



Cite this: *Dalton Trans.*, 2023, **52**, 1841

Received 22nd December 2022,
Accepted 19th January 2023

DOI: 10.1039/d2dt04088c

rscl.li/dalton

Low-concentration CO₂ capture using metal–organic frameworks – current status and future perspectives

Michelle Åhlén, Ocean Cheung * and Chao Xu *

The ever-increasing atmospheric CO₂ level is considered to be the major cause of climate change. Although the move away from fossil fuel-based energy generation to sustainable energy sources would significantly reduce the release of CO₂ into the atmosphere, it will most probably take time to be fully implemented on a global scale. On the other hand, capturing CO₂ from emission sources or directly from the atmosphere are robust approaches that can reduce the atmospheric CO₂ concentration in a relatively short time. Here, we provide a perspective on the recent development of metal–organic framework (MOF)-based solid sorbents that have been investigated for application in CO₂ capture from low-concentration (<10 000 ppm) CO₂ sources. We summarized the different sorbent engineering approaches adopted by researchers, both from the sorbent development and processing viewpoints. We also discuss the immediate challenges of using MOF-based CO₂ sorbents for low-concentration CO₂ capture. MOF-based materials, with tuneable pore properties and tailorable surface chemistry, and ease of handling, certainly deserve continued development into low-cost, efficient CO₂ sorbents for low-concentration CO₂ capture.

Introduction

The ever-increasing combustion of fossil fuels, such as energy generation in coal-fired power plants, cement plants, and oil refineries has contributed towards increasing the atmospheric CO₂ concentration.¹ The atmospheric CO₂ level has gone from the pre-industrial value of 280 ppm to a current level of 418 ppm (December 2022).² The greenhouse effect that is caused by the high atmospheric levels of CO₂ is considered to be one of the main reasons for global warming as well as the associated environmental issues. In addition to the irreversible changes to the climate and environment, a high atmospheric CO₂ concentration is a big risk to human health, for example, it can trigger respiratory illnesses when the atmospheric CO₂ concentration is over 600 ppm.³ Although the move towards non-fossil fuel-based energy generation could be a long-term solution to reduce CO₂ emission due to human activities, carbon capture and storage (CCS) is undoubtedly an important current approach to reduce the CO₂ emission from point sources of CO₂. Alternatively, direct air capture (DAC) of CO₂, which implies not only capturing CO₂ from emission point sources, but rather from the atmosphere, is also an important complementary approach for reducing the atmospheric CO₂ level. DAC is a negative emission technology that has the potential to

lower the atmospheric CO₂ concentration down to 350 ppm.⁴ DAC of CO₂ would mean capturing CO₂ from sources with low-concentrations, or trace amounts of CO₂. This type of CO₂ capture is important for gas purification, indoor air quality control, and a number of industrial processes. DAC of CO₂ has been discussed for indoor settings such as classrooms, hospitals, or offices in order to ensure the well-being of individuals. Air purification devices are needed in ventilators, or in confined spaces such as spacecraft and submarines to keep the CO₂ concentration at a safe level. From an industrial point of view, in order to meet the liquefied natural gas (LNG) specifications, CO₂ concentration in natural gas has to be reduced down to 50 ppm before its liquefaction.⁵ These example application areas show that there is currently a great interest and urgent need for the development of efficient technology for low-concentration CO₂ capture. For the purpose of this perspective, we operationally define “low-concentration” as below approximately 10 000 ppm in CO₂ concentration.

Amine scrubbing is probably the most mature and viable CO₂ capture technology that has been widely applied in natural gas purification and post-combustion capture of CO₂.⁶ It uses aqueous amine solutions to absorb CO₂ from gases *via* chemical reactions, which offers high separation and purification efficiencies even at ultralow CO₂ concentrations (<1000 ppm). However, amine scrubbing suffers from significant drawbacks such as high energy consumption for amine regeneration, risk of amine leakage, and corrosion to the associated equipment. Temperature or pressure swing adsorption (TSA, PSA) technologies have also been developed for CO₂

Division of Nanotechnology and Functional Materials, Department of Materials Science and Engineering, Uppsala University, Ångström Laboratory, SE-751 03 UppsalaBox 35, Sweden. E-mail: ocean.cheung@angstrom.uu.se, chao.xu@angstrom.uu.se; Tel: +46 18 471 3279, +46 18 471 3230



capture. These technologies can be adapted to utilize solid physisorbents. Physisorbents such as porous solids can be engineered to adsorb CO₂ selectively over other gases on the internal surfaces of the sorbent. The adsorbed CO₂ is then released at elevated temperatures and/or reduced pressures and the sorbent is regenerated for subsequent cycles.^{7–9} In contrast to chemisorption processes (using chemisorbents), physisorbents adsorb CO₂ with relatively low enthalpy of adsorption (ΔH_{ads}). The energy cost for regeneration of physisorbents is much lower than for chemisorbents (where chemical bonds form between the sorbent and CO₂). Porous physisorbents including activated carbons,^{10–12} zeolites,^{13–16} silica,^{17,18} and porous organic polymers^{19–21} have been intensively studied for CO₂ capture. Some of these sorbents show great potential for post-combustion carbon capture, where CO₂ partial pressures are 0.05–0.15 bar. However, it appears to be more challenging to use conventional porous physisorbents for low-concentration CO₂ capture, as physisorbents tend to have low CO₂ uptake at low-concentrations (*i.e.* low partial pressures). The advantages of these physisorbents for CO₂ capture at low CO₂ concentrations fade significantly in terms of uptake capacity, selectivity, and adsorption kinetics. In the ideal case, a suitably engineered CO₂ sorbent for DAC would not only have high CO₂ uptake capacity at low CO₂ concentrations, but should also show high CO₂ selectivity under the relevant conditions. This means that the ΔH_{ads} of CO₂ sorption at zero or low loading must be significantly lower than that typically observed for physisorbents (~ -20 to -40 kJ mol⁻¹). However, very low enthalpies of CO₂ sorption (*i.e.* < -60 kJ mol⁻¹), or chemisorption of CO₂, may mean that regeneration of the sorbent will be energy-demanding.

Metal–organic frameworks (MOFs) are a type of porous coordination polymers constructed by linking metal ions or clusters with organic linkers *via* coordination bonds.^{22,23} They usually have ordered porous channels and high specific surface areas. The rich coordination chemistry and large amount of available organic linkers endow MOFs with synthetic and structural diversity. Consequently, more than 90 000+ types of MOFs with defined structures have been synthesized so far.²⁴ The unique advantages of MOFs include tunable pore size and surface chemistry. The possibility to pre-design structures and composition renders MOFs attractive for many applications including low-concentration CO₂ capture.²⁵ For example, by judicious selection of the building units, the size, and shape of the pore-aperture of MOFs can be precisely tuned to achieve a high CO₂ separation efficiency by the molecular sieving effect.^{26–28} Formation of ultramicropores,^{29–31} creation of unsaturated metal centers,^{32,33} and amine grafting^{34,35} are effective approaches to introduce strong CO₂ adsorption sites on MOFs that can increase the binding affinity between the sorbents and CO₂. Such functionalized MOFs usually display high CO₂ uptake capacity and high selectivity, even at very low CO₂ concentrations. In this perspective, we will give an overview of the recent advances in the development and engineering of MOFs for low-concentration CO₂ capture. The relationship between the CO₂ capture per-

formances (*e.g.* uptake capacity, selectivity, enthalpies of adsorption, kinetics, cyclic stability) and the MOF structures, as well as possible approaches to structure and upscale MOF sorbents for applications, will be discussed. We also discuss the prospects and challenges when it comes to the use of MOFs for CO₂ capture from low-concentration sources under different circumstances.

Strategies for enhancing CO₂ capture performance on MOFs

On MOFs, the general approach adopted to increase the CO₂ capture performance, especially for low-concentration CO₂ capture, is by tuning the sorption affinity for CO₂ at low-concentrations.³¹ The CO₂ partial pressures (p_{CO_2}) that is of interest range from ~ 0.4 to 10 mbar. This partial pressure range can be considered as equivalent to 400 ppm (current atmospheric CO₂ concentration) up to 10 000 ppm under atmospheric pressure, respectively. The CO₂ adsorption affinity at low-concentrations can be achieved through three main routes: (1) *via* careful control of the pore architecture of the frameworks, (2) through the introduction of high-energy sorption sites (*e.g.* open-metal sites, anionic groups), and (3) by post-synthetic amine-functionalization. In all cases, the CO₂ sorption isotherm of these MOFs should have a very sharp increase in the very low-pressure region (*i.e.* at 0.4 mbar for atmospheric CO₂ concentration). In this section, these three approaches will be discussed. The CO₂ capture performance of all the sorbents discussed in this paper are summarized in Table 1 for comparison.

Engineering pore architecture for CO₂ sorption

The typically weak CO₂ binding affinity on MOFs is related to physisorption-based processes. The weak interaction between CO₂ and the pore surface of MOFs is also indirectly linked to the low CO₂ selectivity over other gases, including the gaseous constituents of indoor air, such as O₂, N₂, and H₂O.³⁶ The development of MOF sorbents for low-concentration CO₂ capture through pore-size tailoring has been attained with a handful of materials. Ultramicroporous MOFs (*i.e.* frameworks containing pores with apertures $< 5\text{--}7$ Å)³⁷ have remarkable CO₂ uptake capacities at low CO₂ concentrations due to their topological pore structures. The ultramicroporous MOF UTSA-16 is an example of such a structure with narrow pore apertures of 3.30×5.40 Å.³⁸ UTSA-16 was found to be capable of selectively interacting with CO₂ over other gases with a reported CO₂ uptake capacity of ~ 0.95 mmol g⁻¹ at $p_{\text{CO}_2} = 50$ mbar, 298 K.³⁹ The CO₂ molecules were found to interact with the terminal water molecules coordinated to K⁺ ions, through hydrogen bonding, in the interior of the diamond-shaped cages (4.5 Å in diameter) in the framework. This interaction resulted in moderate ΔH_{ads} of CO₂ adsorption of ~ -39.7 kJ mol⁻¹ (at near zero-coverage).⁴⁰ The hydrogen bonding interaction between the oxygen atoms in the CO₂ molecules and the crystallographically independent oxygen



Table 1 Comparison of the physical properties and CO₂ uptake capacities of various MOFs tested for low-concentration CO₂ capture. In studies where CO₂ uptake capacity is not list for very low pressures (*i.e.* <1 mbar), the CO₂ uptake capacity at ~50 mbar is listed

