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# REVIEW



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# Recent achievements toward the development of Ni-based layered oxide cathodes for fast-charging Li-ion batteries

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The driving mileage of electric vehicles (EVs) has been substantially improved in recent years with the adoption of Ni-based layered oxide materials as the battery cathode. The average charging period of EVs is still time-consuming, compared with the short refueling time of an internal combustion engine vehicle. With the guidance from the United States Department of Energy, the charging time of refilling 60% of the battery capacity should be less than 6 min for EVs, indicating that the corresponding charging rate for the cathode materials is to be greater than 6C. However, the sluggish kinetic conditions and insufficient thermal stability of the Ni-based layered oxide materials hinder further application in fast-charging operations. Most of the recent review articles regarding Ni-based layered oxide materials as cathodes for lithium-ion batteries (LIBs) only touch degradation mechanisms under slow charging conditions. Of note, the fading mechanisms of the cathode materials for fast-charging, of which the importance abruptly increases due to the development of electric vehicles, may be significantly different from those of slow charging conditions. There are a few review articles regarding fast-charging; however, their perspectives are limited mostly to battery thermal management simulations, lacking experimental validations such as microscale structure degradations of Ni-based layered oxide cathode materials. In this review, a general and fundamental definition of fast-charging is discussed at first, and then we summarize the rate capability required in EVs and the electrochemical and kinetic properties of Ni-based layered oxide cathode materials. Next, the degradation mechanisms of LIBs leveraging Ni-based cathodes under fast-charging operation are systematically discussed from the electrode scale to the particle scale and finally the atom scale (lattice oxygen-level investigation). Then, various strategies to achieve higher rate capability, such as optimizing the synthesis process of cathode particles, fabricating single-crystalline particles, employing electrolyte additives, doping foreign ions, coating protective layers, and engineering the cathode architecture, are detailed. All these strategies need to be considered to enhance the electrochemical performance of Ni-based oxide cathode materials under fast-charging conditions.

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# 1. Introduction

Innovation in transportation is key to stopping climate change. The Energy Information Administration reported that 37% of the total energy produced in the United States in 2021 was from petroleum sources.<sup>1</sup> The vast majority (91%) of these sources were consumed in the transportation end-user sector such as cars, trucks, and aircraft. Indeed, the transportation sector is the largest contributor (28%) to U.S. greenhouse gas emissions, according to the Environmental Protection Agency, in 2021.<sup>2</sup> Reducing the petroleum reliance in transportation will, therefore, make a direct impact on immediately reducing the entire carbon footprint, shaping a sustainable future.

In light of this clear motivation, the transportation sector is getting electrified. For example, General Motors and Ford, the two largest car makers in the U.S., plan to end their production of engine-powered cars and trucks and will offer electric vehicles (EVs) exclusively by 2035.<sup>3,4</sup> The key enabler of EVs is, undoubtedly, rechargeable batteries. Among the many types of battery, Li-ion batteries (LIBs) that adopt a lithium transition metal oxide (LiTMO<sub>2</sub>) cathode govern the modern EV battery technology with their high energy density and good round-trip efficiency.<sup>5–7</sup>

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As the market for EVs is expanding rapidly, the need for high-performance batteries continues to grow.<sup>8,9</sup>

To enable a high energy density, research efforts to design transition metal compositions for layered oxide cathodes have centered on implementing nickel as a redox center. Ni-based layered oxides such as  $\text{LiNi}_x \text{Co}_y \text{Mn}_z \text{O}_2$  (NCM, x + y + z = 1) and  $LiNi_xCo_yAl_zO_2$  (NCA, x + y + z = 1) have been gradually replacing LiCoO<sub>2</sub>.<sup>10-12</sup> In order to meet the energy density goal of 350 Wh kg<sup>-1</sup> and 750 Wh L<sup>-1</sup> at the cell level, as established by the U.S. Department of Energy (DOE) and U.S. Advanced Battery Consortium (USABC),<sup>13</sup> Ni-based layered oxides with a high Ni content (Ni  $\geq$  80%) are considered suitable for EV applications due to their high specific capacities (200–250 mA h  $g^{-1}$ ), high working voltages (3.6-3.8 V) and relatively low cost.<sup>14,15</sup> The driving mileage of EVs that adopt these Ni-rich cathodes has been substantially improved, comparable with that of internal combustion engine vehicles (ICEVs), thus increasing the market penetration of EVs.16

