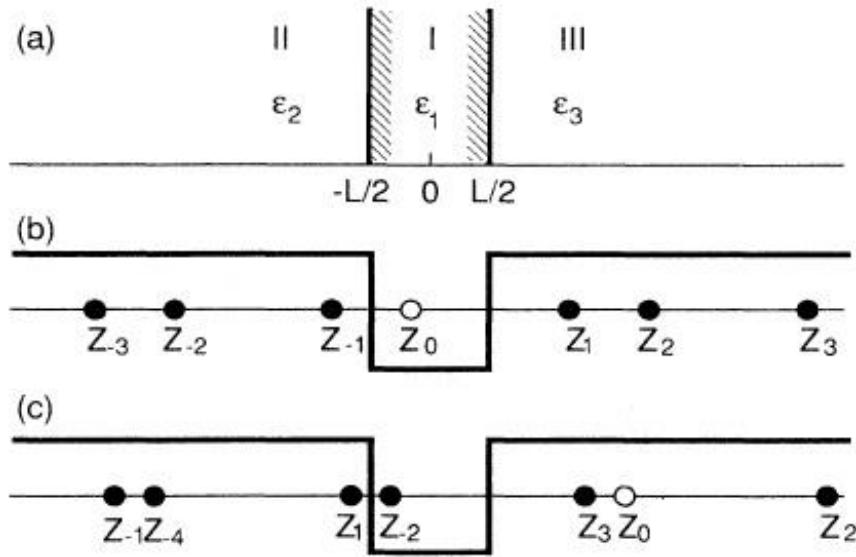


Fig. 2.17 (a) Generalised description of quantum well (region I) and barriers (regions II and III). (b) and (c) show the positions of initial charges (white circle) and image charges created (black circles) due to interfaces. In (b) the initial charge is in the well region, whereas in (c) it is in the barrier region. Reproduced from Ref. [84].

the up-stroke with. The results suggest that defects may have been created with the increase in pressure. As excitons above and below the phase transition seem to involve different electronic states, it has been postulated that the band structure of C₈PI is such that exciton formation involves a direct transition below 12GPa, which changes to an indirect transition above 12GPa [86]. The same experiment was performed on C₄PI with similar results. In C₄PI the phase transition occurred around 10GPa, although some residue of the free exciton band was still seen at 11.2GPa. The difference in transition pressures suggests that the change in exciton energy is at least in part due to a transition in the organic molecules [71].

2.3.5 Influence of organic molecules

How an organic molecule fits into the MQW structure can change the distortion of octahedra, the I-Pb-I bond angle, or the interlayer I-I coupling, all of which lead to a change in the electronic structure [20].

Pb-I-Pb bond angles (i.e. how flat the inorganic sheet is) has a large effect on the band gap of perovskites. Experimental results show that bond angles closer to 180° (undistorted octahedra) lead to smaller band gaps, which agrees with extended Huckel tight-binding calculations used to evaluate band structures of PbI perovskites (results are seen in Fig.

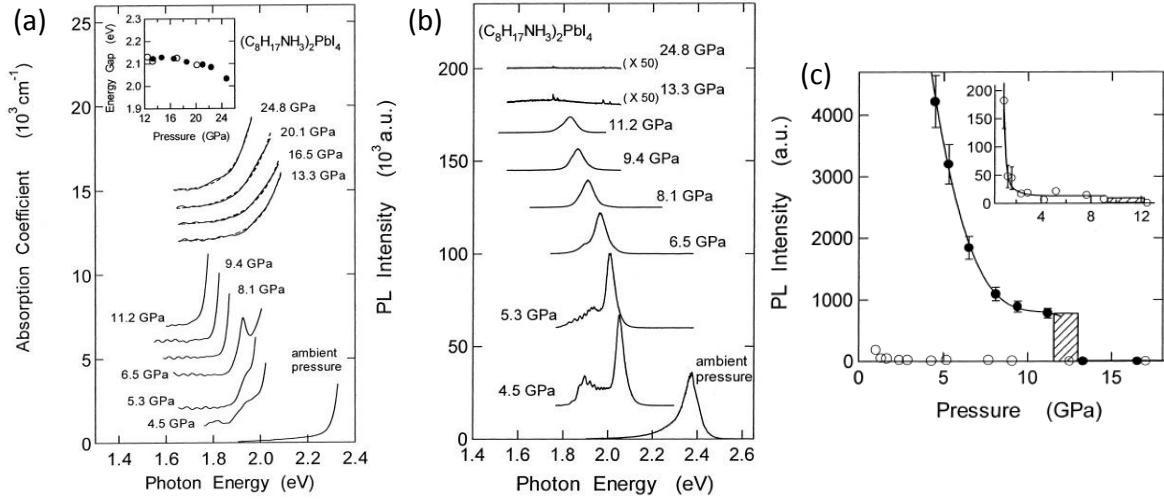


Fig. 2.18 (a) Absorption, and (b) PL spectra of C_8PI under labelled pressure. Exciton energies decrease with increasing pressure, and a phase transition is seen at around 12GPa. The sidebands seen at 4.5 and 5.3GPa spectra are possibly due to bound excitons. The inset of (a) shows the band gap as calculated using Urbach's rule above phase transition. (c) PL intensity against applied pressure for the up-stroke (black circles) and down-stroke (white circles). The shaded box indicates the position of the phase transition. A large hysteresis can be seen as the up-stroke may create defects in the crystal. Reproduced from Ref. [86].

2.19). Although the band gaps are underestimated using this model, the overall correlation between I-Pb-I angle and electronic energy levels are correct [46].

$2,2'$ -biimidazole ($\text{C}_6\text{H}_6\text{N}_4$) was incorporated into the perovskite structure to form $(\text{C}_6\text{H}_8\text{N}_4)\text{PbI}_4$, and the organic molecules can be doubly protonated to 2^+ , or accept two electrons and delocalise the charge across the entire molecule. Since the corrugation of the organic layer is increased if there are strong hydrogen bonds between the organic and inorganic moieties, the weak hydrogen bonds in $(\text{C}_6\text{H}_8\text{N}_4)\text{PbI}_4$ lead to flatter sheets, and thus a smaller band gap [51]. Similarly in $(\text{HO}(\text{CH}_2)_2\text{NH}_3)_2\text{PbI}_4$, where the -OH group is able to hydrogen bond with neighbouring -NH₃ groups or iodine atoms. The extra interactions do not just weaken the NH₃-I hydrogen bonds, but provide a channel for stronger electronic coupling between inorganic layers, leading to a smaller band gap [87].

For a perovskite of form $(\text{R}-(\text{CH}_2)_n\text{NH}_3)_2\text{PbI}_4$, for each -R there appears to be a optimal n that leads to an arrangement that provides the best PL efficiency. In general PL efficiencies are better if the inorganic sheets are close to flat, so as well as hydrogen bonding considerations, the best organic molecules have an -R group that is not too big or small to fit in the spaces between iodine atoms, causing the sheet to deform. Similarly if -R is flexible (e.g. alkyl rather than phenyl), then molecules can fit into interlayer spaces without too much

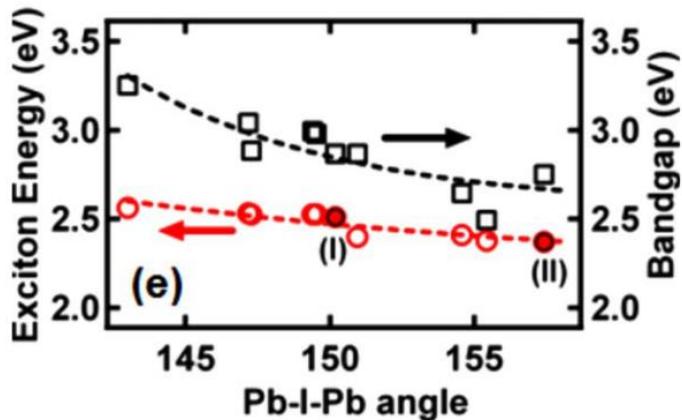


Fig. 2.19 Variation in exciton energy (red) and band gap (black) with I-Pb-I angles calculated using extended Huckel tight binding calculations. Filled in circles represent PL maxima of labelled C₁₂PI phases. The calculations show that exciton and band gap energies both decrease as the Pb-I-Pb bond angle increases. Reproduced from Ref. [46].

deformation of inorganic sheets [40].

As well as effects on the structural leading to optical changes, properties of the organic ligand can be incorporated into the material. For example when chiral organic molecules are used, the perovskite also exhibits optical activity [41]. Similarly tuning the band gap of the organic constituents can lead to charge and energy transfer between the organic and inorganic layers [43, 88, 89].

2.3.6 Applications

Microcavities and photonic crystals

Strong coupling has been observed between PAPI excitons and cavity modes in a distributed feedback microcavity at room temperature [90–92]. The cavity consists of a quartz substrate with PAPI spin coated into the spaces as shown in Fig. 2.20(a) to form parallel wires, and an overcoat of polystyrene added to prevent the degradation of PAPI films. The grating pitch (Γ) was around $0.7\mu\text{m}$, the depth (h) $0.3\mu\text{m}$, PAPI thickness $0.03\mu\text{m}$, and polystyrene thickness $0.5\mu\text{m}$; overall the grating had an area of around $1.5 \times 1.5\text{mm}$. Inside the cavity light is modulated in the x direction, homogeneous in the y direction, and confined in the z direction [91]. When the incident beam is at normal incidence, it is expected that PAPI excitons will couple strongly to the fourth order cavity resonance and form new eigenstates (“cavity polaritons”). Using a variety of grating sizes from 0.62 to $0.72\mu\text{m}$, transmission spectra showed that the upper and lower polariton branches exhibited anti-crossing behaviour, as

expected for strongly coupled modes. Due to the large exciton oscillator strength, the mode splitting was around 100meV, larger than the 9meV observed in GaAs systems in Fabry-Perot microcavities [90]. A strong enhancement of PL intensity of the lower branch polariton is seen if the standing wave cavity mode is in resonance with PAPI excitons, in this case when $\Gamma = 0.68\mu\text{m}$ [91]. No signature of the upper polariton branch was seen since in thermodynamic equilibrium the upper branch would expect to be less populated, however the polariton lifetime may not be long enough for equilibrium to occur. Other suggestions have included a relaxation of the upper branch polaritons towards uncoupled excitonic states, or fast emission of photons between the upper and lower polariton branches [93]. It was thought that the PAPI rods oscillating in phase due to strong coupling with cavity modes at resonance would lead to a macroscopic polarisation and ultrafast resonance, however the polariton lifetime was actually 8ps longer with a grating structure at 40 K. It is possible that this is due to excitons with large wave vectors that cannot couple to the outside without the help of the reciprocal lattice vector [92].

Strong coupling between cavity and PAPI exciton modes were also observed in the Fabry-Perot microcavity [Fig. 2.20(b)] [93, 94]. By adjusting the position of the perovskite layer in the microcavity, the coupling between exciton and photon modes could be controlled, and the splittings seen were between 130-190 nm. Similar CHPI microcavities were constructed and showed splittings of 130meV for a $5\lambda/4$ metal-air microcavity, and of 160meV for a $7\lambda/4$ metal-metal microcavity [19].

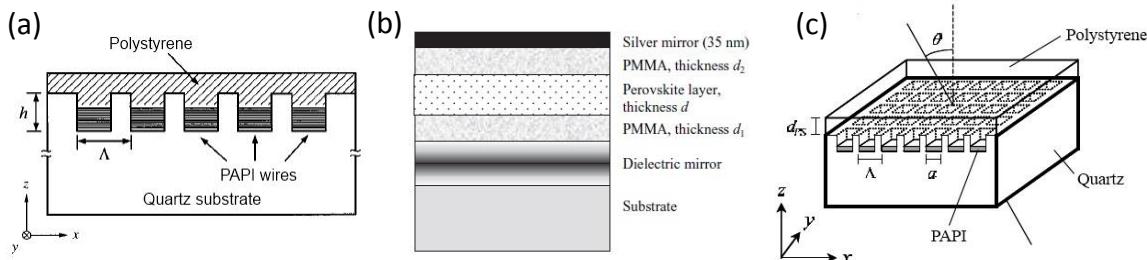


Fig. 2.20 Schematic of (a) a distributed feedback microcavity, (b) Fabry-Perot microcavity, (c) 2D photonic crystal. In (a) and (c) Γ represents the grating pitch, h the grating depth, d_{PS} the height of the polystyrene layer. θ , the angle of incident light, is also labelled. (a) reproduced from Ref. [90], (b) from [93], (c) from [95].

Strong exciton-photon coupling was also seen in a photonic crystal of silica microspheres infiltrated with PAPI (silica opal) [96]. A face centred cubic lattice of silica microspheres with 3D channel voids was created, and a solution of PAPI and DMF were introduced into continuous spaces through capillary forces. The structure was then left for

12hrs at 25°C for the solvent to evaporate. Fig. 2.21(b) shows the reflectivity spectra for a range of incidence angles θ . When $\theta = 10^\circ$, the band near 2.13eV is attributed to the stop band of the photonic crystal, and the band near 2.4eV to excitons in PAPI. At larger angles the stop band blue shifts and asymptotically approaches the almost constant exciton band. Above $\theta = 35^\circ$ another band appears on the higher energy side of the exciton band. The solid black line in Fig. 2.21(c) shows a plot of the expected stop band energy as a result of the photonic structure of a silica opal with a PAPI filling fraction (f_{PAPI}) of 0.06, the dashed black line shows the energy of an exciton for a PAPI sample in free space, and the black circles show experimental peak values from Fig. 2.21(b). The silica opal clearly demonstrates anticrossing behaviour, indicating a strong coupling between the stop band (photon mode) and the exciton mode of PAPI, with a Rabi splitting of 240meV. However there is an exciton peak that seems largely unaffected by the photon modes, and this may be due to some of the bulk PAPI on the surface of the silica opal remaining, or an open photonic gap in the silica opal which leads to insufficient photon confinement, and thus limiting the exciton-photon coupling. Similar strong coupling was seen in the 2D photonic crystal shown in Fig. 2.20(c), where the anticrossing splitting was also 100meV [95].

Optoelectronic devices

Era *et al.* used spin coated PAPI to make an electroluminescent (EL) device. Fig. 2.22(a) contains a schematic of the device, consisting of an indium-tin-oxide (ITO) anode, MgAg cathode, and oxadiazole (OXD7) was used as the electron transport layer. As the device is driven, electrons in the OXD7 layer are injected smoothly into the PAPI layer as there is no energy barrier [Fig. 2.22(b)]. On the other hand, holes injected into PAPI will remain at the PAPI/OXD7 interface due to the barrier potential created by the large OXD7 ionisation potential. Electrons and holes are trapped in the PAPI layer, then recombine to provide luminescence. When driven at liquid nitrogen temperatures, the EL intensity reached a luminescence of more than 10000cdm^{-2} at a current density of 2Acm^{-2} and voltage of 24V. The emission peak at 520nm has a narrow bandwidth, and is very similar to the PL spectrum. However the EL efficiency at room temperature is much smaller than that at liquid nitrogen temperatures, and is mainly caused by thermal ionisation of excitons [97].

Hattori *et al.* made similar EL devices using both PAPI, CHPI, and PBPI ($(\text{C}_6\text{H}_5\text{C}_4\text{H}_8\text{NH}_3)_2\text{PbI}_4$). The authors found that CHPI and PBPI both have much higher PL efficiencies than PAPI. While the EL spectra look very similar to the PL spectra [Fig. 2.23], the PBPI device had the lowest external quantum efficiency η_{ext} (number of emitted photons/number of electrons) and CHPI the highest. The PBPI device was also much more resistive than the others, with

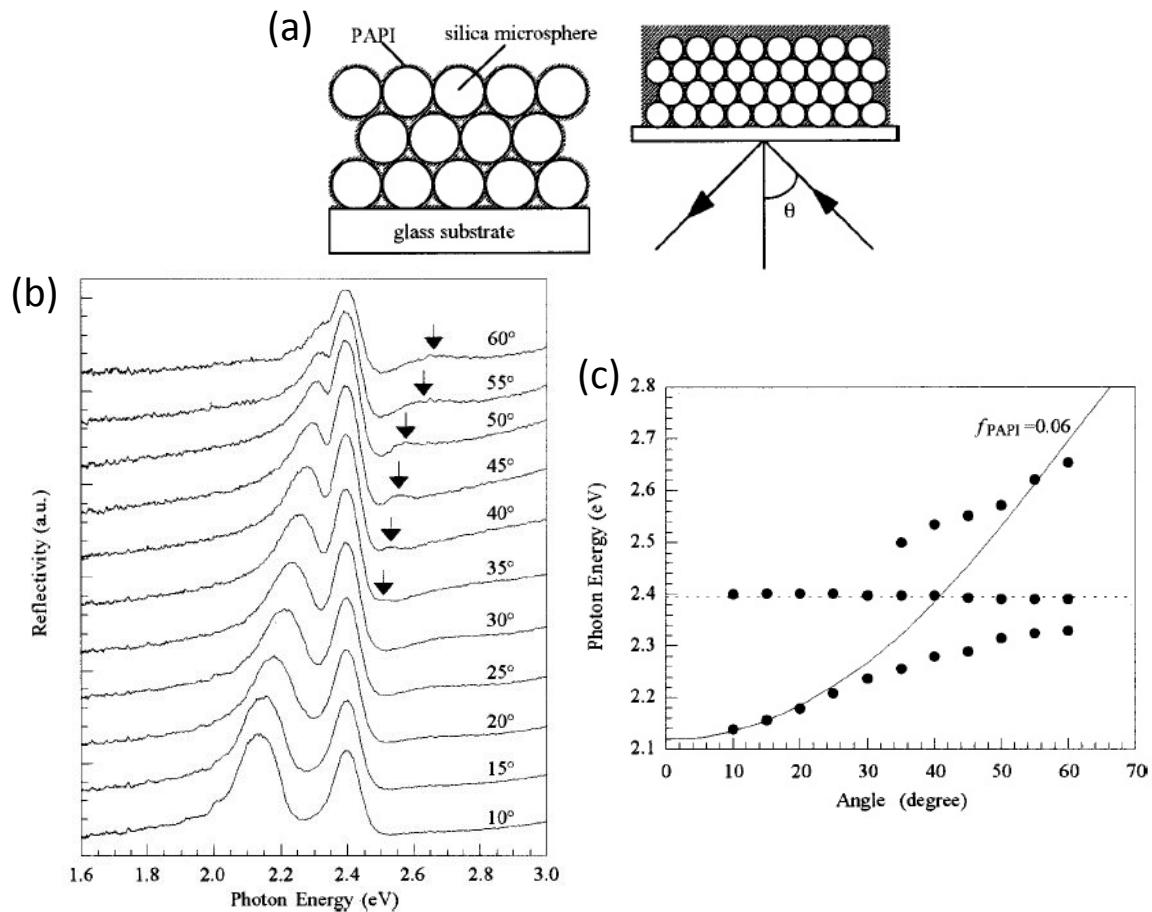


Fig. 2.21 (a) Schematic of the structure of the silica opal infiltrated with PAPI film formed on a glass substrate. θ , the angle of incident light, is also shown. (b) Reflection spectra recorded for a range of θ values, displaced vertically for clarity but plotted on the same scale. The lower polariton branch peaks asymptotically approaches the exciton peaks at 2.4 eV, and small upper polariton branch peaks can be seen at high angles. (c) Peak energies (black circles) against observation angle. The solid black line is the theoretical stop gap energy from Bragg's law, and the dashed black line is the exciton energy of PAPI in free space. Anticrossing behaviour can clearly be observed, and unaffected exciton peaks may be due to some bulk PAPI remaining on the photonic crystal. Reproduced from Ref. [96].

current density around two orders of magnitude less than other devices at same voltage. Both the resistance and the low EL efficiency are probably due to the larger alkyl chain in the organic molecule preventing carrier transport. The external efficiency of the CHPI device was comparable to the highest efficiency reported in EL devices reported at that stage [98].

Matsushima *et al.* used PAPI to create LEDs (schematic shown in Fig. 2.24). PAPI films were deposited onto an OTS coated, CuPc coated, or bare substrate (OTS=octadecyltrichlorosilane,

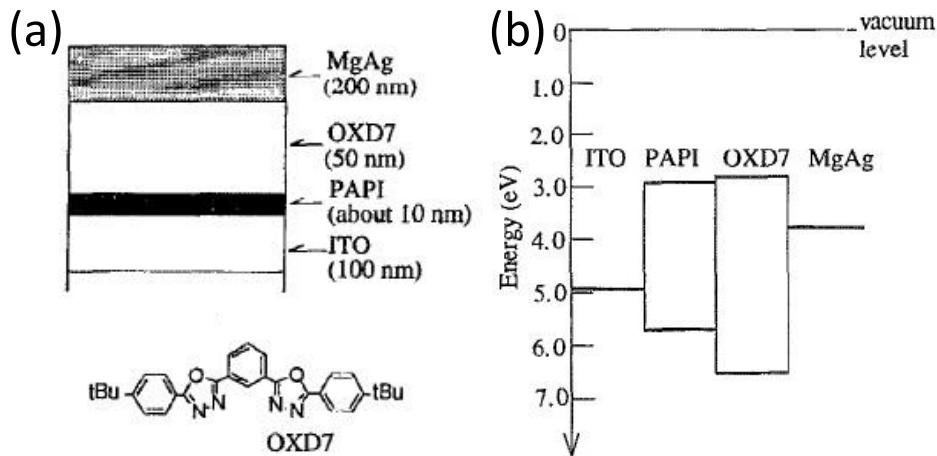


Fig. 2.22 (a) Schematic of the EL device (top) and the molecular structure of OXD7 (bottom). (b) Energy level diagram of EL device. Electrons are smoothly injected from OXD7 to PAPI, however holes remain at the PAPI/OXD7 interface due to large barrier potential. Reproduced from Ref.[97].

CuPc=Cu phthalocyanine), then annealed for 20 mins at 100°C. PAPI films on substrates with an OTS monolayer were more crystalline and had larger absorption intensities compared to bare glass substrates, however films on CuPc covered substrates had the smallest surface roughness. Driven at 105 K, CuPc-coated samples had the best EL efficiency due to a lowering of the hole injection barrier at the ITO/PAPI interface, and also due to smaller leakage currents as a result of smaller surface roughness of PAPI films [99].

The processability of hybrid perovskites from solution make them ideal materials for use in optoelectronic devices. The conductivity of SnI-based perovskites have been investigated, and show that while 2D layered structures are semiconducting, the 3D perovskite structure is a low-carrier density p-type metal with a corresponding increase in the conductivity [100]. Thin film transistors have also been produced from the 2D perovskite, with carrier mobilities of up to $1.4 \text{ cm}^2/\text{Vs}$, better than that of amorphous silicon, on-off ratio > 1000 , and current densities $> 400 \text{ A/cm}^2$ [101? , 102].

Scintillators

Scintillators need to convert radiation energy into photo-emission for the purpose of detecting ionising radiation. They need to have a short luminescence decay time constant in order to react quickly, high resistance to radiation damage, and high efficiency so that a suitable number of photons are created per unit radiation energy of absorbed [103]. Many efficient

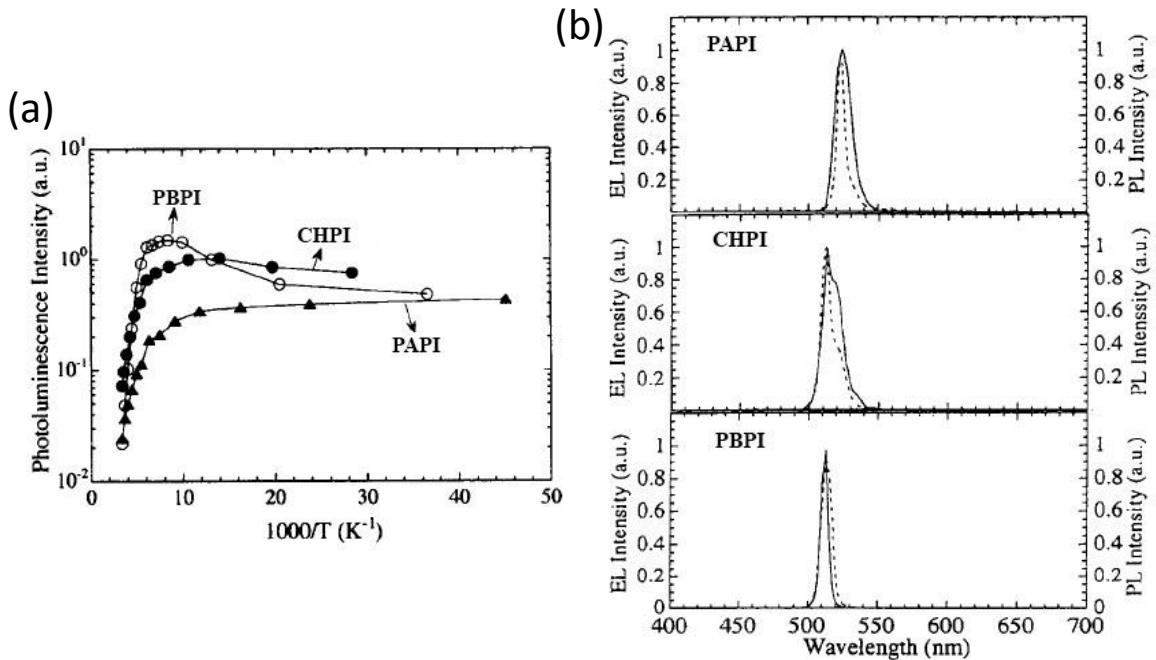


Fig. 2.23 (a) Integrated PL intensity of CHPI, PAPI, and PBPI thin film samples as a function of temperature. (b) EL (solid lines) and PL (dotted lines) spectra of the labelled perovskite. EL devices were made as shown in Fig. 2.22(a), and driven at 110K. PL spectra were recorded using thin film samples at 110K. Although the EL and PL spectra are very similar for all three compounds, the CHPI device had the highest EL efficiency, while PBPI had the worst. Reproduced from Ref. [98].

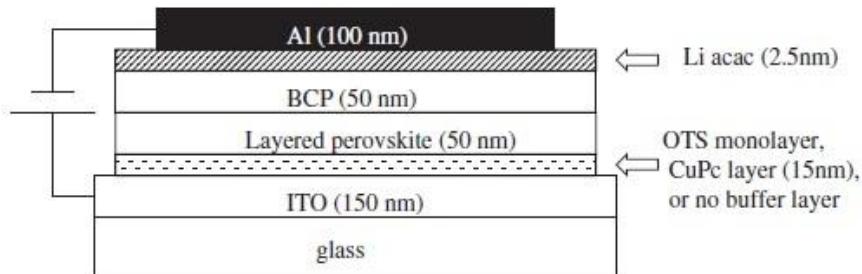


Fig. 2.24 Schematic of the LED made used PAPI films. PAPI was spin coated on either an OTS coated glass substrate, a CuPc coated glass substrate, or a bare glass substrate. Reproduced from Ref. [99].

scintillators (e.g. NaI:TI, CsI:Na) have decay times of 200ns or more, whereas fast scintillators (e.g. BaF₂, CsF) have low light yields of less than 2000 photons per MeV [104]. In order to test its suitability, C₆PI crystals were bombarded by an ultra-short electron beam with pulse width of 1-2ps, generated by a 35MeV linear accelerator. It was found that

C_6PI had a decay constant of 45ps at room temperature [104]. The 3D extension of 2D perovskites, $CH_3NH_3PbBr_3$ was also tested, and the 2D perovskite had a faster decay time as quantum confinements means carrier wavefunctions overlap, giving higher likelihood of decay.

Shibuya *et al.* used different dosages of 2MeV protons to test $\approx 250nm$ C_6PI thin films [103]. Radiation-induced emission spectra showed that the exciton peak did not change position during bombardment, and no additional peaks appeared at any radiation dosage. Three decay components were found: one with decay constant 0.39 ns due to free exciton recombination, then two more with constants of 3.8 and 16 ns due to trapped excitons, with decay constants depending on defect density. Emission intensities of C_6PI excitons attenuated with increased radiation, but the radiation hardness was still enough for practical use. From these results, C_6PI would be a good candidate for scintillator use due to the stability of excitons at room temperature, ease of processing, fast response due to short exciton lifetime, no shift in spectrum due to radiation, and inclusion of high atomic number element of Pb in order to detect low linear energy transfer radiation such as X-rays [105].

2.4 Conclusions

Excitons, bound hydrogen atom-like systems of electrons and holes, produce strong optical signatures in semiconductors. Such effects can be enhanced as a result of a reduction in dimensionality. PbI perovskites are naturally self-assembling materials that create a MQW structure, and whose optical properties are dominated by the excitons produced in the inorganic layers even at room temperature. The great flexibility of the perovskite structure, as well as its processability, provide tunability for use of the material in device applications.

Chapter 3

Plasmonic nanostructures

3.1 Introduction

Surface plasmons are collective electron oscillations in a metal. The form of such oscillations can be found by solving Maxwell's equations, and depends on the geometry of the problem. Such movement of electrons causes large field enhancement, and the resonance frequency is very sensitive to the dielectric environment around the metal. In this Chapter we will solve Maxwell's equations to find the form of plasmon oscillations at a metal-dielectric interface, before exploring how the solution is affected by modifying the metal surface to form a grating. Finally we will explore plasmon oscillations in small metal islands.

3.2 Surface plasmon polaritons

In order to describe the behaviour of electrons in a metal, we can use Maxwell's equations to describe the behaviour of electromagnetic fields:

$$\nabla \cdot \vec{D} = \rho \quad (3.1a)$$

$$\nabla \cdot \vec{B} = 0 \quad (3.1b)$$

$$\nabla \times \vec{E} = -\frac{\partial \vec{B}}{\partial t} \quad (3.1c)$$

$$\nabla \times \vec{H} = \vec{J} + \frac{\partial \vec{D}}{\partial t}. \quad (3.1d)$$

These equations link the fields \vec{D} (dielectric displacement), \vec{E} (electric field), \vec{H} (magnetic field) and \vec{B} (magnetic flux density). The electric charge density is given by ρ , and electric

current density by \vec{J} . For linear, isotropic and non-magnetic materials we have the relationships

$$\vec{D} = \epsilon \epsilon_0 \vec{E} \quad (3.2a)$$

$$\vec{B} = \mu \mu_0 \vec{H}, \quad (3.2b)$$

where ϵ_0, μ_0 are the permittivity and permeability of free space respectively, and ϵ, μ are the relative permittivity and permeability of the material in question. In the case of a non-magnetic medium $\mu = 1$, and the refractive index n is given by $\sqrt{\epsilon}$.

By combining Eqs. (3.1c, 3.1d) and assuming a harmonic time dependence to the electric field with frequency ω such that $\vec{E}(\vec{r}, t) = \vec{E}(\vec{r})e^{-i\omega t}$, we find the Helmholtz equation

$$\nabla^2 \vec{E} + k_0^2 \epsilon \vec{E} = 0, \quad (3.3)$$

where $k_0 = \frac{\omega}{c}$ is the wavevector of the wave in vacuum.