| | CO ₂ partial pressure (mbar) | Temperature (K) | Activation temperature (K) | Uptake (mmol g ⁻¹) | Uptake (cm ³ g ⁻¹) | ΔH_{ads} (kJ mol ⁻¹) | Ref. |
|--|---|-----------------|----------------------------|--------------------------------|---|--|-----------|
| NbOFFIVE-1-Ni | 0.4 | 298 | 378 | 1.30 | 29.14 | ~-50 (1 mmol g ⁻¹ CO ₂ loading) | 44 |
| SIFSIX-3-Ni | 0.4 | 298 | 378 | 0.29 | 6.50 | ~-49.8 (1 mmol g ⁻¹ CO ₂ loading) | 44 |
| ZU-36-Ni (GeFSIX-3-Ni) | 0.4 | 298 | 373 | 1.07 | 23.98 | ~-55.5 (near-zero coverage) | 45 |
| ZU-36-Co (GeFSIX-3-Co) | 0.4 | 298 | 373 | 0.30 | 6.72 | ~-39.1 (near-zero coverage) | 45 |
| SIFSIX-3-Cu | 0.4 | 298 | 323 | 1.24 | 27.79 | -54 (~0.25 mmol g ⁻¹ CO ₂ loading) | 46 |
| TIFSIX-3-Ni | 0.4 | 298 | 347/433 | 1.15 ^d | 25.78 | ~-53 (near-zero coverage) | 31 and 47 |
| NbOFFIVE-1-Ni | 0.4 | 298 | 378 | 1.23 ^d | 27.57 | ~-54.9 (~0.1 mmol g ⁻¹ CO ₂ loading) | 47 |
| SIFSIX-3-Cu-i | 0.4 | 298 | 293 | 0.684 | 15.33 | -32 (near-zero coverage) | 46 and 48 |
| SIFSIX-3-Zn | 0.4 | 298 | 393 | 0.13 | 2.91 | ~-45 (near-zero coverage) | 46 and 49 |
| Mg-MOF-74 | 0.4 | 298 | 453 | 0.14 ^d | 3.14 | ~-41.5 (~0.1 mmol g ⁻¹ CO ₂ loading) | 47 and 50 |
| Zn(ZnOH) ₄ (bibta) ₃ | 0.4 | 300 | 373 | 2.20 | 49.31 | ~-42 (near-zero coverage) ~-71 (~2.0 mmol g ⁻¹ CO ₂ loading) | 51 |
| Pyrazine-functionalized Co-MOF-74 | 0.4 | 298 | 393 | 1.36 | 30.48 | -48.4 (near-zero coverage) | 52 |
| Co-MOF-74 | 0.4 | 298 | 393 | 0.65 ^d | 14.57 | — | 52 |
| ZU-16-Co (TIFSIX-3-Co) | 0.4 | 298 | 373 | 1.05 | 23.53 | — | 53 |
| mmen-Mg ₂ (dobpdc) | 0.4 | 298 | 343 | 2.00 ^d | 44.83 | — | 54 |
| en-Mg ₂ (dobpdc) | 0.4 | 298 | 343 | 2.50 ^d | 56.04 | — | 54 |
| en-Mg ₂ (dobpdc) | 0.39 | 298 | 403 | 2.83 | 63.43 | ~-22.5 (near-zero coverage) ~-49--51 (~1.25-2.0 mmol g ⁻¹ CO ₂ loading) | 55 |
| Mg ₂ (dobdc)(N ₂ H ₄) _{1.8} | 0.4 | 298 | 403 | 3.89 | 87.19 | -90 (Virial) -118 (Clausius-Clapeyron) | 56 |
| SIFSIX-3-Ni | 1 | 298 | 378 | 0.62 | 13.90 | ~-49.8 (1 mmol g ⁻¹ CO ₂ loading) | 44 |
| SIFSIX-3-Cu | 1 | 298 | 378 | 1.72 | 38.55 | ~-53 (~1.05 mmol g ⁻¹ CO ₂ loading) | 44 |
| ZU-36-Ni (GeFSIX-3-Ni) | 1 | 298 | 373 | 1.55 | 34.74 | ~-55.5 (near-zero coverage) | 45 |
| ZU-36-Co (GeFSIX-3-Co) | 1 | 298 | 373 | 0.75 | 16.81 | ~-39.1 (near-zero coverage) | 45 |
| SIFSIX-3-Cu | 1 | 298 | 323 | 1.75 | 39.22 | — | 46 |
| NbOFFIVE-1-Ni | 1 | 298 | 378 | 1.68 | 37.66 | — | 44 |
| TIFSIX-3-Ni | 1 | 298 | 347/433 | 1.50 ^d | 33.62 | ~-53 (near-zero coverage) | 31 and 47 |
| Mg-MOF-74 | 1 | 298 | 453 | 0.33 ^d | 7.40 | ~-41.5 (~0.1 mmol g ⁻¹ CO ₂ loading) | 47 |
| Zn(ZnOH) ₄ (bibta) ₃ | 1 | 300 | 373 | 2.35 ^d | 52.67 | ~-42 (near-zero coverage) ~-71 (~2.0 mmol g ⁻¹ CO ₂ loading) | 51 |
| Pyrazine-functionalized Co-MOF-74 | 1 | 298 | 393 | 2.10 ^d | 47.07 | -48.4 (near-zero coverage) | 52 |
| Co-MOF-74 | 1 | 298 | 393 | 0.90 ^d | 20.17 | — | 52 |
| ZU-16-Co (TIFSIX-3-Co) | 1 | 298 | 373 | 1.55 ^d | 34.74 | — | 53 |
| mmen-Mg ₂ (dobpdc) | 1 | 298 | 343 | 3.00 ^d | 67.24 | — | 54 |
| en-Mg ₂ (dobpdc) | 1 | 298 | 343 | 3.20 ^d | 71.72 | — | 54 |
| en-Mg ₂ (dobpdc) | 1 | 298 | 403 | 3.10 ^d | 69.48 | ~-22.5 (near-zero coverage) ~-49--51 (~0.13-2.0 mmol g ⁻¹ CO ₂ loading) | 55 |
| Mg ₂ (dobdc)(N ₂ H ₄) _{1.8} | 1 | 298 | 403 | 4.35 ^d | 97.50 | -90 (Virial) -118 (Clausius-Clapeyron) | 56 |



Table 1 (Contd.)

| | CO ₂ partial pressure (mbar) | Temperature (K) | Activation temperature (K) | Uptake (mmol g ⁻¹) | Uptake (cm ³ g ⁻¹) | ΔH_{ads} (kJ mol ⁻¹) | Ref. |
|--|--|-----------------|----------------------------|--------------------------------|---|---|-----------|
| SIFSIX-3-Cu | 40 | 298 | 323 | 2.36 ^a | 52.90 | -54 (~0.25 mmol g ⁻¹ CO ₂ loading) | 46 |
| SIFSIX-3-Zn | 40 | 298 | 323 | 2.19 ^a | 49.09 | ~-45 (near-zero coverage) | 46 |
| UTSA-16 | 50 | 298 | 363 | 0.95 ^a | 21.29 | -39.7 (near-zero coverage) | 39 and 40 |
| Zn ₂ (Atz) ₂ Ox (MeOH) | 50 | 293 | 373/353 | 1.10 ^a | 24.66 | -40.8 (near-zero coverage) | 41 and 42 |
| Zn ₂ (Atz) ₂ Ox (H ₂ O) | 50 | 283 | 423 | 2.70 ^a | 60.52 | ~-55 (near-zero coverage) | 43 |
| JLU-MOF56 | 50 | 298 | 303 | 0.25 ^a | 5.60 | ~30 (near-zero coverage) | 57 |
| JLU-MOF57 | 50 | 298 | 303 | 0.09 ^a | 2.02 | ~-32.5 (near-zero coverage) | |
| Cu-F-pym | 50 | 298 | 393 | 0.53 ^a | 11.88 | ~-30 (near-zero coverage) | 58 |
| UTSA-280 | 50 | 298 | 383 | 0.85 ^a | 19.05 | -42.9 (near-zero coverage) | 59 |
| SIFSIX-3-Ni | 50 | 293 | 413 | 2.45 ^a | 54.91 | — | 50 |
| ZU-36-Ni (GeFSIX-3-Ni) | 50 | 298 | 373 | 2.60 ^a | 58.28 | ~-55.5 (near-zero coverage) | 45 |
| ZU-36-Co (GeFSIX-3-Co) | 50 | 298 | 373 | 2.65 ^a | 59.40 | ~-39.1 (near-zero coverage) | 45 |
| Zn(ZnOH) ₄ (bibta) ₃ | 50 | 300 | 373 | 3.00 ^a | 67.24 | ~-42 (near-zero coverage) ~-71 (~2.0 mmol g ⁻¹ CO ₂ loading) | 51 |
| Pyrazine-functionalized Co-MOF-74 | 50 | 298 | 393 | 6.60 ^a | 147.93 | -48.4 (near-zero coverage) | 52 |
| Co-MOF-74 | 50 | 298 | 393 | 5.20 ^a | 116.55 | — | 52 |
| ZU-16-Co (TIFSIX-3-Co) | 50 | 298 | 373 | 2.75 ^a | 61.64 | — | 53 |
| mmen-Mg ₂ (dobpdc) | 50 | 298 | 343 | 3.40 ^a | 76.21 | — | 54 |
| en-Mg ₂ (dobpdc) | 50 | 298 | 343 | 3.60 ^a | 80.69 | — | 54 |
| en-Mg ₂ (dobpdc) | 50 | 298 | 403 | 3.50 ^a | 78.45 | ~-22.5 (near-zero coverage) ~-49--51 (~0.13-2.0 mmol g ⁻¹ CO ₂ loading) | 55 |
| Mg ₂ (dobdc)(N ₂ H ₄) _{1.8} | 50 | 298 | 403 | 5.10 ^a | 114.31 | -90 (Virial method) -118 (Clausius-Clapeyron) | 56 |
| SIFSIX-3-Ni | Lab atmosphere, 49% RH | 296.55 | 413 | 0.18 ^a | 4.03 | — | 50 |
| MIL-101(Cr) | 10 vol% CO ₂ (0.1 atm) 100 ppm SO ₂ , 100 ppm NO, 10% RH | 298 | 473 | 0.48 | 10.78 | — | 60 |

^a Data extracted from the original publication, and may be approximate.

atoms in H₂O was also found to account for 74% of the total CO₂ uptake capacity.³⁸ Similarly, Vaidhyanathan *et al.*^{41,42} and Banerjee *et al.*⁴³ reported on a series of solvothermally synthesized ultramicroporous zinc-based aminotriazolate/oxalate (Atz/Ox) MOFs (Zn₂(Atz)₂Ox (solvent)). Specifically, Zn₂(Atz)₂Ox (MeOH) was observed to possess pore channels with apertures of 3.50 × 4.00 Å along the *a*-axis, 3.90 × 2.10 Å along the *b*-axis, and narrow slit-shaped pores (3.00 × 1.60 Å) along the *c*-axis, respectively. The primary amino-groups on the aminotriazolate ligands were found to protrude into the cube-shaped pore cavities (4.00 × 4.00 × 4.00 Å) along the *a*-axis. The high CO₂ uptake capacity of the framework (~1.1 mmol g⁻¹ at *p*_{CO₂} = 50 mbar, 293 K) was found to be due to both pore-size effects as well as from CO₂-amine interactions. This was also indicated by the relatively low ΔH_{ads} of CO₂, ~-40.8 kJ mol⁻¹ at near zero-coverage.^{41,42} Similarly, Zn₂(Atz)₂Ox (H₂O) was shown to possess the same oxalate-pillared structure as Zn₂(Atz)₂Ox (MeOH) with comparable pore

channels and apertures – 5.35 × 5.35 Å along the *a*-axis, 6.40 × 6.40 Å along the *b*-axis, and 5.80 × 5.25 Å along the [0 1 1] direction, respectively (Fig. 1a).⁴³ However, unlike Zn₂(Atz)₂Ox (MeOH), the water-solvated Zn₂(Atz)₂Ox (H₂O) framework was observed to undergo a subtle CO₂-induced structural rearrangement at *p*_{CO₂} = 200 mbar, 273 K (Fig. 1b). This gate-opening effect was observed alongside the appearance of new adsorption sites (denoted as site I and site II) in the framework, which could be separated by their ΔH_{ads} . The ΔH_{ads} pre-gate opening (*i.e.* corresponding to site I) was shown to increase from -46 kJ mol⁻¹ to -32 kJ mol⁻¹ post-gate opening, where CO₂ adsorption occurred on site II. The ΔH_{ads} for the two sites mainly corresponded to amine-CO₂ (site I) and CO₂-CO₂ (site II) interactions.⁴³

Liu *et al.* also reported on two isomorphous triazolate-based ultramicroporous MOFs, namely, JLU-MOF56 ([Ni₂(μ₂-Cl)(BTBA)₂·DMF]·Cl·3DMF) and JLU-MOF57 ([Co₂(μ₂-Cl)(BTBA)₂·DMF]·Cl·3DMF, where BTBA⁴⁻ = 3,5-bis(triazol-1-yl)



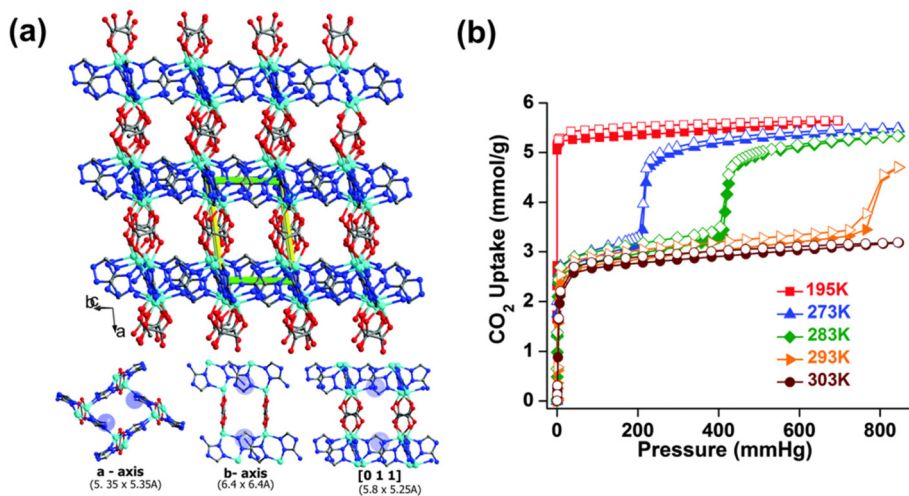


Fig. 1 (a) Three-dimensional structure of $\text{Zn}_2(\text{Atz})_2\text{Ox}$ showing the ultramicroporous channels along a -, b - and c -axis, and (b) CO_2 sorption isotherms of $\text{Zn}_2(\text{Atz})_2\text{Ox}$ at different temperatures.⁴³ Reproduced with permission, Copyright 2015 Royal Society of Chemistry.

