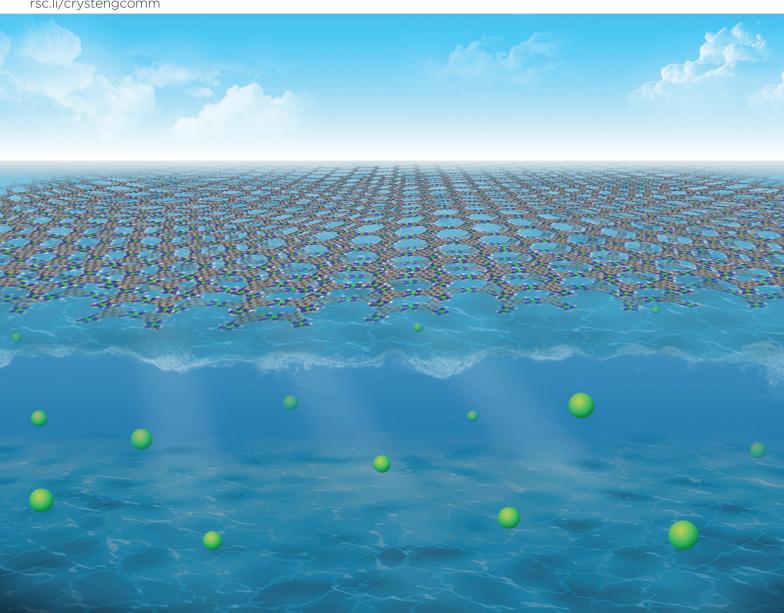
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Influence of varying the nickel salt aqueous subphase on the formation of Ni₃(hexaiminotriphenylene)₂ metal-organic framework nanosheets at the air/liquid interface†

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Metal-organic framework (MOF) nanosheets synthesized at the air/liquid interface exhibit properties, such as electrical conductivity, that are highly dependent on their structural attributes, including morphology, lateral dimensions, thickness, crystallinity, and orientation. Achieving precise control over these features, however, remains a significant challenge. Extending our previous works on the air/liquid interfacial synthesis of uniaxially oriented Ni₃(HITP)₂ nanosheets (HITP-Ni-NS), this study explores the profound influence of the metal salt counterion—a key parameter in MOF crystallization. We present a systematic investigation into how nickel acetate (Ni(OAc)₂), nickel chloride (NiCl₂), and nickel nitrate (Ni(NO₃)₂) precursors affect the resulting nanosheet morphology, thickness, crystallinity, and orientation. Our comparative interfacial syntheses demonstrate that variations in the counterion significantly impact crystal growth kinetics, leading to discernible differences in nanosheet architecture. Notably, while NiCl₂ and Ni(NO₃)₂ precursors result in the incorporation of unreacted HATP ligand stacks and subsequent nanosheet disorder, Ni(OAc)₂ consistently produces HITP-Ni-NS with the greatest thickness and maintains perfect alignment with a preferred ordered crystalline stacking orientation. These observed differences are attributed to variations in the HITP-Ni-NS crystal growth mechanism, likely mediated by the distinct pH of the nickel aqueous subphases. These findings highlight the critical role of the metal salt counterion in directing the growth and ultimately tailoring the functional properties of MOF nanosheets.

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1. Introduction

Metal-organic frameworks (MOFs) represent a class of crystalline porous materials constructed through the coordination of metal ions and organic ligands. 1-5 Their inherent nanoporosity, characterized by uniform pore dimensions and high structural tunability arising from the vast diversity of their constituent components, has positioned MOFs as promising candidates for a wide range applications, including selective adsorption, 6-9 storage, 10,11 separation, 12-14 and catalysis. 9,15 Furthermore, the compositional flexibility of MOFs enables the design of electrically conductive frameworks, expanding their potential devices.16-19 electronic To fully nanotechnological potential of MOFs, particularly

applications such as sensors, transistors, and capacitors, the controlled assembly of their ultrathin films, with precise control over growth direction and thickness, and their integration with other device components are crucial.²⁰ Achieving precise control over the morphology of MOF thin films, including their lateral dimensions and thickness, is paramount for optimizing their performance in various applications.^{21,22}

The air/liquid interface has emerged as a versatile of platform for the fabrication two-dimensional nanomaterials-known as nanosheets—possessing thickness below 100 nm and high aspect ratio. 23-27 Advances in air/liquid interfacial synthesis have broadened the scope of materials amenable to this technique beyond conventional surfactants, enabling the creation of diverse functional nanosheets, including metal oxides^{28,29} and organic³⁰⁻³² and organic-inorganic hybrid materials. 20,27,33,34 Similar to a conventional solvothermal synthesis of MOFs, which is governed by a complex interplay of various factors^{35,36} such as compositional parameters (component structures, reagent concentrations, type of metal salts, solvent, and pH) and process parameters (temperature and time), the air/liquid

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interfacial synthesis of MOFs is also influenced with various factors. Generally, variations in compositional or process parameters, such as surface compression, 37,38 subphase composition and pH, 99-42 concentration of organic ligands, and solvent types, 44 can significantly influence the resulting framework structure and morphology and, consequently, its properties. Therefore, understanding how variations in synthesis conditions at the air/liquid interface can be utilized to tune the morphology of MOF nanosheets represents a critical area of research for advancing their integration into

next-generation nanotechnologies. The fabrication of two dimensional (2D) conductive MOF nanosheets with electrical conductivities exceeding 10⁻⁴ S cm⁻¹ through air/liquid interfacial synthesis has been reported. 21,45-49 Among these materials, Ni₃(HITP)₂ (HITP = 2,3,6,7,10,11-hexaiminotriphenylene), a 2D conductive MOF assembled from triphenylene-derived ligands, has gathered significant attention. 50,51 This MOF, characterized by its hexagonal nanoporous architecture, exhibits notable electrical conductivity in its bulk pellet,50 thin film,50 and microcrystalline forms.⁵¹ The conduction mechanisms in Ni₃(HITP)₂ vary between the inter- and intralayer directions,⁵¹ and the material's bulk electrical properties are substantially modified by the angles and distances of its crystal boundaries.⁵² The ability to produce Ni₃(HITP)₂ nanosheets with well-controlled crystal alignment could significantly advance ongoing research efforts. Our group has previously achieved the successful fabrication of uniaxially oriented nanosheets of Ni₃(HITP)₂ (HITP-Ni-NS) with high electrical conductivity from the reaction between 2,3,6,7,10,11hexaaminotriphenylene (HATP) and Ni2+ ions via the air/ liquid interfacial synthesis.²¹ Furthermore, we have already elucidated the impact of solvent selection,44 the period of standing time for reaction,53 and how the concentrations of the ligand, HATP, 43 affect the characteristics of the resulting nanosheets. This study now turns its attention to another aspect of MOF nanosheet synthesis: the influence of the metal salt and its counterion-a recognized critical factor in both crystallization and crystal growth processes of MOFs. 54-57 Herein, we present an investigation into the effects of counterions of the nickel salt precursor on the morphology, thickness, crystallinity, and orientation of resulting nanosheets (hereafter denoted as HATP_Ni) synthesized with HATP on the air/liquid interface employing three distinct nickel salt species: Ni(OAc)2·4H2O, NiCl2·6H2O, and Ni(NO₃)₂·6H₂O.