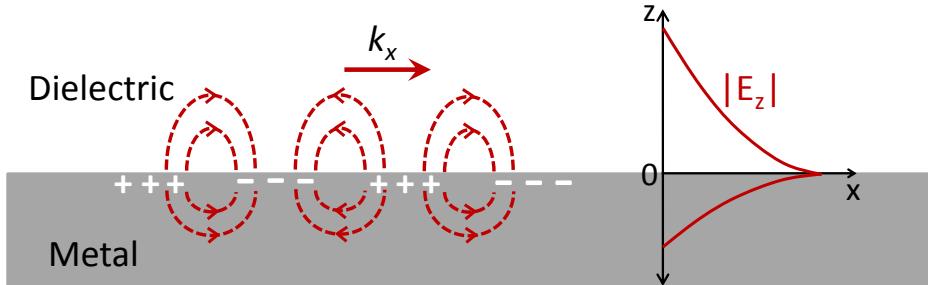


Fig. 3.1 Schematic for SPP oscillations at a metal-dielectric surface, showing the evanescent nature of such plasmons.

Using the geometry of a metal-dielectric interface shown in Fig. 3.1, we look for solutions of waves propagation in the x direction but confined to the interface with evanescent decay in the z direction, such that $\vec{E}(x, y, z, t) = \vec{E}(z) e^{ik_x x}$ where k_x is the propagation constant of the wave. We find one set of solutions that is transverse electric (TE) polarised, with the \vec{E} -field component perpendicular to the direction of travel:

$$H_x = i \frac{1}{\omega \mu_0} \frac{\partial E_y}{\partial z} \quad (3.4a)$$

$$H_z = \frac{k_x}{\omega \mu_0} E_y \quad (3.4b)$$

$$\frac{\partial^2 E_y}{\partial z^2} + (k_0^2 \epsilon_i - k_x^2) E_y = 0. \quad (3.4c)$$

For both solution the subscript i refers to the material the wave is travelling in, d or m for the dielectric and metal respectively. The solutions are waves of the form $e^{ik_xx}e^{k_iz}$, i. e. a wave travelling in x direction that evanescently decays in the z direction. Applying the boundary conditions of E_y and H_x continuity across the metal-dielectric interface, we find the condition

$$A(k_m + k_d) = 0, \quad (3.4)$$

where A is the amplitude of wave in the metal halfspace. Since confinement requires $\text{Re}(k_m, k_d) > 0$, Eq. 3.4 is only fulfilled if $A = 0$ when no wave can be sustained in the metal. The boundary condition indicate similarly the amplitude of the wave must be 0 in the dielectric as well, thus no SPPs exist in TE polarisation.

The transverse magnetic (TM) polarised solution, with the \vec{H} -field component perpendicular to the direction of travel, is:

$$E_x = -i \frac{1}{\omega \epsilon_i \epsilon_0} \frac{\partial H_y}{\partial z} \quad (3.5a)$$

$$E_z = -\frac{k_x}{\omega \epsilon_i \epsilon_0} H_y \quad (3.5b)$$

$$\frac{\partial^2 H_y}{\partial z^2} + (k_0^2 \epsilon_i - k_x^2) H_y = 0. \quad (3.5c)$$

Here continuity of H_y and $\epsilon_i E_z$ across the interface requires

$$\frac{k_d}{k_m} = -\frac{\epsilon_2}{\epsilon_1}, \quad (3.4)$$

so we require $\text{Re}[\epsilon_1]$ and ϵ_2 to be of opposite signs, thus SPPs can only be sustained on a metal-insulator interface. Fulfillment of Eq. 3.5c leads to

$$k_m^2 = k_x^2 - k_0^2 \epsilon_m \quad (3.5a)$$

$$k_d^2 = k_x^2 - k_0^2 \epsilon_d. \quad (3.5b)$$

Combining this with Eq. 3.4 produces

$$k_x = k_0 \sqrt{\frac{\epsilon_m \epsilon_d}{\epsilon_m + \epsilon_d}}, \quad (3.6)$$

the dispersion relation of an SPP on a metal-dielectric interface. Fig. 3.2(a) shows the calculated dispersion using a free electron gas model for the metal-air SPP, with values taken from Ref. [106]). In this model there is no damping in the metal, i. e. $\text{Im}(\epsilon_m) = 0$. Note three

regions in the graph: at high frequencies above the plasma frequency of the electron gas ω_p we have the transparent region (k_x, k_i real) where radiation can penetrate into the metal and excite plasmon polaritons. At low frequencies we have bound surface modes (k_x real, k_i imaginary), and at large k_x the frequency tends to $\omega_{sp} = \frac{\omega_p}{1+\epsilon_d}$. In this limit the plasmons are stationary surface plasmons, and $\epsilon_m + \epsilon_d = 0$. For $\omega_{sp} < \omega < \omega_p$ no propagating modes exist (k_x imaginary). If damping is included in the metal model, then quasi-bound modes can exist in this intermediate region [Fig. 3.2(b), calculated using values for Ag from ?]. Note also damping introduced a finite maximum k_x for the SPP, leading to a finite propagation length $L_x = \frac{1}{2\text{Im}(k_x)}$, typically on the order 10-100 μm at visible wavelengths for Au/Ag [107]. The limit in k_x also produces an upper limit in k_i [Eq. 3.5], thus limiting the skin depth $L_z = \frac{1}{\text{Im}(k_i)}$ to order 10 nm in metals.

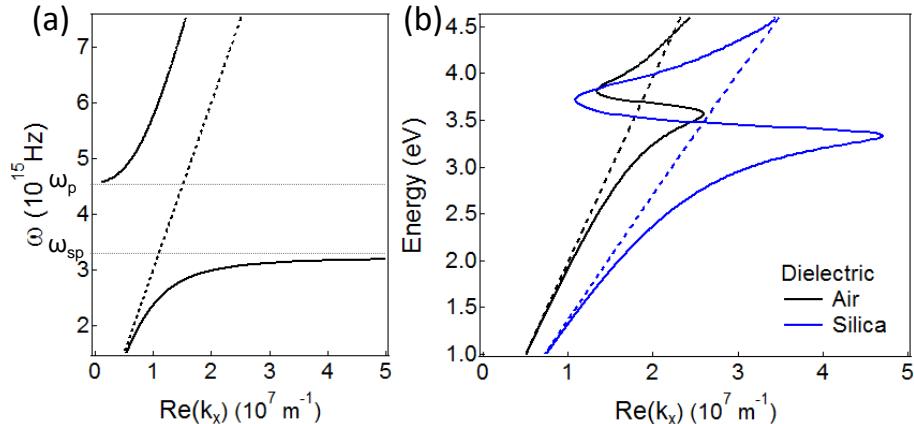


Fig. 3.2 Dispersion of SPPs on metal-dielectric interface for (a) free electron gas and (b) silver (solid lines), and light lines in the dielectric (dashed lines).

3.3 Plasmonic gratings

From Fig. 3.2 we can see that the SPP dispersion always lies on the right of the light line in the dielectric, and this momentum mismatch means that it is not possible to directly excite SPPs using photons. Instead we must use a phase matching technique, for example prism or grating coupling. In prism coupling a three-layer system is employed, so that photons reflected at the higher refractive index interface has sufficient momentum to excite SPPs on the interface between the metal and the lower refractive index material. In grating coupling, due to its periodicity D the grating structure can provide momenta in units of $G_m = \frac{2\pi}{D}$, thereby allowing photons to couple to SPPs. However other types of modes can be observed

in the optical spectra of plasmonic gratings.

3.3.1 First order modes

Using Huygens' construction and considering each point on the grating as a wave scatterer, we reach the well-known grating equation for constructive interference

$$D(\sin \alpha - \sin \beta) = l\lambda, \quad (3.7)$$

where α is the angle of incidence and β the diffracted angle with respect to the grating normal, λ is the wavelength of light and l the order of the diffracted light [Fig. 3.3(a)]. From here we only consider the zeroth diffraction order (i. e. specular reflection) with incidence angle θ and azimuthal angle ϕ . In the order approximation, where we assume different types of diffracted light do not interact, we can distinguish between two types of gratings modes: ‘photonic’ modes caused purely by interference of light due to the periodic structure, and ‘plasmonic’ modes where SPPs are excited on the surface of the grating. We can find the dispersion of such grating modes by considering momentum and energy conservation of incoming/outgoing photons, and thus find

$$k_m = k_i^2 \sin^2 \theta + G_m^2 \pm 2k_i G_m \sin \theta \cos \phi,$$

here the label m indicates the order of grating vector G_m needed for matching. Here the wavevector of the incident light

$$k_i = \frac{\omega}{c} \sqrt{\epsilon_d}, \quad (3.9a)$$

for photons

$$k_m = \frac{\omega}{c}, \quad (3.9b)$$

and for SPPs

$$k_m = \frac{\omega}{c} \sqrt{\frac{\epsilon_m \epsilon_d}{\epsilon_m + \epsilon_d}}. \quad (3.9c)$$

In this case the grating can be thought of as a 1D photonic grating, where the photon/SPP dispersions are displaced by multiples of the grating vector. Although in the first order photonic and plasmonic modes do not interact with each other, SPP modes can strongly couple and create anticrossings in the spectra [108].

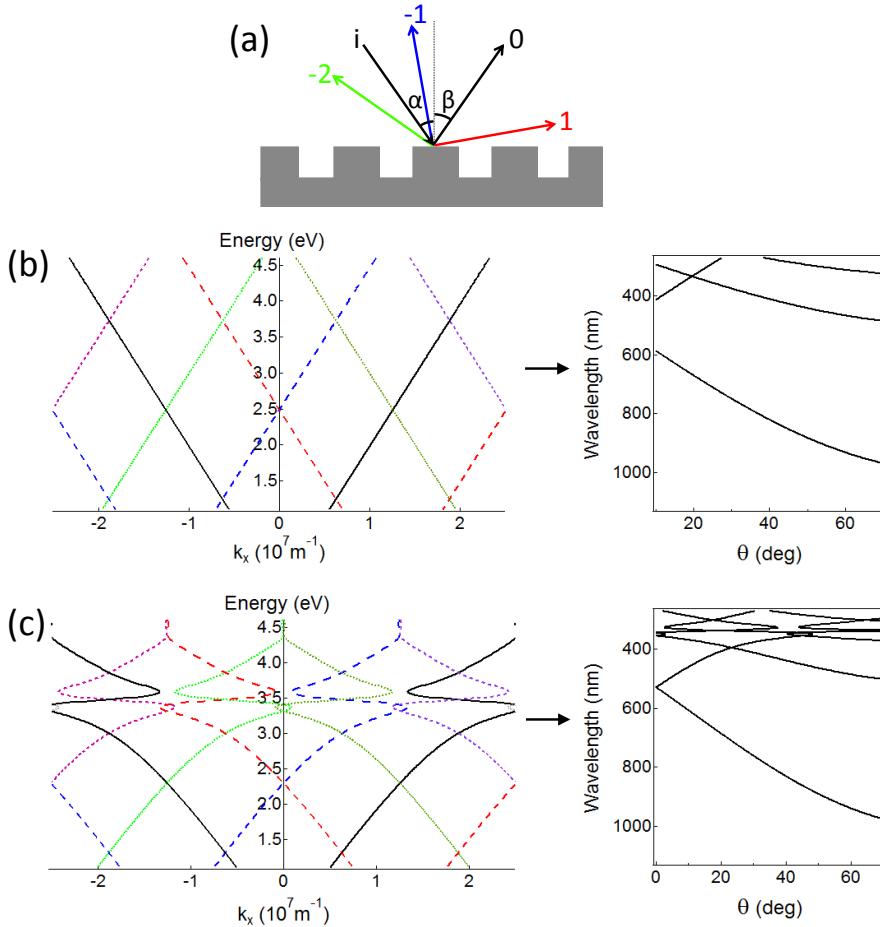


Fig. 3.3 (a) Diffraction from a 1D grating structure, showing the incident light and diffracted orders. Dispersion (left) and modes as a function of incidence angle θ (right) for (b) photonic and (c) plasmonic modes for specularly reflected zeroth order mode for a $D = 500$ nm grating.

3.3.2 Grating anomalies

Anomalies are sharp changes in the response of a grating, and first observed by Wood, who noted “that under certain conditions the drop from maximum illumination to maximum...occurred within a range of wave-lengths not greater than the distance between sodium lines” [109]. Anomalies are second order effects, where different diffracted modes interact with each other. The effect is named after Wood, and in plasmonic gratings occurs as a sharp increase in intensity followed by a dip [Fig. 3.4]. Although overall the Wood’s anomaly is one physical phenomenon, it is possible to separate into two parts for a more intuitive understanding: the threshold (sharp change in intensity) and resonance (dip) anomalies.

The threshold anomaly is a photonic effect, and comes about when an order is diffracted

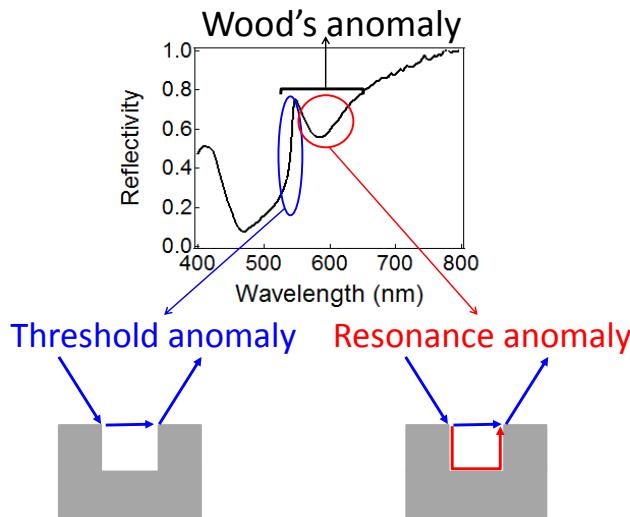


Fig. 3.4 Wood's anomaly in the reflectivity of Ag grating, $D = 417$ nm. Schematics of the threshold and resonance anomalies are included.

along the surface of the grating ($\beta = 90^\circ$). If an order becomes evanescent then the energy available will be redistributed to the other diffractive orders, thus this ‘passing’ order on the edge between diffraction and evanescence leads to a sharp change in the diffraction intensity. These effects are still due to interference from grating structure, and thus occur at the predicted first order photonic modes [Eq. 3.7]. The threshold anomaly can be observed in both polarisation, but generally the strength of the anomaly (i. e. the change from maximum to minimum intensity) is smaller in TE polarisation, particularly in metals, as the E -field is parallel to the grating lines in this case and thus cannot be sustained. The reduction in field intensity leads to less energy to redistribute, hence the effect is less pronounced.

The resonance anomaly is a plasmonic effect, and comes from the excitation of surface waves on the grating. The excited SPPs can interact with the diffractive light, and the addition of the SPP oscillator and background photonic diffracton leads to a Fano resonance, giving the typical peak-trough feature seen. Clearly the resonance anomaly can only be observed when SPPs are excited, e. g. for $\phi = 0$ in TM polarisation, but not TE. Unlike the threshold anomaly, the position and linewidth of the resonance anomaly dip cannot be predicted by Eq. 3.7 as it depends on the nature of the SPP, and is thus sensitive to the geometry of the grating and any surface roughness in the material [Fig. 3.5]. However it is possible to calculate the position of the resonance anomaly by using electromagnetic theory to model the grating (see Refs. [110, 111] and references therein).

Grating anomalies have been observed in both reflection and transmission gratings, and are responsible for the extraordinary transmission observed for some 1D and 2D arrays

[112–118]. Overcoating of a metallic grating affects the strengths and positions of anomalies, in particular the resonance anomaly, where the SPP is very sensitive to the local dielectric environment. In addition, light can excite modes in the coating material, and thus change the direction of the E -field so that anomalies can be excited when they’re not predicted to, e.g. $\phi = 0^\circ$ in TE polarisation [110, 111].

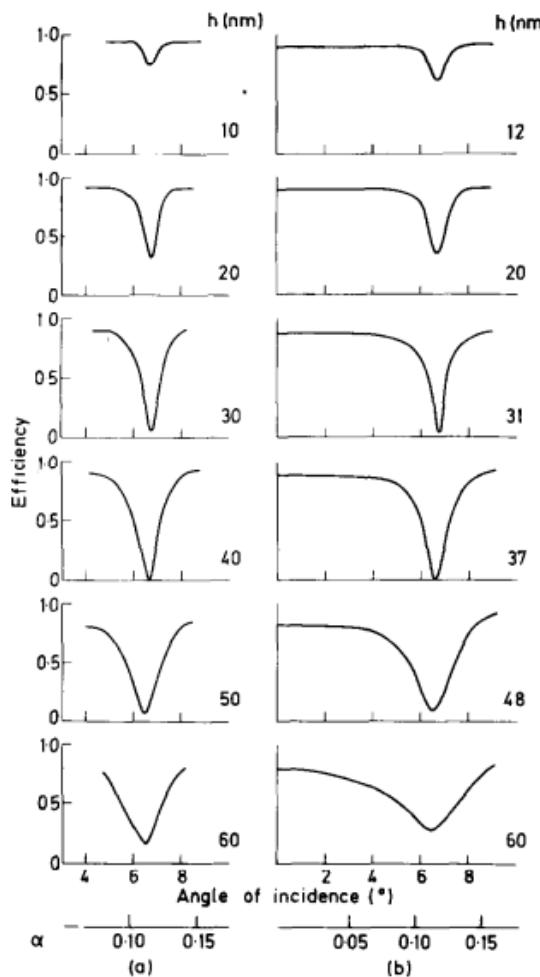


Fig. 3.5 Dependence of resonance anomaly on the grating depth h for sinusoidal grating, $D = 556$ nm at $\lambda = 647$ nm from (a) electromagnetic theory and (b) experiment. Reproduced from Ref. [119].

3.3.3 Localised and guided modes

Gratings can also give rise to optical modes that do not rely on the periodicity of the structure. In particular the grating slits can be thought of as independent, sustaining modes that

do not couple to modes in nearby slits. We can think of the slits as hollow electromagnetic waveguides, and thus support modes that fit the equation [120] [CHECK JACKSON]

$$\frac{\omega^2}{c^2}(n_{eff}^2 - \sin^2 \theta) = \pi^2 \left(\frac{\mu^2}{a_{eff}^2} + \frac{v^2}{b_{eff}^2} \right), \quad (3.10)$$

where n_{eff} is the effective refractive index experienced by the mode in the grating slit, a_{eff} is the effective width and b_{eff} half the effective height for the mode, and μ, v are indices used to label the waveguide mode. Due to the penetration depth of the electromagnetic field in metals, a_{eff} is not the same as the geometric width of the grating slit, and particularly if there is a coating on the metal then a_{eff} and b_{eff} both have some dependence on n_{eff} .

Surface plasmons travelling down the grating slits can interact, particularly if the slit is narrow. In this case we can form symmetric and antisymmetric combinations of the SPP oscillations, essentially forming a metal/insulator/metal waveguide [107]. For rectangular shaped slits, often known as trench waveguide, the highest E -field intensity is often at the top corners of the slits [Fig. 3.6(a)] and thus more extended outside the groove [121–123]. If the groove is V-shaped, the change in effective refractive index due to the change in width leads to reflections, and the highest E -field is localised at the bottom of the grooves [Fig. 3.6(a)] and such oscillations are called channel plasmon polaritons [121, 122, 124? , 125]. Due to their localised nature, the dispersion of such modes is often flat with respect to the angle of incidence θ . Such channel plasmon polaritons can be observed in near-field optical microscopy [Fig. 3.6(b)], and show interference with the light scattered by the structure.

3.4 Localised surface plasmons

3.4.1 Quasi-static approximation

For spherical nanoparticles (NPs), if the diameter $d \ll$, the phase of the E -field is approximately constant across the particle, and we can solve the simplified problem of a sphere in an electrostatic field and include the harmonic time dependence of the field at the end. The geometry is shown in Fig. 3.7(a), with a homogeneous metal particle of diameter d at the origin inside a dielectric medium, and $\vec{E}_0 = E_0 \vec{z}$. Solving the Laplace equation for the

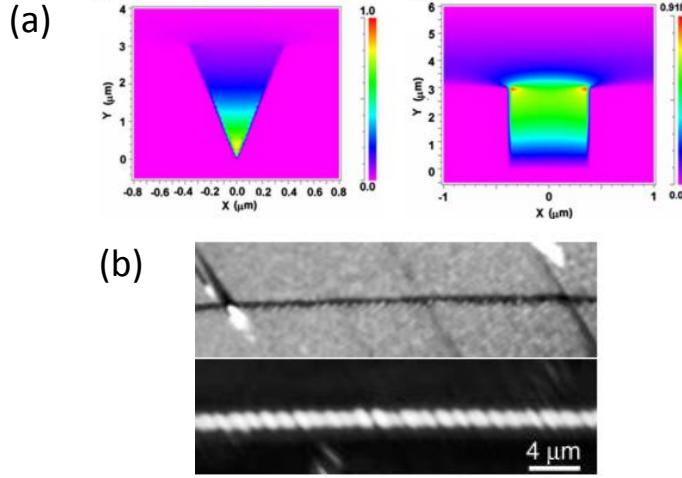


Fig. 3.6 (a) E -field profiles of channel plasmon polariton modes in a V-shaped (left) and trench (right) Au groove, with width $3.75 \mu\text{m}$ and depth $3 \mu\text{m}$ filled with air. (b) Topographical (top) and near-field optical (bottom) images for V-shaped Au slit with width $0.6 \mu\text{m}$ and depth $1 \mu\text{m}$ at $\lambda = 1440 \text{ nm}$. Reproduced from Refs. [121, 122].

potential Φ ($\vec{E} = -\nabla\Phi$), we find

$$\Phi_{in} = -\frac{3\epsilon_d}{\epsilon_m + 2\epsilon_d} E_0 r \cos \theta \quad (3.11\text{a})$$

$$\Phi_{out} = -E_0 r \cos \theta + \frac{\epsilon_m - \epsilon_d}{\epsilon_m + 2\epsilon_d} E_0 \left(\frac{d}{2}\right)^3 \frac{\cos \theta}{r^2} \quad (3.11\text{b})$$

at a distance r from the centre of the sphere, where Φ_{in} and Φ_{out} represent the potentials inside and outside the sphere respectively. Note Eq. 3.11b appears to be the superposition of the applied field E_0 and that of a dipole located at the origin. In this case we can see that the applied field induces a dipole moment \vec{p} in the NP, and also introduce the polarisability α such that

$$\vec{p} = 4\pi\epsilon_0\epsilon_d \left(\frac{d}{2}\right)^3 \frac{\epsilon_m - \epsilon_d}{\epsilon_m + 2\epsilon_d} \vec{E}_0 \quad (3.12\text{a})$$

$$\alpha = 4\pi \left(\frac{d}{2}\right)^3 \frac{\epsilon_m - \epsilon_d}{\epsilon_m + 2\epsilon_d}. \quad (3.12\text{b})$$

A resonance in α is achieved when

$$\epsilon_m(\omega) = -2\epsilon_d(\omega), \quad (3.13)$$

known as the Frölich condition, and provides the resonance frequency of the dipole LSP for a metallic NP. For a free electron gas in air, this condition is achieved at $\omega = \frac{\omega_0}{\sqrt{3}}$, but as can be seen from Eq. 3.13 the position is very sensitive to the dielectric environment around the particle. If harmonic time independence is included, we can see the resonance is caused by an oscillation of electrons in the NP, and as such causes a large field enhancement around the vicinity of the particle [Fig. 3.7(b)]. The LSP resonance also depends on the dielectric function of the metal, and for the same d the Ag resonance is always higher in energy than Au [Fig. 3.7(c)]. Note also the asymmetric shape of the Au Np extinction, which is due to the onset of interband transitions in Au. For arrays of NPs, the resonance frequency in optical spectra is the same as single NPs if the particles are separated by more than $\approx 2d$ [126].

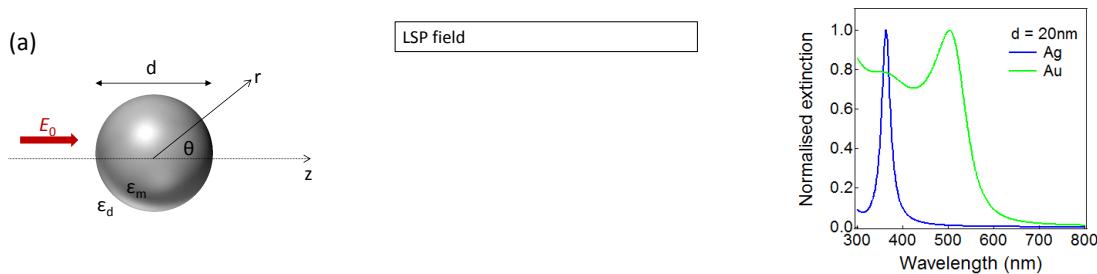


Fig. 3.7 (a) Geometry of homogeneous metal sphere with diameter d placed inside an electrostatic E -field. (b) E -field profile of dipole LSP. (c) Normalised extinction spectra for $d = 20$ nm Ag/Au NPs in air.

The oscillating NP dipole leads to radiation, and can be seen as the scattering of light from the NP. The maximum in polarisability also leads to a resonant enhancement of the scattering C_{scat} and absorption C_{abs} cross section of the particle

$$C_{scat} = \frac{k^4}{6\pi} |\alpha|^2 = \frac{8\pi}{3} k^4 \left(\frac{d}{2}\right)^6 \left| \frac{\epsilon_m - \epsilon_d}{\epsilon_m + 2\epsilon_d} \right|^2 \quad (3.14a)$$

$$C_{abs} = k \text{Im}[\alpha] = 4 \left(\frac{d}{2}\right)^3 \text{Im} \left[\frac{\epsilon_m - \epsilon_d}{\epsilon_m + 2\epsilon_d} \right], \quad (3.14b)$$

and we define extinction $C_{ext} = C_{scat} + C_{abs}$. For very small NPs absorption dominates over scattering due to its d^3 dependence, for example for $d = 30$ nm Ag NPs the extinction is almost entirely due to absorption [Fig. 3.8], while the reverse is true for $d = 90$ nm.

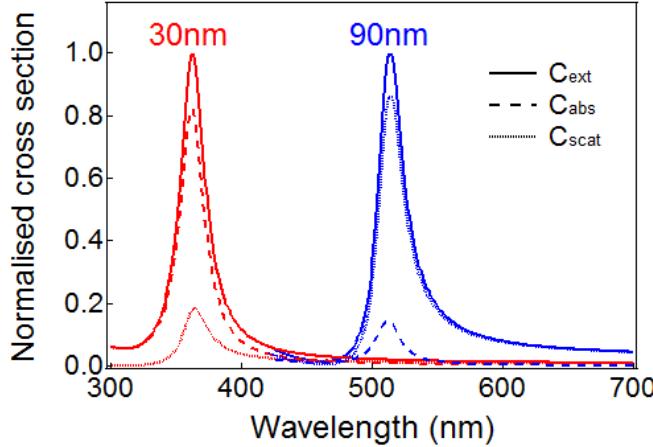


Fig. 3.8 Absorption (dashed lines), scattering (dotted lines) and extinction (solid lines) cross sections for Ag NPs $d = 30$ and 90 nm at $\lambda = 500$ nm in air according to Eq. 3.14, normalised so that the peak extinction is 1. The $d = 90$ nm data has been shifted for clarity.

3.4.2 Size and shape effects

The quasi-static approximation models the NP as an electric dipole whose resonance frequency depends purely on the relative dielectric functions of the metal and surrounding medium. The underlying assumption that the phase of the E -field across the particle is constant is only true for very small particles, and works well for $d < 50$ nm Ag particles [Fig. 3.9]. For larger particles an electrodynamic must be used, for example Mie theory [107, 127] [CHECK BORN WOLF], where the E -field waves are expanded into a superposition of partial waves (normal modes) that are spherical harmonics/radial components of Hertz vectors, and boundary conditions are used to find the absorption and scattering coefficients required. The normal modes are used in order to solve Maxwell's equations for a sphere interacting with an incoming plane wave. This method produces the redshift in LSP dipole resonance with d as seen in experiment [Fig. 3.9], and also shows the emergence of higher order multipole modes for large particles. For example in Ag particles, a quadrupole mode can be excited and is first observed as a shoulder in the extinction spectrum for $d = 90$ nm.

The resonance of NPs is also very sensitive to the shape of the particle, and both quasi-static and Mie theory calculations can be adapted to the boundary conditions of non-spherical geometry. This provides great tunability to the LSP resonances via controlling of particle shape. Deviations from spherical geometry leads to the production of multiple redshifted peaks [Fig. 3.10(a)]. In the case of nanorods, we observe two LSP resonance in the spectra for unpolarised light: a transverse mode associated with electron oscillations along

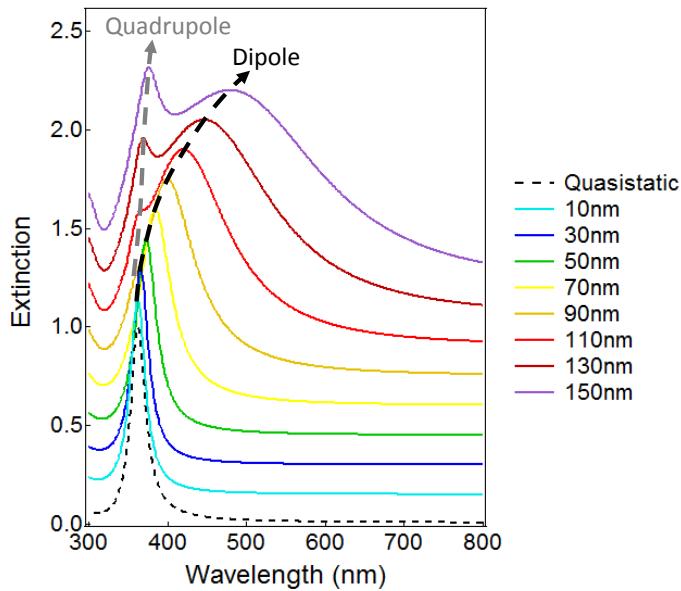


Fig. 3.9 Extinction for NPs with labelled d in air using Mie theory, normalised to the dipole maximum and offset for clarity. The quasi-static approximation resonance is added for comparison.

the short axis, and a longitudinal mode related to electron oscillations along the long axis, the resonance of which depends on the aspect ratio of the nanorod [Fig. 3.10(b)]. Therefore the longitudinal resonance can be tuned through a large range of the electromagnetic spectrum via the growth and assembly of such nanorods [128–130].

3.5 Conclusions

Surface plasmons are collective oscillations of electrons in a metal. Such oscillations show resonant in many geometries, either guided in 2D structure such as planar metal films, or localised 0D nanoparticles. The resonance frequency depends on the relative dielectric functions of the metal and surrounding dielectric, and is very sensitive to the geometry of the nanostructure. Surface plasmons cause large electric field enhancement around the vicinity of the metal, and can thus be used as a sensor. In periodic nanostructures, zone-folding allows incoming/outgoing light to reach parts of the plasmonic dispersion that may not be accessible otherwise, and the interaction between electrons and photons gives rise to many possible modes.