However, the average charging time required for EVs (recharging 80% of their capacity in about 1 h) is incomparable with the short refueling time of ICEVs (refueling can be done in less than 10 min).<sup>17-19</sup> According to the recent guidelines of fast-charging released by the U.S. DOE, the ultimate goal for fast-charging is to refill 60% of the full battery capacity in 6 min or less, which indicates that the charging rate for the battery is greater than 6C.<sup>20</sup> In general, the fast-charging capability of LIBs is limited by the mechanical and electrochemical properties of Ni-based layered oxide cathodes.<sup>21,22</sup> First, the structural instability of highly oxidized Ni-based layered oxides makes it hard for them to accommodate rapid volume changes upon fast-charging.23,24 This results in micro-cracks generated throughout the secondary particles of Ni-based cathodes, creating excess surface areas that facilitate the formation of cathode-electrolyte interphases.<sup>25,26</sup> Second, poor charge transport kinetics of Ni-based layered oxides can induce high cell impedance.<sup>27,28</sup> Especially for the polycrystalline Ni-based

layered oxide materials, low electronic and ionic conductivities of the primary crystals can limit the rate capability.<sup>29,30</sup> Consequently, the Ni-based layered oxide cathodes can degrade faster at fast charging (*e.g.*, 6C) than at slow charging (*e.g.*, 0.5C).<sup>31–33</sup>

In this review, we highlight recent progress in the development of Ni-based LiTMO<sub>2</sub> for EV applications. We will first define fast-charging, discuss the rate capability practically required in EVs, and summarize the electrochemical and kinetic properties of Ni-based layered oxide cathode materials. Then, a comprehensive discussion of the major degradation mechanisms of Ni-based layered oxide cathode materials under fast charging conditions will follow. The state-of-the-art approaches to address cathode degradation and enhance the rate performance of Ni-based layered oxides will be discussed. Finally, an outlook on future directions of designing Ni-based layered cathodes for LIBs for superior fast-charging performance will be presented.

# 2. Features of fast-charging and Ni-based oxide cathode materials

#### 2.1 The definition of fast-charging

Technologies to enable fast-charge or extreme fast-charge (XFC) to recharge batteries in a few minutes have drawn extensive research interest and exhibited a great market prospect. A comparison, available in the literature,<sup>34</sup> of 525-mile intercity travel from Salt Lake City, Utah to Denver, Colorado between internal combustion engine vehicles (ICEVs) and battery electric vehicles (BEVs) is illustrated in Fig. 1A. As the baseline for the comparison, it takes 8 h and 23 min for the ICEV to travel 525 miles with one 15 min refueling stop. For the 300-milerange BEV, it takes 9 h and 16 min to cover the same distance, 53 min more than ICEV, due to a recharging time of 68 min at

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Fig. 1 (A) Comparison of intercity travel from Salt Lake City to Denver.<sup>34</sup> (B) Charging profiles for BEVs with direct-current fast-charging capabilities on the market.<sup>35</sup>

120 kW. In the case of a 200-mile-range BEV equipped with 400 kW XFC, the total travel time is 8 h and 46 min, 30 min less than the 300-mile BEV, even with two recharge stops. Furthermore, the estimated travel time of the 300-mile BEV with 400 kW XFC is 8 h and 31 min, which is only 1.6% longer than the ICEV, with one recharge stop of 23 min at 275 miles. To finish the recharge in 23 min to go for another 250 miles, the charging speed should be faster than 11 miles  $\min^{-1}$ . Of note, the fast-charging goal set by the U.S. DOE is 20 miles min<sup>-1</sup> on average or recharging up to 80% of the battery capacity in 10 min or less.<sup>35</sup> However, the charging speed of most BEVs (e.g., Mitsubishi i-MiEV,<sup>36</sup> Nissan Leaf,<sup>37</sup> BMW i3,<sup>38</sup> and Chevy Bolt<sup>39</sup>) remains below 3 miles min<sup>-1</sup> as indicated in Fig. 1B. The Tesla Model S has achieved a charging rate of 5.6 miles min<sup>-1</sup> with its state-of-the-art 120 kW supercharger,<sup>40,41</sup> but this charging speed still does not meet the fast-charging requirement proposed by the DOE. Thus far, the typical charging rate demonstrated practically is in the ballpark of 2C or even less,<sup>34,42</sup> corresponding to a charge with a power of 50 kW for most BEVs as well as 120 kW for Tesla.<sup>20,42</sup> The higher charging speed may require the voltage rating of BEV to increase from 400 to 1000 V, which augments the charging rate from 1.5 to 6C for the battery as exhibited in Table 1.<sup>20</sup>