2. Experimental

2.1. Materials

2,3,6,7,10,11-Hexaaminotriphenylene·6HCl (HATP·6HCl) was synthesized according to the literature. Nickel(II) acetate tetrahydrate (Ni(OAc)₂·4H₂O, \geq 99%) was obtained from Kanto Chemical or FUJIFILM Wako Pure Chemical Corporation. Nickel(II) chloride hexahydrate (NiCl₂·6H₂O, \geq 99%), nickel(II) nitrate hexahydrate (Ni(NO₃)₂·6H₂O, \geq 99%),

and methanol (MeOH, \geq 99%) were obtained from Kanto Chemical. All chemicals and solvents were used without further purification.

2.2. Preparation of the HATP spread solution

The ligand, HATP-6HCl (abbreviated hereafter as HATP), was dissolved in MeOH at room temperature (20–22 °C) to prepare a spread solution with a concentration of 0.59 mM. To ensure complete dissolution, the solution underwent three iterative cycles, each consisting of a 10 second sonication in a water bath followed by magnetic stirring at 600–700 rpm for 5 min. To mitigate potential degradation, the HATP solution was stored under a nitrogen atmosphere in the dark at 4 °C and used within the day of preparation. Furthermore, the integrity of the solution, with specific attention to degradation products and impurities, was assessed *via* ultraviolet-visible–near infrared (UV-vis–near IR) spectroscopy prior to each use.

2.3. Synthesis of HATP_Ni at the air/liquid interface

Aqueous subphase solutions of Ni(OAc)2·4H2O, NiCl2·6H2O, or Ni(NO₃)₂·6H₂O were prepared in ultrapure water at a concentration of 50 mM. These solutions were sonicated for 60 min at 20-22 °C to facilitate complete dissolution and of utilized on the day preparation. polytetrafluoroethylene (PTFE)-coated Langmuir trough (KSV-NIMA Co. Ltd., $L \times W \times H = 364 \times 75 \times 7 \text{ mm}^3$, surface area between barriers: 318 × 75 mm²) was filled with the subphase solution to form a meniscus above its top edge. Subsequently, the subphase surface underwent aspiration for cleaning, resulting in a final volume of approximately 220 mL. The temperature of the subphase solution in the Langmuir trough was controlled at 20-21 °C using a watercooling circulator.

The HATP spread solution (0.59 mM) was carefully spread onto the aqueous subphase surface in the Langmuir trough over a period of 8 to 12 minutes. The spread volumes were 800 μ L for the Ni(NO₃)₂·6H₂O subphase and 890 μ L for the Ni(OAc)₂·4H₂O and NiCl₂·6H₂O subphases. Subsequently, the subphase temperature was stabilized at 20–21 °C for 60 min. Following equilibration, the surface film was compressed symmetrically from both ends of the trough using PTFE barriers at a compression rate of 10 mm min⁻¹. Surface pressure (π)-mean molecular area (A) isotherms were then recorded using a Wilhelmy plate.

2.4. Deposition of HATP_Ni on solid substrates

HATP_Ni at the air/liquid interface were deposited onto solid substrates (silicon wafers or quartz glass) by horizontally contacting the substrate surface to the interface at room temperature in air (Langmuir–Schaefer technique). For different measurements, specific constant surface pressure was maintained during the transfer process: 5 mN m⁻¹ for AFM, 10 mN m⁻¹ for UV-vis–near IR, and 20 mN m⁻¹ for FTIR and XRD. These surface pressures were precisely controlled

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through the symmetric compression of the Teflon barriers from both ends of the trough at a speed of 3.0 mm min⁻¹. Prior to deposition, silicon and quartz substrates were ultrasonically cleaned in MeOH for 30-60 min. HATP_Ni readily adhere to cleaned solid substrates, a phenomenon largely driven by diverse non-covalent interactions, including van der Waals forces, hydrogen bonding, and/or Coulombic interactions.⁵⁸ Following the transfer of HATP_Ni onto the substrate, the samples were immersed in ultrapure water for 10 seconds to remove any unreacted Ni²⁺ ions. Subsequently, the substrates were dried under a stream of N2 gas or compressed air. The deposited films were then protected from light and stored under argon atmosphere.

2.5. Preparation of HATP nanosheets at the air/pure-water interface and deposition on solid substrates

HATP nanosheets were prepared at the surface of the ultrapure water subphase in the absence of Ni²⁺ ions (hereafter denoted as HATP_water). The preparation process and conditions mirrored those of HATP_Ni, with the sole exception that ultrapure water was used as the subphase. The conditions for depositing the nanosheets onto substrates were also identical to those described for HATP_Ni.

2.6. Brewster angle microscopy at the air/liquid interface

The morphology and surface coverage of HATP Ni and HATP_water at the air/liquid interface were investigated using Brewster angle microscopy, BAM (KSV-NIMA, MicroBAM system), under ambient atmospheric conditions. A laser with a wavelength of 659 nm was directed at the air/liquid interface with an incident angle of 53°. BAM images, with dimension of 3 × 4 mm², were acquired by detecting the reflected laser light using a charge-coupled device (CCD) camera.

2.7. Ultraviolet-visible-near infrared absorption spectroscopy

Ultraviolet-visible-near infrared (UV-vis-near IR) absorption spectra were acquired in the wavelength range of 190-2700 nm at room temperature under atmospheric conditions using a JASCO V-670 spectrophotometer. All measurements for HATP solutions, as well as HATP_Ni and HATP_water deposited on quartz substrates were performed using the transmission mode.

2.8. Fourier transform infrared (FTIR) absorption spectroscopy

Fourier transform infrared (FTIR) spectra of HATP_Ni deposited on high-resistivity Si substrates were recorded using a JASCO FT-IR-6200 spectrometer within the range of 4000–400 cm⁻¹ under vacuum at room temperature. Additionally, FTIR spectra of powder samples of HATP, Ni(OAc)₂·4H₂O, and Ni(NO₃)₂·6H₂O prepared by the KBr pellet technique were obtained under identical measurement conditions.