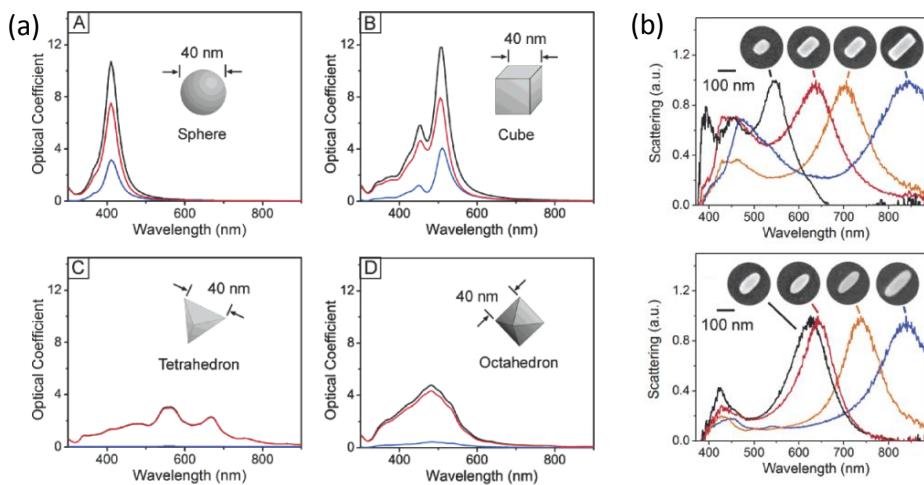


Fig. 3.10 (a) Discrete dipole approximation calculations of the extinction (black), absorption (red) and scattering (blue) of Ag nanoparticles with the geometries shown. (b) SEM images and normalised scattering spectra of individual Ag nanobar (top) nanorice (bottom) structures. Reproduced from Refs. [128, 129].

Chapter 4

Thin films of lead iodide perovskites

4.1 Introduction

Spin coating is a simple process which can be used to deposit thin films onto relatively flat substrates. After the solution has been deposited (usually in excess), the substrate is accelerated to its desired final spin speed and continues to be rotated until a solid film has formed. The process commonly used industrially as it can be used to controllably produce films from nm to μm thickness, covering areas with diameter $\sim 10\text{ cm}$. The thickness and morphology of films can be controlled using parameters such as solution concentration, spin speed and substrate preparation. The material chosen needs to self-assemble to form the correct structure as the solvent evaporates, and thus is suitable for the creation of thin films of PbI perovskites. In this Chapter we will explore the formation of C12PI and CHPI thin films under a variety of experimental conditions.

4.2 Spin coating theory

Although spin coating is experimentally simple, it is a complicated process to model due to the large number of factors involved. Initially, fluid inertia and surface tension are important as the fluid front spreads out in spiral waves. Evaporation also begins at this point, and a small boundary layer is formed at the liquid-gas interface. After the end of this step, a thin and even film has formed on the substrate, and fluid inertia is no longer important. In the next phase thinning is mainly because of fluid flow due to a balance between viscous and centrifugal forces, and the boundary layer in the solution gradually gets thicker. The solute concentration begins to vary throughout the film thickness, and viscosity rises, inhibiting

further flow. Any further fluid loss is then due to solvent evaporation, which dominates thinning in the latter stages, and eventually solute concentration becomes uniform throughout the film. There is also a small atmospheric boundary layer above the solution that can influence heat and mass transfer, as well as exert shear forces at the interface [131–133]. The initial acceleration may also affect final film thickness, as too slow an acceleration can lead to complete solvent evaporation before the final spin speed is reached [134].

Meyerhofer used a model where the solvent evaporation was negligible until the mass loss due to rotational forces fell to the level of the evaporation rate, and calculated that

$$h_f \sim C_0(1 - C_0)^{-1/3} \omega^{-2/3} \left(\frac{\eta_0}{\rho_0} \right)^{1/3} e^{-1/3}, \quad (4.1)$$

where h_f is the final film thickness, C_0 the initial solution concentration, ω the spin speed, η_0 the initial viscosity, ρ_0 the initial solution density, and e the evaporation rate [131]. van Hardeveld used a similar model, except the evaporation was modelled more rigorously in terms of rate of mass transfer at the interface, and calculated that the amount of material deposited was

$$m = C_0 \sqrt[3]{\frac{3\eta e}{2\rho\omega^2}}, \quad (4.2)$$

where m is the amount of material deposited in the final film [132]. The rate determining step for solvent evaporation is found to be mass transfer in the vapour phase, when it is expected that the evaporation rate is proportional to $\omega^{1/2}$. This relationship was verified experimentally by both Meyerhofer and van Hardeveld [131, 132], therefore overall $h_f \sim \omega^{-1/2}$. The dependence of the final film thickness on initial solution concentration is more complicated as η_0 and ρ_0 will also depend on concentration. Lawrence's model considered both the solvent and atmospheric boundary layers, and found that

$$h_f \sim C_0 \left(\frac{\eta_0}{\rho_0} D_0 \right)^{1/4} \omega^{-1/2}, \quad (4.3)$$

where D_0 is the initial solvent diffusivity [133].

Taking into account evaporation, all three models above agree that $h_f \sim \omega^{-1/2}$, but experimentally other values have been found (for example -0.68 and -0.8, see Ref. [133]). Lawrence indicated that an exponent of ω less than -0.5 could be due to films that had not completed the full spinning process, and was thus thicker than the calculations indicated. Shear thinning, where the viscosity of the solution decreases with an increased shear stress, may also be responsible [133].

4.3 Experimental methods

Spin coating solutions are prepared by dissolving the appropriate lead iodide perovskite powder in tetrahydrofuran (THF) with a concentration of 20mg/ml. Silica substrates are sonicated in four separate solvents for approximately 15 minutes each: firstly a deionised water and detergent solution, then deionised water, followed by acetone, and finally isopropanol. Three other substrate preparation techniques are then investigated in order to create the best possible film. (1) CO₂ snowjetting, where a high velocity mix of gaseous and solid carbon dioxide was focused on the substrate, cleaning it due to momentum transfer and solvent action of CO₂ [?]. (2) Silanisation, where substrates are dipped in a 2 vol% solution of aminopropyltriethoxy silane (APTES) in dry acetone for approximately 90 minutes. A self-assembled monolayer of silane molecules forms on the substrate, and in the case of APTES the surface is functionalised with amine groups. (3) Plasma etching, where substrates are treated using a Diener Electronic Femto plasma system for 5 minutes, using an oxygen plasma to react with and clean contaminants on the substrate. The surface is then functionalised with hydroxyl groups and thus becomes more hydrophilic. Films are then characterised using optical microscopy and spectroscopy. The film thickness is determined by AFM measured on the scratch in the film.

4.4 C12PI thin films

BF images at 100× magnification of C12PI films made using a variety of spin coating conditions are shown in Fig. 4.1. Films showed significant dewetting without functionalisation of the substrate due to the hydrophobic nature of the organic molecule [Fig. 4.1(a-f)], and for this reason no films were formed on plasma etched substrates. Fig. 4.1(d) does not exhibit such dewetting as the substrate was heated before the application of the C12PI solution, thus the solvent evaporate before fluid could be removed by the centrifugal force, forming the extremely non-uniform film observed. Surface functionalisation due to APTES improved coverage of C12PI films due to stronger interactions between C12PI constituents and the amine groups. However a further snowjet cleaning steps removed any excess APTES left over from the silanisation step, removing some surface roughness and produced the most uniform C12PI samples.

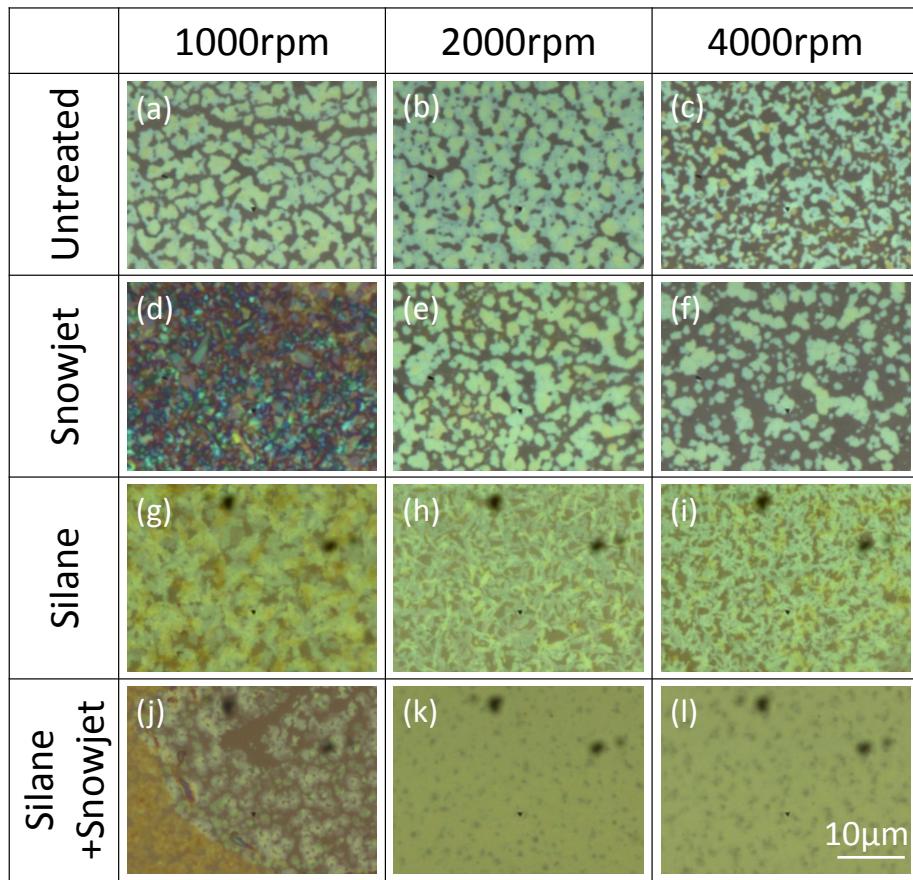


Fig. 4.1 BF images at $100\times$ magnification of spin-coated C12PI films, with spin speed and substrate preparation as labelled. Note that (d) was formed while the substrate was heated, creating the non-uniform film observed.

4.4.1 Spin speed

Optical spectra of C12PI films created on silanised substrates collected over a region with diameter $\sim 20\mu\text{m}$ is shown in Fig. 4.2, and illustrates the general trends seen for all substrate preparations. The extinction appears as a Fano resonance at $\approx 500\text{nm}$ in reflectivity spectra due to interference between its narrow resonance and the continuum background, while a dip appears in the transmittance spectra. Although both phases of C12PI are observed for films on silanised substrates below 2000 rpm, here we only consider areas exhibiting the high energy exciton. As spectra can be directly correlated to the images taken, we observe that the comparable morphologies of the films made above 1000 rpm [Fig. 4.1(g-i)] does indeed produce similar spectra. The film made at 500 rpm is rougher and less uniform in images, a fact reflected by the lowering of the overall reflectivity and Fano resonance shape. As C12PI is a multilayer system, the size of the exciton dip in trans-

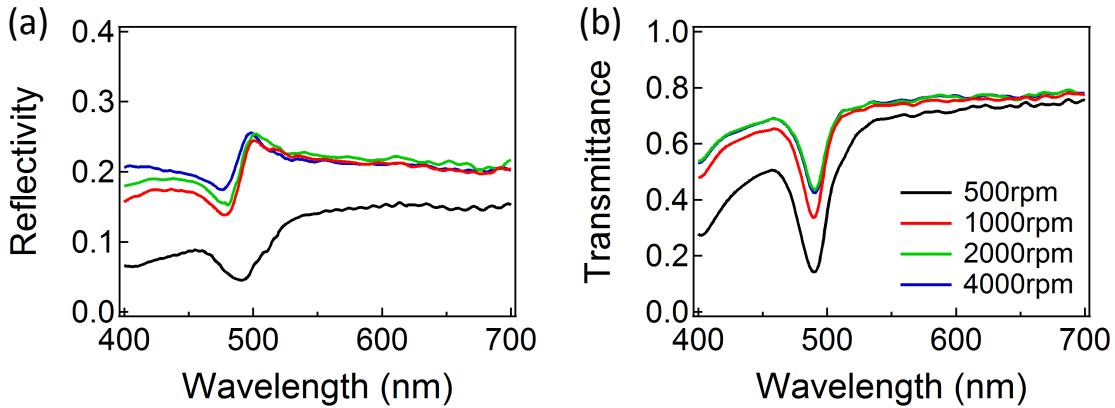


Fig. 4.2 (a) Reflection and (b) transmission spectra for spin coated films on a silanised glass substrate at $5\times$ magnification, collected over a $\sim 20\mu\text{m}$ diameter region.

mission spectra can be used as a gauge of the film thickness, and here we observe that the film thickness decreases with spin speed as expected.

4.4.2 Substrate preparation

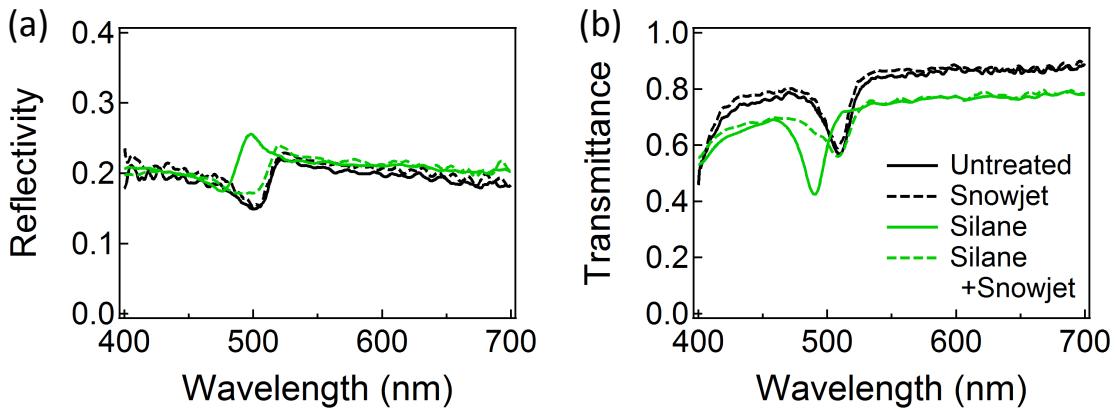


Fig. 4.3 (a) Reflection and (b) transmission spectra for 4000 rpm spin coated films using the labelled substrate preparation at $5\times$ magnification, collected over a $\sim 20\mu\text{m}$ diameter region.

Optical spectra of 4000 rpm C12PI films are shown in Fig. 4.3. The reflectivity spectra are very similar for all substrate preparations, with the exception of the silanised substrates where only the high energy phase was observed. This is most likely due to excess APTES molecules creating surface roughness, thus favouring the more crumpled C12PI phase. Removal of the excess silanes via snowjetting removes such roughness, so only the lower energy exciton is observed. Another difference is seen in the value of the transmittance at

large wavelengths, where improved C12PI film coverage due to silane molecules leads to the lowering of the background transmittance.

4.4.3 Sample degradation

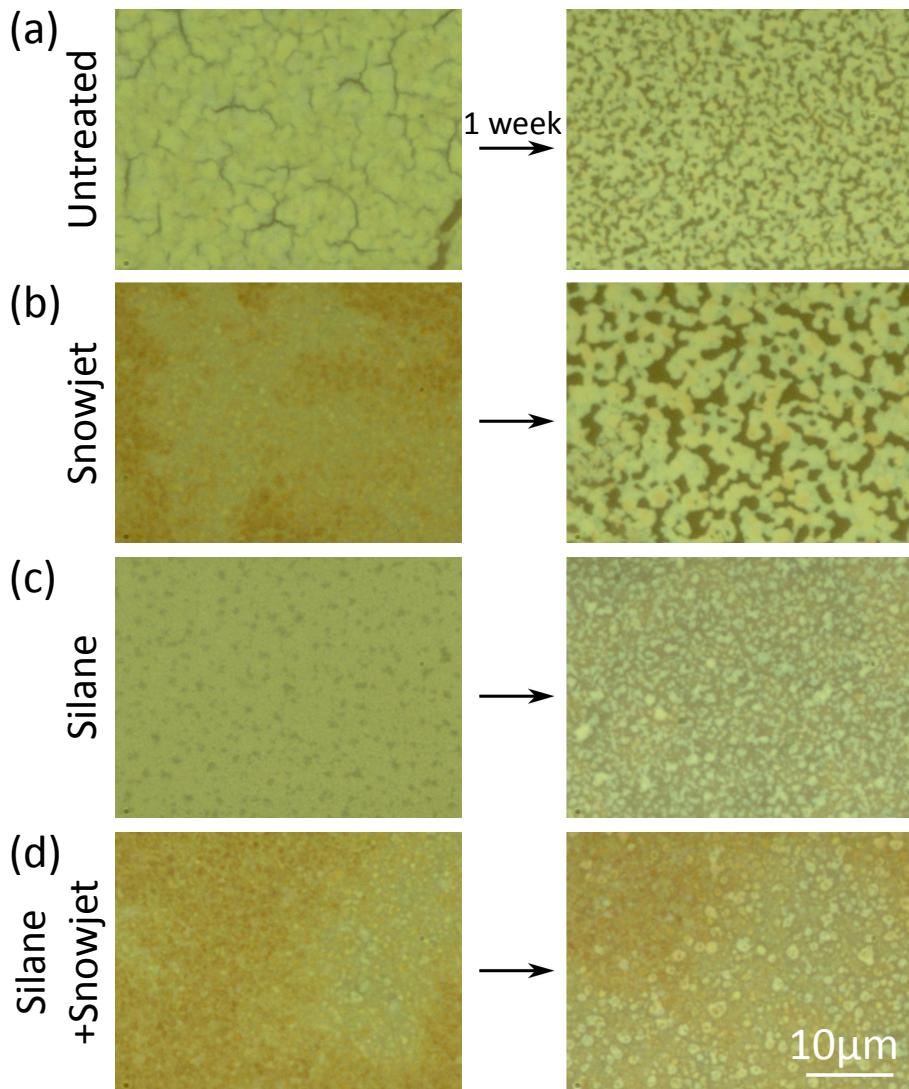


Fig. 4.4 Degradation of C12PI thin films shown in 100 \times magnification BF images. Figures on the right were taken one week after those on the left, during which time samples were placed in standard conditions. All samples were produced at 2000 rpm.

BF images at 100 \times magnification of 2000 rpm films as-made (left) and after week under standard conditions (right) are shown in Fig. 4.4. All films showed signs of dewetting or diffusion during this time, indicating the importance of placing C_nPI films in low humidity

atmospheres after formation, or capping with a polymer layer (e. g. Ref. [46]) to prevent sample degradation due to humidity in the atmosphere.

4.5 CHPI thin films

BF images at $100\times$ magnification of CHPI films made using a variety of spin coating conditions are shown in Fig. 4.5. No dewetting was observed with CHPI due to increased hydrophilicity of the organic molecule. Both snowketting and high spin speeds improved the uniformity of the samples. However, functionalisation of the substrate with APTES greatly improved the film quality, and in this case the improvement with spin speed was minimal. A similar effect was seen in plasma etched substrates.

4.5.1 Spin speed

Optical spectra of CHPI films created on silanised substrates collected over a region with diameter $\sim 20\mu\text{m}$ is shown in Fig. 4.6 for untreated and silanised substrates. For untreated substrates, a change in the shape of recorded spectra between 1000 and 2000 rpm indicate a change in morphology: at low spin speeds a lower reflectivity indicate increase film roughness, and the appearance of a second higher energy exciton can be seen, leading to an increase in the linewidth of the transmission dip. Such extra resonances have been observed for thick perovskite films, and is attributed to stacking faults, strain and structural misalignment in the structure [135]. In contrast, the spectra for silanised substrates are very similar, as expected from their BF images [Fig. 4.5(g-i)]. Although the amplitude of the exciton features change due to film thickness, the similarity in the spectra indicate similar film morphologies. Film thickness measurements for around 10 films made using untreated substrates [Fig. 4.6(e)] show an $\omega^{-0.45}$ dependence to the film thickness, similar to the $\omega^{-0.5}$ relationship predicted by theory.

4.5.2 Substrate preparation

Optical spectra of 4200 rpm CHPI films are shown in Fig. 4.7. The appearance of a second exciton due to structural misalignment is seen in both untreated and snowjetted films. However as indicated by BF images [Fig. 4.5(h,k,n,q)], the spectra and morphologies of films using other substrate preparations are very similar, and the uniformity is greatly improved by functionalisation of the substrate via silanisation or plasma etching. To test the uniformity, spectra of different regions on a 4000 rpm film made on untreated [Fig. 4.7(c)] and

silanised and snowjetted [Fig. 4.7(d)] substrates are taken, and the near-identical spectra for all areas tested on the functionalised substrate is proof of the uniformity observed in BF images.

4.5.3 Humidity

Hydrogen bonding between the organic and inorganic constituents holds the structure together, and as such the self-assembly can be disrupted by water in the atmosphere. Fig. 4.8(a) shows the morphology of a 4000 rpm CHPI film on an untreated substrate at 100 \times magnification in a low humidity atmosphere (October), and a continuous film is formed as expected. However using the same conditions in a high humidity atmosphere leads to dewetting of the film [Fig. 4.8(b)] despite the more hydrophilic organic group. Therefore the level of humidity in the spin coater is very important to film formation, and as much water should be removed from the environment as possible. In order to achieve this, the dehydration agent CaCl_2 is placed inside the spin coater for roughly an hour before film production, and the spin coater pumped with N_2 just before use. The resulting film [Fig. 4.8(c)] is extremely uniform without modification of substrate functionalisation, and produces similar films under all conditions.

4.6 Conclusions

Thin films of PbI perovskites with thickness $\sim 10\text{s nm}$ are produced reliably using spin coating. The morphology of such films depends on the organic molecule within the perovskite structure, and dewetted films are produced for more hydrophobic moieties. However film coverage and uniformity can be improved by functionalisation of the substrate surface using techniques such as silanisation. Higher spin speeds can also be used to produce more uniform films. The film thickness is controlled by the spin speed or initial solution concentration. Formation of the perovskite structure can also be disrupted by water in the atmosphere, and a dehydration agent should be placed in the spin coater to produce a controllably low humidity environments. The simplicity and adaptability of spin coating allows PbI perovskite thin films to be deposited onto nanostructured substrates.

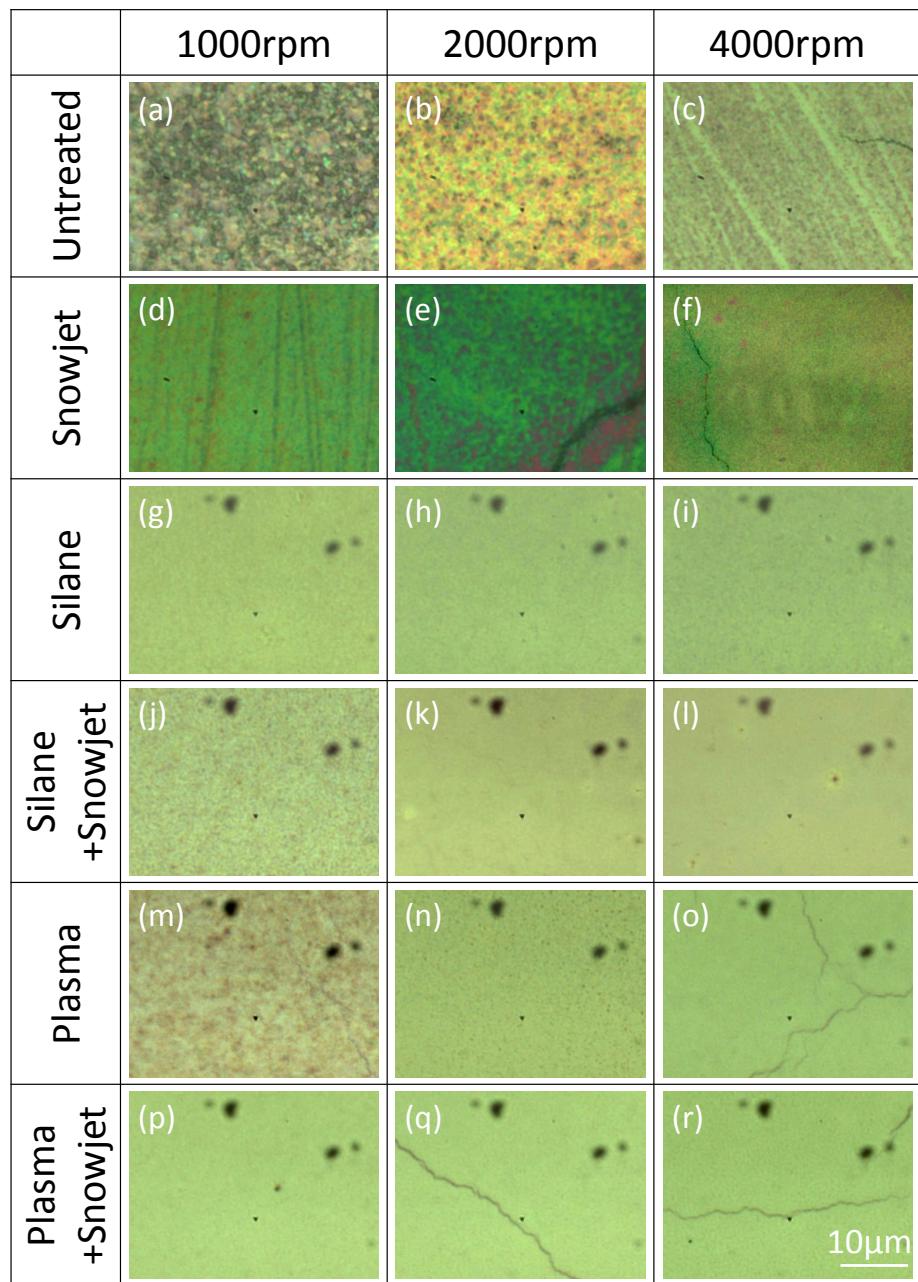


Fig. 4.5 BF images at 100× magnification of spin-coated CHPI films, with spin speed and substrate preparation as labelled.

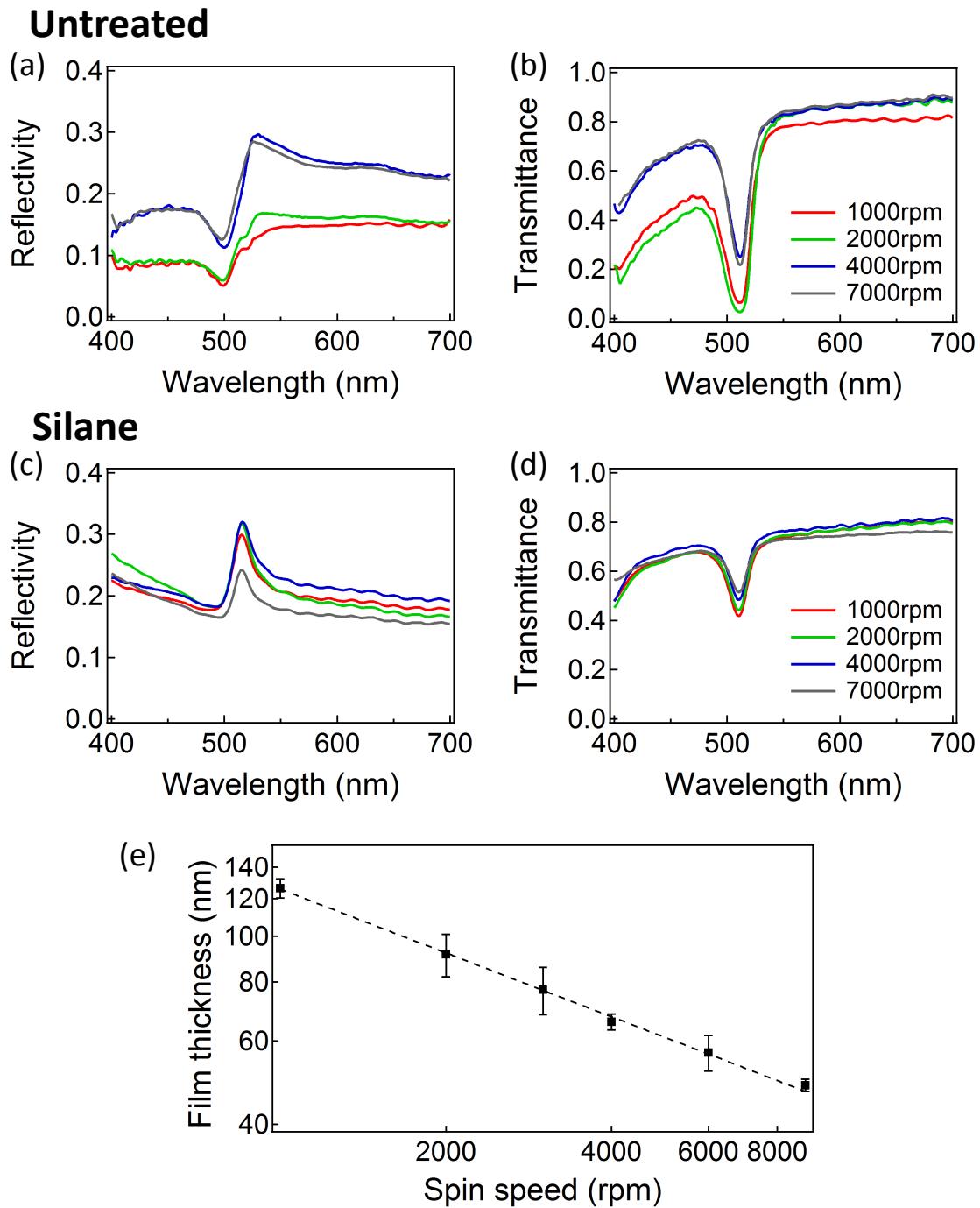


Fig. 4.6 Reflection and transmission spectra collected at $5\times$ magnification over a $\sim 20\ \mu\text{m}$ diameter region, on films prepared on (a,b) untreated and (c,d) silanised substrates. (e) Effect of spin speed on CHPI film thickness prepared on untreated substrates for $30\ \text{mg/ml}$ solutions. The dashed lines indicate fit to the form ax^b , with $b = -0.449 \pm 0.009$.