# 2.2 The electrochemical properties of Ni-based layered oxide materials

LiNiO<sub>2</sub> that is isostructural to LiCoO<sub>2</sub> was investigated as a cathode material to increase the energy density and lower the materials cost for LIBs.<sup>26,43-46</sup> However, LiNiO<sub>2</sub> destabilizes upon delithiation at high voltage, undergoing irreversible phase transformations involving O<sub>2</sub> release.<sup>46-48</sup> To improve the electrochemical stability of LiNiO<sub>2</sub>, Co,<sup>48-50</sup> Mn,<sup>44,51,52</sup> and Al<sup>53-55</sup> were incorporated into LiNiO<sub>2</sub>.<sup>56</sup> It has now been proved that Al<sup>3+</sup> and Mn<sup>4+</sup> do not participate in the redox process but enhance the structural and thermal stability of the



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Table 1 Timeline of BEVs and battery C-rate to support XFC<sup>34</sup>

Charger voltage	400–1000 V		
Charge inlet Vehicle	XFC designed inle 600 V, 400 A, 240 kW 1000 V, 210–280 A	t (s) for 1000 V @40 800 V, 400 A, 320 kW , 210–280 kW	00 A 1000 V, 400 A, 400 kW
Battery	2.0-3.3C	3.3-4.6C	4.6-6C

material.  $\text{Co}^{3+}$  contributes to stabilizing the layered structure, improving the rate capability of the layered oxide cathodes.<sup>57–59</sup> Many studies confirmed that the high Ni content will lead to a concomitant increase in specific capacity at the expense of the electrode stability, as indicated in Fig. 2A.<sup>59–62</sup> This trade-off effect of a high Ni content in the cathode can be understood by the increased high-valence state Ni ions (*i.e.*, Ni<sup>4+</sup>) that readily react with the electrolyte, although these high valence state Ni ions benefit the specific capacity of the cathode.<sup>63–66</sup>

According to the *in situ* XRD measurements (not shown) of various Ni-based oxide cathode materials, the c-lattice parameter varies substantially with the cathode composition.<sup>67</sup> The c-lattice parameter in Fig. 2B decreases substantially at high voltage as the Ni fraction increased. The variation of the lattice parameters results from complex phase transitions upon Li extraction.<sup>67</sup> As shown by the dQ/dV-plots in Fig. 2C–F, NCM materials undergo different phase transitions with the different Ni contents. Upon charge, all NCMs phase-transform

from hexagonal (H1) to monoclinic (M) at ~3.7 V. Ni-rich (80% or more) NCM cathodes show an additional phase transition, M-to-hexagonal (H2) at ~4.0 V and H2-to-hexagonal (H3) phases at ~4.2 V. The H2-to-H3 transition is more evident in lower voltage LiNi<sub>0.95</sub>Co<sub>0.025</sub>Mn<sub>0.025</sub>O<sub>2</sub> at than in LiNi<sub>0.8</sub>Co<sub>0.1</sub>Mn<sub>0.1</sub>O<sub>2</sub>. No significant H2-H3 transition is observed in LiNi<sub>0.6</sub>Co<sub>0.2</sub>Mn<sub>0.2</sub>O<sub>2</sub> as it requires 4.6 V, which is higher than the charge cutoff voltage. Here, it should be emphasized that the shrinkage of the c-lattice parameter mainly originates from the H2-H3 transition at high voltage, leading to a substantial volume change of NCM.<sup>68,69</sup> Thus, the electrochemical and mechanical stability of Ni-rich NCM largely depend on their high-voltage characteristics.<sup>70–72</sup>

#### 2.3 The kinetic properties of Ni-based layered oxide materials

It is critical to understand the kinetic properties of Ni-based layered oxide materials to further optimize the rate performance. Typically, the transport of Li ions within Ni-based layered oxide materials can be categorized into several mechanisms: diffusion in bulk, diffusion across the grain boundaries, and diffusion along the grain planes. For bulk diffusion, there are two Li-ion transport pathways in the layered structure.<sup>73,74</sup> Li-ions diffuse from one octahedral site to the next octahedral site through an intermediate-tetrahedral site (Fig. 3A) or an oxygen dumbbell (Fig. 3B). Specifically, Li-ions prefer to migrate through oxygen dumbbell hopping at the start of the charging process. Then, when more than 1/3 Li-ions are extracted from the cathode, the tetrahedral site hopping plays a dominant role in Li-ion transport.<sup>74</sup> The acti-