benzoate, DMF = N,N' -dimethylformamide).⁵⁷ Both JLU-MOF56 and -57 featured channels with dimensions of $3.50 \times 3.40 \text{ \AA}$, $8.50 \times 2.80 \text{ \AA}$, and $3.50 \times 3.40 \text{ \AA}$ along the a -, b -, and c -axis as well as internal cages 14 \AA in diameter. Despite the presence of uncoordinated N-atom sites, the CO_2 uptake capacities of JLU-MOF56 (0.25 mmol g^{-1} at $p_{\text{CO}_2} = 50 \text{ mbar}$, 298 K) and JLU-MOF57 (0.09 mmol g^{-1} at $p_{\text{CO}_2} = 50 \text{ mbar}$, 298 K) were relatively low at low CO_2 concentrations. This was attributed to the significantly larger dimension of the cages in the framework as compared to the kinetic diameter of CO_2 (3.30 \AA).⁵⁷ Navarro *et al.*⁶¹ and Shi *et al.*⁵⁸ investigated the CO_2 sieving properties of an ultramicroporous MOF possessing appropriately sized channels as well as surface functionalities, namely, Cu-F-pym ($[\text{Cu}(\text{F-pymo})_2(\text{H}_2\text{O})_{1.25}]_n$, where F-pymo = 5-fluorpyrimidin-2-olate). Cu-F-pym exhibited a 3D structure with GIS-related framework topology and possessed helical channels in the ab -plane with an aperture of 2.90 \AA along the c -axis.⁶¹ Selective adsorption of CO_2 ($\sim 0.53 \text{ mmol g}^{-1}$ at $p_{\text{CO}_2} = 53 \text{ mbar}$, 298 K) was observed on Cu-F-pym at ambient temperatures (*i.e.* 293 K) despite the narrow pore aperture. This was attributed to a thermal expansion of the Cu-F-pym framework which enabled the diffusion of CO_2 through the structure.⁶¹ Pore-size tailoring has also been achieved using small non-functionalized linkers. An example of such was presented for the ultramicroporous MOF UTSA-280 ($[\text{Ca}(\text{C}_4\text{O}_4)(\text{H}_2\text{O})] \cdot x\text{H}_2\text{O}$),^{59,62} which utilizes squaric acid to form a 3D framework structure with cylindrical 1D channels ($3.20 \times 4.50 \text{ \AA}$ and $3.80 \times 3.80 \text{ \AA}$ apertures) along the c -axis. The adsorbed CO_2 molecules were found, according to grand canonical Monte Carlo (GCMC) simulations, to interact with the organic linker as well as the coordinating water molecules in the pore channels through van der Waals and electrostatic interactions, giving rise to a ΔH_{ads} of CO_2 of $\sim -42.9 \text{ kJ mol}^{-1}$ from the combined host-guest interactions.⁵⁹

Hybrid ultramicroporous materials (HUMs) have garnered great attention in the last decade due to their unique structural

properties. It is important to note that HUMs may not strictly be classified as MOFs, nevertheless, they will be included in this discussion for comparison. The prototypical HUM structure can broadly be described as 2D square **sql** nets composed of metal-organic units interconnected by pillaring inorganic anions (*e.g.* $[\text{SiF}_6]^{2-}$, $[\text{TiF}_6]^{2-}$, $[\text{NbOF}_5]^{2-}$, and $[\text{GeF}_6]^{2-}$, see Fig. 2).^{45,47,63,64} The inherent structure of HUMs provides them with narrow and highly ordered pore channels that are decorated with polarizing atoms.^{47,64} A comparison between the HUM SIFSIX-3-Ni and TEPA-SBA-15 (tetraethylenepentamine-functionalized mesoporous silica SBA-15) as well as Zeolite-13X was made by Kumar *et al.*^{31,50} The authors showed that SIFSIX-3-Ni had superior CO_2 uptake capacity (0.18 mmol g^{-1}) as compared to Zeolite-13X (0.03 mmol g^{-1}) at 1 bar pure CO_2 with 49% RH (296.55 K). However, the performance of the HUM was observed to be worse than TEPA-SBA-15 (3.59 mmol g^{-1}) under the same conditions. On TEPA-SBA-15, chemisorption of CO_2 accounted for its high CO_2 uptake, especially in the presence of water.⁵⁰ Fine-tuning of the CO_2 uptake properties in SIFSIX-3-M was further attained through the incorporation of different metal cations into the metal-organic unit of the structure. Bhatt *et al.*⁴⁴ observed a narrowing of the square channels in SIFSIX-3-Cu ($d_{\text{F}\dots\text{F}} = 6.483(1) \text{ \AA}$) when compared with SIFSIX-3-Ni ($d_{\text{F}\dots\text{F}} = 6.694(1) \text{ \AA}$) and SIFSIX-3-Zn ($d_{\text{F}\dots\text{F}} = 6.784(1) \text{ \AA}$), due to a reduced distance between adjacent $[\text{SiF}_6]^{2-}$ units. The narrowing of the channel resulted in an enhanced adsorbate-adsorbent interaction at low CO_2 concentrations. This enhanced interaction was also indicated by a decrease in the ΔH_{ads} from -45 kJ mol^{-1} on SIFSIX-3-Ni to -54 kJ mol^{-1} on SIFSIX-3-Cu. As further reported by Bhatt *et al.* the substitution of the pillaring anion from $[\text{SiF}_6]^{2-}$ to $[\text{NbOF}_5]^{2-}$ was also found to decrease the distance between the pendant fluorine moieties due to an increase in the bonding distance for Nb-F ($d_{\text{Nb-F}} = 1.899(1) \text{ \AA}$) as compared to Si-F ($d_{\text{Si-F}} = 1.681(1) \text{ \AA}$). An increase in volumetric CO_2 uptake at $p_{\text{CO}_2} = 0.4 \text{ mbar}$ by 15 to 340% was subsequently observed for





Fig. 2 (a) Schematic illustration of the modularity of pillared square grids that form the pcu topology of HUMs, (b) the prototypal pyrazine (pyz) linked HUM $[\text{Zn}(\text{pyz})_2(\text{SiF}_6)]_n$, and (c) schematic illustration of the building blocks.⁶⁴ Reproduced with permission, Copyright 2021 John Wiley & Sons, Inc.

NbOFFIVE-1-Ni (1.3 mmol g^{-1} at 298 K) over SIFSIX-3-Cu ($\sim 1.25 \text{ mmol g}^{-1}$ at 298 K), SIFSIX-3-Ni ($\sim 0.39 \text{ mmol g}^{-1}$ at 298 K), and SIFSIX-3-Zn ($\sim 0.14 \text{ mmol g}^{-1}$ at 298 K). The increase in CO_2 uptake was attributed to a further decrease in distance between pendant F...F moieties in NbOFFIVE-1-Ni ($d_{\text{F}\cdots\text{F}} = 3.210(8) \text{ \AA}$) compared with SIFSIX-3-Cu ($d_{\text{F}\cdots\text{F}} = 3.483(1) \text{ \AA}$) and SIFSIX-3-Ni ($d_{\text{F}\cdots\text{F}} = 3.694(1) \text{ \AA}$).

Zhang *et al.*⁴⁵ reported the use of $[\text{GeF}_6]^{2-}$ units to slightly reduce the M-F distance ($d_{\text{Ge-F}} = 1.83 \text{ \AA}$) in the inorganic anions in ZU-36-Ni as compared to NbOFFIVE-1-Ni. The formation of an isostructural Co-based HUM, ZU-36-Co, was additionally investigated to assess the pore size and CO_2 uptake capacities in these ZU-36 frameworks. A shortening of the bond distance between the metal cation and the pyrazine linker in the square lattice of ZU-36-Co ($d_{\text{Ni-pyrazine}} = 2.12 \text{ \AA}$) was achieved in ZU-36-Ni ($d_{\text{Ni-pyrazine}} = 2.08 \text{ \AA}$). The decreased metal cation – pyrazine distance led to an enhanced CO_2 uptake capacity in the low-pressure range for ZU-36-Ni (1.07 mmol g^{-1} at $p_{\text{CO}_2} = 0.4 \text{ mbar}$, 1.55 mmol g^{-1} at $p_{\text{CO}_2} = 1 \text{ mbar CO}_2$, 298 K) and corresponded to an over 200% increase from ZU-36-Co (0.30 mmol g^{-1} at $p_{\text{CO}_2} = 0.4 \text{ mbar}$, 0.75 mmol g^{-1} at $p_{\text{CO}_2} = 1 \text{ mbar CO}_2$, 298 K). The CO_2 uptake capacity of ZU-36-Ni was found to be slightly lower than other HUMs such as SIFSIX-3-Cu (1.24 mmol g^{-1} at $p_{\text{CO}_2} = 0.4 \text{ mbar}$, 1.75 mmol g^{-1} at $p_{\text{CO}_2} = 1 \text{ mbar CO}_2$, 298 K) and NbOFFIVE-1-Ni (1.30 mmol g^{-1} at $p_{\text{CO}_2} = 0.4 \text{ mbar}$, 1.68 mmol g^{-1} at $p_{\text{CO}_2} = 1 \text{ mbar CO}_2$, 298 K). The difference in CO_2 uptake may be related to the more optimal F...F distances in SIFSIX-3-Cu and NbOFFIVE-1-Ni than in ZU-36.⁴⁵ Similarly, Kumar *et al.*⁴⁷ investigated another HUM structure that was isorecticular with SIFSIX-3-M, namely TIFSIX-3-Ni. The authors utilized $[\text{TiF}_6]^{2-}$ anionic pillars to further tailor the sorption properties of the SIFSIX-3-M framework. The M-F bond distance in TIFSIX-3-Ni ($d_{\text{Ti-F}} = 1.81 \text{ \AA}$) was found to be similar to ZU-36-Ni ($d_{\text{Ge-F}} = 1.83 \text{ \AA}$) and NbOFFIVE-1-Ni ($d_{\text{Nb-F}} = 1.899(1) \text{ \AA}$). The corresponding CO_2 uptake capacity at $p_{\text{CO}_2} = 0.4 \text{ mbar}$ of TIFSIX-3-Ni ($\sim 1.15 \text{ mmol g}^{-1}$ at 298 K) was not found to differ significantly from NbOFFIVE-1-Ni ($\sim 1.23 \text{ mmol g}^{-1}$ at 298 K) or ZU-36-Ni

(1.07 mmol g^{-1} at 298 K).⁴⁵ The presence of tight CO_2 binding sites was also observed in this framework, as indicated by a low ΔH_{ads} which was compared to other HUMs – TIFSIX-3-Ni, $\sim -49 \text{ kJ mol}^{-1}$ (at $0.1 \text{ mmol g}^{-1} \text{ CO}_2$ loading), NbOFFIVE-1-Ni, $\sim -54.9 \text{ kJ mol}^{-1}$ ($0.1 \text{ mmol g}^{-1} \text{ CO}_2$ loading), ZU-36-Ni, $-55.5 \text{ kJ mol}^{-1}$ (at near-zero loading),⁴⁵ SIFSIX-3-Cu, $\sim -53.0 \text{ kJ mol}^{-1}$ ($0.1 \text{ mmol g}^{-1} \text{ CO}_2$ loading).^{44,50}

The impact of increasing the length of the organic molecule in the metal–organic unit in SIFSIX-3-Cu was further investigated by Shekhah *et al.*⁴⁶ through the substitution of pyrazine in SIFSIX-3-Cu by dipyridylacetylene. This approach resulted in the formation of the isorecticular HUM SIFSIX-2-Cu-i. The authors reported a decrease in CO_2 uptake capacity at $p_{\text{CO}_2} = 0.4 \text{ mbar}$ for SIFSIX-2-Cu-i (0.07 mmol g^{-1} at 298 K) when compared to SIFSIX-3-Cu (1.24 mmol g^{-1} at 298 K) and SIFSIX-3-Zn (0.13 mmol g^{-1} at 298 K). The decrease in CO_2 uptake was attributed to an increase in pore size from 3.5 \AA in SIFSIX-3-Cu to 5.15 \AA in SIFSIX-2-Cu-i. This enlargement of the average distance between the CO_2 molecules and the fluorine atoms in the channels of SIFSIX-2-Cu-i further led to a significantly increased ΔH_{ads} for CO_2 of $\sim -32 \text{ kJ mol}^{-1}$ as compared to SIFSIX-3-Cu (-54 kJ mol^{-1}).⁴⁶ It is therefore clear that the ΔH_{ads} in MOF sorbents play a crucial role in their performance to capture CO_2 . Many sorbents generally exhibit relatively low enthalpies of adsorption ($< -50 \text{ kJ mol}^{-1}$) in the absence of ultramicropores or active functional groups. Thus leading to their poor performance at adsorbing CO_2 at low-concentrations. Various routes for improving the affinity between CO_2 molecules and a framework has been presented in the previous section, however, a compromise is generally required in order for these materials to be utilized in real-world applications. TSA processes were proposed by Lively *et al.*⁶⁵ to be more thermodynamically efficient as compared to PSA processes when utilizing sorbents with lower enthalpies of adsorption (*i.e.* $< -50 \text{ kJ mol}^{-1}$) for dilute steams containing $\sim 100\text{--}1000 \text{ ppm CO}_2$. The PSA process, on the other hand, was found to have low efficiency even for sorbents with relatively low CO_2 affinity (*i.e.* $\Delta H_{\text{ads}} > -35 \text{ kJ mol}^{-1}$). Although sorbents



with low enthalpies of CO₂ adsorption were found to be more suitable for trace CO₂ capture in TSA processes, it is important to note that a too low ΔH_{ads} will significantly increase the regeneration costs.