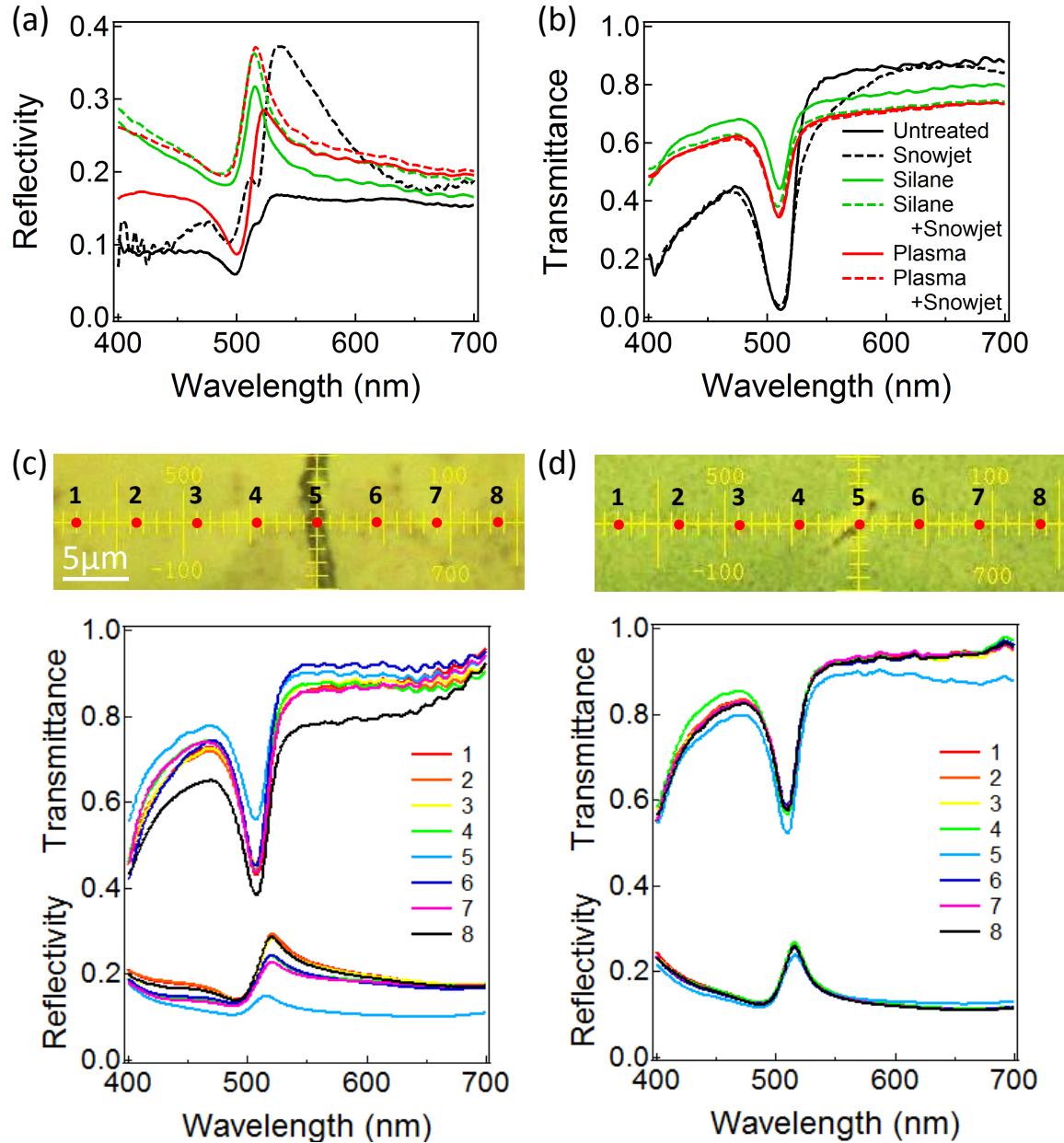


Fig. 4.7 Reflection and transmission spectra collected at $5\times$ magnification over a $\sim 20\ \mu\text{m}$ diameter region, on 2000 rpm films created using the labelled substrate preparation. Reflection and transmission spectra collected at $100\times$ magnification over a $\sim 1\ \mu\text{m}$ diameter region, on 4000 rpm films prepared on (c) untreated and (d) silanised and snowjetted substrate at the positions indicated.

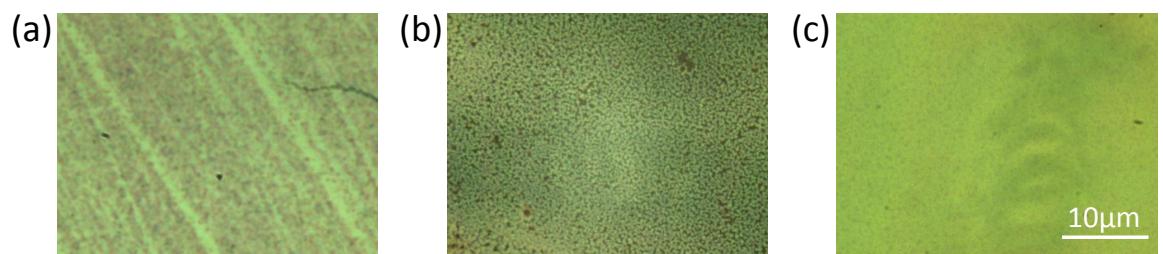


Fig. 4.8 BF images at 100 \times magnification for 4000 rpm films made on untreated substrates in (a) low, (b) high and (c) dehydrated atmospheres in the spin coater.

Chapter 5

Micromechanical exfoliation of lead iodide perovskites

5.1 Introduction

Uniform thin films of PbI perovskites can be created over large areas as a result of spin coating, however it is hard to achieve thicknesses $\lesssim 30\text{ nm}$ and probe the properties of ultra-thin samples. For thinner samples a layer-by-layer deposition technique can be used [56, 59]. Micromechanical exfoliation is another way of achieving such samples: cleaving thicker crystals to separate the layers and form progressively thinner samples.

In recent years much attention has been paid to 2D layered compounds such as graphene or transition metal dichalcogenides. Due to weak van der Waals bonding, it is easy to cleave neighbouring layers and form ultra-thin samples [136–141]. In these materials new optical and electronic properties emerge for mono- or few-layer regions, providing new avenues for material application. In this chapter we report micromechanical exfoliation of 2D PbI perovskites, and explore the few-layer behaviour of such systems via optical spectroscopy.

5.2 Experimental methods

A schematic of the preparation of exfoliated of PbI perovskites is shown in Fig. 5.1. Lead iodide (PbI_2) microcrystals are synthesized using a previously described solvothermal method, and intercalated using an 8 mg/ml organic ammonium iodide/toluene solution to create hexagonal perovskite microcrystals $\sim 30\text{ }\mu\text{m}$ in lateral size [142]. The crystals were then heated at 50°C to completely remove the intercalation solution. We then use a microme-

chanical exfoliation technique to create thinner flakes, transferring the resulting samples onto an oxidized silicon (Si) wafer for further measurements. The thinnest regions are identified using optical microscopy, and then characterized with white light spectroscopy and atomic force microscopy (AFM).

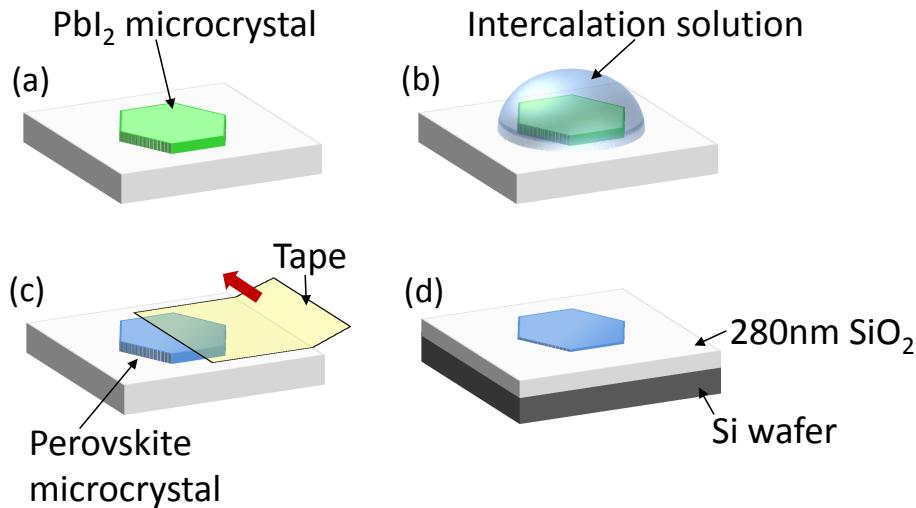


Fig. 5.1 Schematic of the preparation of exfoliated CHPI samples.

5.3 Results

BF images of intercalated CHPI and C12PI microcrystals at $20\times$ magnification are shown in Fig. 5.2(a,b). Reflectivity spectra for such microcrystals show an exciton resonance at $\sim 500\text{nm}$ indicating correct formation of the 2D structure, with Fabry-Perot fringing due to the thickness of the crystals (typically $\sim 1\mu\text{m}$). The top surface of the crystals are rough as a result of etching by the intercalation solution, however as a result of the exfoliation process such surfaces should adhere to the tape and not be present in the transferred flakes for measurement. The exciton wavelength (grey dashed line) varies from the values from literature as it has been shown that a build up of strain in thick layer stacks can lead to structural distortions/rearrangements [19, 135, 142] thus changing the exciton energy.

5.3.1 Exfoliated CHPI microcrystals

BF images of exfoliated flakes of CHPI and C12PI [Fig. 5.2(c,d)] are very similar. Although the crystals may fracture during the exfoliation process, we are able to obtain ultra-thin samples with lateral sizes of $\sim 1 - 10\mu\text{m}$. The excitonic resonance is still seen in such

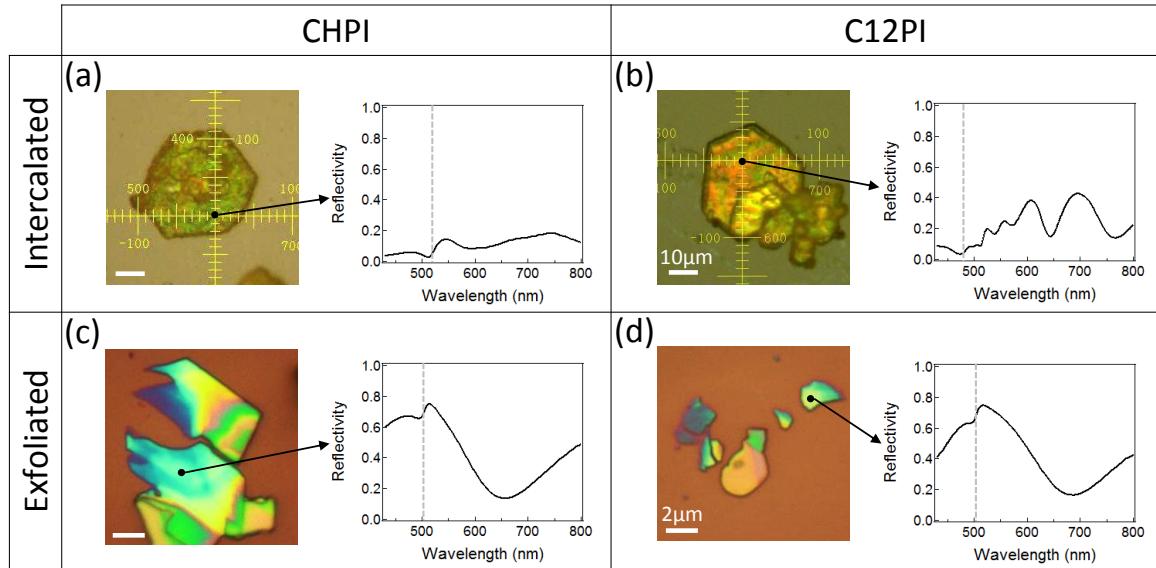


Fig. 5.2 Bright field images of intercalated (a) CHPI and (b) C12PI microcrystals on glass at $20\times$ magnification, and exfoliated (c) CHPI and (d) C12PI flakes at $100\times$ magnification. The reflectivity spectra of the areas indicated on the samples are also given. Grey dashed lines indicate the wavelength of the exciton resonance.

samples, however due to the bistability of excitons in C12PI we use CHPI samples for further analysis.

BF and DF reflection images of a typical CHPI flake are shown at $100\times$ magnification in Figs. 5.3(a,b) respectively. The DF scattering seen from the edges and grain boundaries of the sample is typical for such crystals. The reflectivity spectra for these exfoliated samples [Fig. 5.3(d)] consist of an excitonic Fano resonance at $\lambda_{ex} \approx 504$ nm superimposed on a background of Fabry-Perot fringes. These fringes correspond to the colour of the crystal seen in BF, and come from the path difference experienced by light double passing through the flake. The oxide layer on the substrate is designed to maximize optical contrast for very thin layers, and while optimized for graphene, this 280 nm SiO_2/Si system also works well for CHPI. By using this spectral information in conjunction with AFM measurements [Fig. 5.3(e)], we can correlate the position of Fabry-Perot fringes with thickness t . Thus it is then possible to spectroscopically determine the thickness of CHPI flakes.

A histogram of AFM heights in the boxed area of Fig. 5.3(c) shows three predominant thicknesses, which can be fit to Gaussians separated by steps of 1.6 nm [Fig. 5.3(f)]. This interlayer spacing agrees well with X-ray diffraction measurements, where the periodicity was found to be 1.7-1.8 nm [58, 143]. Due to the presence of molecules adsorbed on the surface of the substrate, the initial step of 2.5 nm is likely due to a monolayer, allowing us

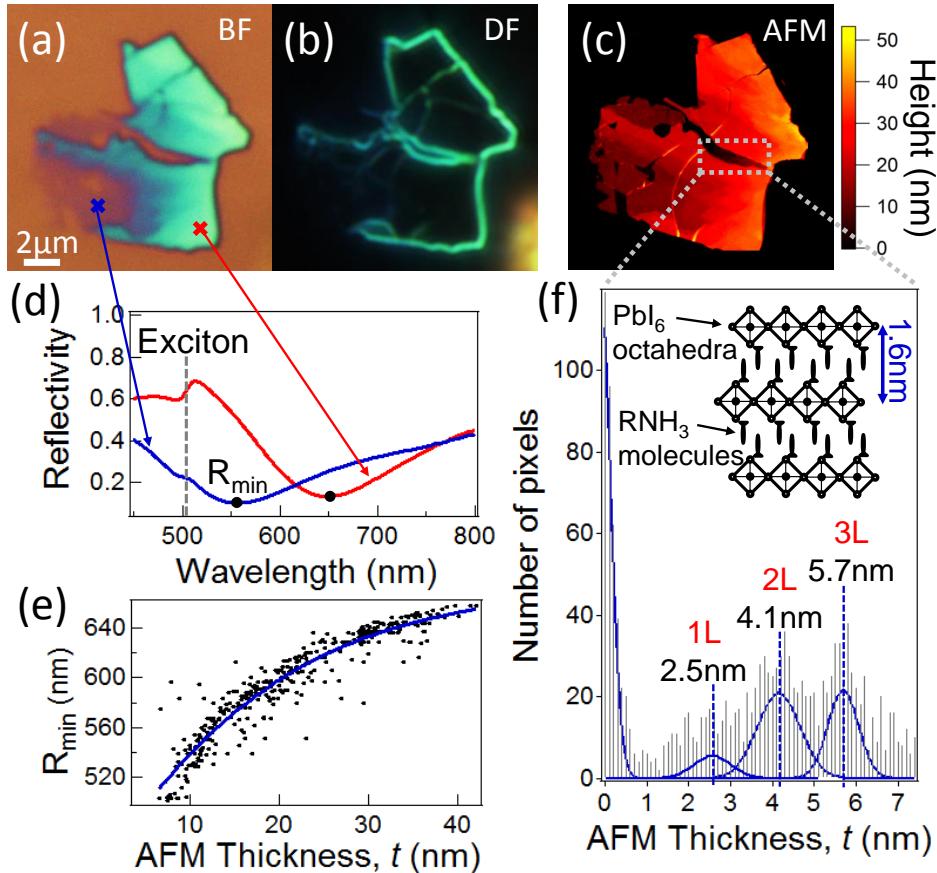


Fig. 5.3 Images in $100\times$ magnification using (a) bright and (b) dark field on an exfoliated CHPI flake; (c) AFM image of the same area. (d) Reflectivity spectra of two regions on the flake. The exciton wavelength is indicated by the dashed line. (e) Relationship between the measured reflectivity minimum (labelled as R_{\min} in (d)) and AFM thickness. (f) Histogram of heights measured in the boxed area of (c). Multipeak fitting to the data (blue lines) gives an interlayer spacing of 1.6 nm. The inset shows the structure of 2D PbI perovskites.

to label the three peaks as 1-, 2-, and 3-layer regions.

Spectroscopic measurements of flake thickness require detailed knowledge of the CHPI refractive index, which is not well known. Instead we extract the required information from reflectivity data using the transfer matrix formulation [144]. In the wavelength range of interest (460 - 750 nm), the dielectric function ε of 2D PbI perovskites can be modelled as the sum of a constant background and two Lorentzian oscillators: the exciton (*ex*) and an additional charge transfer (*CT*) transition at ~ 400 nm [30, 88, 135, 145–147]. Hence

$$\varepsilon = \varepsilon_1 + i\varepsilon_2 + \frac{A_{ex}}{\lambda_{ex}^2 - \lambda^2 + i\Gamma_{ex}\lambda} + \frac{A_{CT}}{\lambda_{CT}^2 - \lambda^2 + i\Gamma_{CT}\lambda}, \quad (5.1)$$

where $\varepsilon_1, \varepsilon_2$ are the background terms, while A_i is the amplitude, λ_i the wavelength, and Γ_i the linewidth of oscillator i . The refractive index ($\tilde{n} = \sqrt{\varepsilon}$) can then be used in the multilayer transfer matrix to calculate the expected reflectivity. The *CT* peak, due to the charge transfer between organic and inorganic layers, is particular sensitive to disorder and the local dielectric environment, and depends on the precise spin coating conditions when comparable thin films are produced [135].

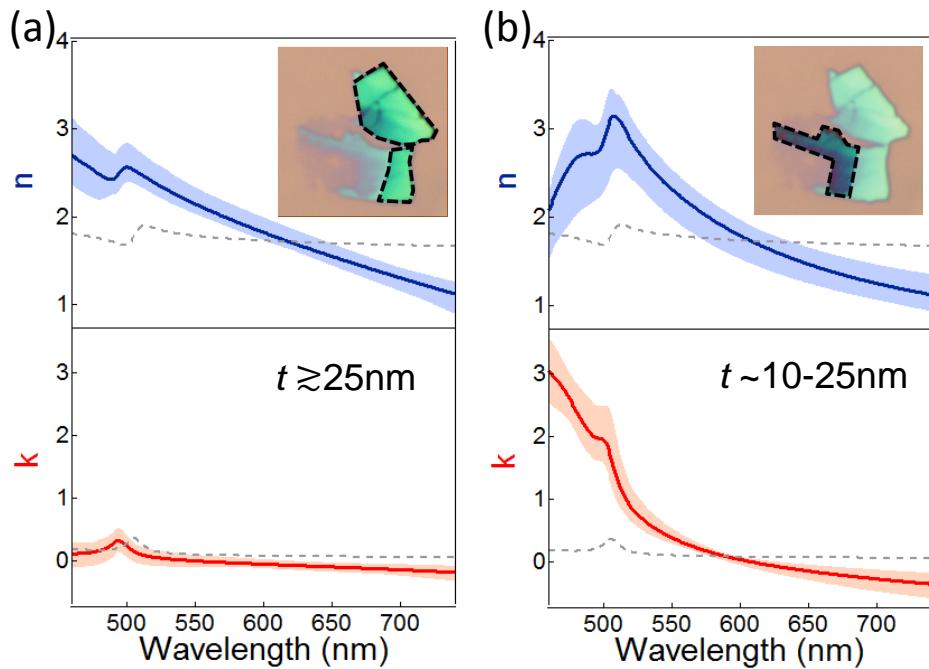


Fig. 5.4 Fitted complex refractive index of CHPI flakes for more than 200 pixels. Two regimes are found: (a) low absorption, occurring at positions of high thickness ($t \gtrsim 25$ nm), and (b) high absorption, at lower thickness ($t \sim 10-25$ nm). The insets indicate typical areas where each regime is found. Shaded regions show the range of values extracted from the fit, and grey dotted lines represent the refractive index of a CHPI film ($t \sim 60$ nm) measured using ellipsometry.

The results of these refractive index fits for more than 200 spectra are shown in Fig. 5.4. The fitting works well for spectra that are not collected at the edges of the flake, therefore the thinnest areas are excluded. Within these regions, two main regimes of refractive index are observed. In both cases the background and exciton oscillator are relatively unchanged. For thicker areas of the sample ($t \gtrsim 25$ nm) a low-absorption regime is observed, where the *CT* oscillator is mainly reflective. For thinner areas ($t \sim 10-25$ nm) a high-absorption regime is seen, where the *CT* oscillator redshifts and becomes more optically active. As discussed below, the thickness range encompassed by the absorbing regime is correlated

with a region of structural reconfiguration. This leads to a change in the energy states of the hybrid perovskite, and modifies the charge transfer between neighbouring organic and inorganic sheets. For comparison, the refractive index of a $t \sim 60$ nm film extracted from ellipsometry (grey dashed line) is also shown in Fig. 5.4. The film absorption (k) is closer to that of thicker flake areas, with a reduced contribution from the *CT* oscillator in the refractive index. X-ray diffraction shows that while distinct layers are formed during spin coating, there is greater structural disorder in each layer when compared with intercalated PbI_2 microcrystals [142]. This interface mismatch can be responsible for a large range of charge transfer environments, leading to the reduced strength of the *CT* resonance that we observe here.

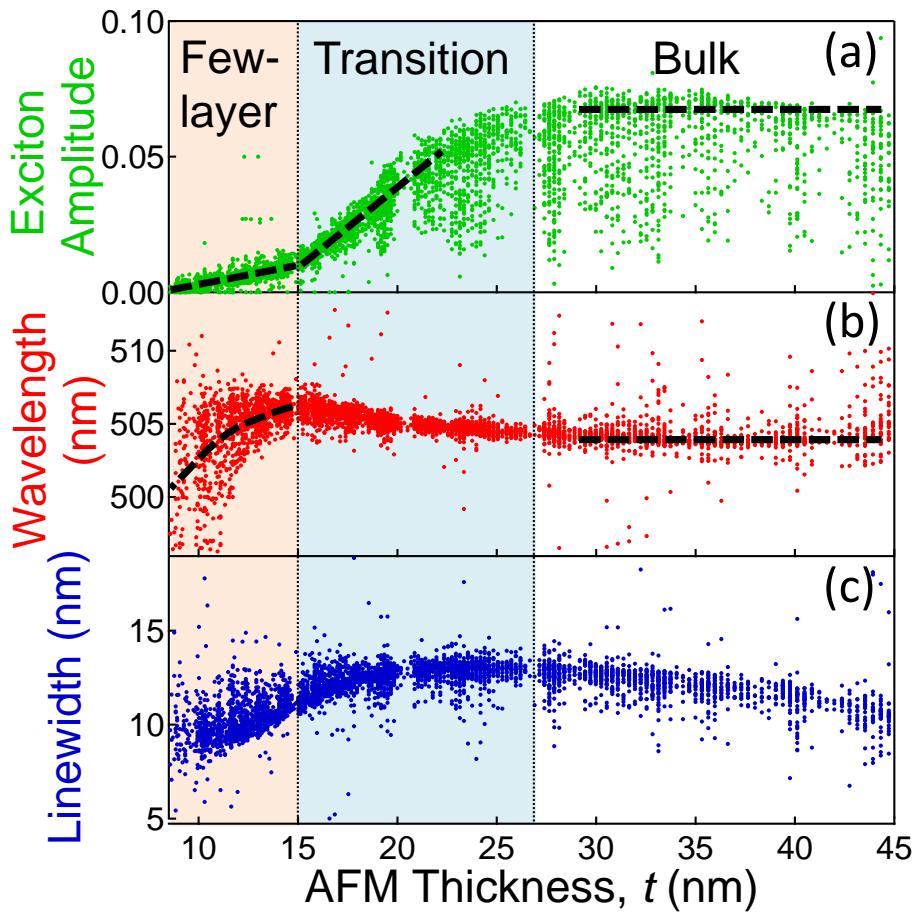


Fig. 5.5 Fitted exciton (a) amplitude, (b) wavelength, and (c) linewidth from reflectivity spectra (see Eq. 5.2). Dashed lines are guides for the eye.

In reflectivity the exciton produces a Fano lineshape due to interference between its narrow resonance and the continuum background. On account of this complication, we

extract information about exciton properties by describing $\Delta R = R_{\text{CHPI}} - R_{\text{substrate}}$ as

$$\Delta R = R_{\text{bkg}} + A \frac{(\lambda - \lambda_{\text{ex}} + q\gamma)^2}{(\lambda - \lambda_{\text{ex}})^2 + \gamma^2}, \quad (5.2)$$

where R_{bkg} represents the continuum background with Fabry-Perot fringing; A , λ_{ex} and γ are the amplitude, wavelength, and linewidth of the exciton respectively, and the parameter q describes the asymmetric shape of the Fano resonance. The results of the fit are shown in Fig. 5.5 for positions across many flakes with different thicknesses. A , λ_{ex} and γ are equivalent to the corresponding terms in Eq. 5.1, while q represents the interference between the exciton, CT and background terms. Near the vicinity of the exciton the effects of the CT and background are not distinguished, therefore Eq. 5.2 allows us to focus exclusive on the exciton components, while Eq. 5.1 gives us the overall refractive index. Since the perovskite resembles a multilayer system, we find the exciton amplitude initially scales linearly with the number of layers as expected, before saturating at $t \approx 27$ nm (15 layers). The large variability of amplitudes at high thickness arises predominantly from spectra taken at edges of flakes. Linear extrapolation of our data indicates the exciton amplitude will drop to zero at $t \sim 7$ nm (3 layers). However layer-by-layer assembly of perovskite films has shown that linear increases in the exciton intensity occur only after the fourth layer, while two monolayers are required to observe room temperature exciton behaviour [59].

From Fig. 5.5, we identify 3 regions of interest. Firstly the ‘bulk’ region ($t \gtrsim 27$ nm), where the exciton wavelength remains roughly constant; secondly the transition region ($t \sim 15–27$ nm), where the wavelength begins to redshift, while the linewidth reaches a maximum; and finally the few-layer region, where the wavelength blueshifts below the bulk limit, along with a decrease in the linewidth. This data helps us understand the changes happening at a structural level: disorder causes inhomogeneous broadening of the exciton resonance [148–150], while the exciton energy is directly related to the angle between PbI_6 octahedra in the inorganic layers [46]. In ‘bulk’ CHPI, the exciton has a wavelength of 504 nm and a spectral width of ≈ 10 nm. Close to the thickness transition region, the system is seen to become more disordered as the PbI sheets rearrange, becoming flatter and more strained. Finally at small t , the few-layer regime reveals how the layers relax and crumple again to reach the lowest energy configuration. Extrapolation of the fitted exciton wavelength to monolayer thickness (3 nm) leads to a wavelength of ~ 495 nm, comparable to the value of ~ 490 nm reported for PbI_2 thin films [151, 152]. We were unable to spectroscopically probe areas with $t < 8$ nm (4 layers) as they lie on the edges of flakes, and are around 100 nm in size. Since the lateral resolution of reflectivity measurements is $1 \mu\text{m}$, these spec-

tra are averaged with the much bigger signals from thicker areas. In order to achieve more sizeable monolayer regions, large-area samples are desirable for exfoliation, for instance using solution-grown single crystals. Exfoliating onto flexible polymer substrates may also improve our capability for attaining large monolayer regions as this can reduce fracture of crystals. However our measurements clearly show that these organic-inorganic hybrid perovskites change their electronic properties as the thickness is reduced, and this is connected to changes in the strain, disorder and layer structure.

5.4 Conclusions

In conclusion, we report the exfoliation of 2D organic-inorganic perovskites. Monolayers are observed, and the interlayer distance was found to be 1.6 nm. As with other 2D materials, the thinnest regions (<8 layers) behave differently from the bulk material due to the influence of strain on the layer structure. We note however that the active excitonic layers are already electronically isolated in these hybrids, so changes in the band structure as observed in dichalcogenide systems are not expected. Instead, the effects seen are due to the re-organisation of organic molecules around the inorganic sheets. This suggests that pre-organisation of the intercalating molecules is key to controlling material structure at the monolayer scale, which may be accessible through chemical growth rather than exfoliation. This work suggests the potential to construct optoelectronic devices for monolayers of these hybrid materials, offering new routes to emission.

Chapter 6

Exciton-localised surface plasmon interactions

6.1 Introduction

Localised surface plasmons are caused by the collective motion of electrons at the surface of nanoparticles, creating large field enhancements in the vicinity of the nanoparticle. For this reason such structures can be used in Surface Enhanced Raman Spectroscopy [153–155]. The resonance wavelength of such oscillations are very sensitive to the local dielectric environment, and can therefore be used for sensing devices [156–159]. The resonance wavelength can also be tuned using the shape and size of the particle, and this flexibility allows for a tunable plasmonic resonance that can couple to excitons in PbI perovksites.

In this Chapter we describe the creation of nano-islands of gold and silver overcoated with a perovskite layer. We use structural and optical characterisation techniques to see how excitons are affected by electrons in the metal.

6.2 Metal island films

The properties of a thin film grown on the substrate depends on the interactions between the film and substrate atoms, as well as deposition conditions [160]. For evaporation techniques, the deposition is a heterogeneous nucleation process, and requires high vapour pressure of the thin film material. Various growth modes are possible, but for noble metal films deposited on glass the metal atom-atom interaction is stronger than the metal atom-substrate interaction, causing island formation on the substrate. With increased deposition time such

islands can coalesce, and the grain boundaries may be preserved or a continuous structure may form. The morphology of the film depends on the diffusion of metal atoms on the substrate surface, and can therefore be affected by the deposition rate, substrate temperature, and subsequent annealing of the film.

The optical properties of such metal island films (MIFs) are the same as what's expected for a nanoparticle array, and if the islands are well separated (on the order of the island diameter) then there is no coupling between the particle resonances, and we expect to see a single resonance in the optical spectra. The resonance position of such arrays depend on the island morphology, and can be controlled by the deposited film thickness [161–165]. The resonance wavelength can be determined by modelling the system as a series of dipoles embedded in some medium [166–169].

6.2.1 Experimental methods

Glass substrates used are prepared as described in Sec. 4.3. Metal deposition is performed using an Edwards resistance evaporator, in a pressure of $\sim 4 \times 10^{-6}$ mbar with a deposition rate of $\sim 0.5 \text{ \AA/s}$. The substrates are not heated, and the deposited film thickness t is determined by a 6 MHz quartz crystal microbalance. Due to oxidation, Ag samples are placed in a cabinet with a nitrogen atmosphere, typically within 15 minutes of creation. Such Ag samples are only removed from the cabinet for further processing/characterisation. Both annealed Au and Ag films are made by heating the deposited films at 200°C for 24 hours. In order to create the CHPI overcoating, CHPI/THF solution is spin coated onto the nanostructured films under a dehydrated atmosphere, forming a layer thickness $\sim 100 \text{ nm}$. The samples are then characterised using AFM, SEM, and white light microscopy. Unpolarised reflection (R) and transmission (T) spectra of the sample are taken of the samples by making 400 scans over an area $0.5 \times 0.5 \text{ mm}^2$. Due to the uniformity of samples, all the spectra taken in the scan are then averaged to produce the spectra shown, where extinction is defined as $1 - R - T$.