**Fig. 2** (A) A map summarizing the trend relationship between the discharge capacity, thermal stability, and capacity retention of NCM materials.<sup>60</sup> (B) *c*-Axis lattice parameter of Ni-rich materials as a function of the cell voltage where *x* indicates the content of Ni in NCM.<sup>67</sup> (C)–(F) dQ/dV profiles of Ni-rich materials as a function of the cult voltage materials from 0.6 to 0.95.<sup>67</sup>



**Fig. 3** (A) Tetrahedral site pathway and (B) oxygen dumbbell pathway for Li-ion diffusion in an NCM layered structure.<sup>74</sup> (C) The lattice of the NCM layered structure.<sup>74</sup> (D) Li diffusion across (top) and within (bottom) the grain boundary.<sup>76</sup> (e) First cycle GITT curves (bottom) and the corresponding  $D_{Li+}$  (top) of NCM811.<sup>79</sup>

vation energy for Li-ion transport in the suggested pathways is closely associated with the volume of Li-O tetrahedra and octahedra in the Li slab.<sup>73,75</sup> In Fig. 3C, the stacking of Li slabs and transition metal (TM) slabs alternates, and  $d_1$  and  $d_2$  are the height of the TM and Li slabs, respectively. As a result, the high Li-ion diffusivity with low activation energy along the diffusion pathway is mainly ascribed to the large Li slab spacing. Meanwhile, Li ions can also diffuse along the grain boundaries and across the grain boundaries, as indicated in Fig. 3D.<sup>76</sup> Grain boundaries with specific orientations such as  $\Sigma_2(1\overline{1}0\overline{4})$  and  $\Sigma_3(\overline{1}10\overline{2})$  and hence different microstructures regulate Li migration in different ways, which leads to different Li diffusivities and overall conductivities.<sup>76</sup> In most cases, Liion diffusion at grain boundaries can be facilitated, compared with that in the bulk, due to the kinetically favorable Li-ion transport environment (i.e., reduced energy barrier for migration) at/near the grain boundaries.<sup>77,78</sup> The thicker disordered phases at/near the grain boundaries of NCM cathode materials generally lead to 10 times higher Li transport than that in bulk.76

In addition, the state of charge (SoC) of the NCM cathode dictates the Li kinetics.<sup>79</sup> As shown in Fig. 3E, at the beginning of the charging process, D<sub>Li+</sub> gradually increases from 10<sup>-9</sup>  $\text{cm}^2 \text{ s}^{-1}$  and reaches  $10^{-8} \text{ cm}^2 \text{ s}^{-1}$  at Li = 0.5 upon delithiation. Then it decreases back to  $10^{-9}$  cm<sup>2</sup> s<sup>-1</sup> at Li = 0.1. At the end of charging,  $D_{Li^+}$  rapidly decreases to  $10^{-10}$  cm<sup>2</sup> s<sup>-1</sup>. The sluggish Li kinetics at the initial SoC of the cathode are mainly ascribed to the H1 phase that contains the high Li content with a few Li vacancies and a narrow Li slab spacing.<sup>29,79</sup> As charging continues, new Li vacancies are generated and the spacing of Li slabs gradually increases, facilitating Li diffusion. At the end of the charge, the transition metal ions are in the highest valence state, which attracts the electron cloud from oxygen and decreases the O-O repulsion, leading to the sharp volume decrease of the unit cell.<sup>79</sup> Thus, D<sub>Li+</sub> is generally small at the start of the charging process and at the overcharging state. Upon discharge, D<sub>Li+</sub> remains high at 10<sup>-7</sup>- $10^{-8}$  cm<sup>2</sup> s<sup>-1</sup> and decreases to  $10^{-11}$  cm<sup>2</sup> s<sup>-1</sup> at the end of discharge. High  $D_{Li^+}$  in the beginning of discharge reflects the high concentration gradient across the cathode–electrolyte interface, and sluggish diffusion at the end of discharge results from Li redistribution in the bulk, as is commonly observed in Li cathodes.<sup>80,81</sup>