CO₂ sorption on high-energy sorption sites

Coordinatively unsaturated or open metal sites in MOFs have additionally been used to tune the CO₂ uptake capacity at low CO₂ concentrations. Strong interactions between such sites which exhibit electrostatic fields enable them to interact with the π -orbitals of polarizable guest molecules such as CO₂. The interaction between CO₂ and these high-energy adsorption sites is pivotal for low-concentration CO₂ capture.^{66,67} Notably, MOFs such as M-MOF-74 (e.g. M = Mg²⁺, Fe²⁺, Co²⁺, Zn²⁺, Ni²⁺), HKUST-1, and MIL-101 have been proposed for CO₂ capture at low CO₂ concentrations. Liu *et al.*⁶⁰ investigated the CO₂ capture performance of MIL-101(Cr), a mesoporous MOF containing Cr(III) with open metal sites. This MOF was observed to have a CO₂ uptake of 0.48 mmol g⁻¹ at 298 K and 10% RH from a gas stream containing 100 000 ppm ($\sim p_{\text{CO}_2} = 0.1$ bar) CO₂ as well as 100 ppm SO₂, 100 ppm NO.⁶⁰ Kumar *et al.*⁴⁷ investigated the sorption properties of Mg-MOF-74 and HKUST-1 and compared them with TIFSIX-3-Ni and NbOFFIVE-1-Ni. At $p_{\text{CO}_2} = 0.4$ and 1 mbar, the CO₂ uptake capacities of Mg-MOF-74 (0.14 mmol g⁻¹ at 0.4 mbar, 0.33 mmol g⁻¹ at 1 mbar, 298 K) and HKUST-1 (no uptake at 0.4 mbar, 0.13 mmol g⁻¹ at 1 mbar, 298 K) were significantly lower as compared to the HUMs. The ΔH_{ads} for Mg-MOF-74 (~ -41.5 kJ mol⁻¹ at 0.1 mmol g⁻¹ CO₂ loading) and HKUST-1 (~ -23 kJ mol⁻¹ at 0.1 mmol g⁻¹ CO₂ loading) further confirmed that the presence of open-metal sites in these frameworks can create strong binding sites for CO₂ as compared to conventional physisorption-based interactions. On the other hand, the narrow pore structure of HUMs,^{44,47,64} such as those found in NbOFFIVE-1-Ni,⁴⁴ still offers higher energy CO₂ sorption sites ($\Delta H_{\text{ads}} \sim -54.9$ kJ mol⁻¹) than the open-metal sites on some MOFs.

Post-synthetic modification of MOFs for CO₂ capture

The introduction of functional groups, commonly through post-synthetic modifications, offers several advantages for increasing the binding interaction with CO₂. Typical functional groups include basic groups such as primary and secondary amines, polarizing halogen atoms, and larger hydrocarbon chains. Not only can these functional groups facilitate Lewis acid–base reactions (*i.e.* chemisorption-based adsorption processes), they can also increase the presence of strong electrostatic interactions, and possibly introduce steric effects to enhance adsorbate–adsorbent van der Waals interactions. As such, numerous studies on functionalized MOFs for trace CO₂ capture have been carried out. For example, Bien *et al.*⁵¹ utilized a mild ligand exchange procedure to introduce nucleophilic sites in [Zn(ZnO₂CCH₃)₄(bibta)₃] (bibta²⁻ = 5,5'-bibenzotriazolate) to produce [Zn(ZnOH)₄(bibta)₃], as presented in the schematics shown in Fig. 3.⁵¹ The post-synthetic modification resulted in significantly enhanced CO₂ uptake. The CO₂

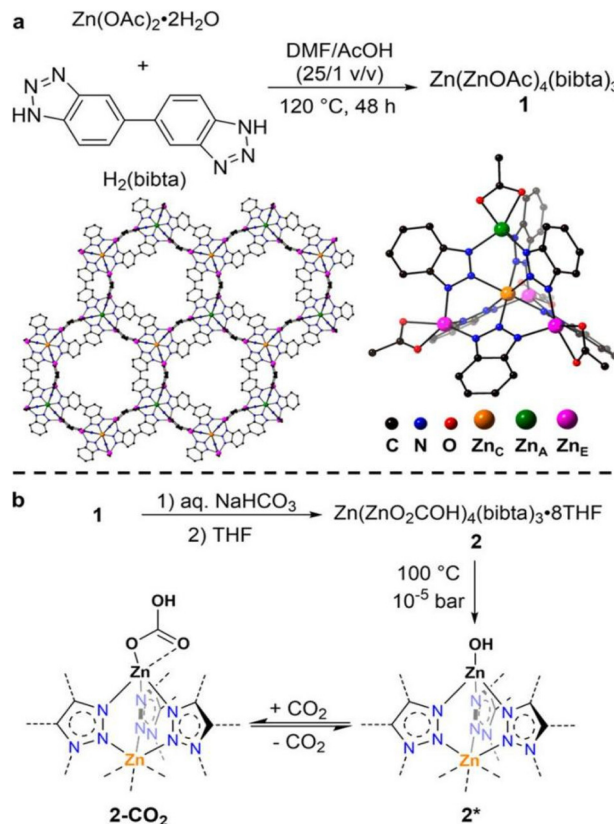


Fig. 3 (a) Synthesis structure of the Zn(ZnOAc)₄ SBUs. (b) Synthesis of Zn(ZnOH)₄(bibta)₃ and mechanism of reversible CO₂ fixation.⁵¹ Reproduced with permission, Copyright 2013 American Chemical Society.

uptake capacity of the modified [Zn(ZnOH)₄(bibta)₃] was 2.20 mmol g⁻¹ (0.4 mbar, 300 K), which was a significant improvement as compared to the negligible uptake on [Zn(ZnO₂CCH₃)₄(bibta)₃] under the same conditions. The chemisorption of CO₂ in [Zn(ZnOH)₄(bibta)₃] was further confirmed by the low ΔH_{ads} for the MOF (-71 kJ mol⁻¹ at ~ 2.0 mmol g⁻¹ CO₂ loading) thus indicating that CO₂ fixation likely occurred *via* a reversible Zn-OH/Zn-O(COOH) route (Fig. 3).⁵¹ Hu *et al.*⁵² also successfully obtained a pyrazine-functionalized Co-MOF-74 by post-synthesis modification. The effective pore size of the Co-MOF-74 was reduced from 11–12 Å down to <7 Å,⁶⁸ at the same time, Lewis basic sites were introduced from the non-coordination N-atoms on the pyrazine molecules. CO₂ uptake for the pyrazine-functionalized Co-MOF reached 1.26 mmol g⁻¹ (0.4 mbar at 298 K), which was significantly higher than its non-functionalized counterpart (~ 0.65 mmol g⁻¹ at 0.4 mbar and 298 K). The uptake capacity on the functionalized MOF was however slightly lower as compared to other high-performing MOFs such as TIFSIX-3-Ni (~ 1.1 mmol g⁻¹ at 298 K)³¹ and ZU-16-Co (~ 1.05 mmol g⁻¹ at 298 K).⁵³ Correspondingly, the ΔH_{ads} of the pyrazine-functionalized Co-MOF-74 was higher (-48.4 kJ mol⁻¹ at zero loading) than other frameworks with highly tailored pore structures (e.g. NbOFFIVE-1-Ni, -54.9 kJ mol⁻¹ at 0.1 mmol g⁻¹ CO₂



Perspective

loading)⁴⁷ or chemisorption-based sorbents (e.g. [Zn(ZnOH)₄(bibta)₃], -71 kJ mol⁻¹ at ~2.0 mmol g⁻¹ CO₂ loading).^{51,52} The moderate ΔH_{ads} of the functionalized MOF was however noticeably lower than the ΔH_{ads} of typical physisorbents, but not as low as the ΔH_{ads} of chemisorbents. This gives pyrazine-functionalized Co-MOF-74 an advantage over other sorbents with respect to energy costs for regeneration.

Similarly, MOFs with coordinatively unsaturated sites have commonly been used to post-synthetically introduce amine moieties. Lewis acid–base reactions between the CO₂ molecules and the accessible amine groups (e.g. primary or secondary amines) on the pore surfaces generally lead to the formation of carbamic acid followed by ammonium carbamate in the presence of humidity.^{35,69} Notable examples include M-MOF-74, Mg₂(dobpdc) (dobpdc⁴⁻ = 4,4'-dioxido-3,3'-biphenyldicarboxylate), and MIL-101(Cr).⁷⁰ Park *et al.* grafted various diamines on to the pores of Mg₂(dobpdc) to increase the CO₂ binding affinity.⁵⁴ Linear and branched diamines with ethylene and propylene linkages were introduced post-synthetically onto Mg₂(dobpdc) (Fig. 4, 5a, and b). The CO₂ uptake capacity of *N*-isopropylethylenediamine-appended Mg₂(dobpdc) (mmen-Mg₂(dobpdc)) and ethylenediamine-appended MOFs at 400/1000 ppm were approximately 2.30/3.00 mmol g⁻¹ and 2.50/3.20 mmol g⁻¹, respectively, at 298 K. The bulky *N*-isopropylethylenediamine introduced steric hindrance through the branched isopropyl-substituent, which may have kinetically restricted the CO₂ diffusion and the accessibility of the amine sites thus resulting in a slightly lower CO₂ uptake.⁵⁴ Similarly, Lee *et al.*⁵⁵ observed that the CO₂ uptake of ethylenediamine-appended Mg₂(dobpdc) (en-Mg₂(dobpdc)) was 2.83 mmol g⁻¹ at 0.39 mbar, 298 K, which was 1.4 times higher than the *N*-isopropylethylenediamine-functionalized counterpart of the MOF (mmen-Mg₂(dobpdc)) at 2.00 mmol g⁻¹ (0.39 mbar, 298 K). The higher uptake capacity of en-Mg₂(dobpdc) as compared to mmen-Mg₂(dobpdc) at low CO₂ pressures was hypothesized by Lee *et al.*⁵⁵ and McDonald *et al.*³⁴ to be related to two factors: (1) the higher accessibility of the primary amine moieties in en-Mg₂(dobpdc), and (2) due to a large increase in entropy connected with the reorganiz-

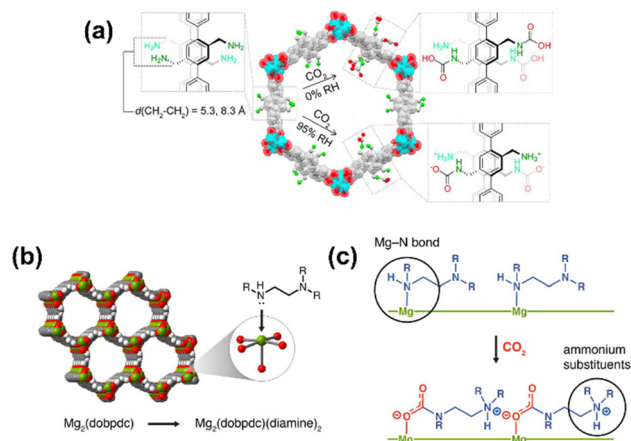


Fig. 5 (a) Chemisorption species post-CO₂ adsorption in IRMOF-74-III-(CH₂NH₂)₂.³⁵ (b) Representative structure of the metal–organic framework Mg₂(dobpdc). Green, red, gray, and white spheres represent Mg, O, C, and H atoms, respectively. (c) Depiction of cooperative CO₂ insertion into a row of Mg²⁺–diamine sites to form ammonium carbamate chains along the pore axis.⁸⁸ Reproduced with permission, Copyright 2017 American Chemical Society.

ation of the secondary amines required for chemisorption of CO₂. These two factors led to preferential CO₂ adsorption onto low-energy sites (*i.e.* not associated with amine groups) and weak amine sites. The ΔH_{ads} between ~1.25–2.0 mmol g⁻¹ CO₂ loading was estimated to range from -49 to -51 kJ mol⁻¹ and corresponded well with the enthalpy of formation (ΔH_f) of carbamic acid (-52.8 kJ mol⁻¹). Additionally, a pressure-induced phase change was also observed in mmen-Mg₂(dobpdc), giving rise to a sharp increase in CO₂ uptake at ~0.2 mbar (298 K). This phenomenon was attributed to a cooperative CO₂ adsorption process wherein the deprotonation of the metal-bound amine by an adjacent non-coordinating amine moiety triggered a nucleophilic addition of CO₂. The resulting formation of a carbamate-ammonium ion pair in turn had a destabilizing effect on the metal-bound amine on the neighboring molecule. This destabilization, in turn, initiated the adsorption of



Fig. 4 (Left) Schematic of synthesis of Mg₂(dobpdc) and (middle) the amine-functionalization process to produce mmen-Mg₂(dobpdc), (right) interaction between pre-treated (degassed) mmen-Mg₂(dobpdc) and CO₂ molecules. Green, red, and gray spheres represent Mg, O, and C atoms respectively; H atoms are omitted for clarity.³⁴ Reproduced with permission, Copyright 2012 American Chemical Society.