6.2.2 Au metal island films

SEM images of evaporated Au films on glass show the formation of a continuous film when 30 nm is deposited [Fig. 6.1(c)], however the film is very rough and non-uniform. As the deposited thickness decreases [Figs. 6.1(a,b)] the films become smoother, however due to the weak Au-glass interaction dewetting is observed. As a result of annealing the Au atoms diffuse on the substrate and form distinct islands [Figs. 6.1(d-f)]. The islands

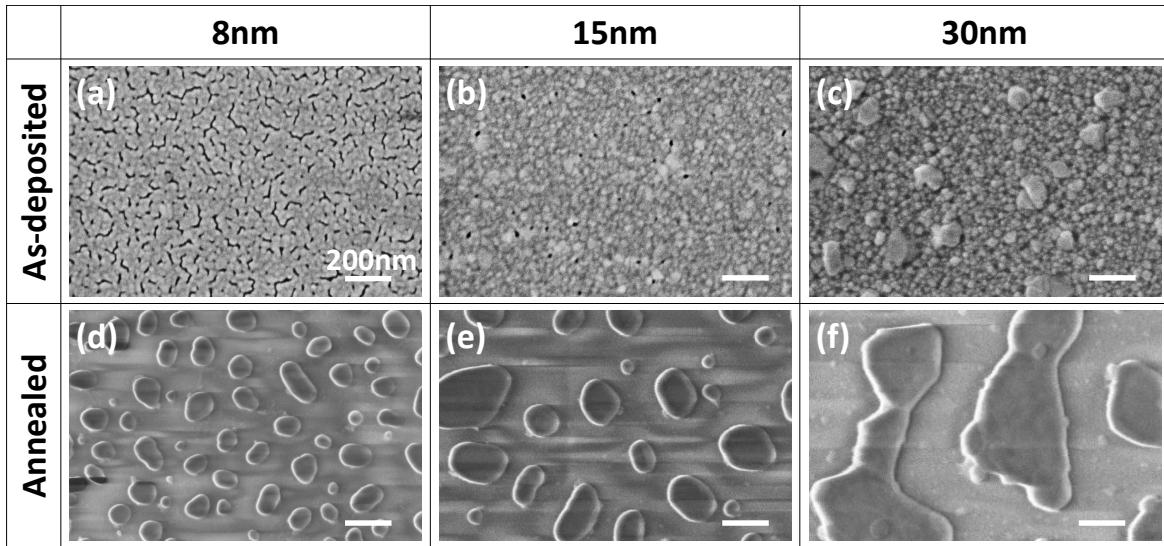


Fig. 6.1 SEM images of (a-c) as-deposited and (d-f) annealed evaporated Au films. The initial deposited film thickness is labelled.

become smaller in both lateral size and height [Fig. 6.2], and their shapes more uniformly ellipsoidal as t decreases, and for 8 nm we observe islands with size 50 – 100 nm, height \sim 70 nm, and separation 100 – 200 nm. The decrease in size can also be seen in 100 \times magnification DF images, where the scattering from the islands due to the LSP resonances become progressively broadband and appears white for 30 nm, but becomes progressively redder as the island size decreases [Fig. 6.2]. The change in the LSP resonance can be quantified optically using optical spectra. Fig. 6.3 shows the extinction spectra for 8 nm as-deposited and annealed Au films. We observe a resonance at 570 nm in the extinction in the as-deposited film, however this is due to the roughness of the film and therefore has a large width of 245 nm. In the annealed film we observe a resonance at 550 nm with linewidth 50 nm due to the LSPs in the nano-islands, and using Mie theory this corresponds to a spherical particle with a diameter of 100 nm.

6.2.3 CHPI-coated Au metal island films

BF images at 100 \times magnification show the formation of CHPI on t =8 nm Au samples (green areas in Figs. 6.3(b,c)), however the CHPI film is much rougher than on glass [Fig. 6.3(a)], and some dewetting can be seen. The exciton resonance at 505 nm is observed for all three films, confirming the formation of the correct MQW structure. We observe a redshift in the LSP resonance of the Au nano-islands for annealed films due to the CHPI: the resonance changes from 550 nm to 735 nm, with a considerable increase in the linewidth due to the

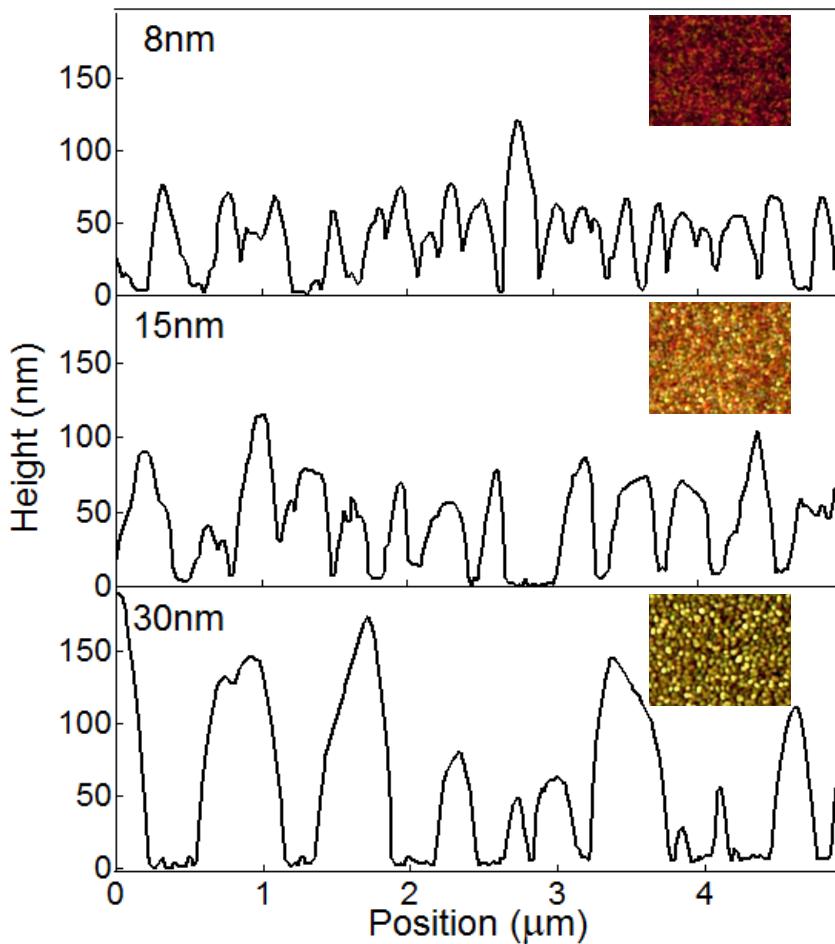


Fig. 6.2 AFM profiles of annealed Au metal island films. The deposited film thickness is labelled. Insets show 100 \times magnification DF images of the samples.

non-uniform CHPI coating. The redshift predicted by Mie theory for a spherical particle embedded in a medium with the refractive index of CHPI is 625 nm, and the discrepancy may be due to the ellipsoidal shape of the Au islands in the MIF. Note however that excitons in CHPI are completely unaffected by the Au films, keeping the same wavelength and linewidth, although the overall magnitude of the extinction has increased due to the Au films.

6.2.4 Ag metal island films

The morphology of Ag evaporated films glass show similar trends to Au films: rough films lead to smoother dewetted films with decreasing t , and the formation of MIFs when the

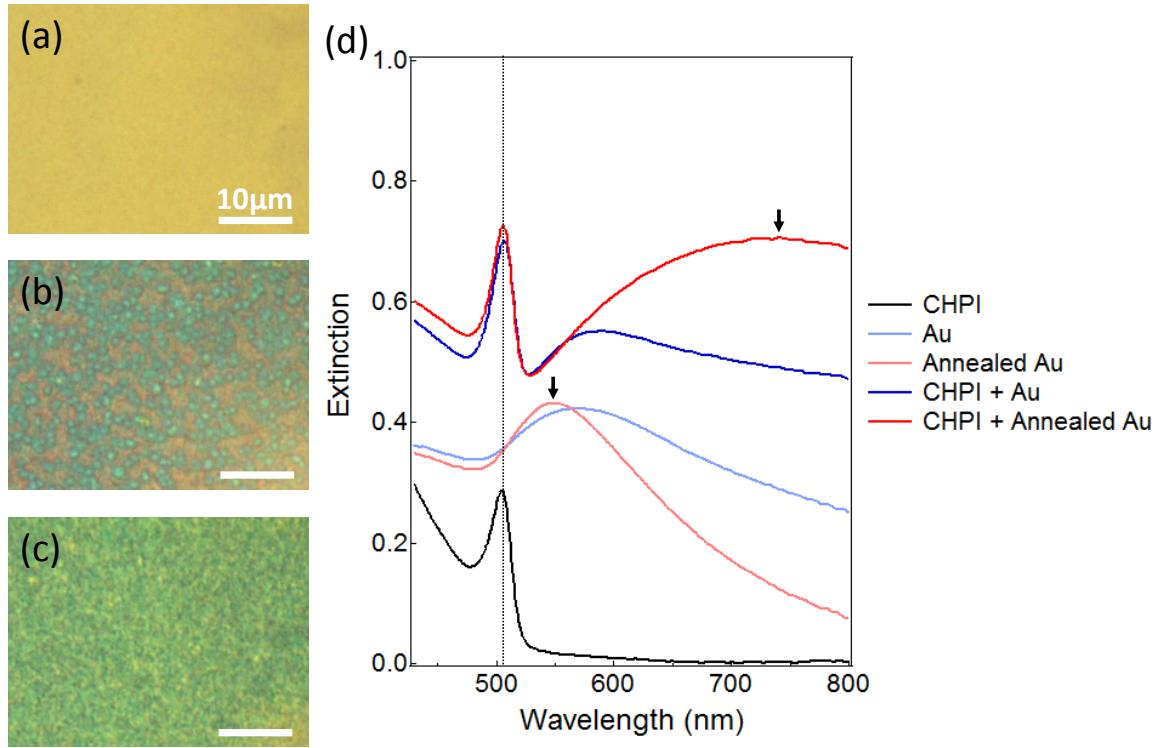


Fig. 6.3 BF images at $100\times$ m magnification for CHPI films on (a) glass, (b) 8 nm evaporated Au film, and (c) 8 nm annealed Au metal island film. (d) Extinction spectra at $5\times$ magnification. The exciton wavelength of the CHPI film is marked by the dashed line, and SPP resonances by arrows. The CHPI + (annealed) Au spectra are offset for clarity.

as-deposited films are annealed [Fig. 6.4]. However the interactions between Ag atoms and glass is clearly stronger than with Au, as for 30 nm films annealing only causes coalescence and larger grains, not well separated islands. Annealed 8,Ag MIFs consist of ellipsoidal islands with lateral size 40 – 100 nm, height \sim 80 nm, and separation 50 – 150 nm. DF images [Fig. 6.5] also show broadband white scattering for larger islands.

As seen in the SEM images, for $t > 8$ nm as-deposited Ag films are essentially continuous with some dewetting, and the extinction spectra are similar to that of a bulk Ag film, with low extinction at visible wavelengths, increasing extinction up to the band gap \sim 300 nm [Fig 6.6(a)]. However resonances can be observed for lower t films, particular for 2 nm as-deposited films with a resonance at 560 nm and linewidth 175 nm, suggesting the formation of islands even without annealing (see below).

After annealing, the LSP resonance of the Ag islands dominated the extinction spectra [Fig. 6.6(b)]. Comparing the extinction peaks [Fig. 6.6(c)], we see that the positions of the extinction peaks do not change significantly with t , however there is a clear decrease in

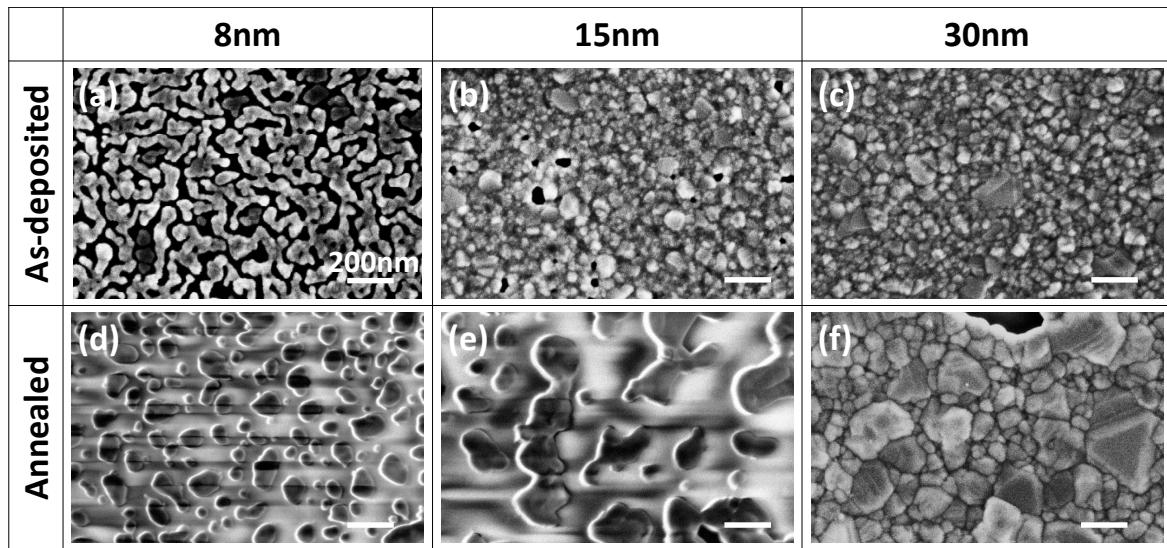


Fig. 6.4 SEM images of (a-c) as-deposited and (d-f) annealed evaporated Ag films. The initial deposited film thickness is labelled. Dark areas/streaks are seen in (a, d, e) due to charging of the sample.

the linewidth of the $t=2$ nm film compared to the others. The relative stability of the LSP resonance position suggests that the average island size does not change with t , however for larger t we may have a larger deviation in the shape and sizes of the islands, leading to a superposition of many resonance wavelengths and higher order LSP modes. The resonance wavelength of ~ 440 nm for the Ag islands is larger than the predicted value of an 100 nm spherical particle from Mie theory (380 nm), which may be due to both the ellipsoidal nature of the Ag islands, as well a potential oxidation layer formed on the surface of the islands.

6.2.5 CHPI-coated Ag metal island films

CHPI coated Ag films behave similarly for all t , so here we use $t=8$ nm as an example. BF images at $100\times$ magnification show CHPI films deposited on as-deposited and annealed Ag films appear very similar to CHPI films deposited on glass [Fig. 6.7(a-c)], although some non-uniformity is observed in the case of annealed Ag MIF. The extinction spectra of CHPI coated as-deposited Ag film is very similar to the spectra of CHPI film on glass (exciton at 505 nm). However the Ag islands of the annealed film causes a blueshift of the exciton wavelength to 500 nm, we see no appearance of the island LSP resonance, and there is an overall increase in the exciton extinction peak as a result of the Ag islands. Together this indicates weak coupling between the Ag LSP and excitons and enhancement in exciton absorption due to the near-field of the LSP.

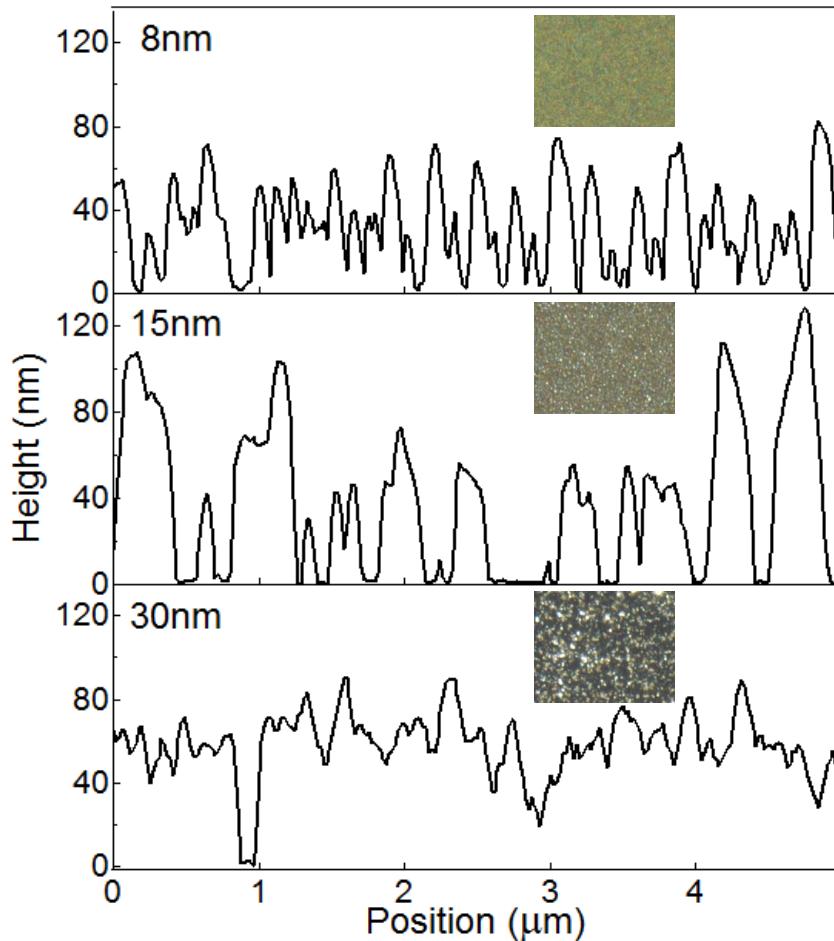


Fig. 6.5 AFM profiles of annealed Ag metal island films. The deposited film thickness is labelled. Insets show 100 \times magnification DF images of the samples.

6.2.6 Ag islands on CHPI films

Instead of coating Ag MIF films with CHPI, we also fabricated samples of Ag islands on top of a CHPI film on glass. Since the organic molecules undergo a melting transition $\sim 80^\circ\text{C}$ [44], we thermally evaporate only 2 nm of Ag to prevent any degradation of the CHPI film due to heat. The AFM profile of a 2 nm as-deposited Ag film on glass [Fig. 6.8(b)] shows the formation of separated metal islands, with lateral size ~ 30 nm and height ~ 6 nm. The AFM profile of a Ag MIF on CHPI is dominated by the surface roughness of the CHPI film (~ 5 nm, *cf* Fig. 6.8(a)), however some high frequency noise on the order of the island sizes can also be seen. However the island features are not distinct from the CHPI film, which suggests the particles may be embedded in the CHPI film. Looking at the extinction spectra [Fig. 6.8(d)], the strong exciton peak indicates that the MQW structure remains despite the

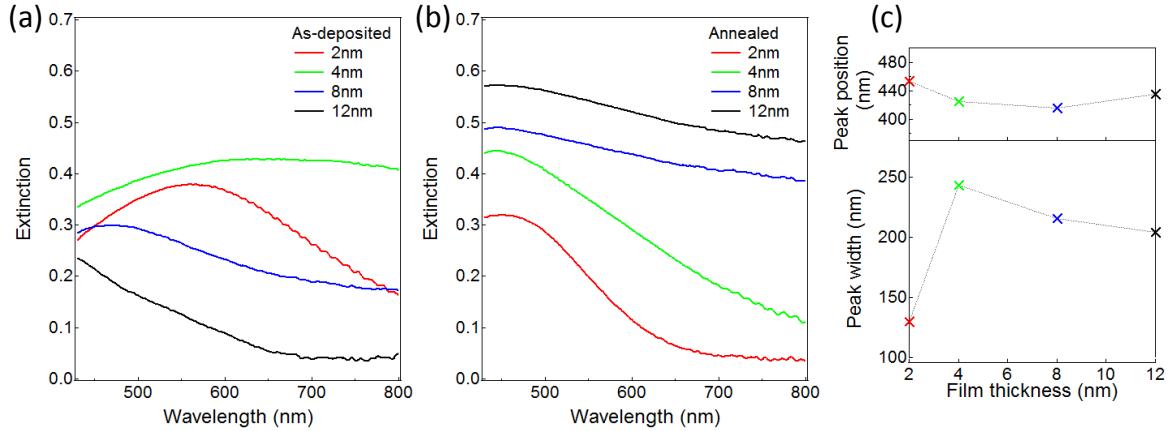


Fig. 6.6 Extinction spectra collected at $5\times$ magnification for (a) as-deposited Ag films and (b) annealed Ag metal island films with the labelled thickness. (c) Extinction peak position and linewidth for the spectra in (b).

thermal evaporation of Ag, and the interactions between the LSP of the Ag islands and the excitons leads to weak coupling, with a blueshift in the exciton wavelength of 4 nm, similar to what was observed in the CHPI-coated Ag islands [Fig. 6.7(d)].

6.3 Nanosphere lithography

Nanosphere lithography involves the use of closely-packed 2D arrays of nano-/microparticles as lithography masks. Evaporation of metal through the mask followed by removal of the spheres leaves behind an array of metal islands on the substrate [170]. The geometry of the island array depends on the diameter of spheres D used in the mask. For a colloidal monolayer, the island diameter is 0.223 , while the inter-island separation is $0.58D$ [171]. The islands formed are triangular in shape due to the interstices between spheres, however for small D the islands can become more spherical [172]. Array geometry can also be controlled by changing the evaporation angle [173].

Much like MIFs, the metallic islands produce a LSP peak in optical spectra, where the resonance wavelength depends on the size and shape of islands, as well as the dielectric environment [156], and as such can be modelled as an array of dipoles [158, 174]. However nanosphere lithography provides better control of the island geometry due to the periodicity of the microsphere mask, so should provide a sharper LSP resonance when compared with MIFs.

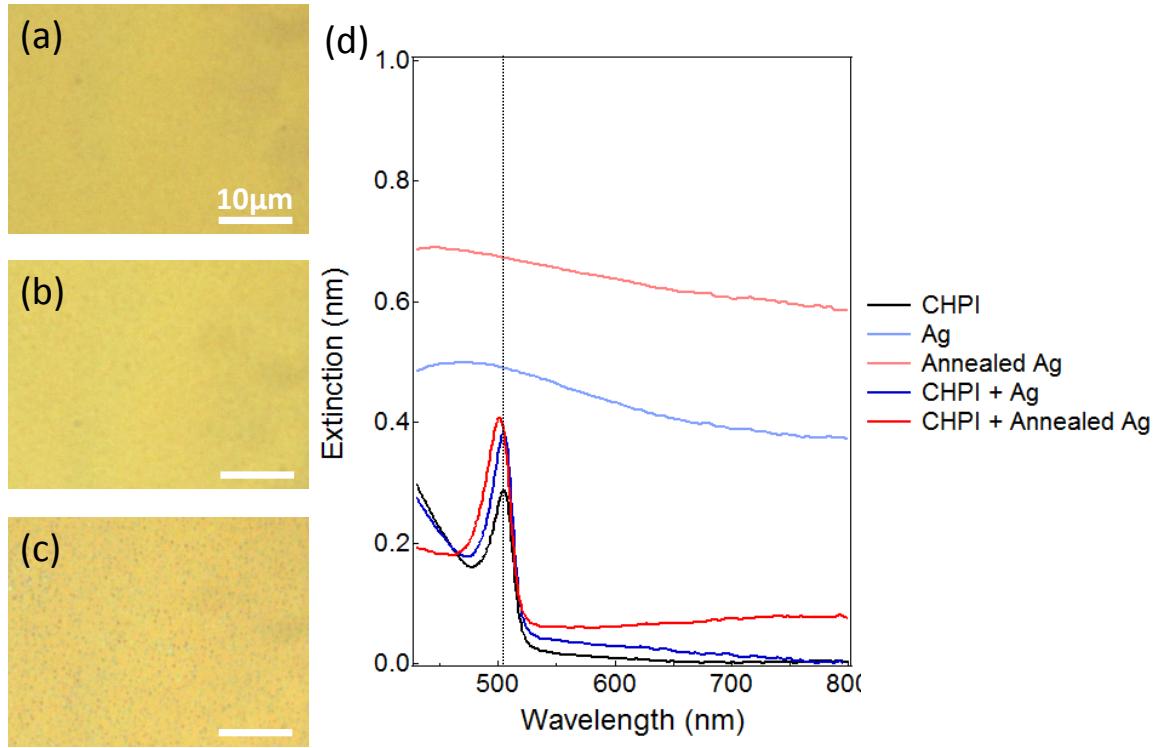


Fig. 6.7 BF images at $100\times$ m magnification for CHPI films on (a) glass, (b) 8 nm evaporated Ag film, and (c) 8 nm annealed Ag metal island film. (d) Extinction spectra at $5\times$ magnification. The exciton wavelength of the CHPI film is marked by the dashed line. The (annealed) Ag spectra are offset for clarity.

6.3.1 Experimental methods

We use 460 nm diameter polystyrene (PS) microspheres from Sigma Aldrich to create the colloidal monolayer. The 10% microsphere solution is then diluted in a 1:1 mix with absolute ethanol. Glass substrates are cleaned as described in Sec. 4.3 then plasma etched for 1 minute to create a hydrophilic surface. The substrates are placed at a 10° angle before a deionised water droplet is applied to cover the glass surface. A 2wt% solution of sodium dodecylsulphate in water ($< 0.5\ \mu\text{l}$) is then applied to the water surface, where the amphiphilic surface molecules acts to reduce the surface charges on PS microspheres. A pipette is used to spread the PS microsphere onto the droplet surface [Fig. 6.9(a)], and the water allowed to evaporate under standard conditions. 50 nm of Au is then deposited on the samples using an electron-beam evaporator system under a pressure $\sim 5 \times 10^{-6}$ Torr at a rate of 1\AA/s . The PS microspheres are dissolved by placing the sample in a solution of dichloromethane for 30 minutes, then sonicating the solution for 5 minutes. CHPI-coating is then produced by spin coating a CHPI/THF solution onto the island samples under a dehydrated atmosphere.

Optical characterisation is performed by taking 400 scans over a $50 \times 50 \mu\text{m}^2$ region, then averaged to produce the spectra shown.

6.3.2 Au islands

Closely-packed 2D arrays of PS microspheres are formed using this technique [Fig. 6.9(b)], and the ordering is best at contact lines due to the coffee ring effect. After removal of PS, triangular islands are left behind on the glass substrate [Fig. 6.9(c)] with lateral size $\sim 90 \text{ nm}$. We can see from Fig. 6.9(c) that even in the best areas we do not uniformly find well-separated triangular islands. Due to small deviations in the microsphere packing bowtie shaped islands can form, or lines of Au in a gap between domains. However in the extinction spectra of such island samples [Fig. 6.10] we do observe an LSP resonance at 590 nm with a linewidth of 80 nm.

6.3.3 CHPI-coated Au islands

Similar to the CHPI-coated Au MIF, strong exciton peaks in the extinction spectra of CHPI-coated Au islands indicate formation of the correct MWQ structure. The exciton wavelength is unaffected by the Au and remains at 505 nm. As before, the LSP resonance redshifts due to the CHPI coating, and the linewidth broadens to $\sim 150 \text{ nm}$. We observe a systematic increase of the LSP redshift as a result of increasing spin speed, and attribute this to more complete CHPI encapsulation of the Au islands due to larger forces induced by the higher spin speeds.

6.4 Conclusions

Evaporation of noble metals can be used to create ellipsoidal nanoparticles on a glass substrate due to poor adhesion of the metal atoms. Optically the electrons in such particles can oscillate in phase to create LSP resonances. The resonance wavelength can be tuned using the shape and size of the particles, and can be changed by changing the dielectric environment around the particles. When the LSP resonance is far off-resonance with excitons, as is the case with Au, the perovskite thin film merely acts as a high refractive index medium and causes a redshift in the LSP wavelength. However if the two resonances are closer together in energy, as with Ag islands, weak coupling can occur, causing a blueshift in the exciton wavelength by 5 nm, and enhancing the exciton absorption due to the enhanced electric field

of the LSP. Such absorption enhancement has been investigated and is of particular interest for the design of solar cells [175–178].

In order to observe strong coupling between LSPs and excitons, one problem may be with the width of the LSP resonance. The large variation in shape and size of metal particles in MIFs could potentially be an issue. However from our experiments the controllable islands created using nanosphere lithography do not create a marked improvement in the optical signature of the LSP resonance. A potential future exploration could be to use chemically created metallic NPs, which are often pre-screened for size. However a method of controllable assembling a dense array of individual NPs from solution will need to be explored.

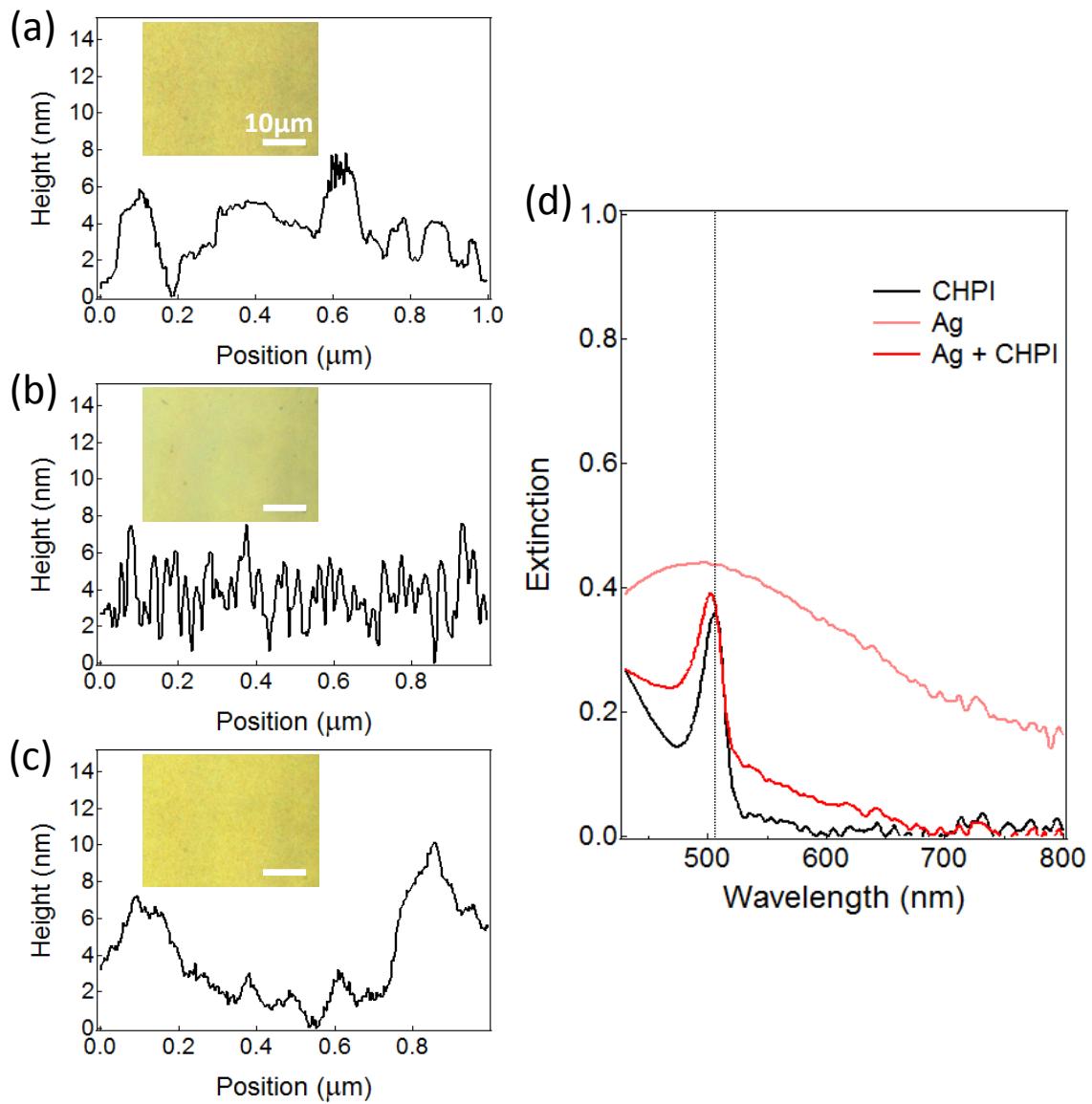


Fig. 6.8 AFM profiles of (a) CHPI film on glass, (b) 2 nm evaporated Ag film on glass, and (c) 2 nm evaporated Ag film on CHPI. The insets show 100 \times magnification BF images of the samples. (d) Extinction spectra at 5 \times magnification. The exciton wavelength of the CHPI film is marked by the dashed line. The Ag spectrum is offset for clarity.