# 3. Degradation mechanism under fast-charging

Charging batteries in a shorter time inevitably requires a higher current density than the usual. Therefore, batteries at fast charging experience severe operation conditions (e.g., high voltage) that likely accelerate their degradation. Indeed, the degradation of Ni-based layered oxide cathodes is a critical obstacle to achieving capacity retention (*i.e.*, a long cycle life) for practical applications.<sup>82–84</sup> This section discusses the effect of Li transport kinetics and surface oxygen loss to further understand the degradation mechanisms of Ni-based layered oxide cathodes.<sup>85–87</sup>

#### 3.1 Kinetic inhomogeneity at the multi-scale

3.1.1 Inhomogeneity at the electrode-scale. The LIB cathode consists of multiple components (i.e., the active material, carbon additive, and binder), which need to be coherently interoperated to obtain a desirable battery performance. The complex, heterogeneous construction makes it difficult to probe the cathode at various length- and timescales. The morphological and chemical characteristics of Nibased layered oxide at the microscale have been recently investigated by Yang et al.<sup>88</sup> They leveraged hard X-ray phase-contrast nanotomography to compare the morphology of more than 600 particles in the NCM622 cathode cycled 10 times with a 5C rate (Fig. 4A). Three representative particles were marked in red, green, and blue, as shown in Fig. 5A, corresponding to the severely damaged, mildly damaged, and least damaged particles, respectively. The coexistence of differently damaged particles implies the inhomogeneity of the NCM



Fig. 4 (A) 3D rendering of micro-/nanotomography images of an NCM electrode and three of the representative particles highlighted in red, green, and blue, indicating severely damaged, mildly damaged, and least damaged particles, respectively.<sup>88</sup> (B) Slices over two particle layers that are close to the ten-cycled electrode's top surface (near the separator) and the aluminum current collector (bottom of the ten-cycled electrode).<sup>88</sup> (C) The spatial distributions of the severely damaged particles measured by nano-holotomography in the 10-cycled and 50-cycled electrodes and the probability distributions of the distance between two neighboring severely damaged particles in 10-cycled and 50-cycled electrodes.<sup>89</sup> (D) Particles sampled from fast-cycled (2C) cells and slow-cycled (0.05C) cells during the second cycle at states-of-charge corresponding to average lithium fractions of 0.75  $\rightarrow$  0.50  $\rightarrow$  0.75  $\rightarrow$  1.00, respectively.<sup>91</sup>

particles in the cathode. They also observed that particles located near the separator are likely damaged more than those near the current collector, as illustrated in Fig. 4B. Li et al.<sup>89</sup> monitored the dynamic evolution of the chemical and morphological characteristics of thousands of individual NCM particles through nano-holotomography at different cycles and SoCs. In Fig. 4C, the concentration of severely damaged particles (marked in red) increases, while the distance of neighboring particles considerably reduces, as the cycle number increases from 10 to 50. These variations likely disintegrate and distort the cathode due to different mechanical properties of the NCM particles (elastic modulus,  $E \sim 140$  GPa) and carbon/binder domains ( $E \sim 2$  GPa).<sup>89</sup> The transport of Li ions and electrons across the interfaces between active materials and carbon/binder domains can be disrupted, leading to a poor percolation of charge transport pathways. Mistry et al.90 indicated the inhomogeneous electrode would lead to the preferential intercalation of the NCM materials, which may induce detrimental heat spots in the electrode and trigger safety issues (e.g., fire or explosion) of the entire cell.<sup>90</sup>

To explain the electrode-level inhomogeneity and understand the degradation mechanisms at fast charge, Park *et al.*<sup>91</sup> exploited autocatalytic and autoinhibitory reactions to prepare NCM particles with compositional variation. Fig. 4D shows the Li content in the cathode particles at two different rates. They found that many particles remain lithiated at 2C unlike the uniform delithiation observed at 0.05C. This rate-dependency is attributed to the different Li kinetics of NCM particles with different compositions, leading to inhomogeneous Li distribution (*i.e.*, non-uniform SoC) over multiple particles upon galvanostatic delithiation.<sup>91</sup> This can polarize the cell substantially, limiting capacity at the charge cutoff voltage.