2.9. Atomic force microscopy

Atomic force microscopy (AFM) was employed using an SPA400 (SII) to obtain surface images of HATP_Ni and HATP_water deposited on low-resistivity Si substrates. Sample imaging was performed in the dynamic force mode at room temperature in air. All images were acquired using a Si microcantilever with Al back coating (Si-DF20). For each sample, a minimum of four AFM images were obtained and subjected to statistical analysis to determine the nanosheet thickness.

2.10. X-ray crystallography

Grazing incidence (GI) synchrotron X-ray diffraction (XRD) and out-of-plane (θ -2 θ geometry) XRD measurements were conducted on HATP_Ni deposited on Si substrates with a multiaxis diffractometer at the BL19B2 beamline in SPring-8. XRD measurements were also performed using a SmartLab (Rigaku) with a Cu K_{α} source operating at a wavelength (λ) of 1.5418 Å. Both GI in-plane and θ -2 θ geometric XRD scans were conducted. For both configurations, the glancing angle (ω) was optimized to maximize the peak intensity corresponding to the 100 reflection (GI-XRD) or to the 001 reflection (out-of-plane XRD). A Soller collimator (0.114°) was positioned in front of a D/tex detector. A nanosheet deposited on a high-resistivity Si substrate was used for these measurements.

3. Results and discussion

3.1. Formation of HATP_Ni at the air/liquid interface

The synthesis of HATP_Ni at the air/liquid interface involves the reaction of HATP ligands with Ni2+ ions present in an aqueous subphase (Fig. 1). The HATP_Ni comprises planar honeycomb networks of HITP-Ni-NS, with layers stacked via π - π interactions. ^{21,52} To investigate the influence of nickel salts, three aqueous subphase solutions containing different Ni²⁺ sources were prepared: Ni(OAc)₂·4H₂O, NiCl₂·6H₂O, or Ni(NO₃)₂·6H₂O. A methanol solution of HATP ligands was then carefully spread onto each subphase with the spreading rate controlled. Upon completion of the HATP solution spreading onto the aqueous subphase, the interface was

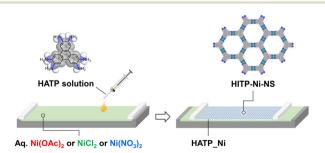


Fig. 1 Schematic illustration of the synthetic procedure for HATP_Ni at the air/liquid interface, utilizing various nickel salt precursors. HATP_Ni is characterized by planar honeycomb networks of HITP-Ni-NS, with π - π interactions facilitating the stacking of each layer.

maintained undisturbed for 60 min. Subsequently, the synthesized HATP_Ni on the subphase surface was gathered by controlled compression using PTFE barriers and deposited onto solid substrates for further analysis. HATP_Ni fabricated for each Ni²⁺ containing subphase was labelled as HATP_Ni(OAc)₂, HATP_NiCl₂, and HATP_Ni(NO₃)₂. For comparative purposes, the HATP solution was also spread onto a surface of the ultrapure water subphase, followed by the collection and deposition of the resulting HATP_water onto solid substrates.

The formation of **HATP_Ni** at the air/liquid interface was monitored with surface pressure (π) – mean molecular area (A) isotherms (Fig. 2, left panel). Representative π –A isotherms for **HATP_Ni** synthesized on subphases of Ni(OAc)₂, NiCl₂, and Ni(NO₃)₂, alongside that of **HATP_water**, are presented in Fig. 2a, d, g, and j, respectively. Mechanical compression of the surface with two barriers resulted in a monotonic increase in π , indicating the occurrence of the nanosheet formation process. Notably, the initial onset of the steep π increase (A values) was consistent across all samples. The first inflection point was assigned to the sharp transition at \sim 0–3 mN m⁻¹. To accurately determine the second inflection points, which presented a gradual transition, a

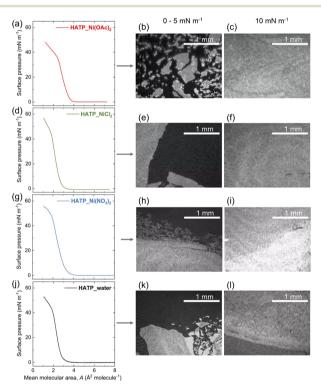


Fig. 2 In situ observation of the formation, morphology, and surface coverage of HATP_Ni at the air/liquid interface with π -A isotherms and Brewster angle microscopy (BAM). The representative π -A isotherms of (a) HATP_Ni(OAc)₂ (red), (d) HATP_NiCl₂ (green), (g) HATP_Ni(NO₃)₂ (blue), and (j) HATP_water (black). BAM images of the corresponding π -A isotherms at π = 0-5 mN m⁻¹ and 10 mN m⁻¹ on the millimetre scale: (b and c) HATP_Ni(OAc)₂, (e and f) HATP_NiCl₂, (h and i) HATP_Ni(NO₃)₂, and (k and l) HATP_water as a reference material. A scale bar is 1 mm.

first-derivative analysis of the π -A isotherm was performed. This analysis was subsequently refined by visual inspection to pinpoint abrupt changes in the slope of the isotherms and their first derivative. While the first inflection points ($\sim 0-3$ mN m⁻¹) and slopes of the isotherms exhibited no significant variation, the second inflection points in the π -A isotherms displayed considerable difference ranging from 31 to 51 mN m⁻¹. Specifically, the surface pressure at the second inflection point was lowest for HATP_water ($\pi = 36 \pm 2 \text{ mN m}^{-1}$), followed by HATP_Ni(OAc)₂ ($\pi = 38.3 \pm 0.9 \text{ mN m}^{-1}$). In contrast, significantly higher values were observed for **HATP_NiCl**₂ ($\pi = 48.8 \pm 1.5 \text{ mN m}^{-1}$), and **HATP_Ni(NO**₃)₂ (π = 50.4 ± 1.3 mN m⁻¹). We noted that the slightly increased uncertainty associated with the second inflection points of HATP_water, relative to other samples, primarily stemmed from the fluctuating nature of HATP domains during film compression. We hypothesize that the elevated surface pressure for HATP_NiCl2 and HATP_Ni(NO3)2 indicates increased mechanical strength of their respective nanosheet thin films under surface compression, potentially attributable to heterogeneous compositions within these films. The interfaces between these different domains, as well as the interactions within and between the different components, probably contribute significantly to the film's resistance to compression. The observed variation in the second inflection points suggests a correlation with the distinct roles of the counterions of nickel precursors in the formation of the **HATP_Ni** structure under compression.