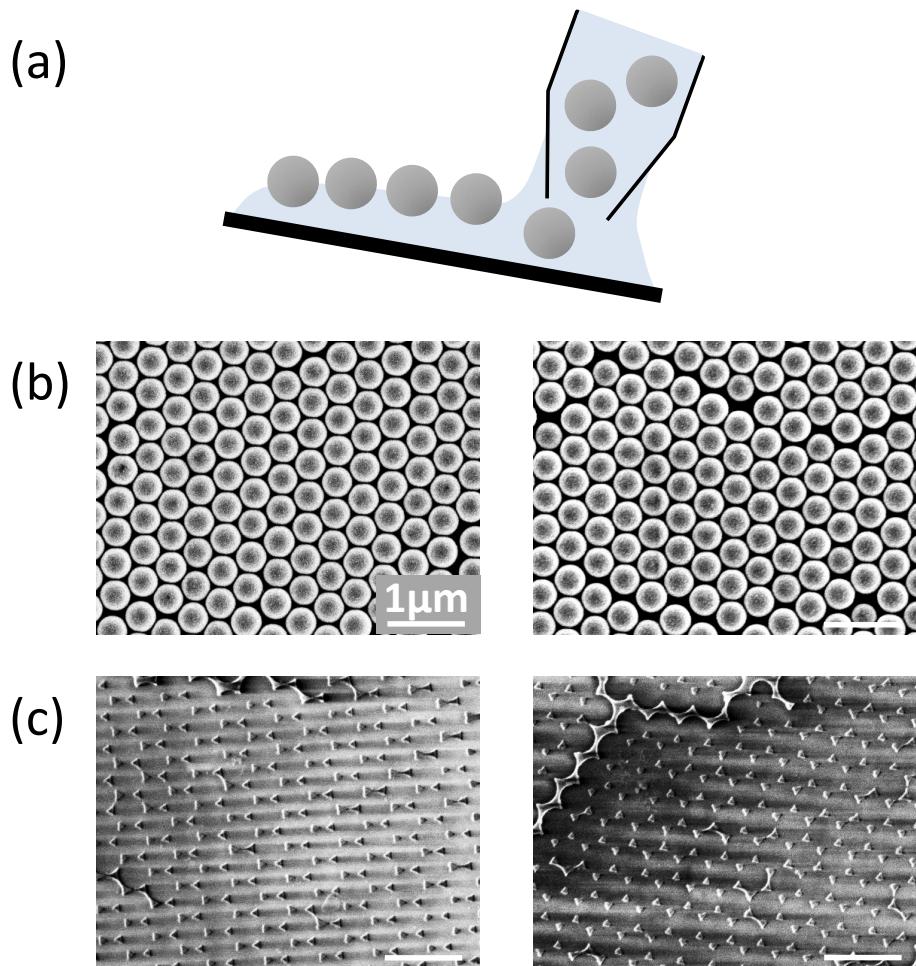


Fig. 6.9 (a) Schematic illustration of the transfer of 460 nm PS colloids to a water droplet via a pipette. (b) SEMs of colloid monolayer formed after the water droplet evaporates under standard conditions. (c) SEMs of triangular islands formed after evaporation of Au onto colloid monolayers and dissolving the PS.

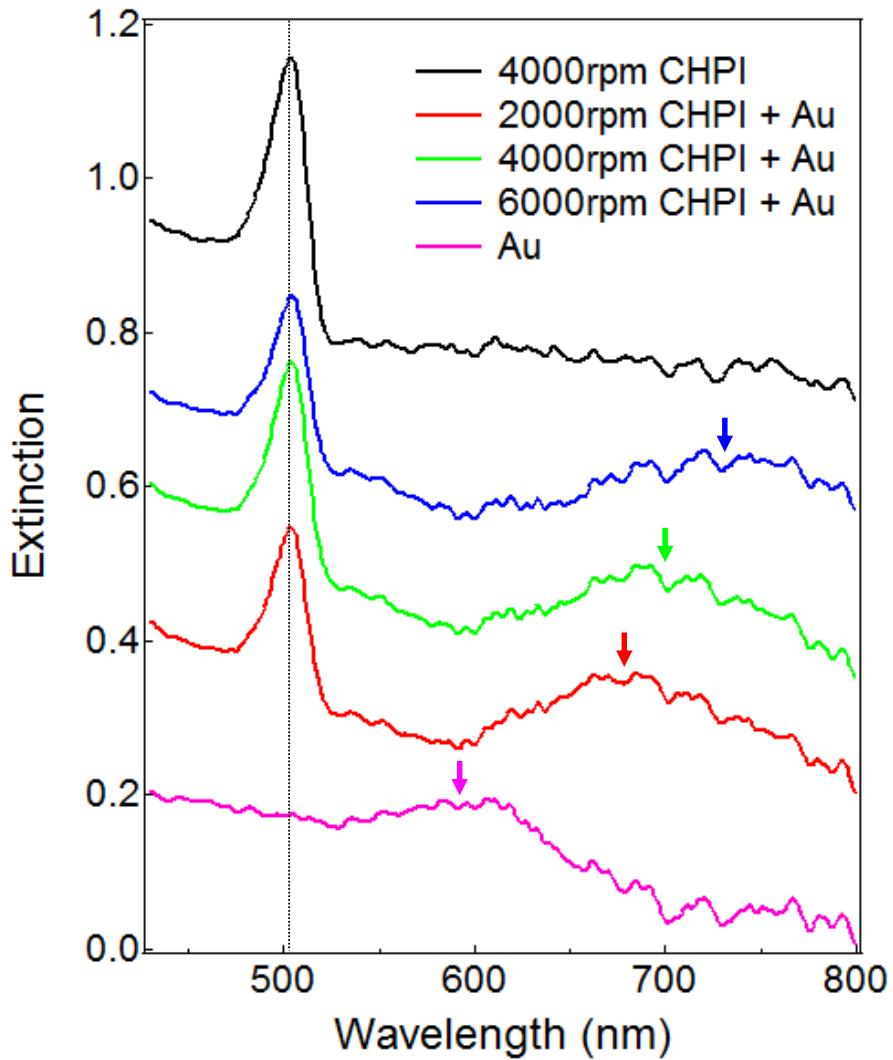


Fig. 6.10 Extinction spectra for CHPI-coated Au islands as labelled. Arrows indicate the positions of LSP resonances, and a dashed line indicates the exciton wavelength.

Chapter 7

Exciton-surface plasmon polariton interactions

7.1 Introduction

Gratings are structures with periodicity in one dimension. Despite its simple structure, such samples can display many optical signatures, from laterally localised modes such as waveguide modes or channel plasmons, or diffractive modes that occur due to interference effects. For a grating made of a plasmonic metal, diffractive modes can be simplistically separated into ‘photonic’ modes due to such interference, and ‘plasmonic’ modes where SPPs on the metal also interact with the diffracted light. The dispersion and efficiency of modes in optical spectra depend on the coupling with incoming/outgoing photons, and as such is very sensitive to factors such as the polarisation of light, changes in geometry and the refractive index of any coating materials.

In this Chapter the optical behaviour of both plasmonic and non-plasmonic gratings, and use CHPI-coated Ag gratings to understand the interactions between excitons and SPPs on the metal surface.

7.2 Experimental methods

A schematic of the fabrication of dielectric-coated metal gratings is shown in Fig. 7.1(a). Gratings are fabricated in ethylene tetrafluoroethylene (ETFE) from nanopatterned silicon stamps using nanoimprinting. A sheet of ETFE (thickness 0.8 mm) is placed on a silicon stamp, and heated to a temperature of 200°C before applying a pressure of 30 Bar for

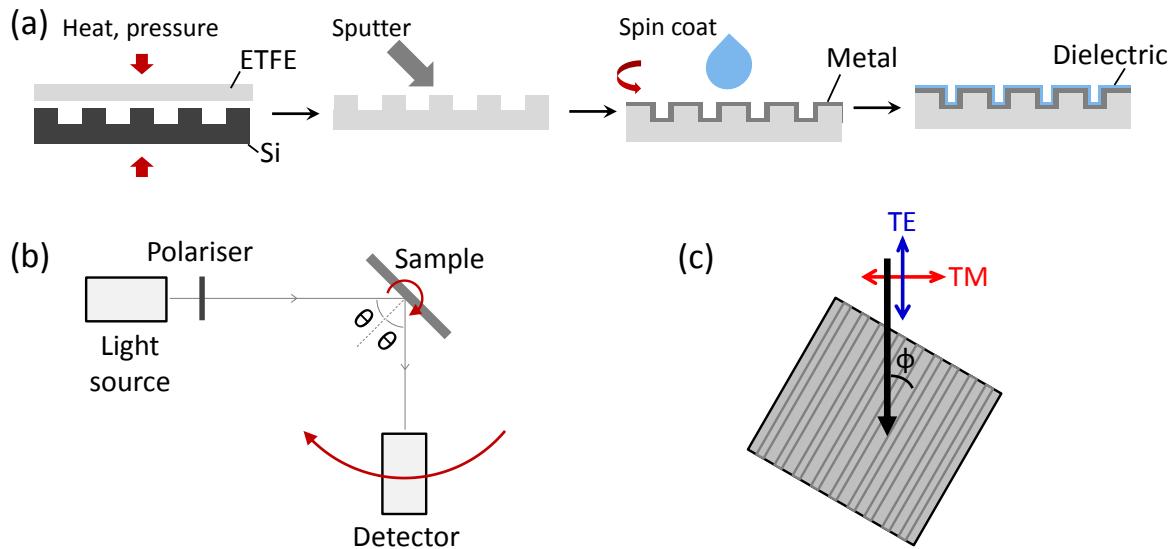


Fig. 7.1 (a) Schematic of dielectric-coated metal grating fabrication. (b) Schematic of angle-dependent reflectivity measurement setup. (c) Relationship between the polarisation of incoming light and azimuthal angle ϕ .

300s. The ETFE is then cooled to 90° while maintaining the same pressure, before release from the stamps. An optically opaque metal layer (~ 120 nm thick Ti/Ag) is deposited onto the polymer to form metal gratings. Chemically synthesised CHPI powder is dissolved in tetrahydrofuran and spin coated onto the gratings in a dehydrated atmosphere to produce a conformal coating. For PS-coated gratings, $M_w = 500000$ PS powder is dissolved in toluene and spin coated onto the gratings. All samples are kept under a nitrogen atmosphere to prevent oxidation. Measurements by SEM and AFM of the metal and dielectric-coated gratings are used to extract the dimensions of the nanostructures. Specular reflection measurements are made as a function of the incident polar (θ) and azimuthal (ϕ) angles using a polarised broadband white light source (215 – 2500 nm), with the geometry shown in Fig. 7.1(b,c). The sample properties are uniform over cm^2 areas, with small variations due to the depth and morphology of the coatings.

7.3 Dielectric gratings

7.3.1 ETFE gratings

AFM scans of the imprinted $D = 417$ nm ETFE grating structure [Fig. 7.2(a)] show the formation of a square-wave grating with depth 140 nm and slit width 200 nm [Fig. 7.2(b)]. TM and TE polarised reflectivity scans show $m = -1$ photonic modes according to Eq. [grating]

(grey dashed lines on Figs. 7.2[c,e]), which appear as dips in the reflectivity [Figs. 7.2(d,f)]. In TE polarisation we also see the appearance of a redshifted photonic mode (grey dot-dashed line on Fig. 7.2(e)), corresponding to $n = 1.4$ in Eq.[grating]. This fits well with the ETFE refractive index reported [179], thus attribute this redshifted mode to light which has penetrated the transmissive ETFE. For both polarisations the grating modes are no longer visible for $\phi > 60^\circ$. Note also a dip in the reflectivity of TM scans at $\theta \approx 50^\circ$ due to the Brewster angle of ETFE.

7.3.2 CHPI-coated ETFE gratings

The exciton resonance at 505 nm dominates both the TM and TE reflectivity scans of CHPI-coated $D = 417$ nm ETFE gratings [Figs. 7.3(a,c)]. Diffractive photonic grating modes are also visible (marked by grey dot-dashed lines), and appear as Fano resonances in reflectivity due to interaction with the broad reflectivity background of the CHPI film [Figs. 7.3(b,d)]. The modes correspond $m = \pm 1$ photonic modes are at the same positions as the $n = 1.4$ redshifted modes in Fig. 7.2(e) and are thus attributed to the same source. In TM polarisation the diffractive modes are strongest for $\phi = 90^\circ$, while for TE they are strongest at $\phi = 0^\circ$. [Reason?] The Brewster angle in TM scans has changed to $\theta \approx 62^\circ$ due to the larger refractive index of CHPI. Most notably, the grating modes do not interact with the excitons, even when crossings occur.

7.4 Non-plasmonic gratings

7.4.1 Ti gratings

SEM image of the $D = 417$ nm Ti grating shows the roughness of the sputtered Ti film [Fig. 7.4(a)], while AFM measurements reveal a the formation of a more trapezoid grating profile, likely due to the initial imprinted EFTE grating rather than the sputtering process. Heat from the sputtering process also appeared to change the grating periodicity D due to shrinking of ETFE, as the photonic grating modes in reflectivity scans [Figs. 7.4(c,e)] are best fit to $D = 410$ nm in Eq.[Grating]. Aside from the change in geometry, the appearance of $m = -1$ grating modes are very similar to the spectra of ETFE gratings, with modes appearing as dips in the reflectivity spectra. However the coupling to outgoing light is much weaker in TE polarisation, as shown by the comparative sizes of the dips in Figs. 7.4(d,f).

7.4.2 CHPI-coated Ti gratings

Polarised reflectivity spectra of CHPI-coated planar Ti film [Fig. 7.5] shows the appearance of an exciton resonance at 505 nm, showing that the excitons are unaffected by the metal film below. The experimental data fits well to transfer matrix simulations of 70 nm CHPI-coated 120 nm Ti film, and the differences observed can be attributed to non-uniformity and roughness in both the CHPI and Ti films.

AFM measurements of CHPI-coated $D = 417$ nm Ti grating show no appearance of the grating structure [Fig. 7.6(a)], and from this we conclude that the grating structure has been completely immersed in the non-uniform CHPI film [Fig. 7.6(b)]. As with CHPI-coated ETFE gratings, the exciton resonance at 505 nm dominates reflectivity spectra for both TM and TE polarisation [Figs. 7.6(c,e)]. Although very weak dips can be seen in reflectivity to indicate diffractive $m = -1$ grating modes in TM polarisation [Fig. 7.6(d)], coupling of TE-polarised light to grating modes is so weak that spectra look essentially like that of the CHPI-coated planar Ti film [Fig. 7.6(f)]. In both cases there are no interactions between CHPI excitons and modes of the Ti grating.

7.5 Plasmonic gratings

7.5.1 Ag gratings

Fig. 7.7 shows reflectivity scans for Ag gratings with periodicities $D = 556, 417, 278$ nm. The spectra for all three gratings show the same features: in TM polarisation a sharp threshold anomaly is observed, whose dispersion follows Eq.[grating] for the $m = \pm 1$ modes (grey dashed lines). A redshifted dip follows for the resonance anomaly, indicating the presence of excited SPPs (black dashed lines). In TE polarisation we don't observe any anomaly features due to the inability to excite SPPs, instead we see the $m = \pm 1$ photonic modes.

Concentrating on the $D = 417$ nm grating, we see that the sputtered Ag also shows some roughness [Fig. 7.8(a)], and AFM measurements indicate a square-wave grating with depth 140 nm and slit width 130 nm [Fig. 7.8(b)]. In reflectivity [Fig. 7.8(c,e)] the threshold (grey dashed lines) and resonance (black dashed lines) anomalies shift as expected from Eq.[grating] with ϕ in TM polarisation, and appear as the expected sharp change in intensity/dip in spectra [Fig. 7.8(d)]. The resonance anomaly becomes weaker with increasing ϕ and is no longer observed at $\phi = 60^\circ$ as it becomes harder to couple to SPPs in this polarisation. For this same reason no anomalies are observed in TE polarisation at low ϕ , where the photonic modes appear as rather weak peaks in the spectra [Fig. 7.8(f)]. We would expect to

observe anomalies at $\phi = 90^\circ$ in TE polarisation, however the energy of the mode is too high for our measurement range here. The broad dip seen at $\phi = 60^\circ$ in both polarisations and $\phi = 90^\circ$ in TE polarisation is assigned to a Fabry-Perot interference mode of light reflected from the top and bottom surfaces of the grating. The mode doesn't change in position with ϕ as expected, and extrapolates to ~ 600 nm at $\theta = 0^\circ$, which fits the height of the gratings as seen in AFM measurements.

We use finite element method (FEM) to model the electromagnetic nearfield of the grating modes to understand their behaviour. The modelled spectra for $\phi = 0^\circ \lambda = 600$ nm agrees very well with the features of the experimental data [Fig. 7.9(a)], but has a larger reflectivity overall due to roughness on the experimentally created Ag gratings. The low reflectivity region at high θ is caused by low efficiency of the specularly reflected order. Instead the strongest coupling is to the -1 diffracted order, as evidenced by the wavefronts of E -field intensity in a diagonal direction in Fig. 7.9(b). The modelled H_z field component for $\theta = 30^\circ \lambda = 700$ nm shows at various points in the optical cycle [Fig. 7.9(c)] shows that the resonance anomaly does indeed behave like SPP on the surface of the metal, in this case travelling towards the left of the grating.

The position of the threshold anomaly is fixed by the periodicity of the structure, however as the resonance anomaly is a Fano resonance due to the interference between diffracted light and SPPs we expect the position of this to depend much more sensitively on the geometry of the grating. Fig. 7.10 shows the structures and TM reflectivity scans at $\phi = 0^\circ$ for three difference gratings, ranging from a square-wave grating [Fig. 7.10(a)] to a more sinusoidal grating [Fig. 7.10(b,c)], and these geometric differences are caused by small changes in the nanoimprinting conditions. We see that the sharp threshold anomalies (grey dashed lines) remain in the same positions for all three gratings, barring small changes in the periodicity due to the shrinkage of ETFE. However the width and positions of the resonance anomaly dips vary greatly with geometry. As shown in the changing the grating geometry during modelling, square-wave gratings produce the widest resonance anomalies, and sinusoidal gratings the sharpest resonances. We also observe a dispersionless mode in the reflectivity at ~ 450 nm for the more sinusoidal gratings. This could be due to the presence of a channel plasmon, which requires a narrowing of the gap as seen in Fig. 7.10(c) in particular.

7.5.2 PS-coated Ag gratings

Fig. 7.11(a) shows the AFM image of a PS-coated Ag grating, where the grating structure is clearly observable, however only with an average height of 5 nm. From this we deduce

the Ag grating is almost submerged beneath the non-uniform PS layer [Fig. 7.11(b)]. The presence of the PS overcoating increases the complexity of the reflectivity spectra by allowing more modes to be accessed. In TM polarisation, both photonic (grey dashed lines) and redshifted plasmonic modes (black dashed lines) can be observed [Fig. 7.11(c)]. Both of these modes are also observed in TE polarisation [Fig. 7.119e)], although the photonic mode is much weaker. At high ϕ in TM polarisation, a potential second set of plasmonic modes can be seen (black dot-dashed lines), indicating two refractive index environments, perhaps the top and bottom surface of the grating where the PS coverages are different. We also see a broader mode at ~ 520 nm for $\phi = 0^\circ \theta = 10^\circ$, which redshifts slightly with increasing ϕ while maintaining a similar shape (purple dashed line). A similar mode can be observed in TE polarisation, although not at exactly the same position.

Using FEM modelling, we observe two types of modes at $\phi = 0^\circ$ at TM polarisation. A mode at higher energy has field intensity concentrated at the top surface of the grating, evanescently decaying from the Ag surface [Fig. 7.12(a)]. Taking snapshots throughout the optical cycle, the mode appears to be a quasiparticle travelling along the top surface of the grating as expected in the $\phi = 90^\circ$ geometry [Fig. 7.12(b)]. By varying the geometry of the grating, we see that the mode decreases in energy as D increases, increases in energy with the slit width, and is unaffected by grating height, thus showing the behaviour expected for an SPP mode. The field intensity of the lower energy mode is mainly concentrated in the slit of the grating [Fig. 7.12(c)], and again appears to travel along the slit throughout the optical cycle [Fig. 7.12(d)]. The energy of the mode is unaffected by D or the grating height, and decreases as the slit width increases. This behaviour for a mode waveguided by the grating slit, and according to Eq.[waveguide] the dispersion fits that of a TE_{10} mode.

Both SPP and waveguided modes are very sensitive to the dielectric environment as shown by Eqs.[grating,waveguide], and Fig. 7.13 shows the change in these grating modes with increasing PS coverage. We can see both the narrower SPP resonances (black dashed lines) and broader waveguided modes (purple dashed lines) redshift with increasing PS coverage as expected. For the structure in Fig. 7.13(c) these two modes actually overlap, although no interaction occurs between them.

7.5.3 CHPI-coated Ag gratings

Fig. 7.14 shows TM reflectivity scans for CHPI-coated Ag gratings with periodicities $D = 556, 417, 278$ nm at $\phi = 0$ and 90° . The spectra for $D = 556$ and 417 nm are particularly similar, showing two exciton modes (red arrows), which strongly couple to the SPP grating

mode (black) as the oscillations become resonant at $\phi = 90^\circ$. Two excitons can also be observed for $D = 278$ nm, however the SPP mode is at a higher energy than for higher D and thus the coupling occurs at $\phi = 0^\circ$.

TM polarised reflectivity scans of a CHPI-coated Ag grating at $\phi = 0^\circ$ [Fig. 7.15(a)] show two dispersionless exciton modes at 480 and 500 nm (marked by arrows) far off resonance with grating modes. The persistent presence of a second exciton is only detected when SPPs can be excited, i. e. in TM polarisation [Fig. 7.15(a)] but not TE [Fig. 7.15(b)], nor in CHPI-coated planar Ag films [Fig. 7.15(c)]. It is also not observed for CHPI-coated non-plasmonic gratings, as seen in the CHPI-coated ETFE and Ti gratings. From Fig. 7.15 we see that SPP excitation leads to the observation of an additional redshifted exciton with a splitting of 100 meV. Its appearance only when SPPs are present rules out any influence from modified CHPI assembly in the grooves, which are in any case hundreds of times larger than the PbI layer spacing. We note slight changes in the CHPI coverage alter the positions and intensities of dispersive grating modes [*cf* Fig. 7.17(c), with a thinner CHPI coating], however the exciton modes remain essentially unchanged.

It is well known that the emitted energy of a dipole (exciton) is lowered when placed in front of a metallic surface due to interactions between the dipole and the reflected electromagnetic field [180–185]. Using the method of images, we can replace the metal and describe instead the coupling between an exciton in the CHPI (ϵ_1) and its image exciton in the metal (ϵ_2), modified by their respective dielectric environments. Chance *et al.* [183] showed the redshift in the emitted energy of an exciton (ΔE_{ex}) oriented parallel to the interface can be approximated by

$$\Delta E_{ex} \sim \left(\frac{k_1}{l} \right)^3 \text{Re} \left\{ \frac{\epsilon_2 - \epsilon_1}{\epsilon_2 + \epsilon_1} \right\} \Gamma_0, \quad (7.1)$$

where l is the distance between the exciton and a metal surface, k_1 is the wavenumber of light in CHPI, and Γ_0 is the inverse exciton radiative lifetime without the metal. From this we can see the l^{-3} dependence of the redshift as shown in Fig. ??(d), where the experimentally observed $\Delta E_{ex} \sim 100$ meV corresponds to $l \sim 22$ nm, close to the experimentally-determined CHPI thickness. Clearly ΔE_{ex} is also affected by the dielectric response of CHPI and Ag, and from Eq. 7.1 we see that ΔE_{ex} is maximised if $\epsilon_2 + \epsilon_1 \rightarrow 0$, i. e. when emission is resonant with an SPP on the metal-dielectric interface.

The role of the SPP in this case is to outcouple the signal of the redshifted exciton. There are three main decay channels for dipole emission near a metal surface: direct emission to photons, emission to SPPs, and nonradiative processes such as the excitation of electron-

hole pairs and lossy surface waves on the metal. Other nonradiative paths via defects or phonons are independent of l and will be ignored in this analysis. Emission into SPPs provides an extra radiative decay channel as this signal can be extracted to the far field via the periodic nanostructure, and this mechanism has been used to improve the luminescence efficiency of light emitting devices [186, 187]. The relative decay probability for each process is calculated as a function of l [185] and shown in Fig. 7.16(a). Up to a CHPI thickness of 25 nm, SPP mediated emission is the most important radiative decay channel with a maximum emission probability at 22 nm, matching the experimentally observed ΔE_{ex} . Even for thicker CHPI films we expect the exciton modes to remain at the same positions, because SPP emission becomes weak at large l where ΔE_{ex} is negligible.

In our MQW perovskite system, localised excitons in periodically-spaced nearby QWs are optically coupled together to form collective exciton-polariton states an average distance l from the Ag surface [188–191]. Therefore in CHPI-coated Ag gratings we observe both in-plane exciton-polaritons, and out-of-plane interactions that lead to ‘image-biexcitons’, which are outcoupled via SPP emission with a binding energy of 100 meV at room temperature [Fig. 7.16(b)]. For our grating system, the exciton and SPP modes become closer in energy with increasing ϕ [see below and Fig. 7.17(c)], and as a result splitting between the exciton modes (indicated by arrows in Fig. 7.17(c)) increases to around 185 meV at $\phi = 90^\circ$. The azimuthal dependence of the exciton splitting reflects the tuneable modification of the Coulomb interaction in this geometry, but however requires further theoretical development.

AFM image of the CHPI-coated Ag grating clearly shows the grating structure, although the CHPI coating is rather non-uniform [Fig. 7.17(a)]. Using AFM measurements, we find the CHPI forms a conformal coating around the Ag grating with thickness ~ 25 nm [Fig. 7.17(c)]. In TE polarisation, we only observe the presence of one exciton, without signature of any grating modes, similar to the CHPI-coated Ti gratings. In TM polarised reflectivity scans, as well as strong excitons (red arrows) we also observe $m = \pm 1$ photonic and plasmonic grating modes [Fig. 7.17(c)]. As the SPP modes become resonant with the exciton and image exciton, the light-matter modes strongly couple and produce an anti-crossing in the reflectivity of 0.25 eV. Extracting the mode positions from the $\phi = 90^\circ$ scan [Fig. ??(d)] allows them to be fit to a three oscillator model using the Hamiltonian

$$\hat{H} = \begin{pmatrix} E_{ex} & 0 & \Omega_{ex}/2 \\ 0 & E_{bx} & \Omega_{bx}/2 \\ \Omega_{ex}/2 & \Omega_{bx}/2 & E_{pl} \end{pmatrix}, \quad (7.2)$$

where E_{ex} , E_{bx} and E_{pl} are the energies of the exciton-polariton, image-biexciton and plasmonic grating modes respectively, while Ω_{ex} and Ω_{bx} represent the interaction between the SPP and exciton/image-biexciton. From this we find Rabi splittings of $\Omega_{ex} = 150$ meV and $\Omega_{bx} = 125$ meV. These are greatly enhanced because of the large confinement of the plasmonic optical field in the thin PbI QW layers. The Rabi splitting is given by $\Omega \propto \sqrt{f_{osc}N_{QW}/V}$, where the oscillator strength (f_{osc}) of the CHPI is assumed to be similar for coupling to photons or plasmons, the number of QWs (N_{QW}) is proportional to the CHPI thickness, and the mode volume (V) is here proportional to the optical mode size. Comparing to Fabry-Perot planar CHPI microcavities in strong coupling [19] which have CHPI thickness of 72 nm, cavity length of 407 nm, and a Rabi frequency of $\Omega_{FP} = 65$ meV, the simple scaling above predicts $\Omega_{SPP} \sim \Omega_{FP} \sqrt{(22/72).(407/22)} = 156$ meV, in excellent agreement with our measurements. Using SPPs to strongly couple to the excitons thus dramatically reduces the cavity length, thus enhancing the light-matter coupling.

We calculate the full eigenstates of the system using FEM simulations. These confirm the anticrossings observed, and provide the optical field profiles. In the case of strong coupling at $\phi = 90^\circ$, the time-averaged near-field shows strongest intensity inside the CHPI which coats the bottom surface of the grating, with a rapid evanescent decay away from the interface [Figs. ??(b,c)]. The mode is thus both laterally confined by the grating as well as being trapped inside the surface layers where it couples to the excitons.

The SPP E -field direction is primarily perpendicular to the metal-dielectric interface, while excitons in CHPI QWs are polarised parallel to this interface [19, 49]. Simulated $\phi = 90^\circ$ spectra for in- and out-of-plane exciton dipoles are shown in Figs. ??(d,e) respectively. While strong coupling is seen for both dipole orientations, the bare exciton is only seen for the in-plane dipole. It thus appears that the coupling between the excitons and their images are responsible for mixing the dipole orientations, enabling the strong coupling with the SPP mode. The polariton states mix excitons within the perovskite which are delocalised across many PbI monolayers, with SPPs which are tightly confined to the CHPI layer above the Ag grating and laterally localised in the grating slits by the coupling of standing waves. Such light-matter polaritonic quasiparticles thus combine organic, inorganic and plasmonic components in an unusual fashion.

7.6 Conclusions

A simple 1D plasmonic periodic structure can give host to a wide range of grating modes: photonic modes due to the interference of light, excitement of SPPs on the surface of the

metal, or laterally localised modes such as channel plasmons or waveguide modes. The positions and efficiencies of many of these modes depends sensitively on the geometry of the grating and the dielectric environment provided by any overcoating materials. Thus it is not always easy to predict what can be observed in optical spectra.

In CHPI-coated Ag gratings, we observe evidence of image-biexcitons with binding energy 100 meV at room temperature. Such quasiparticles arise from the interaction between excitons and their images in the metal, and are outcoupled from the grating structure via SPP emission. These out-of-plane biexciton states mediate coupling between in-plane QW excitons and out-of-plane SPP grating modes. This enables the observation of strong coupling at room temperature with Rabi splittings of 150 and 125 meV for the exciton and image-biexciton respectively. Both the biexciton binding energy and strong coupling Rabi splitting is tuneable by small changes in the structure of the coated gratings.

Strong coupling has previously been observed between inorganic or organic excitons and Au nanoslit gratings at low temperature. The coupling constants in these systems are much smaller compared to CHPI at room temperature: 55 meV for 50 nm J-aggregate films at 77 K [192], and 8 meV for 10 nm GaAs QWs at 10 K [193]. One key difference is that for the III-V semiconductors the QWs have to be spaced at least 20 nm from the metal surface to maintain their optical quality. In contrast our 25 nm thick CHPI film is prepared directly on the metal, and still gives strongly radiative exciton modes because the organic sandwich protects the PbI QW layers. Theoretically Fig. ??(d) shows that excitons remain radiative via SPP coupling for film thickness above 10 nm. Hence the perovskite system is well suited to manipulate light-matter interactions. Such modification of the exciton and SPP states is of great interest for future optoelectronic devices, involving ultrathin semiconductors including the layered van der Waals materials based on graphene and dichalcogenides.