**3.1.2** Inhomogeneity at the particle-scale. In addition to the inhomogeneity at the electrode level, the heterogeneous distribution of the charge in a single particle is observed. Chueh and coworkers used a synchrotron-based transmission X-ray study to show depth-dependent SoC in a secondary particle and found more than 10% variation of Li concentration within the particle.<sup>92</sup> This uneven charging of the secondary particles is attributed to the anisotropic volume change of the primary particles.<sup>92</sup> In addition, the higher current density may intensify the local inhomogeneity of the charge distribution and chemical environment (*e.g.*, valence state of transition metals) over a secondary particle. Xu *et al.*<sup>93</sup> suggested two types of inhomogeneous charge distribution (rod- and gravel-like) in NCM particles, as shown in Fig. 5A. Both rod-



**Fig. 5** (A) 3D Ni valence state distribution, and 2D nanodomain valence gradient of the rod-NCM and gravel-NCM.<sup>93</sup> (B) The 3D rendering of the particles in the pristine state and after 50 cycles at 1C, 2C, 5C, and 10C, and the corresponding 2D virtual slices of the particles.<sup>94</sup> (C) The 3D rendering of the segmentation results and the calculated charge distribution over two regions of interest with the carbon/binder domain set to be transparent for better visualization of the NCM particle (orange) and the voids (gray-blue).<sup>98</sup>

(upper) and gravel-like (bottom) NCM particles show more oxidation in the surface regions than in the bulk due to the greater electrical and ionic conductivity near the surface. In addition, they specified the spatial distribution of Ni oxidation states in both NCM particle models. Rod-type NCMs show a linear, parallel configuration of oxidized Ni, while the gravel-NCM has a random arrangement of the 2D nanodomain valence gradient vectors. The two distinct Ni valence patterns correspond to different crystalline orientations that determine the redox reaction pathway and charge distribution in the secondary particles. This inhomogeneity at the particle-scale illustrates that the overcharged regions in a particle are susceptible to oxygen loss at a high SoC, which will be discussed later.

In order to further investigate the morphology degradation of the particle in terms of the charging rate, Xia *et al.*<sup>94</sup> depicted the 3D rendering of the particle through X-ray micro/ nano tomography before and after 50 cycles under different rates (0.5C, 1C, 2C, 5C, 10C) In Fig. 5B, 2D virtual cross-sectional images at the center of the particles exhibit morphological defects. The degree of particle fracture observed is proportional to the charge rate. The particle fracture can be ascribed to the oxygen loss from the unstable H3 phase due to the partially overcharged region in the cathode. The detailed mechanism of the oxygen loss and related capacity degradation will be provided in the following section (3.2). They also demonstrated that large secondary particles have a low tolerance against crack formation compared with small particles at fast charge. Generally, crack formation in cathode particles at a high rate is irreversible, one of the main reasons for the capacity fade upon fast-charging.95-97 Through the combination of machine learning and high-resolution hard X-ray nano-tomography, Jiang et al.98 found the detachment of inactive materials (i.e., additives and binders) from NCM particles in the cathode. Fig. 5C shows a reconstructed 3D model that reflects the physical contact between the NCM particle and the surrounding conductive carbon/binder domain matrix. The detachment of the active NCM particles from the carbon/ binder domain was observed at the particle level and marked as blue void regions. This results from uneven anisotropic volume change during repetitive charge-discharge processes,

increasing the impedance across the surface of the NCM particle.

#### 3.2 Oxygen loss due to thermodynamic instability

It has been broadly reported that high-voltage instability of the H3 phase of NCM leads to unfavorable oxygen evolution, transition metal dissolution, and irreversible phase transformation (Fig. 6A).<sup>68,99,100</sup> Charge compensation at high voltage can involve not only transition metal redox but also oxygen redox. Especially, oxidized Ni<sup>4+</sup> in NiO<sub>6</sub> octahedra can be stabilized by oxidizing the surrounding oxygen. The increasing amount of overcharged (at high SoC) particles/regions with a high current density, as discussed above, is thus susceptible to severe oxygen loss in the Ni-based layered oxide cathode. At a high voltage, lattice oxygen may have several forms, including singlet oxygen ( $^{1}O_{2}$ ), peroxide anodic radicals (O–O<sup>–</sup>), and peroxide bonds between transition metal ions (MO–OM), all of which have been experimentally identified.<sup>68,101–103</sup>