Complementary in situ BAM analysis (Fig. 2, right panel) provided real-time insights into the morphology and surface coverage of HATP_Ni at the air/liquid interface. Prior to the sharp increase in surface pressure ($\pi < 5$ mN m⁻¹), BAM images of all HATP_Ni samples (Fig. 2b, e and h) and HATP_water (Fig. 2k) revealed a morphology characterized by the coexistence of small and large island-like assemblies separated by significant inter-assembly gaps. Upon compression to a higher surface pressure ($\pi = 10 \text{ mN m}^{-1}$), BAM images indicated increased contact and reduced interassembly gaps for all HATP_Ni samples (Fig. 2c, f and i) and HATP_water (Fig. 21). Notably, under compression, all observed HATP_Ni samples displayed relatively uniform thin film morphologies of without prominent features, suggesting homogeneity at the macroscopic scale. While the macroscopic morphology and surface coverage observed via BAM and the overall shape of the π -A isotherms did not exhibit significant differences among the HATP_Ni synthesized with various nickel salt precursors, notable variations were observed in the surface pressure at the second inflection during film compression. Specifically, HATP_NiCl₂ and HATP_Ni(NO₃)₂ displayed significantly higher surface pressures at this point. This implies that the kinetics of HITP-Ni-NS crystal growth and/or assembly processes are significantly influenced by the selection of nickel salt precursors. We hypothesize that the elevated mechanical strength observed in HATP_NiCl₂ HATP_Ni(NO₃)₂ under surface compression stems from the

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potential presence of unreacted HATP ligands within the synthetic systems. These unreacted ligands may facilitate the formation of vertically aligned HATP molecules through π - π stacking. The subsequent incorporation of these unreacted HATP domains between the HITP-Ni-NS assemblies could enhance the film's resistance to compression. This proposed structural characteristic appears to be less prevalent in HATP_Ni(OAc)2, where a more complete reaction between HATP ligands and Ni²⁺ is inferred, potentially leading to a more homogeneous film structure.

3.2. Ex situ surface observation of HATP_Ni

Atomic force microscopy (AFM) was employed to further investigate the nanoscale surface coverage and morphology of HATP_Ni synthesized with various nickel salt precursors. Specifically, HATP_Ni and HATP_water deposited on Si substrates at $\pi = 5$ mN m⁻¹ were analysed. Topographical AFM images, representative thickness profiles selected from the images along the red line, and thickness histograms are shown in Fig. 3. The topographical images of HATP_Ni synthesized with Ni2+ reveal the hierarchical assembly of

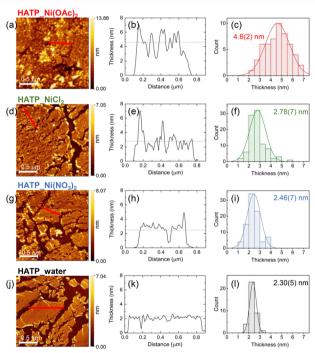


Fig. 3 Ex situ atomic force microscopy (AFM) for surface observation of the nanoscale surface coverage and morphology of thin films deposited on Si substrates at π = 5 mN m⁻¹. (a), (d), and (g) show $\textbf{HATP_Ni} \ \text{synthesized using Ni(OAc)}_{2}, ^{43} \ \text{NiCl}_{2}, \ \text{and Ni(NO}_{3})_{2} \ \text{precursors},$ respectively, and (j) displays HATP_water prepared under identical deposition conditions. (b), (e), (h), and (k) display representative thickness profiles measured along the red lines depicted in the corresponding AFM topographical images. Vertical dotted lines indicate the average thickness, calculated from at least four nanosheet images. (c), (f), (i), and (l) display the corresponding thickness histograms and average thickness values for HATP_Ni(OAc)2, $\label{eq:hatp_NiCl2} HATP_Ni(NO_3)_2, \ and \ HATP_water, \ respectively.$

primary HITP-Ni-NS crystals (Fig. 3a, d and g), characterized by round-shaped nanosheets with lateral dimensions of the order of tens of nanometers, into larger secondary assemblages. In contrast, HATP_water, lacking these primary HITP-Ni-NS crystals, exhibits only a homogeneous distribution of nanosheets (Fig. 3j). Notably, HATP_Ni(OAc)2 exhibits the highest concentration of these primary HITP-Ni-NS crystals. Subsequent thickness analysis was performed on the AFM images of each sample, and representative thickness profiles are presented in Fig. 3b, e, h and k. The thickness distribution for HATP_Ni synthesized with various nickel salt precursors was further quantified through histograms, and the average thickness values were compared (Fig. 3c, f, i and l). The average thicknesses of HATP_Ni(OAc)2, HATP_NiCl2, and HATP_Ni(NO₃)₂, as well as HATP_water, were determined to be 4.6(2), 2.78(7), 2.46(7), and 2.30(5) nm, respectively. Detailed examination of the thickness distribution revealed that HATP_NiCl2 and HATP_Ni(NO3)2 exhibited narrow distributions, displaying a similar trend to that observed for HATP_water. Conversely, HATP_Ni(OAc)2 showed a broader thickness distribution and the largest average nanosheet thickness. We hypothesize that the reaction rate of crystal growth of HITP-Ni-NS for HATP_Ni(OAc)2 is faster than those of HATP_NiCl2 and HATP_Ni(NO3)2. These findings indicate that the choice of the nickel salt precursor possibly impacts the crystal growth kinetics of HITP-Ni-NS, thereby governing the thickness and uniformity of the synthesized HATP_Ni.

3.3. Electronic states and coordination bonds of HITP-Ni-NS

UV-vis-near-IR spectroscopy was employed to analyse HATP_Ni samples prepared with different nickel precursors. The resulting spectra (Fig. 4a and b) show characteristic absorption bands at ~624 nm, attributed to d-d* transitions

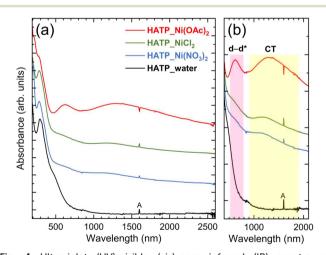


Fig. 4 Ultraviolet (UV)-visible (vis)-near infrared (IR) spectra of nanosheets deposited on quartz substrates. (a) Spectra of HATP_Ni synthesized using various nickel salt precursors and HATP_water. (b) Enlarged view of absorption bands associated with d-d* transition at 624.5 nm and charge transfer (CT) between HITP and Ni $^{2+}$ at $\sim\!1200$ nm. Absorbance spectra were normalized at 280 nm. Feature 'A' indicates instrumental artifacts arising from light source switching