another CO₂ molecule through the same process, as described in Fig. 5c.⁷¹ Furthermore, adjacent non-coordinating amines were found to interact through hydrogen bonding in the absence of CO₂⁷² which may prevent these groups from participating in the CO₂ adsorption process at pressures below ~0.2 mbar.^{56,71}

Liao *et al.*⁵⁶ proposed that the incorporation of smaller diamine-species could increase the intermolecular distance between the uncoordinated amines and thus reduce the energetic favorability of hydrogen bonding between the moieties. An enhanced CO₂ uptake capacity in the hydrazine (N₂H₄)-functionalized Mg₂(2,5-dihydroxy-1,4-benzenedicarboxylate - dobdc), [Mg₂(dobdc)(N₂H₄)_{1.8}] of 3.89 mmol g⁻¹ at ~0.4 mbar (298 K) was observed. The CO₂ uptake of the [Mg₂(dobdc)(N₂H₄)_{1.8}] was higher than mmen-Mg₂(dobdc) (2.0 mmol g⁻¹ ~0.4 mbar at 298 K)⁵⁵ as well as en-Mg₂(dobdc) (2.85 mmol g⁻¹ ~0.4 mbar at 298 K).⁵⁵ The low Δ*H*_{ads} (-118 kJ mol⁻¹ at zero-coverage), in conjunction with observations from *in-situ* infrared spectroscopy, supported the formation of carbamate, which was found to be possible due to the long intramolecular distances between neighboring amines. An absence of proton-transfer during the adsorption process was also observed from the spectroscopic measurements.⁵⁶

The three different approaches used to engineer MOFs for low-concentration CO₂ capture discussed here mainly aim to increase the interaction between CO₂ and the adsorption site. All three approaches have been shown to work effectively. The balance between having sufficiently low Δ*H*_{ads} of CO₂ to selectively capture CO₂ at low-concentration, with low enough activation energy for sorption to take place at ambient temperature, and allowing desorption to take place with minimal energy input is perhaps the biggest challenge that deserves continued attention.

CO₂ selectivity, adsorption kinetics, and sorbent stability

As discussed, an ideal sorbent for CO₂ capture at low CO₂ concentrations should have strong binding interactions with CO₂, high CO₂ uptake capacities at low-concentrations, and relatively low Δ*H*_{ads} of CO₂. The utilization of MOFs as CO₂ sorbents in real-life applications is strongly dependent on (1) the selective adsorption of CO₂ over other gaseous constituents in air (particularly H₂O), (2) the framework stability in the presence of guest species, as well as (3) the Δ*H*_{ads} which plays a pivotal role in sorbent regeneration.

Various factors affect the gas selectivity of a sorbent, which include the topological structure and surface chemistry of the materials, and the physicochemical properties of the gas in question. The low CO₂ partial pressure at low CO₂ concentrations (*i.e.* <1 mbar), in conjunction with the significant presence of other gas molecules (*e.g.* H₂O, N₂, O₂ - physical properties of these gases are shown in Table 2) present a remarkably challenging problem for the selective capture of CO₂ in these conditions. Kinetic (also known as partial molecular sieve action) and thermodynamic effects likely play a crucial role in the overall observed CO₂ selectivity of a sorbent.⁷⁷ Shekhah *et al.* observed significantly stronger and more rapid

Table 2 Physical properties of gases commonly found in a mixture with CO₂ in low-concentration CO₂ capture-related application sources^{73–76}

| Gas molecule | Kinetic diameter (Å) | Polarizability (10 ⁻²⁵ cm ³) | Dipole moment (10 ¹⁸ esu ⁻¹ cm ⁻¹) | Quadrupole moment (10 ²⁶ esu ⁻¹ cm ⁻²) |
|------------------|----------------------|---|--|--|
| CO ₂ | 3.30 | 29.1 | 0 | 4.30 |
| H ₂ O | 2.65 | 14.5 | 1.85 | — |
| N ₂ | 3.68 | 17.4 | 0 | 1.52 |
| O ₂ | 3.46 | 15.8 | 0 | 0.39 |

adsorption of CO₂ in SIFSIX-3-Cu than O₂, CH₄, and H₂. The strong and fast adsorption of CO₂ was attributed to a combined kinetic and thermodynamic effect. Partial sieving was achieved in the HUM due to the small size of the CO₂ molecules in conjunction with the large quadrupole moment and polarizability of CO₂. These factors combined enhanced the favorable interaction between CO₂ and the adsorption sites on the HUM.⁴⁶

On the other hand, amine-modified MOFs rely heavily on the accessibility of the amine-sites on the pore surface as well as the formation of necessary intermediate species that govern the chemisorption process. The formation of carbamate/ammonium ion pairs in MOFs, such as that seen on mme-Mg₂(dobpdc) have been found to proceed through the formation of an intermediate species and subsequently followed by a transition state (Fig. 5 and in ref. 71). Although energy is required for the chemisorption process to take place, high temperatures will also simultaneously drive the reaction back towards the starting species, as the formation of the intermediate species is entropically unfavored. Thus, the formation rate of the product is limited by the intermediate species and the rate-limiting step in the reaction is attributed to the chemisorption process.⁷⁸ Stability of the frameworks in the presence of moisture presents an obstacle from a structural integrity point of view and can severely limit the applicability of MOFs in CO₂ capture, especially from humid gas mixtures with low CO₂ concentrations. The overall chemical and thermal stability of MOFs could be improved by utilizing high-valence metals (*e.g.* Zr⁴⁺ and Al³⁺), which tend to form strong coordination bonds with carboxylate-based organic ligands, or similarly, lower-valence metals (*i.e.* soft metals such as Zn²⁺ and Cu²⁺) which form strong coordination bonds with soft basic ligands. These approaches allow the formation of robust frameworks with enhanced stability that can cope with issues related to *e.g.* ligand displacement by water molecules.⁷⁹ Furthermore, the incorporation of ligands with hydrophobic functional groups (*e.g.* fluorine-containing,⁸⁰ alkyl-,⁸¹ or ethyl ester-groups⁸²) or groups that introduce steric effects^{81,82} may impede the diffusion of water molecules through the framework and prevent hydrolysis. Therefore, careful consideration of the chemical compositions of the frameworks is necessary for the utilization of the MOFs in many real-world applications.



The susceptibility of amine species to thermal and oxidative degradation has also been well-documented in solution, *e.g.* in the case of monoethanolamine (MEA) which is commonly used in amine scrubbing for CO₂ capture and separation.^{6,83–86} A handful of studies on amine-grafted solid sorbents, such as MCM-41 silica (TRI-PE-MCM-41), has shown primary amine to be more resilient towards oxidative degradation at high temperatures (>100 °C) as compared to secondary amines.⁸⁷ A cooperative degradation mechanism has also been suggested to take place in diamine-supported MCF silica involving the secondary and terminal primary amines.⁸⁸ However, the stability of amine-modified MOFs has, on the other hand, been studied to a lesser extent. Siegelman *et al.*⁸⁹ observed an improved oxidative stability of 2-ampd-Mg₂(dobpdc) (where 2-ampd = 2-(aminomethyl)piperidine) when exposed to a dry gas mixture (~21% O₂ and ~79% N₂ at 1 bar) at 100 °C for 5 h, as compared to many other silica-based materials functionalized with secondary amines. A negligible reduction in CO₂ uptake capacity was observed with the modified MOF as compared to the pristine material. Furthermore, no oxidative by-products of 2-ampd were detected by either ¹H NMR or IR spectroscopy, in contrast to amine-modified silicas that are prone to excessive oxidative degradation when treated in similar environments. The improved stability of 2-ampd-Mg₂(dobpdc) was attributed to the distance between the metal sites in the framework, which separates the amine species by ~7 Å in the pore channels, thus preventing oxidation reactions from taking place between adjacent amine moieties.⁸⁹ Furthermore, the incorporation of branched diamines (*e.g.* 1,1-dimethylethylenediamine and *N*-ethylethylenediamine) was found to improve the stability of functionalized Mg₂(dobpdc) as compared to primary diamines. Signifying that alkyl substituents may play a significant role in improving framework stability in the presence of oxygen and water vapor.^{90,91}

Cost of sorbent production and scalability

Although MOFs show great potential in various application fields, several restrictions such as high synthesis cost, difficulties in scalability, and processability of MOFs need to be addressed before their commercialization and industrial applications can be achieved. In addition to the exploration of MOFs with new structures and applications, great efforts have been devoted to the development of cost-effective, green, and scalable approaches for the synthesis of MOFs. These approaches are especially important for candidate sorbents that are highly feasible for practical applications;^{92–96} for example, M-MOF-74, UiO-66, HKUST-1, and MIL-series MOFs are relevant for trace CO₂ capture.

Metal ions and organic linkers are the two components for building a MOF structure. Metal nitrates and chlorides with high solubility and weak interfacial interactions are the most common metal sources for the synthesis of coordination complexes including MOFs.⁹² However, the by-products containing chloride and nitrate ions are highly corrosive and toxic in general. Recent studies demonstrated that inexpensive metal acetates, hydroxides, and oxides can be used as precursors for

low-impact synthesis of MOFs.⁹⁷ For example, MIL-53(Al)-NH₂ layers formed directly on the surface or in the channels of anodized aluminium oxide (AAO) substrates by solvothermal reactions, in which the aluminium oxides served as metal precursors to release aluminium ions for the construction of MOF structures.^{98–100} The mixed-matrix membranes (MMMs) based on the AAO-MOF nanocomposites exhibited high CO₂/N₂ selectivities of 34–39 with extremely high CO₂ permeance of up to 3000 GPU (1 GPU = 3.35 × 10⁻¹⁰ mol cm⁻² s⁻¹ Pa⁻¹) that can potentially be developed for low-concentration CO₂ capture. In another example, Majano *et al.* demonstrated the conversion of Cu(OH)₂ into HKUST-1 at room temperature in an aqueous ethanoic solution with high space-time-yields (STY) up to 18 127 kg m⁻³ d⁻¹.¹⁰¹ More importantly, the non-soluble nature of Cu(OH)₂ could avoid the release of copper ions into the solution that may ease the purification of the recycled solution. This synthesis approach opens new opportunities for the industrial production of HKUST-1 with low cost, low energy consumption, and minimal environmental impact. In addition, metal acetylacetonates (acac) with low toxicity are considered as green reagents in industrial processes that have been also used for the synthesis of various MOFs. Avci-Camur *et al.* synthesized several Zr(IV)-carboxylate based MOFs from Zr(acac)₄ with relatively high yields in water, including UiO-66-NH₂, UiO-66-(OH)₂, UiO-66-COOH, and UiO-66-(COOH)₂.¹⁰²

Compared to the low cost and adequate availability of most metal precursors, the organic linkers involved in the synthesis of MOFs are usually costly. For example, polytopic carboxylic acids, the most common linkers for MOF synthesis, are prepared from petrochemical feedstocks through several steps involving the use of massive amounts of organic solvents and the generation of toxic by-products. Therefore, it is greatly desired to develop green MOF synthesis routes from renewable, affordable, and non-toxic linkers. Recently, several studies focused on the synthesis of bio-MOFs using biomolecules and biomass-derived organic compounds.¹⁰³ Biomolecules such as amino acids, nucleobases, proteins, peptides, cyclodextrins, tannins, and saccharides, which contain coordination sites of carboxylate groups or nitrogen and oxygen atoms with lone pairs of electrons have been successfully used for synthesizing a range of bio-MOFs.¹⁰⁴ Fumaric acid, a naturally available organic compound containing two carboxylic acid groups, has been constructed into several fumarate-based MOFs. For example, the aluminum fumarate MOF (Basolite A520), an analogue of MIL-53(Al), has been commercialized by BASF with high STYs of up to 5300 kg m⁻³ d⁻¹ could be achieved.^{105,106} Such bio-MOFs with good biocompatibility and low toxicity offer great hope for the development of many biological and medical applications.¹⁰⁷ In addition, several recent studies have investigated the potential of bio-MOFs for CO₂ capture.¹⁰⁸ Some of the bio-MOFs showed relatively high CO₂ capacity and selectivity at the conditions that are relevant to post-combustion capture of CO₂. For example, Basolite A520 with remarkable water stability and decent CO₂ uptake (2.1 mmol g⁻¹ at 1 bar, 303 K) is suitable for CO₂ capture under wet conditions.¹⁰⁹ Bio-MOF-11, a cobalt-



adeninate obtained by the solvothermal reaction between cobalt acetate and adenine (a type of nucleobase), displayed a high CO₂ uptake of 6.0 mmol g⁻¹ at 1 bar and 273 K and high calculated CO₂/N₂ selectivities up to 81 at 273 K.¹¹⁰ However, the performance of bio-MOFs for low-concentration CO₂ capture is typically somewhat moderate when compared to the abovementioned top performing MOF-based CO₂ sorbents. We anticipate that future studies focusing on amine-modification of bio-MOFs will make these materials suitable for low-concentration CO₂ capture.