3.2.1 Phase transformation. Phase transitions from layered to spinel and from spinel to rock-salt result from oxygen loss. These transitions become evident with large delithiation.<sup>86</sup> Oxygen evolved from the cathode oxidizes the electrolyte, forming carbon dioxide  $(CO_2)$  exothermically, the sign of battery failure. The effect of the gas generation on battery degradation will be detailed in section 3.2.2. Layered NiO<sub>2</sub> (Fig. 7A) in a highly delithiated state can transform to spinel Ni<sub>3</sub>O<sub>4</sub> (Fig. 7B) by oxygen loss with the formation of active oxygen intermediates.14 Oxygen vacancies formed by oxygen loss at high voltage could significantly decrease a Ni migration barrier to the adjacent, empty Li site (Fig. 7C), leading to the Ni/Li intermixing. This eventually alters the surface structure from the layered to the rock-salt (Fig. 7D).<sup>14</sup> Without exception, the oxygen evolution reaction for all Ni-based layered oxides in the cathode was observed near the start of the H2 to the H3 phase transition, corresponding to the 75-80% SoC. Generally,

these active oxygen intermediates can react with the electrolyte immediately. Also, the spinel and rock-salt phases are localized near the surface of NCM particles.<sup>104</sup> According to electron energy-loss spectroscopy (EELS), Yang et al.<sup>86</sup> argued that the phase transition can further extend to the bulk of NCM particles as a result of fast charging. In Fig. 7E, the O pre-Kedge peak completely disappears deep in the particle (~30 nm) at a high rate. High-resolution transition electron microscopy (HRTEM) in Fig. 7F shows that the thick rock-salt phase was indeed formed on NCM811. This irreversible phase transformation can generate local stress along the grain boundaries.<sup>105,106</sup> The stress concentration may accelerate the particle fracture. As the fractured surface of the NCM particle allows the liquid electrolyte to infiltrate, the newly exposed surface also undergoes the irreversible phase transitions associated with oxygen loss.<sup>107</sup>

3.2.2 Gas release. Gas release from the Ni-based layered oxide cathode is one of the main causes of the thermal runaway of Li-ion batteries. The generated gaseous products mainly consist of O2, CO2, and CO (Fig. 8A).<sup>108</sup> As discussed previously, fast-charging can develop non-uniformity of Li composition from a particle to an electrode level, leading to local overcharge in the NCM cathode. According to electrochemical reaction pathways described in Fig. 8A, oxidized ethylene carbonate (EC) can generate CO (and/or  $CO_2$ ) and protic species (R-H<sup>+</sup>) and release electrons at the particle/electrolyte interface upon charging when the cell voltage is higher than the electrochemical stability limit of the carbonate electrolyte. To maintain the charge balance, Li ions should be extracted from the electrolyte and reinserted into the Ni-based layered oxide cathode. This self-discharge reaction at overcharge substantially decreases the coulombic efficiency of the cell.

The gas generation from the charged NCM811 cathode was investigated by differential electrochemical mass spectrometry (DEMS).<sup>108</sup> In Fig. 8B, O<sub>2</sub>, CO<sub>2</sub> and CO generation corresponds



Fig. 6 (A) Schematic illustration of oxygen loss and the related effects.<sup>68</sup> (B) Correlation of oxygen evolution with SoC, potential, and differential capacity of an NCM811 cathode, where the pink shaded area corresponds to the H2  $\rightarrow$  H3 transition region while the green area indicates the stable region of H1 and H2.<sup>68</sup>



**Fig. 7** Schematic diagram of ordered and disordered phases and their structural transitions in layered lithium metal oxides. (A) Ordered  $R\bar{3}m$  structure, (B) Cation mixing phase with an  $Fm\bar{3}m$  structure, (C)  $R\bar{3}m$  structure with Li vacancies at a highly charged state and (D) cation mixed phase with partial TM ions in the Li layer.<sup>14</sup> (E) EELS and (F) HR-TEM images of high-loading NCM811.<sup>86</sup>



**Fig. 8** (A) Schematic description of the proposed electrochemical and chemical electrolyte oxidation pathways occurring at high potentials.<sup>108</sup> (B) Cell voltage vs. time of an NCM811-graphite cell over four charge/discharge cycles and evolution of CO<sub>2</sub> (dark blue), H<sub>2</sub> (green), C<sub>2</sub>H<sub>4</sub> (orange), CO (blue), and O<sub>2</sub> (black, 10-fold magnified) as a function of time.<sup>108</sup>

to the H2–H3 phase transition at  $\sim$ 4.2 V. Oxygen oxidation in the overcharged H3 phase at high rates leads to oxygen gas evolution that reduces the Ni,<sup>109</sup> forming the surficial spinel

and/or rock-salt structures as discussed in section (3.2.1). Note that overcharged regions in the cathode may release oxygen and generate CO,  $CO_2$  and  $H_2O$  chemically (Fig. 8A).  $H_2O$  gen-

erated can accelerate transition metal dissolution from Nibased layered oxide (see section 3.2.3).