in Ni²⁺, and at ~1255 nm, corresponding to charge transfer (CT) between Ni²⁺ and HITP. These spectral features indicate the presence of Ni²⁺ ions, an alteration in the ligand field symmetry of their d-orbital upon coordination to HITP ligands, and the establishment of electronic interactions through coordination bonds under all examined synthesis conditions. Notably, the CT transition exhibited its highest intensity in HATP Ni(OAc)2 compared to HATP NiCl2 and HATP_Ni(NO₃)₂, which implies a greater extent of Ni²⁺-HITP coordination within the HITP-Ni-NS for HATP_Ni(OAc)2. For HATP_Ni(OAc)2, HATP_NiCl2, and HATP_Ni(NO3)2, the peak height ratios of the charge transfer (CT) band (~1255 nm) to the unreacted HATP ligand band (288 nm) were calculated to be 0.47, 0.17, and 0.17, respectively. This trend correlates with a higher synthetic yield of HATP_Ni(OAc)2 relative to the Fourier samples. transform infrared (FTIR) spectroscopy was also employed to investigate the coordination bonds (Fig. S1†). The FTIR spectra of HATP_Ni(OAc)2, HATP_NiCl2, and HATP_Ni(NO3)2 revealed a redshift in the NH stretching peaks (~3500-3250 cm⁻¹, Fig. S1a†) and a dominant CN stretching band (~1330 cm⁻¹) relative to the NH bending band ($\sim 1640 \text{ cm}^{-1}$) (Fig. S1b†). These observations provide evidence for the formation of HITP-Ni-NS within all investigated HATP_Ni. The more pronounced decrease in the NH bending peak observed for compared HATP_Ni(OAc)₂ to HATP_NiCl₂ HATP_Ni(NO₃)₂ suggests a higher degree of N-Ni bond formation in HATP_Ni(OAc)2. Analyses by both UV-vis-near IR and FTIR spectroscopy confirm the successful formation of HITP-Ni-NS in all HATP_Ni samples under investigation. Notably, the synthetic yield of HITP-Ni-NS was greatest when

Ni(OAc)₂ served as a nickel salt precursor. The formation of the HITP and Ni²⁺ complex at the air/ liquid interface is facilitated by the deprotonation of HATP into HITP.21 A relatively alkaline reaction environment is conductive to the acid-base equilibrium of HATP shifting towards its deprotonated HITP form, which consequently enhances coordination between HITP and Ni2+ for the formation of HITP-Ni-NS.59 The pH values of the aqueous subphase were measured to be 7.4, 5.6, and 4.8 for syntheses employing Ni(OAc)2, NiCl2, and Ni(NO3)2, respectively. In the absence of pH control agents, the formation of HATP_Ni was demonstrably influenced by the inherent pH of the subphase, which varied significantly depending on the nickel salt precursor utilized. Moreover, the unique counterions present in each nickel salt exert an influence on the hydrolysis equilibria of the metal ions within the subphase. Notably species such as Cl⁻ and NO₃⁻ possess the capacity to compete with HITP for coordination sites on the metal ions,60 thereby altering the local concentration of reactive Ni2+ species. The competition reaction could not be definitively confirmed in our current investigations, as Ni2+ and its associated counterions were fully removed during the nanosheet purification steps (Fig. S1a†). Nevertheless, its occurrence is considered highly plausible. Consequently, HATP_NiCl₂ and HATP_Ni(NO₃)₂,

which formed under relatively acidic conditions, exhibited lower concentrations of deprotonated HITP, leading to slower crystal growth rates of HITP-Ni-NS³⁶ and substantial incorporation of unreacted HATP domains. The enhanced uniformity in nanosheet thickness observed with AFM (Fig. 3e and h) for in HATP_NiCl₂ and HATP_Ni(NO₃)₂ compared to HATP_Ni(OAc)₂ (Fig. 3b) is likely attributable to a higher incorporation of unreacted HATP domains during their syntheses. Conversely, the nucleation rate of HITP-Ni-NS in the HATP_Ni(OAc)₂ was accelerated, yielding a considerable number of nanocrystals that occasionally promoted stacking through various mechanisms.

3.4. Crystalline structure and orientation

The influence of nickel salt precursors on the crystallinity and orientation of HITP-Ni-NS in the synthesized HATP_Ni was investigated by GI in-plane XRD measurements. Fig. 5 displayed GI in-plane XRD profiles for HATP_Ni(OAc)₂, HATP_NiCl₂, and HATP_Ni(NO₃)₂. The GI in-plane XRD profile of HATP_Ni(OAc)₂ revealed Bragg reflections exclusively indexed to the *hk*0 planes of a metrically hexagonal lattice, thereby confirming the uniaxial orientation of the HITP-Ni-NS.^{21,43} Conversely, the XRD profiles obtained from HATP_NiCl₂ and HATP_Ni(NO₃)₂ exhibited significantly attenuated *hk*0 reflection peaks, alongside a more pronounced Bragg reflection indexed as 001. This spectral difference implies a comparatively lower degree of structural orientation in these samples relative to HATP_Ni(OAc)₂.

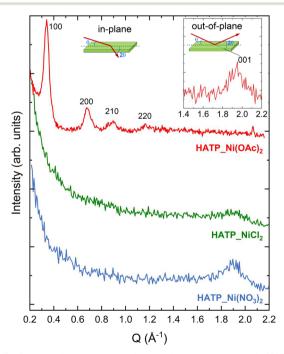


Fig. 5 Structural and orientational characterization of HATP_Ni deposited on Si substrates. GI in-plane XRD profiles for HATP_Ni(OAc) $_2$ (red), HATP_NiCl $_2$ (green), and HATP_Ni(NO $_3$) $_2$ (blue). The inset profile shows the out-of-plane XRD profile for HATP_Ni(OAc) $_2$ of the 001 peak at \cong 1.9 Å.

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Furthermore, an increased ratio of the hk0 reflection peak intensity to the 001 peak indicates that the lateral domain sizes of the HITP-Ni-NS crystals are substantially larger in HATP_Ni(OAc)₂ than in HATP_NiCl₂ and HATP_Ni(NO₃)₂. The orientation of HATP_Ni was further investigated through out-of-plane XRD analysis. A consistent single peak was detected at $Q \sim 1.9 \text{ Å}^{-1}$ across all **HATP_Ni** samples, including HATP_Ni(OAc)₂ (Fig. 5 inset, Fig. S2†), HATP_NiCl₂, and HATP_Ni(NO₃)₂ (Fig. S2†). This observation indicates a highly ordered π - π stacked crystal arrangement oriented vertically to the surface.