In addition to the metal sources and organic linkers, the selection of an appropriate solvent is vital for MOF synthesis, especially for the development of green and scalable synthesis approach. The physical properties of the solvent such as polarity, boiling point, viscosity, as well as the cost and environmental impact should be taken into consideration. DMF with a high boiling point and high polarity is widely employed for the synthesis of numerous MOFs under solvothermal conditions. However, the use of DMF not only increases the cost but also generates hazardous amines upon heating, which is not favourable for industrial-scale synthesis of MOFs. Remarkably, recent studies demonstrated that various MOFs including M-MOF-74, NH₂-MIL-53(Al), HKUST-1 can be synthesized in green solvents such as ethanol and water, and in some cases even at room temperature and ambient pressure.^{96,111–115} For example, Huo *et al.* developed a facile, rapid, and industrially relevant approach for the synthesis of HKUST-1 in water at room temperature with high STY of >2000 kg m⁻³ d⁻¹. D'Amato *et al.* performed the reaction of cerium ammonium nitrate and tetrafluoroterephthalic acid in water and afforded a MOF with MIL-140 topology.¹¹⁶ The obtained MOF exhibited an unusual S-shape CO₂ isotherm displaying a steep adsorption increase at pressure <0.03 bar at 273 K, which was an indication of the specific interactions of the quadrupolar CO₂ molecule (Table 2) with the MOF surface. The high CO₂ adsorption capacity at the low partial pressure, the exceptionally high CO₂/N₂ selectivity of up to 1900, as well as the green synthesis approach, shortlists this MIL-140 type MOF for possible application in low-concentration CO₂ capture. It is worth mentioning that MOFs with good CO₂ uptake capacity (2.5 mmol g⁻¹ at 1 bar and 273 K) can be synthesized with bio-molecules in water at room temperature, as demonstrated by the synthesis of bismuth-based MOFs [Bi₂O(H₂O)₂(C₁₄H₂O₈)_n·*n*H₂O], which was made using ellagic acid as the organic linker.¹¹⁷ Still, solvent-based synthesis will inevitably produce liquid waste containing metal ions and organics that are harmful to the environment. The ideal case would be for the metal precursors and the organic linkers to be entirely converted into desired MOFs *via* solid-state reactions with no waste/by-products (including solvent waste). Mechanochemistry has been shown as a green chemistry approach for the synthesis of a variety of functional porous materials such as covalent organic frameworks,¹¹⁸ zeolites,¹¹⁹ as well as MOFs.¹²⁰ For example, various MOFs have been successfully synthesized with high yields by ball milling or grinding of the starting materials at room temperature without any

solvent.¹²¹ In some cases, trace-amount of solvent is needed to increase the reactivity of the reagent. The obtained MOFs have high specific surface areas and high crystallinity that are comparable to those prepared by conventional methods. For example, Julien *et al.* demonstrated the synthesis of highly porous and crystalline Zn-MOF-74 on a gram scale by mechanochemical milling of zinc oxide and 2,5-dihydroxyterephthalic acid without using bulk solvents. Their synthesis approach offers a fast, efficient, cost-effective, and environmentally friendly synthesis method in comparison with conventional solvothermal synthesis.¹²² The milling reaction was monitored by real-time *in-situ* X-ray powder diffraction technique and it revealed a stepwise reaction mechanism. The formation of Zn-MOF-74 done by Julien *et al.* proceeds *via* a close-packed intermediate. Zhang *et al.* developed a solvent-free mechanochemical-assisted approach to the synthesis of ZU-36-Ni (GeFSIX-3-Ni).⁴⁵ Given the ultramicroporous structure, the obtained MOF displayed excellent performance for low-concentration CO₂ capture, exhibiting high CO₂ uptake of 1.07 mmol g⁻¹ at 0.4 mbar and benchmark CO₂/N₂ separation selectivity of 4300 at 273 K. More interestingly, Chen *et al.* showed that mechanochemical methods can be applied in the synthesis of three-component MOFs consisting of mixed organic linkers. The method enables large-scale synthesis of ultramicroporous MOFs of Zn-atz-ipa and Zn-datz-ipa (atz = 3-amino-1,2,4-triazole, datz = 3,5-diamino-1,2,4-triazole, ipa = isophthalic acid) at room temperature with high STY up to 4800 kg m⁻³ d⁻¹.¹²³ Although not all studies discussed in this section have focused on low-concentration CO₂ capture, we believe that the different approaches could be applied to the development of MOFs targeted for low-concentration CO₂ capture.

Structuring of sorbents for real-life application

Prior to practical CO₂ capture applications, the as-synthesized MOFs sorbents have to be processed into certain shapes (*e.g.*, pellets, granules, films, fibers). However, synthesis of MOFs typically yields solid crystalline powders that can be brittle, insoluble, and infusible. These features of the as-synthesized MOFs provide challenges in processing and shaping these sorbents and have significantly hampered their practical applications. The employment of conventional powder processing methods to shape MOF micro/nanocrystals, such as the use of binders, will block the porous channels and thus compromise the adsorption performances of the MOF sorbents.¹²⁴ For instance, Chang *et al.* developed a wet granulation method for shaping a family of MOF nano/microcrystals into mechanically strong granules with the assistance of 5 wt% of mesoporous RHO alumina as the inorganic binder. The shaping process involved wetting and mixing, granulation, and drying procedures, during which the interaction of hydroxyl groups on the MOF and the binder particles assisted in the formation of a monolithic structure. Similarly, Mathe *et al.* employed sucrose as an organic binder to process zirconium-based MOF powder into pellets *via* steps of mixing, sieving, and granulation. The MOF pellets can be produced on a kilogram scale in a relatively short operation time of 30 min. However, up to



50% of the porosity was lost during the shaping due to the pore-blocking effect.¹²⁵ Very recently, Lee *et al.* reported the scalable synthesis of highly uniform Mg-based MOF (Mg₂(dobpdc)) crystals along with the open-porous fiber networks for CO₂ capture.¹²⁶ Specifically, MgO particles were initially loaded into the porous poly(ether imide) (PEI) fiber matrix, and the obtained MgO/PEI precursor fiber was reacted with the organic linker by solvothermal treatment to convert the MgO particles into Mg₂(dobpdc) crystals. The Mg₂(dobpdc)/PEI composite was further soaked in diamine solution to form the fiber sorbents diamine-Mg₂(dobpdc)/PEI. Similarly, Quan *et al.* developed a straightforward and scalable approach to fabricate hollow fiber sorbents by incorporation of diamine-Mg₂(dobpdc) into poly(ethersulfone) (PES) substrate through a conventional “dry-jet, wet-quench” method.¹²⁷ These porous fiber sorbents showed unprecedented cyclic CO₂ capacities in conditions that are relevant to both post-combustion capture of CO₂ and low-concentration CO₂ capture. In addition, biopolymers such as cellulose, chitosan, have been recently studied as substrates to process MOFs into free-standing films and foams.^{128–130} The use of such renewable and biodegradable substrates in MOF processing opens new possibilities to develop sustainable CO₂ sorbents based on MOF composites.

The emerging 3D printing technology, also known as additive manufacturing, could potentially overcome the drawbacks such as loss of surface area and clogging associated with applying conventional granulation and pelletization techniques in MOF processing.^{131–134} The technique offers an opportunity to process MOFs into desired shapes and geometries in an easy-to-handle form. More importantly, the printed MOF composites could have high MOF loading with preserved MOF structure and properties. To date, various printing techniques including fused filament fabrication (FFF), digital light processing (DLP), selective laser sintering (SLS), and direct ink writing (DIW) have been employed to formulate MOFs for CO₂ adsorption and separation studies. Rezaei *et al.* employed DIW method to fabricate Ni-MOF-74 and UTSA-16(Co) powders into mechanically stable monoliths with the assistance of clay as a binder and poly(vinyl alcohol) (PVA) as a plasticizer.¹³⁵ The 3D-printed MOF monoliths had high MOF loading of up to 85 wt% and high surface areas of up to 737 m² g⁻¹. Remarkably, the CO₂ adsorption capacities of the MOFs were fully retained during the printing. The MOF monoliths showed relatively high CO₂ uptakes of >1.3 mmol g⁻¹ at a low p_{CO_2} = 5 mbar and 298 K. In a similar approach, the same group prepared amine-functionalized MIL-101 monoliths, which showed high CO₂ uptakes of up to 1.6 mmol g⁻¹ at p_{CO_2} = 3 mbar at 298 K.¹³⁶ It is noteworthy that the choices of solvent and binder are extremely important to form stable MOF monoliths *via* the DIW printing approach. Indeed, there are significant challenges for 3D printing MOFs at the industrial scale with concerns of cost and material stability. It is hoped that these issues will be addressed by future studies and 3D printing will be a powerful approach for processing various sorbents. We expected that the knowledge transfer from 3D printing to MOF

processing would overcome the difficulties in shaping and structuring MOFs and promote their practical applications in trace CO₂ capture as well as various separation processes.

Conclusion and future outlook

Significant efforts in developing technologies to combat the high emission of CO₂ from point sources have been put in by researchers all over the world in the last few decades. CO₂ capture technologies have been implemented with the use of liquid-based amine scrubbing processes. In recent years, there has been notable development on adsorption-based technologies as an alternative to amine scrubbing. At the same time, low-concentration CO₂ capture, including direct air capture (DAC) and trace CO₂ capture (*i.e.* from very low partial pressures) through adsorption-based technologies have been put forward as the next step in reducing the atmospheric concentration of CO₂. However, for efficient low-concentration CO₂ capture, the ideal sorbent must be able to selectively capture CO₂ with high CO₂ capacity. This means that the sorbent must have a high affinity for CO₂ (*i.e.* low enthalpy of CO₂ sorption), but not so high that the regeneration of the sorbent becomes energy intensive. A summary of the recent advances in the development of metal-organic frameworks (MOFs) and the closely related hybrid ultramicroporous materials (HUMs) was presented here. These potential sorbents have been engineered through different approaches to have many of the desirable properties needed for a good DAC CO₂ sorbent. Such approaches include pore size tuning, engineering open metal sites, and the introduction of functional groups with high affinities for CO₂ sorption. Significant advances have been made with respect to sorbent properties in recent years. Challenges remain in further improving the CO₂ uptake performance of these MOF sorbents, adopting sustainable, and energy-efficient green synthesis routes as well as structuring sorbents *via* innovative methods such as 3D printing.

Conflicts of interest

The authors declare that there are no conflicts of interest.

Acknowledgements

The authors thank the Swedish Research Council (VR, grant no. 2020-04029), the Åforsk foundation (grant no 19-493 and 22-54) and the Swedish Foundation for Strategic Environmental Research (Mistra) (project name: Mistra TerraClean, project number 2015/31) for their financial support.

References

- 1 R. S. Haszeldine, *Science*, 2009, **325**, 1647–1652.