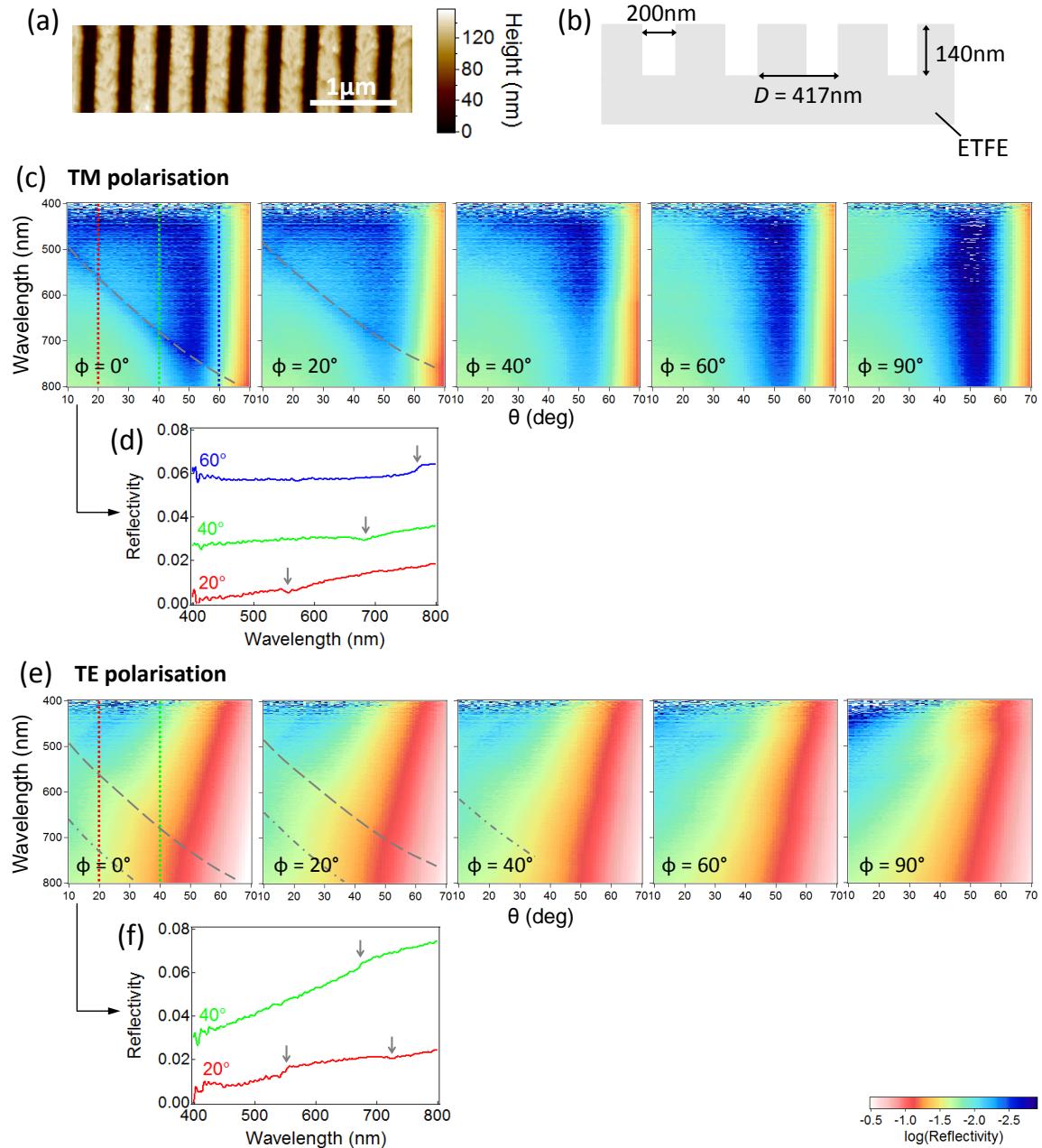


Fig. 7.2 (a) AFM image and (b) schematic of ETFE grating structure. (c) TM polarised reflectivity scans of $D = 417\text{ nm}$ ETFE grating at labelled ϕ , and (d) reflectivity spectra at indicated θ values for $\phi = 0^\circ$. (e,f) Same as above for TE polarisation. Photonic grating modes are indicated by grey lines/arrows on reflectivity scans/spectra respectively.

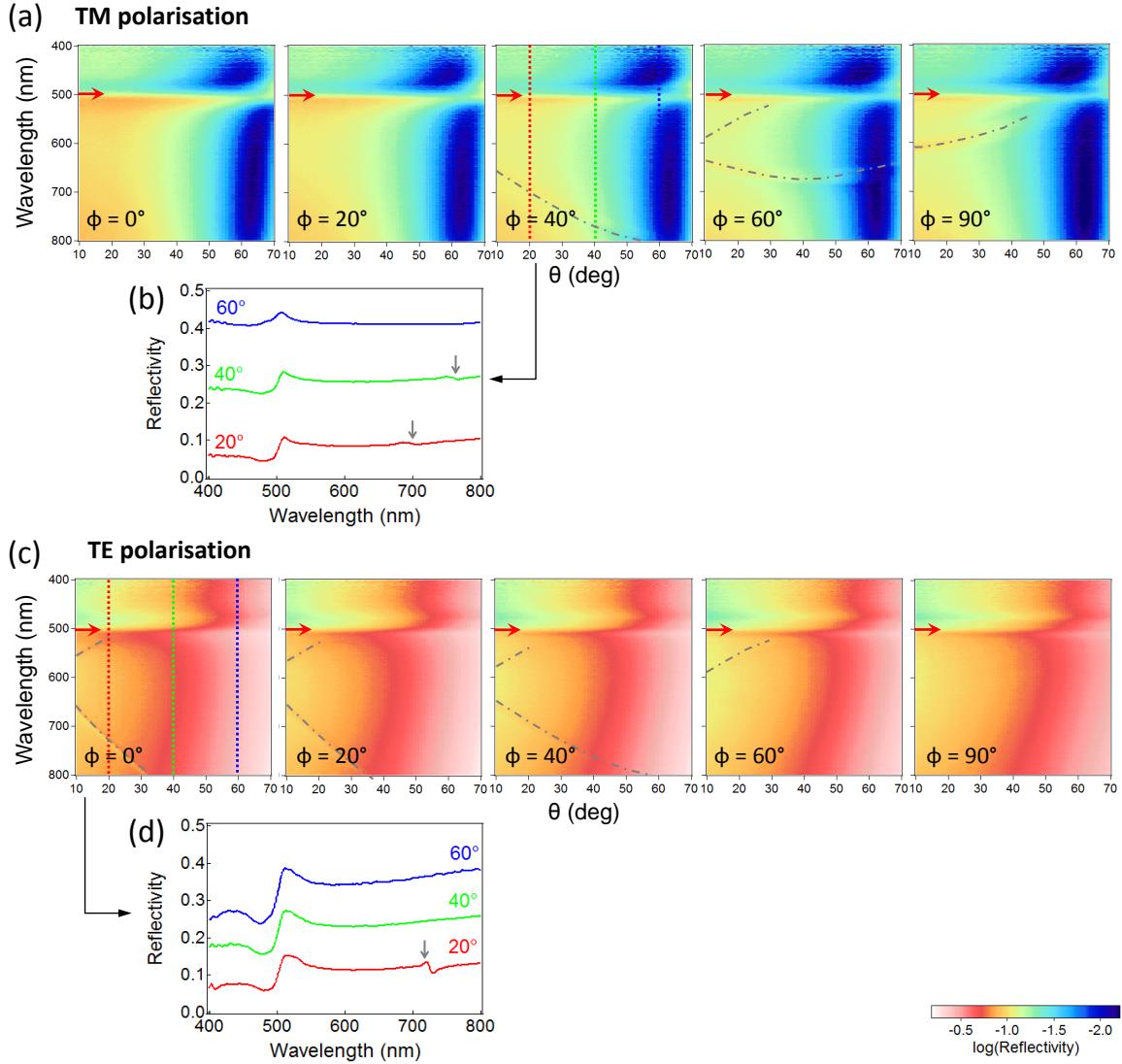


Fig. 7.3 (a) TM polarised reflectivity scans of $D = 417$ nm CHPI-coated ETFE grating at labelled ϕ , and (b) reflectivity spectra at indicated θ values for $\phi = 0^\circ$. (c,d) Same as above for TE polarisation. Photonic grating modes are indicated by grey lines/arrows on reflectivity scans/spectra respectively, and excitons by red arrows.

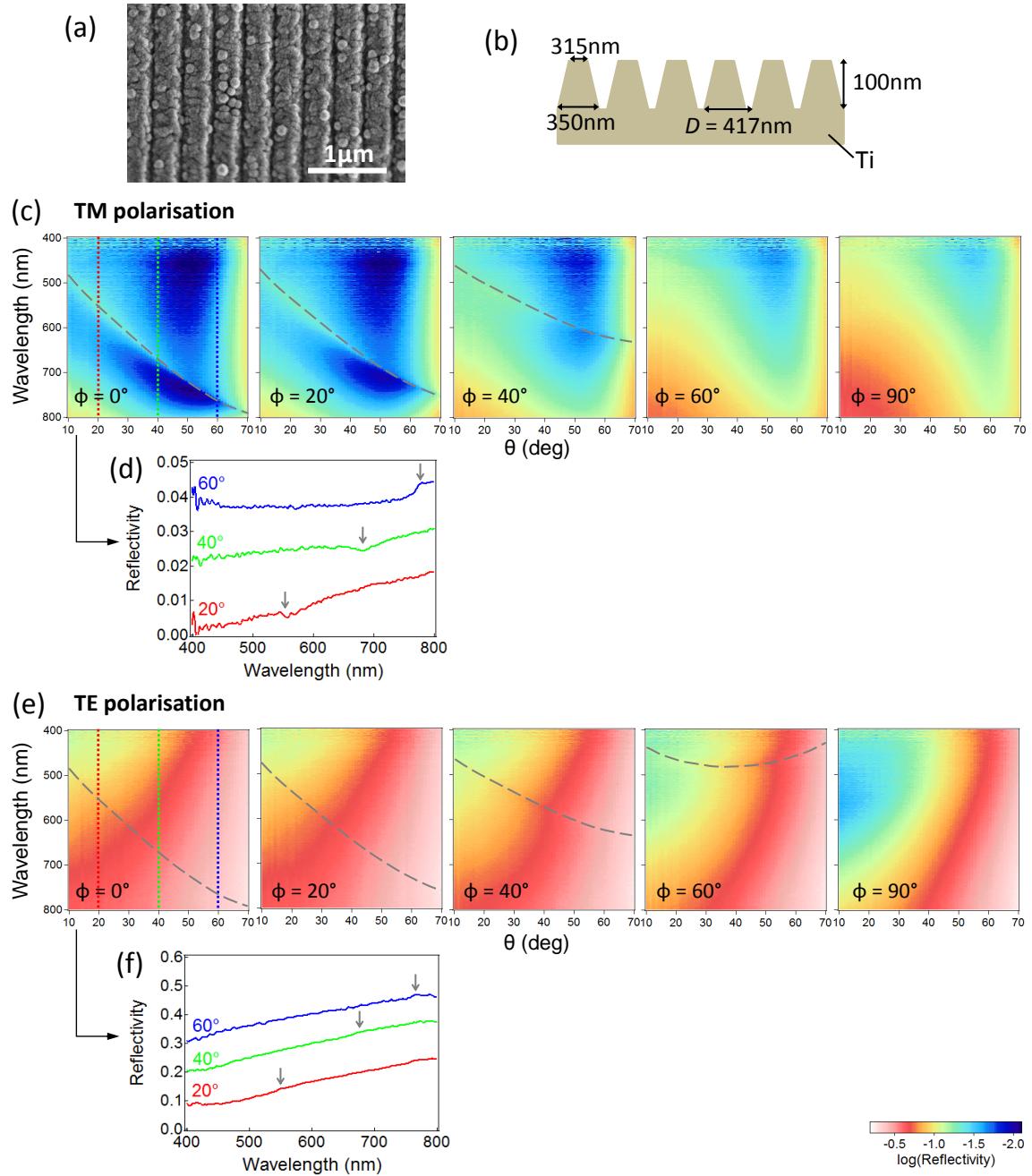


Fig. 7.4 (a) SEM image and (b) schematic of Ti grating structure. (c) TM polarised reflectivity scans of $D = 417\text{ nm}$ Ti grating at labelled ϕ , and (d) reflectivity spectra at indicated θ values for $\phi = 0^\circ$. (e,f) Same as above for TE polarisation. Photonic grating modes are indicated by grey lines/arrows on reflectivity scans/spectra respectively.

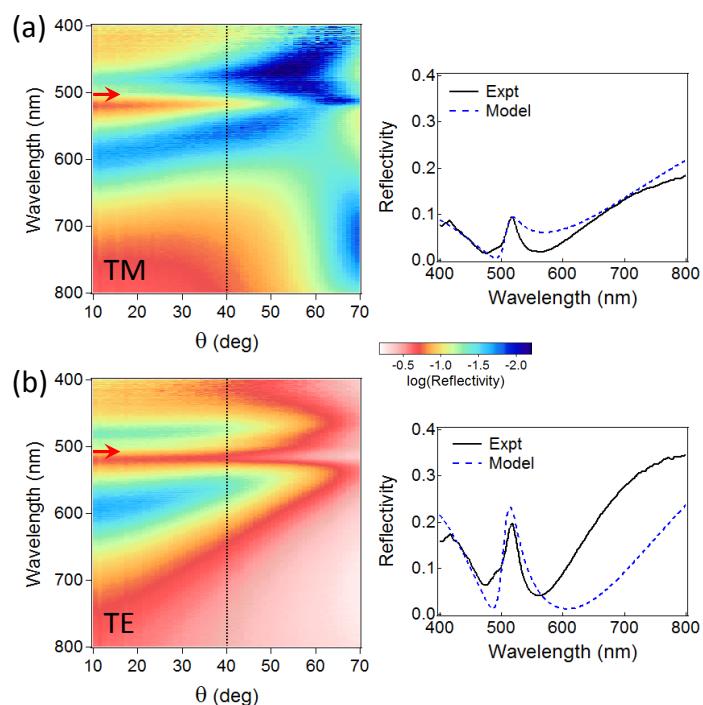


Fig. 7.5 Specular reflectivity scans of 70 nm CHPI film on 120 nm planar Ti film with (a) TM and (b) TE polarised light. Excitons are marked by red arrows. Spectra at $\theta = 40^\circ$ are plotted with those predicted by transfer matrix simulations (right).

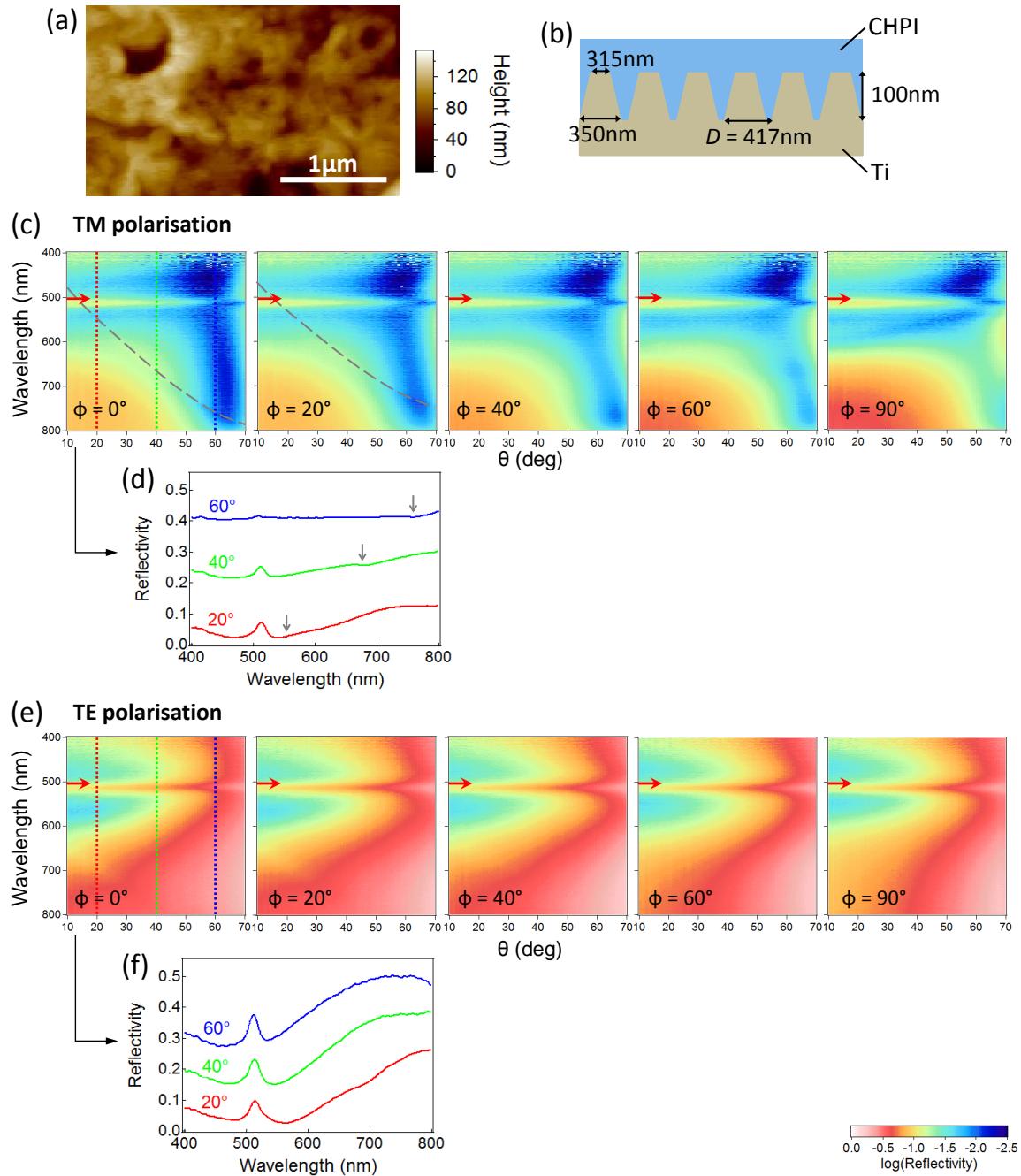


Fig. 7.6 (a) AFM image and (b) schematic of CHPI-coated Ti grating structure. (c) TM polarised reflectivity scans of $D = 417$ nm CHPI-coated Ti grating at labelled ϕ , and (d) reflectivity spectra at indicated θ values for $\phi = 0^\circ$. (e,f) Same as above for TE polarisation. Photonic grating modes are indicated by grey lines/arrows on reflectivity scans/spectra respectively, and excitons by red arrows.

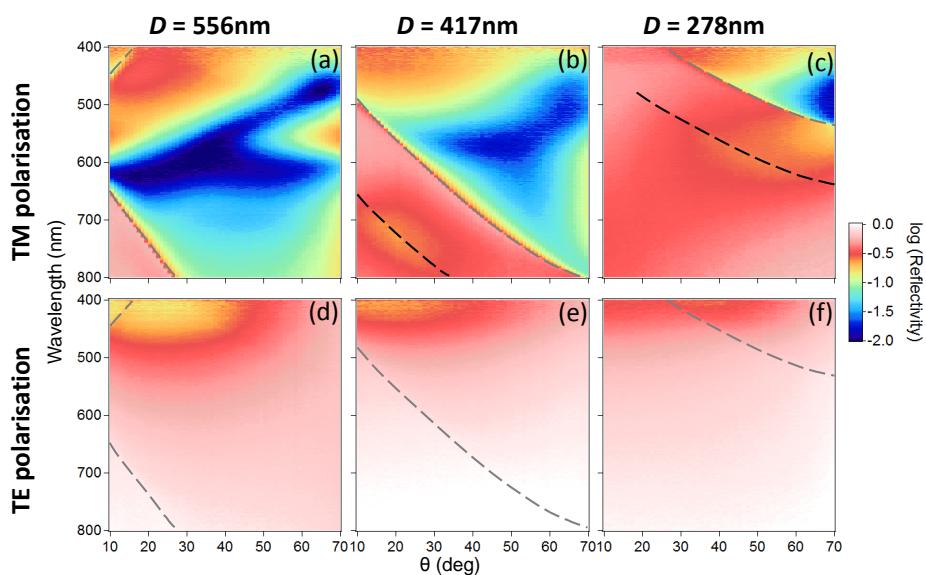


Fig. 7.7 Specular reflectivity scans of uncoated Ag gratings with the labelled periodicity D and polarisation of light. Photonic grating modes (threshold anomalies) are marked by grey dashed lines, and plasmonic grating modes (resonance anomalies) are marked by black dashed lines.

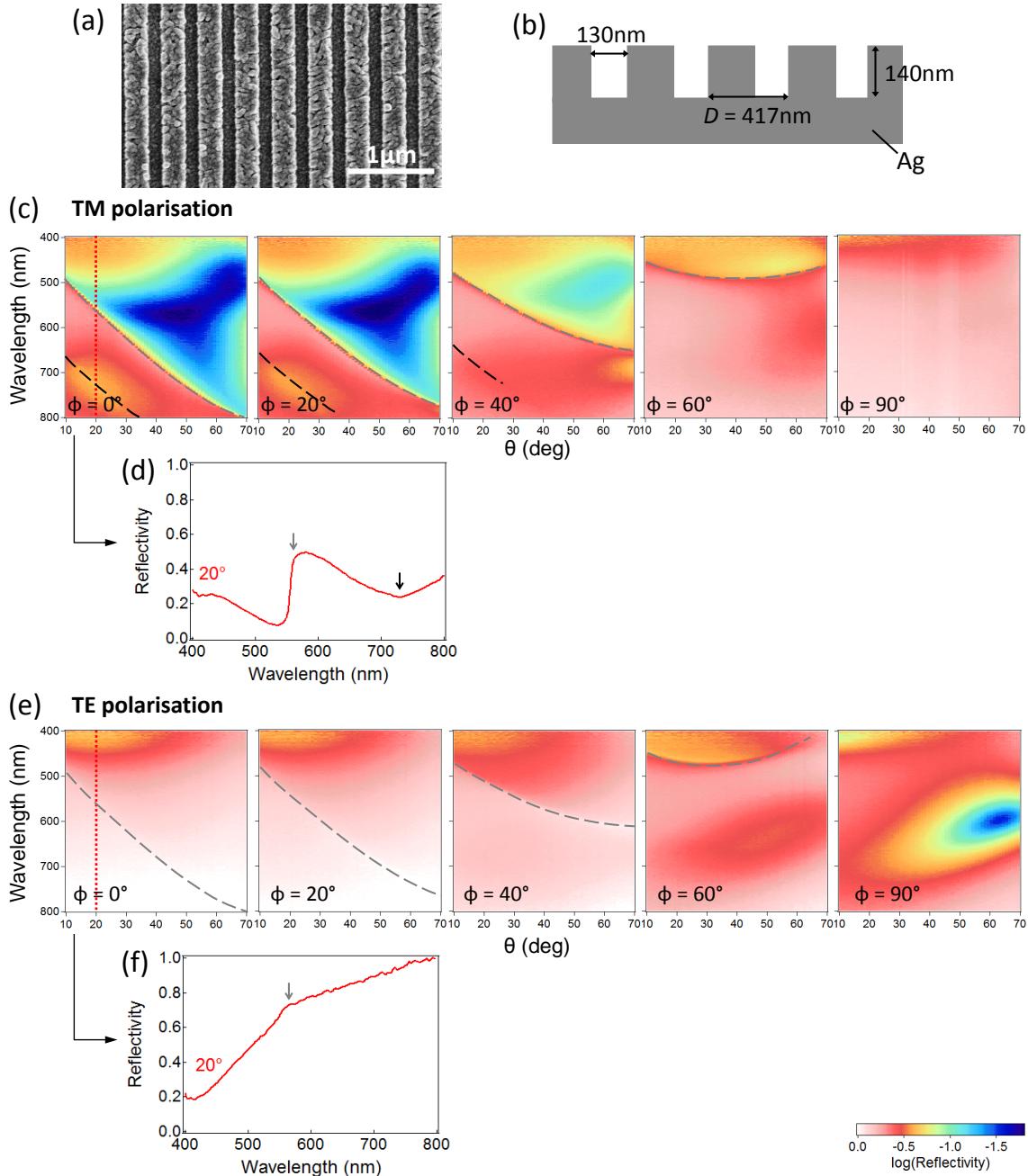


Fig. 7.8 (a) SEM image and (b) schematic of Ag grating structure. (c) TM polarised reflectivity scans of $D = 417$ nm Ag grating at labelled ϕ , and (d) reflectivity spectra at indicated θ values for $\phi = 0^\circ$. (e,f) Same as above for TE polarisation. Photonic grating modes (threshold anomalies) are indicated by grey lines/arrows on reflectivity scans/spectra respectively, and plasmonic gratings modes (resonance anomalies) by black lines/arrows.

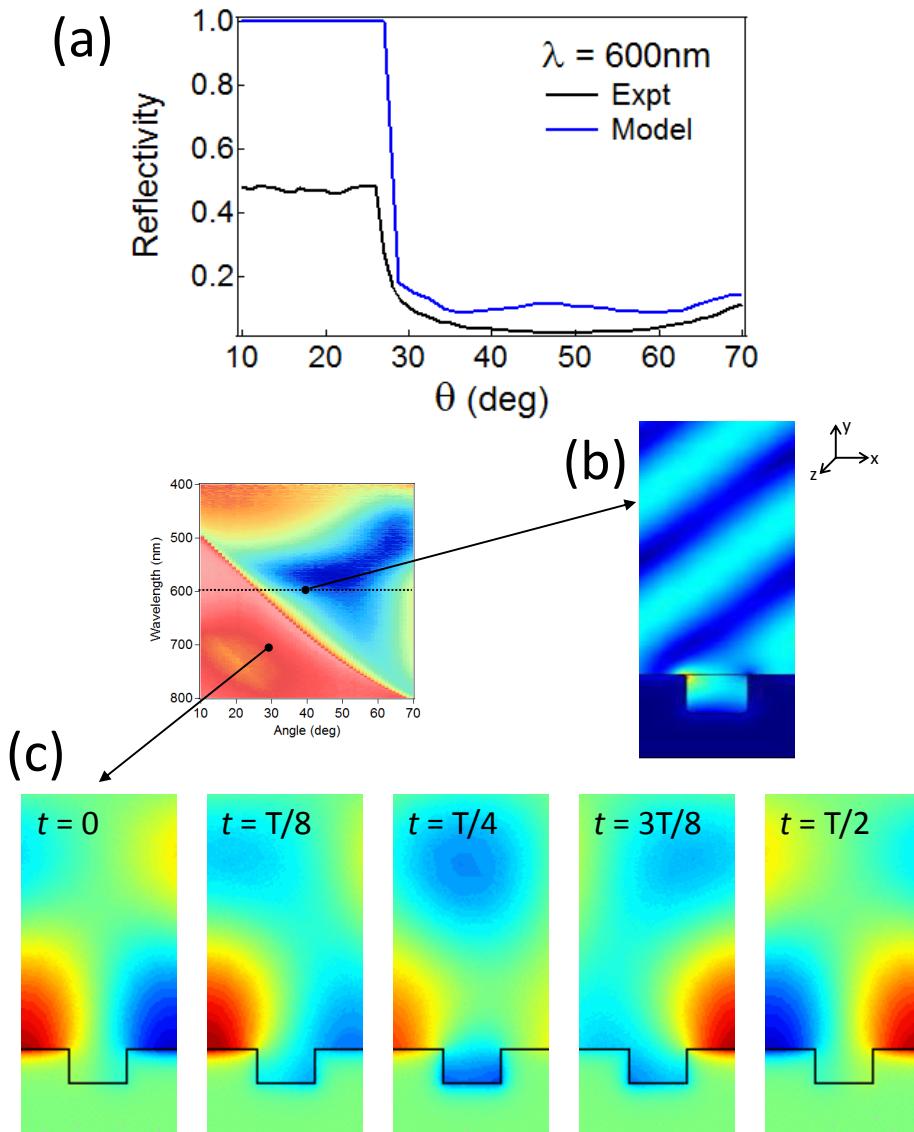


Fig. 7.9 (a) Experimental and modelled TM reflectivity spectra for $\phi = 0^\circ$. (b) Time averaged E -field intensity ($\vec{E} \cdot \vec{E}$) profile for $\lambda = 600\text{ nm}$ $\theta = 40^\circ$. (c) H_z nearfield profile at labelled points of the optical cycle for $\lambda = 700\text{ nm}$ $\theta = 30^\circ$.

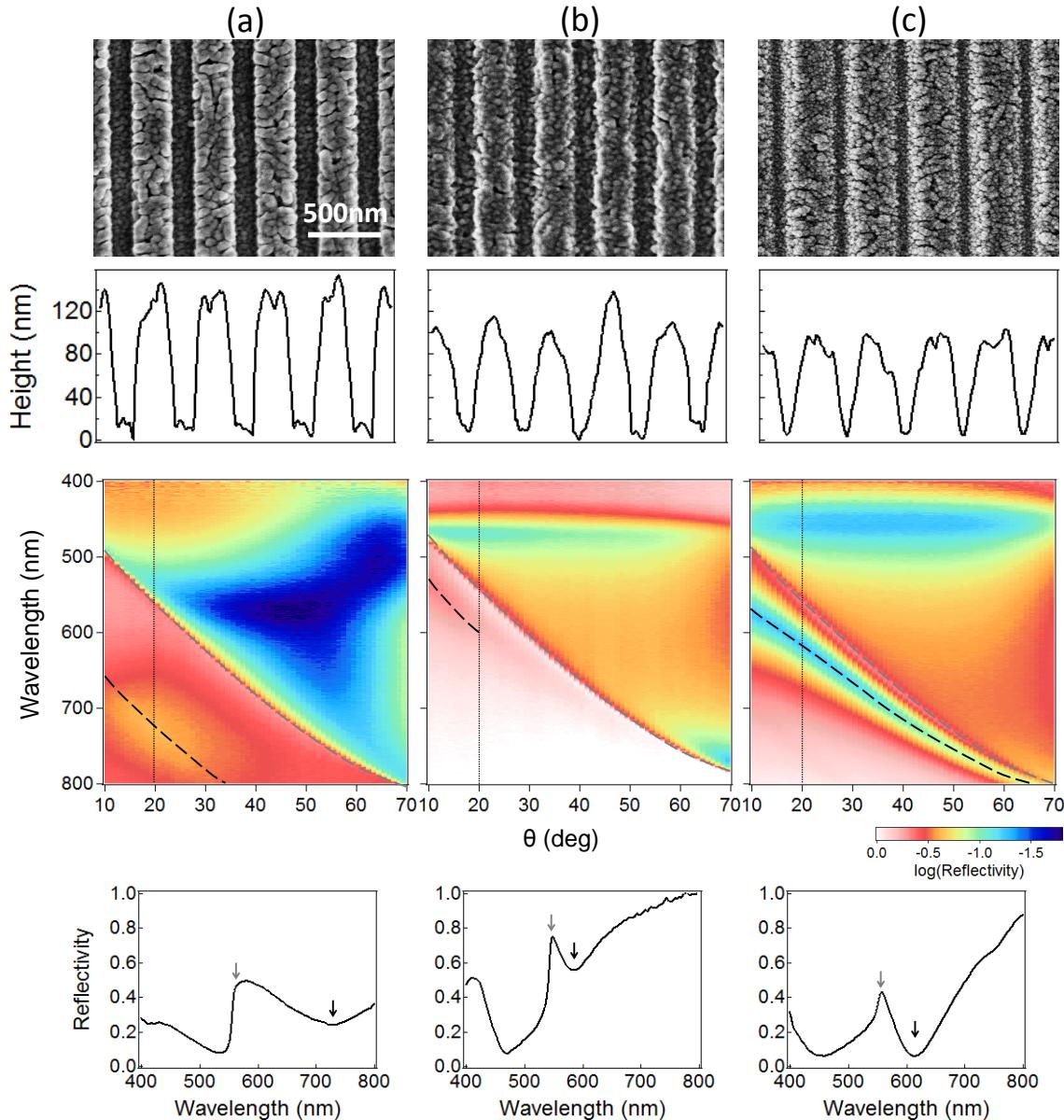


Fig. 7.10 Effect of grating geometry on the optical spectra of $D = 417$ nm Ag gratings. (From top) SEM image, AFM profile, TM specular reflectivity scans at $\phi = 0^\circ$, and reflectivity spectra at $\theta = 20^\circ$ for three gratings. Photonic grating modes (threshold anomalies) are marked by grey dashed lines/arrows, and plasmonic grating modes (resonance anomalies) by black dashed lines/arrows on reflectivity scans/spectra respectively.