The proposed HITP-Ni-NS crystalline structure, derived from the GI in-plane XRD measurement, consistently manifests a uniaxial orientation, with in-plane lattice parameters of $a = b \sim 21.4$ Å (Fig. 6a). This layered motif is further supported by out-of-plane XRD measurement, which attribute to π - π stacking between individual nanosheets, resulting in an interlayer distance of $d \sim 3.2$ Å (Fig. 6a), consistent with a previous report.21 Crystallite size analysis of HATP_Ni was subsequently conducted by employing the full width at half maximum (FWHM) of the 001 reflection. The crystallite size derived from the 001 reflection in the out-of-plane XRD profiles provided specific information regarding the stacking arrangement and thickness of the nanosheet layers. Additionally, we aimed to elucidate the molecular stacking of unreacted HATP domains within the nanosheet via analysis of the 001 peak of the GI in-plane XRD profiles. However, due to the highly preferential orientation of HATP_Ni(OAc)2, which is established during air/liquid interface growth, the 001 reflection (indicative of perpendicular stacking order) is entirely absent in its GI inplane XRD profile. For these reasons, the crystallite size analysis utilized the GI in-plane profiles for HATP_NiCl2

and HATP_Ni(NO₃)₂ and the out-of-plane profile for HATP_Ni(OAc)2. Notably, the FWHM values derived from the GI in-plane profiles for HATP_NiCl₂ and HATP_Ni(NO₃)₂ exhibited good agreement with those obtained from their respective out-of-plane XRD profiles. The crystallite sizes for HATP_Ni(OAc)₂, HATP_NiCl₂, and HATP_Ni(NO₃)₂ were determined to be 3.99(14) nm, 2.6(2) nm, and 2.1(2) nm, respectively. Notably, the largest crystallite size observed for HATP_Ni(OAc)₂ indicates that the synthesis conditions employed for the Ni(OAc)₂ precursor were most conducive to crystallite growth. This trend correlates with the increased nanosheet thickness as determined by AFM measurements.

3.5. Proposed crystalline structure and orientation of HATP Ni synthesized using various nickel salt precursors

The formation of HATP_Ni(OAc)2 at the air/liquid interface has been proposed via a two-stage mechanism:⁵³ (i) interfacial deprotonation of HATP followed by the spontaneous nucleation and growth of primary HITP-Ni-NS nanocrystals through coordination bonding between the deprotonated HITP ligand and Ni2+ ions at the interface, and (ii) subsequent lateral assembly of these primary nanocrystals into larger nanosheets, accompanied by the dissolution of unreacted HATP ligands into the aqueous subphase. Notably, the standing time of 60 min, used in this work, was determined to be critical for achieving optimal interfacial nanosheet quality. Under these conditions, any HITP-Ni-NS formed within the subphase from dissolved HATP did not float to and adhere to the pre-formed interfacial HITP-Ni-NS, thus preventing the formation of undesirable multi-layered or aggregated structures.

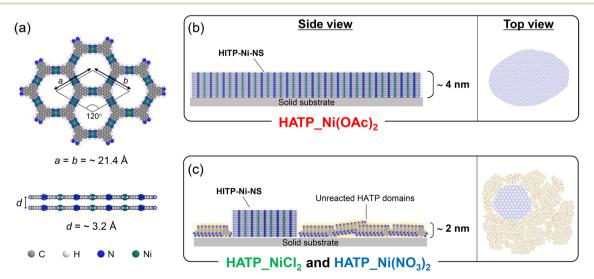


Fig. 6 Proposed crystalline structures of HATP_Ni synthesized at the air/liquid interface and deposited on a solid substrate. (a) A simulated crystallite structure illustrating the basal plane projection of HITP-Ni-NS and its stacking arrangement.²¹ Atomic representations: carbon (grey), hydrogen (white), nitrogen (blue), and Ni²⁺ ions (green). Schematic illustrations (side and top views) depict the morphology of HATP_Ni synthesized with different nickel precursors at the air/liquid interface: (b) Ni(OAc)2, and (c) NiCl2 and Ni(NO3)2. Notably, syntheses employing NiCl2 and Ni(NO₃)₂ resulted in a considerable incorporation of stacking HATP ligands to form unreacted HATP domains.

Alterations in crystal growth kinetics, morphology, thickness, and crystal orientation of HITP-Ni-NS were observed to be dependent on the specific nickel salt utilized precursors in the aqueous subphase: Ni(OAc)₂·4H₂O, NiCl₂·6H₂O, and Ni(NO₃)₂·6H₂O. The rates of HATP deprotonation and subsequent HITP-Ni-NS nucleation were found to follow the descending order: $Ni(OAc)_2 > NiCl_2 > Ni(NO_3)_2$. These kinetic parameters are demonstrably influenced by both the pH of aqueous subphase and the competitive coordination of Cl and NO₃ with Ni²⁺ ions. The combination of a rapid nucleation rate, an elevated concentration Ni²⁺ deprotonated HITP, and a satisfactory local concentration collectively promotes the formation of primary nanocrystals and a significant yield of HITP-Ni-NS nanocrystals. Moreover, a lower concentration of residual HATP ligands in the Ni(OAc)₂ subphase system facilitates the self-assembly of these primary nanocrystals into larger nanosheets with a high degree of crystal orientation. This favourable environment for lateral growth and subsequent increased stacking is observed in HATP_Ni(OAc)₂ (Fig. 6b), resulting in a larger average crystallite size (~4 nm). Based on the interlayer distance of ~3.2 Å, the estimated average number of stacked layers in HATP_Ni(OAc)₂ is approximately 12. Conversely, the slightly acidic pH of the Ni²⁺ subphase when employing NiCl₂ and Ni(NO₃)₂ does not effectively promote HATP deprotonation, resulting in a slower nucleation rate, a lower degree of lateral growth, and less stacking of HITP-Ni-NS for HATP_NiCl₂ $HATP_Ni(NO_3)_2$. Consequently, fraction a unreacted HATP ligand likely aggregated and became incorporated within the forming HITP-Ni-NS crystals at the air/liquid interface, or dissolved into the subphase. The crystallite sizes for HATP_NiCl₂ (~2.6 nm) and $HATP_Ni(NO_3)_2$ (~2.1 nm) as evaluated from the 001 peak in GI in-plane XRD is potentially influenced by the incorporation of unreacted HATP domains. The stacked residual unreacted HATP molecules on the substrate surface are similar in orientation (predominantly vertical or tilted) to those found in HATP_water. Considering the approximated 2 nm molecular size of HATP, the higher crystallite measured for HATP_NiCl₂ HATP_Ni(NO₃)₂ than this value suggests the co-existence

4. Conclusions

unreacted HATP domains.