- 2 N. O. O. A. US, *Department of Commerce, Global Monitoring Laboratory – Carbon Cycle Greenhouse Gases, GML*, https://gml.noaa.gov/ccgg/trends/gl_trend.html, (accessed on December 5, 2022).
- 3 S. Mukherjee, A. Kumar and M. J. Zaworotko, in *Metal-Organic Frameworks (MOFs) for Environmental Applications*, ed. S. K. Ghosh, Elsevier, Amsterdam, 2019, ch. 2, pp. 5–61, DOI: [10.1016/B978-0-12-814633-0.00003-X](https://doi.org/10.1016/B978-0-12-814633-0.00003-X).
- 4 R. Custelcean, *Chem. Sci.*, 2021, **12**, 12518–12528.
- 5 T. Maldal and I. M. Tappel, *Energy*, 2004, **29**, 1403–1411.
- 6 G. T. Rochelle, *Science*, 2009, **325**, 1652–1654.
- 7 R. Ben-Mansour, M. A. Habib, O. E. Bamidele, M. Basha, N. A. A. Qasem, A. Peedikakkal, T. Laoui and M. Ali, *Appl. Energy*, 2016, **161**, 225–255.
- 8 N. Hedin, L. Andersson, L. Bergström and J. Yan, *Appl. Energy*, 2013, **104**, 418–433.
- 9 R. L. Siegelman, E. J. Kim and J. R. Long, *Nat. Mater.*, 2021, **20**, 1060–1072.
- 10 M. Sevilla and A. B. Fuertes, *Energy Environ. Sci.*, 2011, **4**, 1765–1771.
- 11 G. Singh, K. S. Lakhi, S. Sil, S. V. Bhosale, I. Kim, K. Albahily and A. Vinu, *Carbon*, 2019, **148**, 164–186.
- 12 C. Xu and M. Strømme, *Nanomaterials*, 2019, **9**, 103.
- 13 O. Cheung and N. Hedin, *RSC Adv.*, 2014, **4**, 14480–14494.
- 14 S. Kumar, R. Srivastava and J. Koh, *J. CO₂ Util.*, 2020, **41**, 101251.
- 15 M. Xu, S. Chen, D.-K. Seo and S. Deng, *Chem. Eng. J.*, 2019, **371**, 693–705.
- 16 A. A. Dabbawala, I. Ismail, B. V. Vaithilingam, K. Polychronopoulou, G. Singaravel, S. Morin, M. Berthod and Y. Al Wahedi, *Microporous Mesoporous Mater.*, 2020, **303**, 110261.
- 17 W. Chaikittisilp, R. Khunsupat, T. T. Chen and C. W. Jones, *Ind. Eng. Chem. Res.*, 2011, **50**, 14203–14210.
- 18 C. Chen, S. Zhang, K. H. Row and W.-S. Ahn, *J. Energy Chem.*, 2017, **26**, 868–880.
- 19 W. Wang, M. Zhou and D. Yuan, *J. Mater. Chem. A*, 2017, **5**, 1334–1347.
- 20 L. Zou, Y. Sun, S. Che, X. Yang, X. Wang, M. Bosch, Q. Wang, H. Li, M. Smith, S. Yuan, Z. Perry and H.-C. Zhou, *Adv. Mater.*, 2017, **29**, 1700229.
- 21 C. Xu, G. Yu, J. Yuan, M. Strømme and N. Hedin, *Mater. Today Adv.*, 2020, **6**, 100052.
- 22 H. Furukawa, K. E. Cordova, M. O’Keeffe and O. M. Yaghi, *Science*, 2013, **341**, 1230444.
- 23 S. Kitagawa, R. Kitaura and S.-i. Noro, *Angew. Chem., Int. Ed.*, 2004, **43**, 2334–2375.
- 24 S. M. Moosavi, A. Nandy, K. M. Jablonka, D. Ongari, J. P. Janet, P. G. Boyd, Y. Lee, B. Smit, H. J. Kulik, S. M. Moosavi, A. Nandy, K. M. Jablonka, D. Ongari, J. P. Janet, P. G. Boyd, Y. Lee, B. Smit and H. J. Kulik, *Nat. Commun.*, 2020, **11**, 4068.
- 25 J. Liu, Y. Wei and Y. Zhao, *ACS Sustainable Chem. Eng.*, 2019, **7**, 82–93.
- 26 Q. Wang, J. Bai, Z. Lu, Y. Pan and X. You, *Chem. Commun.*, 2016, **52**, 443–452.
- 27 R.-B. Lin, Z. Zhang and B. Chen, *Acc. Chem. Res.*, 2021, **54**, 3362–3376.
- 28 M. Åhlén, A. Jaworski, M. Strømme and O. Cheung, *Chem. Eng. J.*, 2021, **422**, 130117.
- 29 Z. Zhang, S. B. Peh, R. Krishna, C. Kang, K. Chai, Y. Wang, D. Shi and D. Zhao, *Angew. Chem., Int. Ed.*, 2021, **60**, 17198–17204.
- 30 S. Nandi, P. De Luna, T. D. Daff, J. Rother, M. Liu, W. Buchanan, A. I. Hawari, T. K. Woo and R. Vaidhyanathan, *Sci. Adv.*, 2015, **11**, e1500421.
- 31 S. Mukherjee, N. Sikdar, D. O’Nolan, D. M. Franz, V. Gascón, A. Kumar, N. Kumar, H. S. Scott, D. G. Madden, P. E. Kruger, B. Space and M. J. Zaworotko, *Sci. Adv.*, 2019, **5**, eaax9171.
- 32 J. A. Mason, K. Sumida, Z. R. Herm, R. Krishna and J. R. Long, *Energy Environ. Sci.*, 2011, **4**, 3030–3040.
- 33 P. D. C. Dietzel, V. Besikiotis and R. Blom, *J. Mater. Chem.*, 2009, **19**, 7362–7370.
- 34 T. M. McDonald, W. R. Lee, J. A. Mason, B. M. Wiers, C. S. Hong and J. R. Long, *J. Am. Chem. Soc.*, 2012, **134**, 7056–7065.
- 35 R. W. Flaig, T. M. O. Popp, A. M. Fracaroli, E. A. Kapustin, M. J. Kalmutzki, R. M. Altamimi, F. Fathieh, J. A. Reimer and O. M. Yaghi, *J. Am. Chem. Soc.*, 2017, **139**, 12125–12128.
- 36 D. G. Madden, H. S. Scott, A. Kumar, K.-J. Chen, R. Sanii, A. Bajpai, M. Lusi, T. Curtin, J. J. Perry and M. J. Zaworotko, *Philos. Trans. R. Soc., A*, 2017, **375**, 20160025.
- 37 K. Adil, Y. Belmabkhout, R. S. Pillai, A. Cadiou, P. M. Bhatt, A. H. Assen, G. Maurin and M. Eddaoudi, *Chem. Soc. Rev.*, 2017, **46**, 3402–3430.
- 38 S. Xiang, Y. He, Z. Zhang, H. Wu, W. Zhou, R. Krishna and B. Chen, *Nat. Commun.*, 2012, **3**, 954.
- 39 A. Masala, F. Grifasi, C. Atzori, J. G. Vitillo, L. Mino, F. Bonino, M. R. Chierotti and S. Bordiga, *J. Phys. Chem. C*, 2016, **120**, 12068–12074.
- 40 A. Masala, J. G. Vitillo, F. Bonino, M. Manzoli, C. A. Grande and S. Bordiga, *Phys. Chem. Chem. Phys.*, 2016, **18**, 220–227.
- 41 R. Vaidhyanathan, S. S. Iremonger, K. W. Dawson and G. K. H. Shimizu, *Chem. Commun.*, 2009, 5230–5232.
- 42 R. Vaidhyanathan, S. S. Iremonger, G. K. H. Shimizu, P. G. Boyd, S. Alavi and T. K. Woo, *Science*, 2010, **330**, 650–653.
- 43 A. Banerjee, S. Nandi, P. Nasa and R. Vaidhyanathan, *Chem. Commun.*, 2016, **52**, 1851–1854.
- 44 P. M. Bhatt, Y. Belmabkhout, A. Cadiou, K. Adil, O. Shekhah, A. Shkurenko, L. J. Barbour and M. Eddaoudi, *J. Am. Chem. Soc.*, 2016, **138**, 9301–9307.
- 45 Z. Zhang, Q. Ding, S. B. Peh, D. Zhao, J. Cui, X. Cui and H. Xing, *Chem. Commun.*, 2020, **56**, 7726–7729.
- 46 O. Shekhah, Y. Belmabkhout, Z. Chen, V. Guillermin, A. Cairns, K. Adil and M. Eddaoudi, *Nat. Commun.*, 2014, **5**, 4228.
- 47 A. Kumar, C. Hua, D. G. Madden, D. O’Nolan, K.-J. Chen, L.-A. J. Keane, J. J. Perry and M. J. Zaworotko, *Chem. Commun.*, 2017, **53**, 5946–5949.



- 48 P. Nugent, Y. Belmabkhout, S. D. Burd, A. J. Cairns, R. Luebke, K. Forrest, T. Pham, S. Ma, B. Space, L. Wojtas, M. Eddaoudi and M. J. Zaworotko, *Nature*, 2013, **495**, 80–84.
- 49 K. Uemura, A. Maeda, T. K. Maji, P. Kanoo and H. Kita, *Eur. J. Inorg. Chem.*, 2009, 2329–2337.
- 50 A. Kumar, D. G. Madden, M. Lusi, K.-J. Chen, E. A. Daniels, T. Curtin, J. J. Perry IV and M. J. Zaworotko, *Angew. Chem., Int. Ed.*, 2015, **54**, 14372–14377.
- 51 C. E. Bien, K. K. Chen, S.-C. Chien, B. R. Reiner, L.-C. Lin, C. R. Wade and W. S. W. Ho, *J. Am. Chem. Soc.*, 2018, **140**, 12662–12666.
- 52 P. Hu, H. Liu, H. Wang, J. Zhou, Y. Wang and H. Ji, *J. Mater. Chem. A*, 2022, **10**, 881–890.
- 53 Z. Zhang, Q. Ding, J. Cui, X. Cui and H. Xing, *Sci. China Mater.*, 2020, **64**, 691–697.
- 54 J. Park, J. R. Park, J. H. Choe, S. Kim, M. Kang, D. W. Kang, J. Y. Kim, Y. W. Jeong and C. S. Hong, *ACS Appl. Mater. Interfaces*, 2020, **12**, 50534–50540.
- 55 W. R. Lee, S. Y. Hwang, D. W. Ryu, K. S. Lim, S. S. Han, D. Moon, J. Choi and C. S. Hong, *Energy Environ. Sci.*, 2014, **7**, 744–751.
- 56 P.-Q. Liao, X.-W. Chen, S.-Y. Liu, X.-Y. Li, Y.-T. Xu, M. Tang, Z. Rui, H. Ji, J.-P. Zhang and X.-M. Chen, *Chem. Sci.*, 2016, **7**, 6528–6533.
- 57 S. Liu, S. Yao, B. Liu, X. Sun, Y. Yuan, G. Li, L. Zhang and Y. Liu, *Dalton Trans.*, 2019, **48**, 1680–1685.
- 58 Y. Shi, Y. Xie, H. Cui, Z. A. Allothman, O. Alduhaish, R.-B. Lin and B. Chen, *Chem. Eng. J.*, 2022, **446**, 137101.
- 59 R.-B. Lin, L. Li, A. Alsalmeh and B. Chen, *Small Struct.*, 2020, **1**, 2000022.
- 60 Q. Liu, L. Ning, S. Zheng, M. Tao, Y. Shi and Y. He, *Sci. Rep.*, 2013, **3**, 2916.
- 61 J. A. R. Navarro, E. Barea, A. Rodríguez-Diéguez, J. M. Salas, C. O. Ania, J. B. Parra, N. Masciocchi, S. Galli and A. Sironi, *J. Am. Chem. Soc.*, 2008, **130**, 3978–3984.
- 62 R.-B. Lin, L. Li, H.-L. Zhou, H. Wu, C. He, S. Li, R. Krishna, J. Li, W. Zhou and B. Chen, *Nat. Mater.*, 2018, **17**, 1128–1133.
- 63 S. Subramanian and M. J. Zaworotko, *Angew. Chem., Int. Ed. Engl.*, 1995, **34**, 2127–2129.
- 64 S. Mukherjee, N. Kumar, A. A. Bezrukov, K. Tan, T. Pham, K. A. Forrest, K. A. Oyekan, O. T. Qazvini, D. G. Madden, B. Space and M. J. Zaworotko, *Angew. Chem., Int. Ed.*, 2021, **60**, 10902–10909.
- 65 R. P. Lively and M. J. Realff, *AIChE J.*, 2016, **62**, 3699–3705.
- 66 T. Quainoo, S. N. Lavan and Z.-F. Liu, *J. Mater. Res.*, 2021, **37**, 334–345.
- 67 Y. Xie, H. Cui, H. Wu, R.-B. Lin, W. Zhou and B. Chen, *Angew. Chem., Int. Ed.*, 2021, **60**, 9604–9609.
- 68 I. Strauss, A. Mundstock, M. Treger, K. Lange, S. Hwang, C. Chmelik, P. Rusch, N. C. Bigall, T. Pichler, H. Shiozawa and J. Caro, *ACS Appl. Mater. Interfaces*, 2019, **11**, 14175–14181.
- 69 A. C. Forse, P. J. Milner, J.-H. Lee, H. N. Redfearn, J. Oktawiec, R. L. Siegelman, J. D. Martell, B. Dinakar, L. B. Zasada, M. I. Gonzalez, J. B. Neaton, J. R. Long and J. A. Reimer, *J. Am. Chem. Soc.*, 2018, **140**, 18016–18031.
- 70 A. Sinha, L. A. Darunte, C. W. Jones, M. J. Realff and Y. Kawajiri, *Ind. Eng. Chem. Res.*, 2017, **56**, 750–764.
- 71 T. M. McDonald, J. A. Mason, X. Kong, E. D. Bloch, D. Gygi, A. Dani, V. Crocellà, F. Giordanino, S. O. Odoh, W. S. Drisdell, B. Vlasisvljevich, A. L. Dzubak, R. Poloni, S. K. Schnell, N. Planas, K. Lee, T. Pascal, L. F. Wan, D. Prendergast, J. B. Neaton, B. Smit, J. B. Kortright, L. Gagliardi, S. Bordiga, J. A. Reimer and J. R. Long, *Nature*, 2015, **519**, 303–308.
- 72 N. Planas, A. L. Dzubak, R. Poloni, L.-C. Lin, A. McManus, T. M. McDonald, J. B. Neaton, J. R. Long, B. Smit and L. Gagliardi, *J. Am. Chem. Soc.*, 2013, **135**, 7402–7405.
- 73 Y. A. Majid, S. K. Garg and D. W. Davidson, *Can. J. Chem.*, 1968, **46**, 1683–1690.
- 74 G. Sethia, R. S. Somani and H. C. Bajaj, *RSC Adv.*, 2015, **5**, 12773–12781.
- 75 X. Wang, L. Zhong, Y. Cressault, A. Gleizes and M. Rong, *J. Phys. D: Appl. Phys.*, 2014, **47**, 495201.
- 76 O. M. Yaghi, M. J. Kalmutzki and C. S. Diercks, in *Introduction to Reticular Chemistry: Metal-Organic Frameworks and Covalent Organic Frameworks*, ed. O. M. Yaghi, M. J. Kalmutzki and C. S. Diercks, Wiley-VCH Verlag GmbH & Co. KGaA, Weinheim, Germany, 2019, ch. 14, pp. 313–314, DOI: [10.1002/9783527821099.ch14](https://doi.org/10.1002/9783527821099.ch14).
- 77 J.-R. Li, R. J. Kuppler and H.-C. Zhou, *Chem. Soc. Rev.*, 2009, **38**, 1477–1504.
- 78 J. D. Martell, P. J. Milner, R. L. Siegelman and J. R. Long, *Chem. Sci.*, 2020, **11**, 6457–6471.
- 79 X. Zhang, B. Wang, A. Alsalmeh, S. Xiang, Z. Zhang and B. Chen, *Coord. Chem. Rev.*, 2020, **423**, 213507.
- 80 C. Yang, U. Kaipa, Q. Z. Mather, X. Wang, V. Nesterov, A. F. Venero and M. A. Omary, *J. Am. Chem. Soc.*, 2011, **133**, 18094–18097.
- 81 Y. Chen, B. Wang, X. Wang, L. H. Xie, J. Li, Y. Xie and J. R. Li, *ACS Appl. Mater. Interfaces*, 2017, **9**, 27027–27035.
- 82 J. M. Taylor, R. Vaidhyanathan, S. S. Iremonger and G. K. Shimizu, *J. Am. Chem. Soc.*, 2012, **134**, 14338–14340.
- 83 S. A. Bedell, C. M. Worley, K. Darst and K. Simmons, *Int. J. Greenhouse Gas Control*, 2011, **5**, 401–404.
- 84 K. Veltman, B. Singh and E. G. Hertwich, *Environ. Sci. Technol.*, 2010, **44**, 1496–1502.
- 85 K.-S. Zoannou, D. J. Sapsford and A. J. Griffiths, *Int. J. Greenhouse Gas Control*, 2013, **17**, 423–430.
- 86 H. Lepaumier, D. Picq and P.-L. Carrette, *Ind. Eng. Chem. Res.*, 2009, **48**, 9068–9075.
- 87 A. Heydari-Gorji, Y. Belmabkhout and A. Sayari, *Microporous Mesoporous Mater.*, 2011, **145**, 146–149.
- 88 P. Bollini, S. Choi, J. H. Drese and C. W. Jones, *Energy Fuels*, 2011, **25**, 2416–2425.
- 89 R. L. Siegelman, T. M. McDonald, M. I. Gonzalez, J. D. Martell, P. J. Milner, J. A. Mason, A. H. Berger, A. S. Bhowm and J. R. Long, *J. Am. Chem. Soc.*, 2017, **139**, 10526–10538.