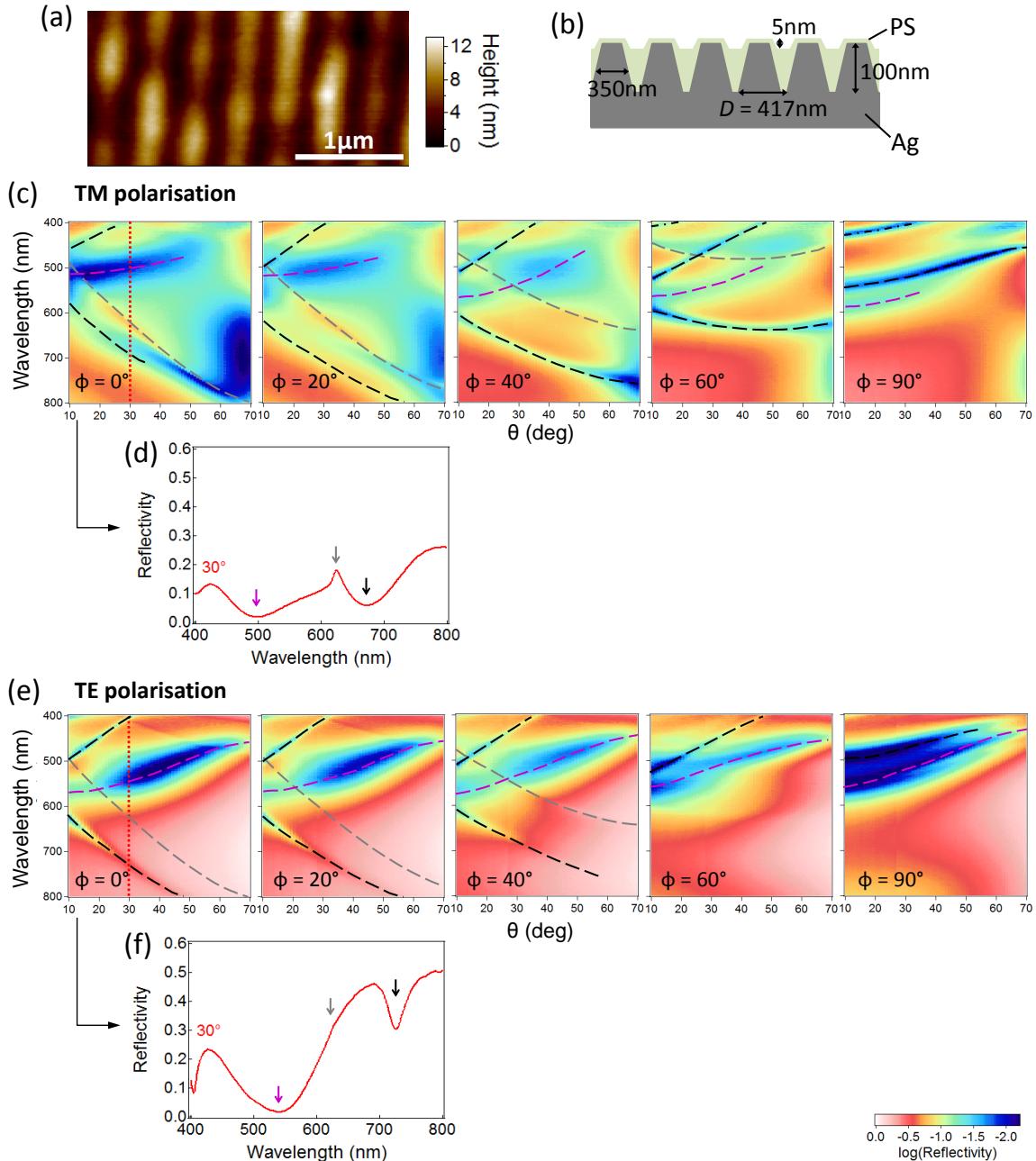


Fig. 7.11 (a) AFM image and (b) schematic of PS-coated Ag grating structure. (c) TM polarised reflectivity scans of $D = 417 \text{ nm}$ PS-coated Ag grating at labelled ϕ , and (d) reflectivity spectra at indicated θ value for $\phi = 0^\circ$. (e,f) Same as above for TE polarisation. Photonic grating modes are indicated by grey lines/arrows on reflectivity scans/spectra respectively, plasmonic gratings modes by black lines/arrows, and waveguide modes by purple lines arrows.

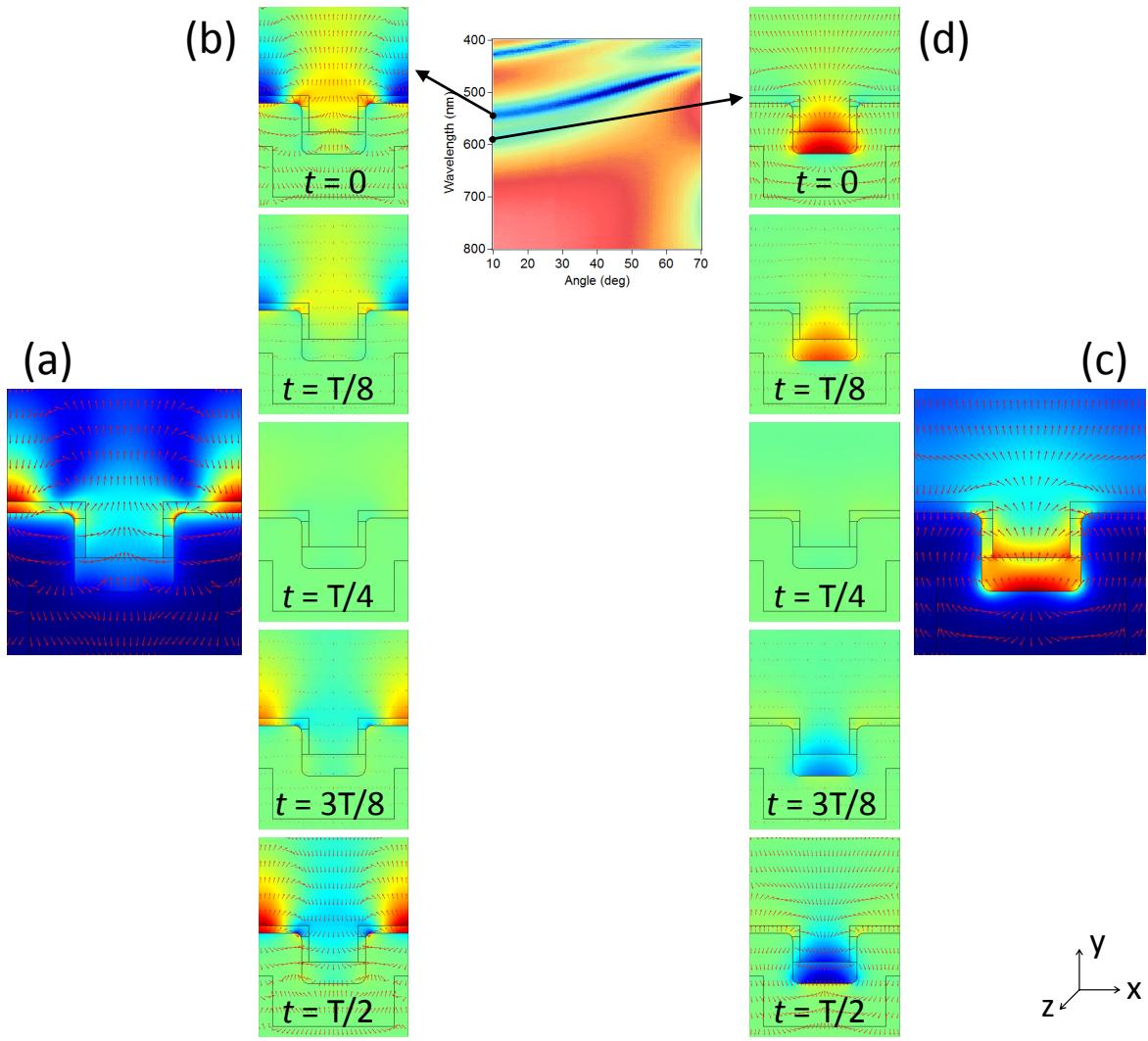


Fig. 7.12 (a) Time averaged and (b) snapshots at time t in the optical cycle T of E_y nearfield intensity of the higher energy (plasmonic) mode at $\phi=0^\circ$, $\theta=0^\circ$. (c,d) same as above for the lower energy (waveguide) mode. The arrows on the diagrams indicate the Poynting vector (?).

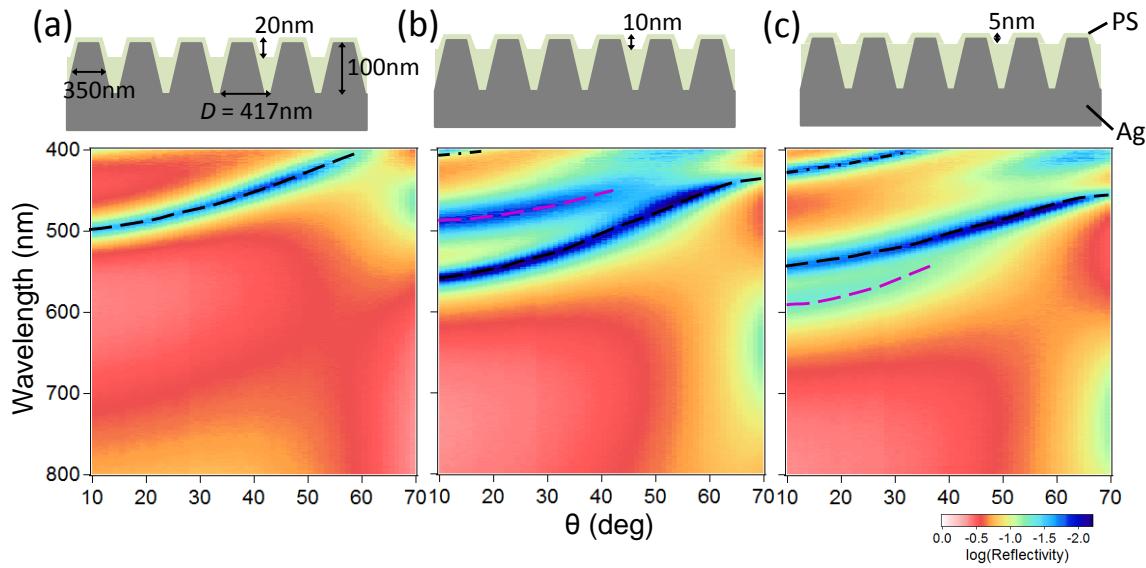


Fig. 7.13 Schematic of PS-coated $D = 417$ nm Ag grating structure (above), and TM polarised reflectivity scans at $\phi = 90^\circ$. Plasmonic modes are indicated by black lines, and waveguide modes by purple lines.

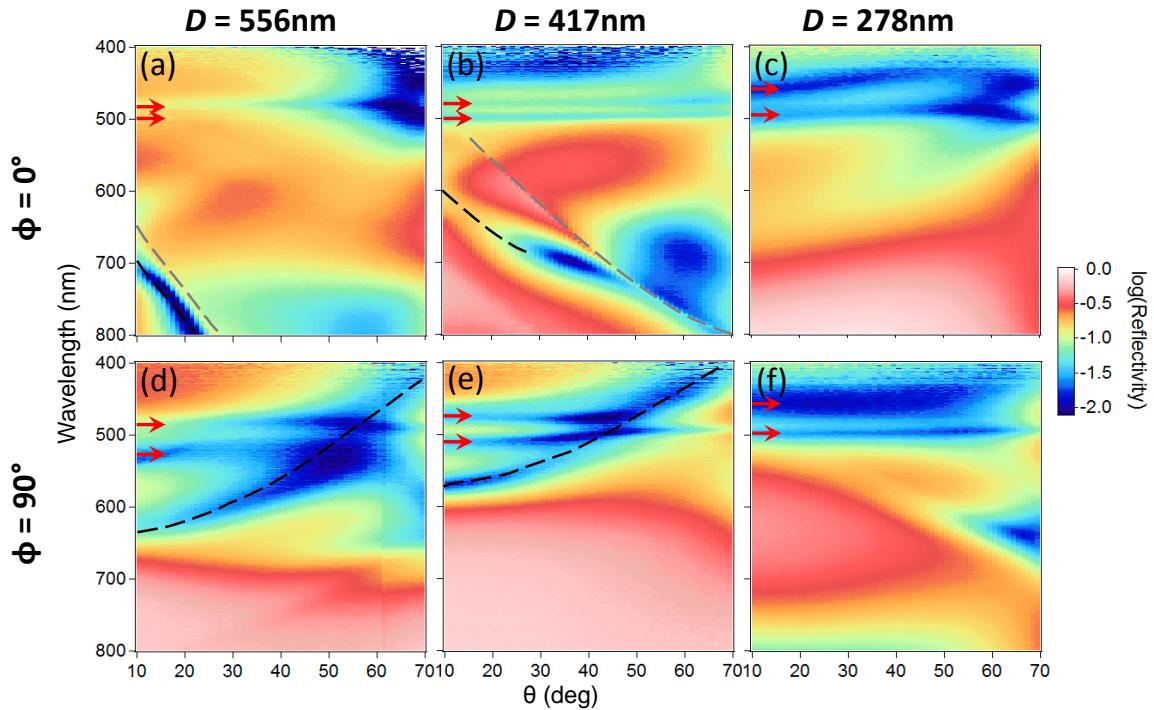


Fig. 7.14 TM specular reflectivity scans of CHPI-coated Ag gratings with the labelled periodicity D and at the labelled ϕ . Photonic grating modes are marked by grey dashed lines, plasmonic grating modes are marked by black dashed lines, and excitons indicated by red arrows.

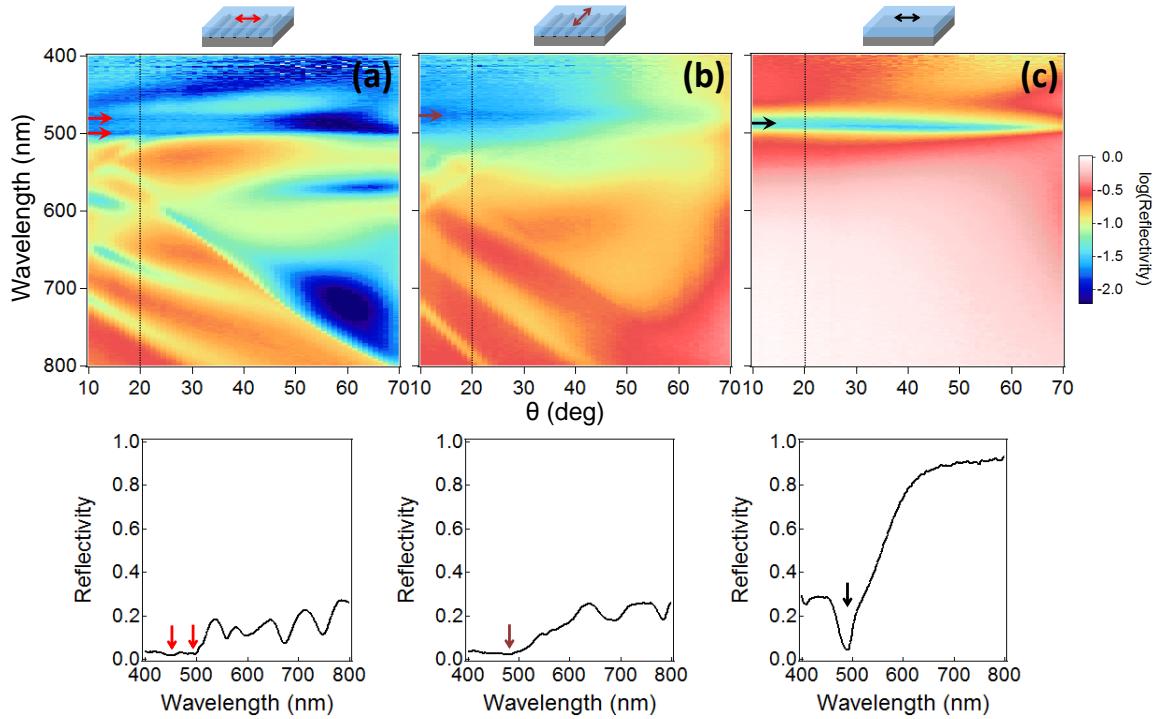


Fig. 7.15 Specular reflectivity scans (above) at $\phi = 0^\circ$: CHPI-coated Ag grating with (a) TM and (b) TE polarised light, and (c) CHPI-coated 120 nm planar Ag film with TM polarised light; the reflectivity spectra at $\theta = 20^\circ$ is shown below. The electric field orientation is shown above each scan, and positions of exciton modes indicated by arrows.

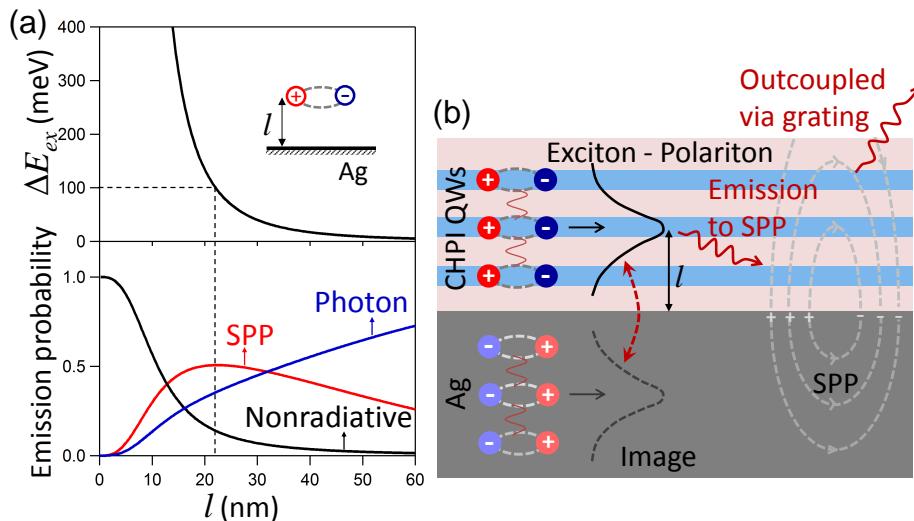


Fig. 7.16 (a) Change in emitted energy (top) and relative decay probabilities (bottom) of an exciton with energy 2.6 eV placed distance l from the Ag surface. The dashed line indicates the experimentally measured redshift. (b) Schematic mechanism for SPP-mediated emission of image-biexciton.

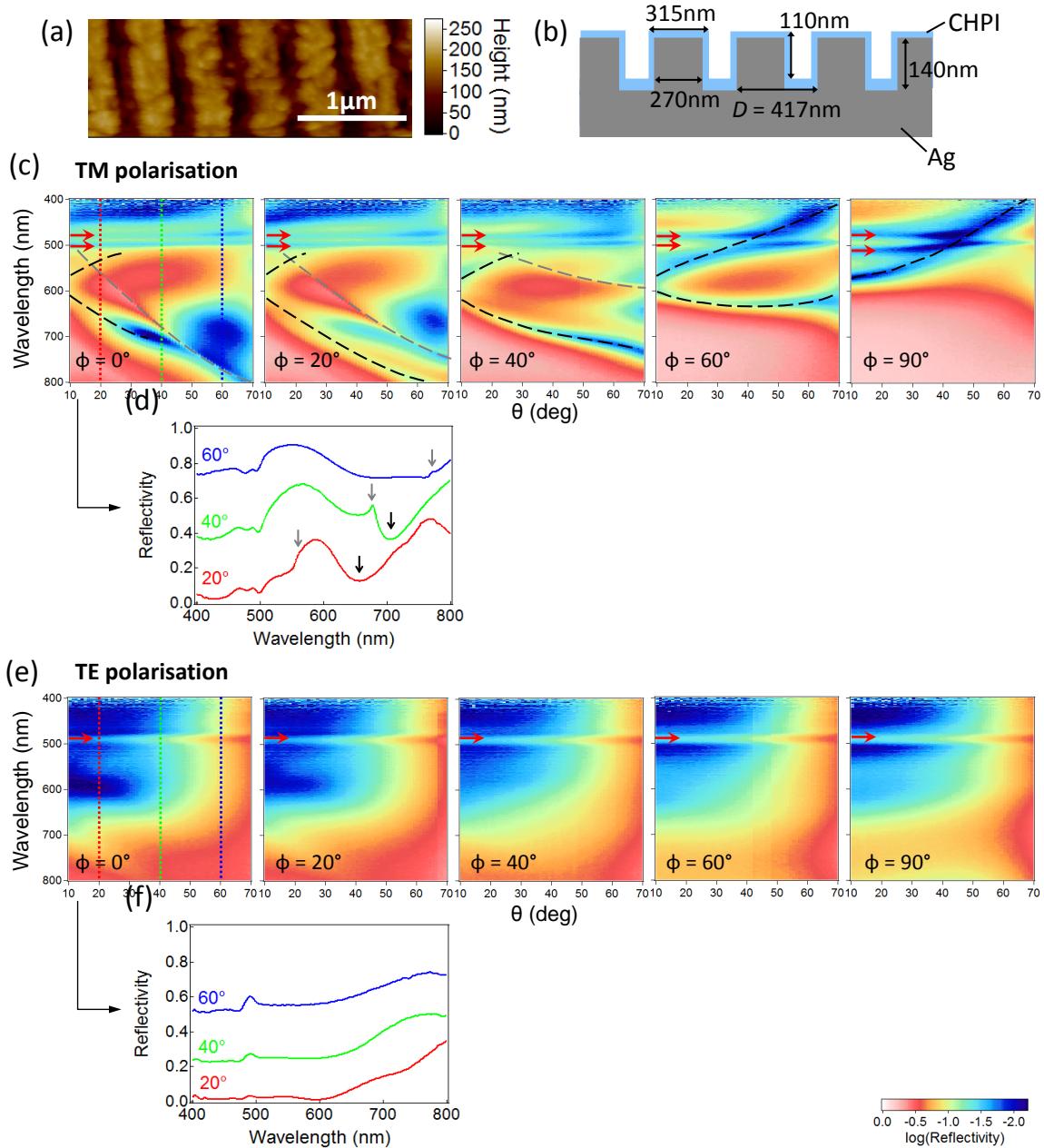


Fig. 7.17 (a) AFM image and (b) schematic of CHPI-coated Ag grating structure. (c) TM polarised reflectivity scans of $D = 417$ nm Ag grating at labelled ϕ , and (d) reflectivity spectra at indicated θ value for $\phi = 0^\circ$. (e,f) Same as above for TE polarisation. Photonic grating modes are indicated by grey lines/arrows on reflectivity scans/spectra respectively, plasmonic gratings modes by black lines/arrows, and excitons by red arrows.

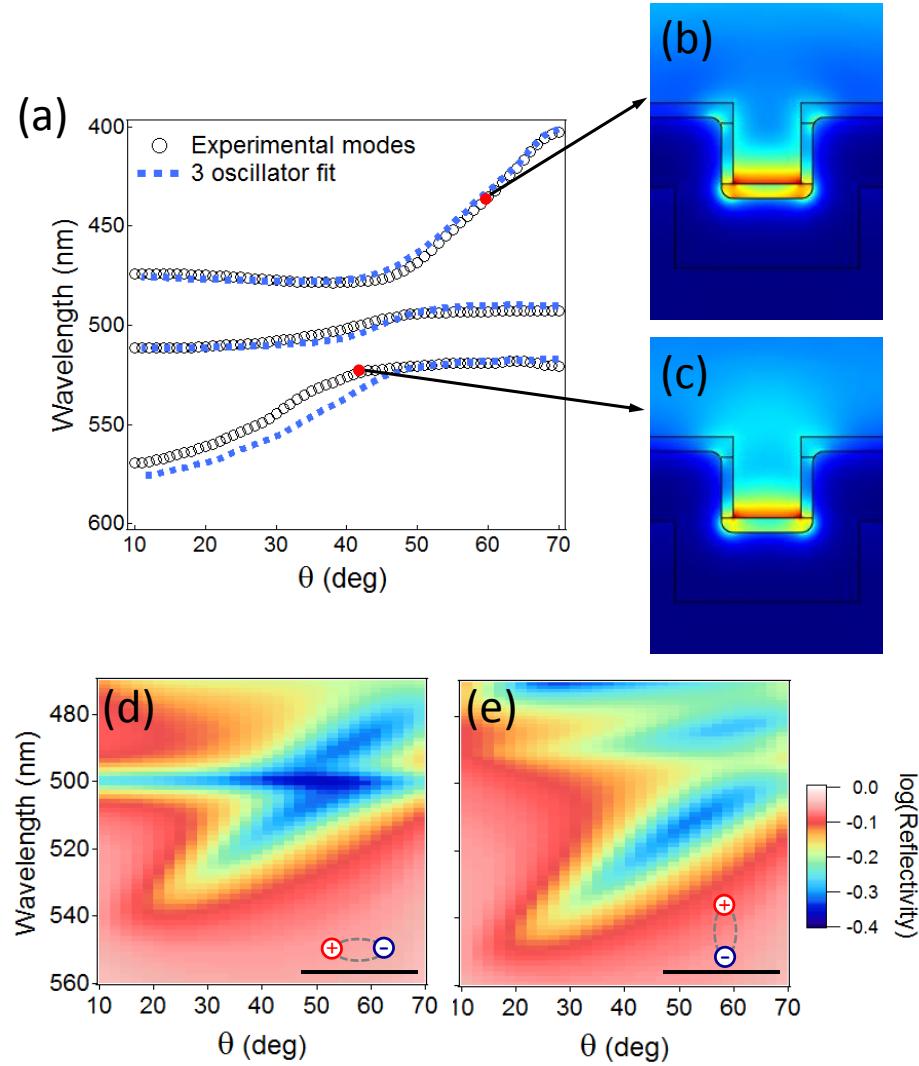


Fig. 7.18 (a) Extracted spectral mode positions for $\phi = 90^\circ$ reflection dips (open circles), and fit from three oscillator coupling model (dashed lines). (b,c) Time-averaged E -field intensity profiles ($\vec{E} \cdot \vec{E}$) as indicated. (d,e) Simulated reflection spectra for (d) in-plane and (e) out-of-plane exciton dipoles.

Chapter 8

Conclusions and further work

Metal-halide based organic-inorganic perovskites are self-assembling semiconductors with many flexible and tunable properties. The exact structure formed depends on the stoichiometric mix of organic and inorganic constituents, and this thesis explored the optical behaviour of thin 2D perovskite samples. Lead iodide perovskites ($(\text{RNH}_3)_2\text{PbI}_4$) forms a layered structure with alternating inorganic $[\text{PbI}_6]$ octahedral networks and organic molecular sheets. Due to the difference in the band gaps of the organic and inorganic constituents a multiple quantum well structure is formed, so that excitons are created and trapped in the inorganic sheets. This reduction in dimensionality (quantum confinement), as well as the reduced refractive index of organic molecules (dielectric confinement), lead to large exciton binding energy. As such the optical properties of such perovskites are dominated by exciton effects even at room temperature. The tunability of the exciton energy via structural changes or external stimuli, as well as the many fabrication techniques available, make such materials promising candidates for use in future devices.

In Chapter 4 we looked at the fabrication of thin perovskite films by spin coating a solution of the perovskite constituents dissolved in a suitable solvent. The morphology of the thin films were strongly affected by the organic moiety of the perovskite, the substrate functionalisation and the atmosphere in which the spin coating took place. The thickness of the films was controlled by the spin speed and the initial solution concentration. We found we could reliably create thin films with thickness $\sim 30 - 150$ nm, uniform on μm lengthscales by spin coating in a dehydrated atmosphere and by functionalising the surface with an aminosilane.

In Chapter 5 we created ultrathin samples of perovskites using exfoliation, and probed the optical effects of such samples. By changing the substrate to improve optical contrast we were able to view layers just a few monolayers thick microscopically. We were able to pro-

duce monolayer-thick perovskite samples, and found a layer thickness in CHPI of 1.6 nm. The optical spectra were dominated by the exciton and charge transfer between organic and inorganic layers, and we observed three main regimes of behaviour. For areas >27 nm (15 layers) we observed ‘bulk’ thin film behaviour, similar to spin coated films. A structural transition region was found between 15 – 25 nm, where flattening and straining of the inorganic layers led to a redshift and broadening of the exciton peak, with a corresponding redshift and increase in optical activity of the charge transfer. Finally for the thinnest samples (<15 nm, 8 layers), relaxation of the inorganic layers produce a blueshift and decrease in linewidth of the exciton peak.

In Chapter 6 we explored the interaction excitons with localised surface plasmons. Metal island structure were created via the evaporation of noble metals, both with and without templating, then coated with perovskite films. In Au nanostructures, where the plasmon is far off-resonance with the exciton, the perovskite coating caused a redshift and broadening of the localised surface plasmon. However for Ag islands the two oscillators were more in resonance, and thus we observed weak coupling with a ~ 5 nm blueshift of the exciton, as well as an increase in the extinction due to the field enhancement of the localised surface plasmon.

In Chapter 7 we investigated the coupling between excitons and modes on 1D gratings. Both plasmonic and non-plasmonic gratings were created, capable to sustaining a range of diffractive, surface plasmon polariton and localised modes. The energies and strengths of these modes were found to be sensitive to grating geometry, dielectric coatings and light polarisation, and we were able to create modes resonant with the exciton by change the periodicity and incident/azimuthal angle of light. We found no changes to the exciton except in the case of perovskite-coated Ag gratings. In this case we observed the appearance of a second exciton mode due, the image biexciton, formed as a result of the interaction between the exciton and its image in the metallic mirror, outcoupled via the exciton surface plasmon polariton. The binding energy of the image biexciton is 100 meV, and tunable via a change in the electric environment around the exciton, e. g. resonance with a surface plasmon polariton. We also observed strong coupling between the exciton, biexciton, and surface plasmon polariton mode on the grating, with Rabi splittings of 150 and 125 meV for the exciton and biexciton respectively. The out-of-plane image biexciton mediated coupling between perpendicular exciton and surface plasmon modes. Finite element method modelling showed that the surface plasmon mode was trapped in the bottom of the grating slits, and the field enhancement in the perovskite layer as a result enhanced the coupling strength and reduced the cavity length needed when compared to exciton-photon strong coupling. This Rabi

splitting in two orders of magnitude larger than what has been observed in inorganic GaAs quantum wells, and is observable even at room temperature.

Although the optical properties of 2D lead iodide perovskites are fairly well understood, fewer studies have focused on studying their transport activities. Although carrier can clearly be transported in the plane of the inorganic sheets, the structure of the electroluminescent devices [Fig. 2.23] suggests there may be some charge transfer between the organic and inorganic layers, the exact nature of which is not well understood. The transport properties of perovskites must be studied more in order to compare them to materials currently used in optoelectronic devices.

More fundamentally, the driving force behind the self-assembly of the perovskite structure is not known. Particularly important is the understanding of the crystal formation on a molecular level from a liquid phase, for example in spin coating. As a result the formation of layers over structured substrates, for example in the grating system, is not known. This layer orientation has great implications in the optical properties of the nanostructures in question. Better understanding would allow us to control and tailor the perovskite formation to suit the needs of the system.

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