Our investigation elucidated the influence of various nickel salt precursors on the morphology, surface coverage, thickness, crystallinity, and orientation of **HATP_Ni** nanostructures synthesized at the air/liquid interface. Through comparative analyses employing $Ni(OAc)_2 \cdot 4H_2O$, $NiCl_2 \cdot 6H_2O$, and $Ni(NO_3)_2 \cdot 6H_2O$ in the aqueous subphase, we demonstrated that counterions significantly modulate

of larger HITP-Ni-NS crystallites with the stacked residual

HITP-Ni-NS crystal kinetics. This modulation gives rise to marked variations in the morphology, thickness, and crystal orientation of the resulting HATP_Ni nanostructures. We theorize that these differences stem from distinct crystal growth mechanisms for HITP-Ni-NS, which are influenced, at least in part, by variations in the aqueous subphase's pH, the deprotonation rate of HATP, and the local concentration of reactive Ni²⁺ ions. These precursor-dependent factors consequently exert a critical influence on the nucleation and growth kinetics, as well as the preferred self-assembly orientation of HITP-Ni-NS at the air/liquid interface. The absence of incorporated domains of unreacted HATP ligands during interfacial synthesis is attributed to the enhancement of structural order. The use of Ni(OAc)2 as a nickel salt precursor resulted in HATP_Ni nanosheets that exhibit the most highly ordered crystalline stacking structure of HITP-Ni-NS. These nanosheets are also free of impurities and achieve the greatest HITP-Ni-NS thickness among all precursors investigated.

Data availability

The data supporting this article have been included as part of the ESI.†

Author contributions

Kazuaki Tachimoto: sample synthesis, characterization, data analysis, and investigation. Kanokwan Jumtee Takeno: data analysis, validation, investigation and writing. Rie Makiura: conceptualization, direction, supervision, data analysis, validation, investigation, writing and funding acquisition.

Conflicts of interest

The authors declare no conflict of interest.

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References

1 M. Kondo, T. Yoshitomi, H. Matsuzaka, S. Kitagawa and K. Seki, *Angew. Chem., Int. Ed. Engl.*, 1997, **36**, 1725–1727.

2 H. Li, M. Eddaoudi, M. O'Keeffe and O. M. Yaghi, Nature,

CrystEngComm

- 1999, **402**, 276–279.
- 3 O. M. Yaghi, M. O'Keeffe, N. W. Ockwig, H. K. Chae, M. Eddaoudi and J. Kim, *Nature*, 2003, **423**, 705–714.
- 4 S. Kitagawa, R. Kitaura and S. Noro, *Angew. Chem., Int. Ed.*, 2004, 43, 2334–2375.
- 5 S. Horike, S. Shimomura and S. Kitagawa, *Nat. Chem.*, 2009, **1**, 695–704.
- 6 D. Britt, D. Tranchemontagne and O. M. Yaghi, *Proc. Natl. Acad. Sci. U. S. A.*, 2008, 105, 11623–11627.
- 7 K. Yang, F. Xue, Q. Sun, R. Yue and D. Lin, *J. Environ. Chem. Eng.*, 2013, 1, 713–718.
- 8 K. Vellingiri, J. E. Szulejko, P. Kumar, E. E. Kwon, K.-H. Kim, A. Deep, D. W. Boukhvalov and R. J. C. Brown, *Sci. Rep.*, 2016, **6**, 27813.
- 9 P. Li, S. Kim, J. Jin, H. C. Do and J. H. Park, Appl. Catal., B, 2020, 263, 118284.
- 10 J. P. Marco-Lozar, J. Juan-Juan, F. Suárez-García, D. Cazorla-Amorós and A. Linares-Solano, *Int. J. Hydrogen Energy*, 2012, 37, 2370–2381.
- 11 M. K. Taylor, T. Runčevski, J. Oktawiec, M. I. Gonzalez, R. L. Siegelman, J. A. Mason, J. Ye, C. M. Brown and J. R. Long, J. Am. Chem. Soc., 2016, 138, 15019–15026.
- 12 Y. Peng, Y. Li, Y. Ban, H. Jin, W. Jiao, X. Liu and W. Yang, *Science*, 2014, **346**, 1356–1359.
- 13 X. Wang, C. Chi, K. Zhang, Y. Qian, K. M. Gupta, Z. Kang, J. Jiang and D. Zhao, *Nat. Commun.*, 2017, 8, 14460.
- 14 P. Cheng, Y. Huang, C. Wu, X. Wang, X. Fu, P. Li, Y. Liu and S. Xia, J. Membr. Sci., 2021, 640, 119812.
- S. Yuan, L. Zou, H. Li, Y. Chen, J. Qin, Q. Zhang, W. Lu, M. B. Hall and H. Zhou, *Angew. Chem., Int. Ed.*, 2016, 55, 10776–10780.
- 16 A. A. Talin, A. Centrone, A. C. Ford, M. E. Foster, V. Stavila, P. Haney, R. A. Kinney, V. Szalai, F. El Gabaly, H. P. Yoon, F. Léonard and M. D. Allendorf, *Science*, 2014, 343, 66–69.
- 17 X. Xu, J. Tang, H. Qian, S. Hou, Y. Bando, M. S. A. Hossain, L. Pan and Y. Yamauchi, ACS Appl. Mater. Interfaces, 2017, 9, 38737–38744.
- 18 L. S. Xie, G. Skorupskii and M. Dincă, Chem. Rev., 2020, 120, 8536–8580.
- 19 S. Tao, J. Wang and J. Zhang, ACS Nano, 2025, 19, 9484–9512.
- 20 R. Makiura, S. Motoyama, Y. Umemura, H. Yamanaka, O. Sakata and H. Kitagawa, *Nat. Mater.*, 2010, **9**, 565–571.
- 21 T. Ohata, A. Nomoto, T. Watanabe, I. Hirosawa, T. Makita, J. Takeya and R. Makiura, ACS Appl. Mater. Interfaces, 2021, 13, 54570–54578.
- 22 T. Lee, J. Kim, C. Park, H. Kim, M. Kim, H. Park, I. Kim, J. Ko, K. Pak, S. Q. Choi, I. Kim and S. Park, Adv. Mater., 2022, 34, 2107696.
- 23 J. A. Zasadzinski, R. Viswanathan, L. Madsen, J. Garnaes and D. K. Schwartz, *Science*, 1994, 263, 1726–1733.
- 24 T. Nakamura, H. Tachibana, M. Yumura, M. Matsumoto, R. Azumi, M. Tanaka and Y. Kawabata, *Langmuir*, 1992, **8**, 4–6.
- 25 R. Makiura, K. Tsuchiyama and O. Sakata, *CrystEngComm*, 2011, 13, 5538–5541.