- 90 H. Jo, W. R. Lee, N. W. Kim, H. Jung, K. S. Lim, J. E. Kim, D. W. Kang, H. Lee, V. Hiremath, J. G. Seo, H. Jin, D. Moon, S. S. Han and C. S. Hong, *ChemSusChem*, 2017, **10**, 541–550.
- 91 W. R. Lee, J. E. Kim, S. J. Lee, M. Kang, D. W. Kang, H. Y. Lee, V. Hiremath, J. G. Seo, H. Jin, D. Moon, M. Cho, Y. Jung and C. S. Hong, *ChemSusChem*, 2018, **11**, 1694–1707.
- 92 S. Kumar, S. Jain, M. Nehra, N. Dilbaghi, G. Marrazza and K.-H. Kim, *Coord. Chem. Rev.*, 2020, **420**, 213407.
- 93 A. Garcia Marquez, P. Horcajada, D. Grosso, G. Férey, C. Serre, C. Sanchez and C. Boissiere, *Chem. Commun.*, 2013, **49**, 3848–3850.
- 94 T. He, X. Xu, B. Ni, H. Wang, Y. Long, W. Hu and X. Wang, *Nanoscale*, 2017, **9**, 19209–19215.
- 95 Y. Pan, Y. Liu, G. Zeng, L. Zhao and Z. Lai, *Chem. Commun.*, 2011, **47**, 2071–2073.
- 96 J. Huo, M. Brightwell, S. El Hankari, A. Garai and D. Bradshaw, *J. Mater. Chem. A*, 2013, **1**, 15220–15223.
- 97 G. Zhan and H. C. Zeng, *Chem. Commun.*, 2016, **53**, 72–81.
- 98 C. Yim, M. Lee, M. Yun, G.-H. Kim, K. T. Kim and S. Jeon, *Sci. Rep.*, 2015, **5**, 10674.
- 99 K. Xie, Q. Fu, C. Xu, H. Lu, Q. Zhao, R. Curtain, D. Gu, P. A. Webley and G. G. Qiao, *Energy Environ. Sci.*, 2018, **11**, 544–550.
- 100 Y. Yu, X.-J. Wu, M. Zhao, Q. Ma, J. Chen, B. Chen, M. Sindoro, J. Yang, S. Han, Q. Lu and H. Zhang, *Angew. Chem.*, 2017, **129**, 593–596.
- 101 G. Majano and J. Pérez-Ramírez, *Adv. Mater.*, 2013, **25**, 1052–1057.
- 102 C. Avci-Camur, J. Perez-Carvajal, I. Imaz and D. MasPOCH, *ACS Sustainable Chem. Eng.*, 2018, **6**, 14554–14560.
- 103 I. Imaz, M. Rubio-Martínez, J. An, I. Solé-Font, N. L. Rosi and D. MasPOCH, *Chem. Commun.*, 2011, **47**, 7287–7302.
- 104 S. L. Anderson and K. C. Stylianou, *Coord. Chem. Rev.*, 2017, **349**, 102–128.
- 105 E. Alvarez, N. Guillou, C. Martineau, B. Bueken, B. Van de Voorde, C. Le Guillouzer, P. Fabry, F. Nouar, F. Taulelle, D. de Vos, J.-S. Chang, K. H. Cho, N. Ramsahye, T. Devic, M. Daturi, G. Maurin and C. Serre, *Angew. Chem., Int. Ed.*, 2015, **54**, 3664–3668.
- 106 B. Yilmaz, N. Trukhan and U. Müller, *Chin. J. Catal.*, 2012, **33**, 3–10.
- 107 P. Horcajada, R. Gref, T. Baati, P. K. Allan, G. Maurin, P. Couvreur, G. Férey, R. E. Morris and C. Serre, *Chem. Rev.*, 2012, **112**, 1232–1268.
- 108 A. Zulys, F. Yulia, N. Muhadzib and N. Nasruddin, *Ind. Eng. Chem. Res.*, 2020, **60**, 37–51.
- 109 J. A. Coelho, A. M. Ribeiro, A. F. P. Ferreira, S. M. P. Lucena, A. E. Rodrigues and D. C. S. d. Azevedo, *Ind. Eng. Chem. Res.*, 2016, **55**, 2134–2143.
- 110 J. An, S. J. Geib and N. L. Rosi, *J. Am. Chem. Soc.*, 2010, **132**, 38–39.
- 111 M. Sánchez-Sánchez, N. Getachew, K. Díaz, M. Díaz-García, Y. Chebude and I. Díaz, *Green Chem.*, 2015, **17**, 1500–1509.
- 112 L. H. Wee, M. R. Lohe, N. Janssens, S. Kaskel and J. A. Martens, *J. Mater. Chem.*, 2012, **22**, 13742–13746.
- 113 J.-L. Zhuang, D. Ceglarek, S. Pethuraj and A. Terfort, *Adv. Funct. Mater.*, 2011, **21**, 1442–1447.
- 114 J. Shi, J. Zhang, D. Tan, X. Cheng, X. Tan, B. Zhang, B. Han, L. Liu, F. Zhang, M. Liu and J. Xiang, *ChemCatChem*, 2019, **11**, 2058–2062.
- 115 I. Pakamoré, J. Rousseau, C. Rousseau, E. Monflier and P. Á. Szilágyi, *Green Chem.*, 2018, **20**, 5292–5298.
- 116 R. D'Amato, A. Donnadio, M. Carta, C. Sangregorio, D. Tiana, R. Vivani, M. Taddei and F. Costantino, *ACS Sustainable Chem. Eng.*, 2019, **7**, 394–402.
- 117 E. S. Grape, J. G. Flores, T. Hidalgo, E. Martínez-Ahumada, A. Gutiérrez-Alejandre, A. Hautier, D. R. Williams, M. O'Keeffe, L. Öhrström, T. Willhammar, P. Horcajada, I. A. Ibarra and A. K. Inge, *J. Am. Chem. Soc.*, 2020, **142**, 16795–16804.
- 118 G. Das, D. B. Shinde, S. Kandambeth, B. P. Biswal and R. Banerjee, *Chem. Commun.*, 2014, **50**, 12615–12618.
- 119 D. N. Rainer and R. E. Morris, *Dalton Trans.*, 2021, **50**, 8995–9009.
- 120 A. M. Fidelli, B. Karadeniz, A. J. Howarth, I. Huskić, L. S. Germann, I. Halasz, M. Etter, S.-Y. Moon, R. E. Dinnebier, V. Stilinović, O. K. Farha, T. Friščić and K. Užarević, *Chem. Commun.*, 2018, **54**, 6999–7002.
- 121 S. Głowniak, B. Szcześniak, J. Choma and M. Jaroniec, *Mater. Today*, 2021, **46**, 109–124.
- 122 P. A. Julien, K. Užarević, A. D. Katsenis, S. A. J. Kimber, T. Wang, O. K. Farha, Y. Zhang, J. Casaban, L. S. Germann, M. Etter, R. E. Dinnebier, S. L. James, I. Halasz and T. Friščić, *J. Am. Chem. Soc.*, 2016, **138**, 2929–2932.
- 123 T. Gao, H.-J. Tang, S.-Y. Zhang, J.-W. Cao, Y.-N. Wu, J. Chen, Y. Wang and K.-J. Chen, *J. Solid State Chem.*, 2021, **303**, 122547.
- 124 Z. Wang, L. Liu, Z. Li, N. Goyal, T. Du, J. He and G. K. Li, *Energy Fuels*, 2022, **36**, 2927–2944.
- 125 J. Ren, N. M. Musyoka, H. W. Langmi, A. Swartbooi, B. C. North and M. Mathe, *Int. J. Hydrogen Energy*, 2015, **40**, 4617–4622.
- 126 Y. H. Lee, Y. Kwon, C. Kim, Y.-E. Hwang, M. Choi, Y. Park, A. Jamal and D.-Y. Koh, *JACS Au*, 2021, **1**, 1198–1207.
- 127 W. Quan, H. E. Holmes, F. Zhang, B. L. Hamlett, M. G. Finn, C. W. Abney, M. T. Kapelewski, S. C. Weston, R. P. Lively and W. J. Koros, *JACS Au*, 2022, **2**, 1350–1358.
- 128 S. Sultan, H. N. Abdelhamid, X. Zou and A. P. Mathew, *Adv. Funct. Mater.*, 2019, **29**, 1805372.
- 129 S. Zhou, X. Kong, B. Zheng, F. Huo, M. Strømme and C. Xu, *ACS Nano*, 2019, **13**, 9578–9586.
- 130 X.-H. Ma, Z. Yang, Z.-K. Yao, Z.-L. Xu and C. Y. Tang, *J. Membr. Sci.*, 2017, **525**, 269–276.
- 131 E. R. Kearns, R. Gillespie and D. M. D'Alessandro, *J. Mater. Chem. A*, 2021, **9**, 27252–27270.
- 132 S. Lawson, X. Li, H. Thakkar, A. A. Rownaghi and F. Rezaei, *Chem. Rev.*, 2021, **121**, 6246–6291.
- 133 C. A. Grande, R. Blom, V. Middelkoop, D. Matras, A. Vamvakeros, S. D. M. Jacques, A. M. Beale, M. Di



- Michiel, K. A. Andreassen and A. M. Bouzga, *Chem. Eng. J.*, 2020, **402**, 126166.
- 134 K. A. Evans, Z. C. Kennedy, B. W. Arey, J. F. Christ, H. T. Schaefer, S. K. Nune and R. L. Erikson, *ACS Appl. Mater. Interfaces*, 2018, **10**, 15112–15121.
- 135 H. Thakkar, S. Eastman, Q. Al-Naddaf, A. A. Rownaghi and F. Rezaei, *ACS Appl. Mater. Interfaces*, 2017, **9**, 35908–35916.
- 136 S. Lawson, C. Griffin, K. Rapp, A. A. Rownaghi and F. Rezaei, *Energy Fuels*, 2019, **33**, 2399–2407.