- 26 O. N. Oliveira, L. Caseli and K. Ariga, Chem. Rev., 2022, 122, 6459–6513.
- 27 R. Makiura, Coord. Chem. Rev., 2022, 469, 214650.
- 28 F. Wang, J.-H. Seo, G. Luo, M. B. Starr, Z. Li, D. Geng, X. Yin, S. Wang, D. G. Fraser, D. Morgan, Z. Ma and X. Wang, *Nat. Commun.*, 2016, 7, 10444.
- 29 Y. Wang, Y. Shi, Z. Zhang, C. Carlos, C. Zhang, K. Bhawnani, J. Li, J. Wang, P. M. Voyles, I. Szlufarska and X. Wang, *Chem. Mater.*, 2019, 31, 9040–9048.
- 30 R. Makiura, K. Tsuchiyama, E. Pohl, K. Prassides, O. Sakata, H. Tajiri and O. Konovalov, *ACS Nano*, 2017, **11**, 10875–10882.
- 31 S. Li and H.-G. Liu, Langmuir, 2020, 36, 10876-10884.
- 32 K. Yamanami, Y. Fujita, K. Matsui, R. Asari, T. Kusawake and T. K. Shimizu, *Langmuir*, 2022, 38, 1910–1914.
- 33 M. K. Bera, T. Mori, T. Yoshida, K. Ariga and M. Higuchi, ACS Appl. Mater. Interfaces, 2019, 11, 11893–11903.
- 34 I. Aratani, Y. Horii, Y. Kotani, H. Osawa, H. Tanida, T. Ina, T. Watanabe, Y. F. Yano, A. Mizoguchi, D. Takajo and T. Kajiwara, *J. Mater. Chem. C*, 2024, **12**, 724–735.
- 35 E. Biemmi, S. Christian, N. Stock and T. Bein, *Microporous Mesoporous Mater.*, 2009, **117**, 111–117.
- 36 J. Łuczak, M. Kroczewska, M. Baluk, J. Sowik, P. Mazierski and A. Zaleska-Medynska, *Adv. Colloid Interface Sci.*, 2023, 314, 102864.
- 37 D. Ishikawa, T. Mori, Y. Yonamine, W. Nakanishi, D. L. Cheung, J. P. Hill and K. Ariga, *Angew. Chem., Int. Ed.*, 2015, 54, 8988–8991.
- 38 I. Ciria-Ramos, I. Tejedor, L. Caparros, B. Doñagueda, O. Lacruz, A. Urtizberea, O. Roubeau, I. Gascón and M. Haro, *Dalton Trans.*, 2023, 52, 7196–7207.
- 39 F. Bonosi, A. Renault and B. Berge, *Langmuir*, 1996, **12**, 784–787.
- 40 A. Dhanabalan, N. Prasanth Kumar, S. Major and S. S. Talwar, *Thin Solid Films*, 1998, 327–329, 787–791.
- 41 G. B. Khomutov, A. M. Tishin, S. N. Polyakov and J. Bohr, *Colloids Surf.*, *A*, 2000, **166**, 33–43.
- 42 K. Gong, S.-S. Feng, M. L. Go and P. H. Soew, *Colloids Surf.*, *A*, 2002, **207**, 113–125.
- 43 K. Tachimoto, T. Ohata, K. J. Takeno, A. Nomoto, T. Watanabe, I. Hirosawa and R. Makiura, *Langmuir*, 2023, 39, 8952–8962.
- 44 T. Ohata, K. Tachimoto, K. J. Takeno, A. Nomoto, T. Watanabe, I. Hirosawa and R. Makiura, *Bull. Chem. Soc. Jpn.*, 2023, 96, 274–282.
- 45 K. Hoshiko, T. Kambe, R. Sakamoto, K. Takada and H. Nishihara, *Chem. Lett.*, 2014, 43, 252–253.
- 46 T. Pal, T. Kambe, T. Kusamoto, M. L. Foo, R. Matsuoka, R. Sakamoto and H. Nishihara, *ChemPlusChem*, 2015, 80, 1255–1258.
- 47 V. Rubio-Giménez, M. Galbiati, J. Castells-Gil, N. Almora-Barrios, J. Navarro-Sánchez, G. Escorcia-Ariza, M. Mattera, T. Arnold, J. Rawle, S. Tatay, E. Coronado and C. Martí-Gastaldo, *Adv. Mater.*, 2018, 30, 1704291.
- 48 V. Rubio-Giménez, N. Almora-Barrios, G. Escorcia-Ariza, M. Galbiati, M. Sessolo, S. Tatay and C. Martí-Gastaldo, *Angew. Chem.*, *Int. Ed.*, 2018, 57, 15086–15090.

49 Z. Wang, L. S. Walter, M. Wang, P. St, B. Petkov, H. Liang, N. N. Qi, M. Nguyen, H. Hambsch, M. Zhong, S. Wang, L. Park, K. Renn, T. Watanabe, S. C. B. Taniguchi, T. Mannsfeld, U. Heine, S. Kaiser, R. T. Zhou, X. F. Weitz and R. Dong, J. Am. Chem. Soc., 2021, 143, 13624–13632.

- D. Sheberla, L. Sun, M. A. Blood-Forsythe, S. Er, C. R. Wade,
 C. K. Brozek, A. Aspuru-Guzik and M. Dincă, *J. Am. Chem. Soc.*, 2014, 136, 8859–8862.
- 51 R. W. Day, D. K. Bediako, M. Rezaee, L. R. Parent, G. Skorupskii, M. Q. Arguilla, C. H. Hendon, I. Stassen, N. C. Gianneschi, P. Kim and M. Dincă, ACS Cent. Sci., 2019, 5, 1959–1964.
- 52 M. E. Foster, K. Sohlberg, M. D. Allendorf and A. A. Talin, *J. Phys. Chem. Lett.*, 2018, **9**, 481–486.
- 53 T. Ohata, A. Nomoto, T. Watanabe, I. Hirosawa, T. Makita, J. Takeya and R. Makiura, *J. Colloid Interface Sci.*, 2023, **651**, 769–784.

- 54 P. Díaz, J. Benet-Buchholz, R. Vilar and A. J. P. White, *Inorg. Chem.*, 2006, 45, 1617–1626.
- 55 H.-L. Jiang and Q. Xu, *CrystEngComm*, 2010, **12**, 3815–3819.
- 56 A. Schejn, L. Balan, V. Falk, L. Aranda, G. Medjahdi and R. Schneider, *CrystEngComm*, 2014, **16**, 4493–4500.
- 57 D. Bara, E. G. Meekel, I. Pakamorė, C. Wilson, S. Ling and R. S. Forgan, *Mater. Horiz.*, 2021, 8, 3377–3386.
- 58 W. Zhao, T. Chen, W. Wang, B. Jin, J. Peng, S. Bi, M. Jiang, S. Liu, Q. Zhao and W. Huang, *Sci. Bull.*, 2020, 65, 1803–1811.
- 59 Y. Qian, G. Xue, L. Chen, G. Xu and G. Wang, *Adv. Mater.*, 2024, **36**, 2310795.
- 60 F. Bigdeli, M. N. A. Fetzer, B. Nis, A. Morsali and C. Janiak, J. Mater. Chem. A, 2023, 11, 22105–22131.