**3.2.3 Transition metal dissolution.** Owing to the harsh conditions of fast-charging, the erratic H3 phase can be accumulated in the cathode. It is generally accepted that the transition metal ion dissolution accelerated by  $H_2O$  generation would result in capacity fading due to the deficiency of the Liion insertion sites in Ni-based particles.<sup>110–112</sup>  $H_2O$  can hydrolyze LiPF<sub>6</sub> in the electrolyte, producing HF that continuously corrodes the cathode surface. This corrosion process is described as follows:

$$LiPF_6 + H_2O = LiF + HF + POF_3$$
(1)

$$HF + Li_{x}TMO_{2} = MF + H_{2}O + LiF$$
(2)

The dissolved transition metal ions further migrate to the anode and will be reduced to metal that makes the solid–electrolyte interphase (SEI) electronically conductive. As the anode cannot be fully passivated, SEIs will continuously grow upon repetitive charge and discharge, depleting Li ions and electrolyte molecules. The cathode/anode ratio would be altered by the Li ion loss due to the extracted Li ions being unable to intercalate into the anode, which leads to the overcharging of the cathode. Ko *et al.*<sup>113</sup> investigated how the transition metal composition of a Ni-rich NCM particle changes over cycles through EDS mapping. Compared with pristine NCM in Fig. 10A, the transition metal content decreases as the cycle number increases in Fig. 9B and C. Notably, the transition metal ions are more likely released from the cracked NCM particle. Using *operando* X-ray absorption spectroscopy, Jung

*et al.*<sup>114</sup> found increasing transition metal concentration at the graphite anode when the NCM622 cathode was cycled at >4.5 V, associated with the H3 phase formation, in Fig. 9D.

# 4. Approaches to improve the electrochemical properties of Ni-based layered oxide cathodes

Approaches to improving the electro-chemo-mechanical stability of Ni-based layered oxide cathodes at fast charge has centered on various strategies to engineer charge carrier kinetics and suppress oxygen evolution.<sup>32,115,116</sup> In this section, we will highlight effective approaches that can successfully modify the physical, chemical, and morphological properties of NCM cathodes to achieve high reversible capacity.

#### 4.1 Single crystalline particles

Ni-rich NCM undergoes anisotropic volume change at fast charge, leading to particle microcracking that creates a surface newly exposed to the electrolyte, susceptible for gas evolution reactions.<sup>117,118</sup> Thus, controlling the orientation of NCM crystallites can be an effective means to preserve particle integrity.<sup>119</sup> Single crystalline NCM cathodes have demonstrated improved mechanical and thermal stability.<sup>120–122</sup> Ma and coworkers engineered the particle size of single-crystal NCM622 to find a balance between rate capability and cycling stability.<sup>123</sup> They synthesized single-crystal NCM622, in which the average particle size is 2  $\mu$ m, by a molten-salt method, as



Fig. 9 Elemental mapping of Ni, Co, and Mn in LiNi<sub>0.87</sub>Co<sub>0.09</sub>Mn<sub>0.04</sub>O<sub>2</sub> particles in the (A) pristine, (B) cracked, and (C) broken states.<sup>113</sup> (D) Effect of operating voltage on transition metal ion dissolution observed from the graphite side of an NCM622 Li-ion cell.<sup>114</sup>



**Fig. 10** (A) Band-contrast EBSD maps and (B) the EBSD orientation (Euler angles) map of the single-crystal NCM622 particles.<sup>123</sup> (C) SEM image of the as-synthesized single crystal NCM622 particles.<sup>123</sup> (D) STEM-EDS mapping of the particle showing the homogeneous distribution of Ni, Co, and Mn elements.<sup>123</sup> (E) Rate capability of single crystal and polycrystalline NCM622 particles in terms of cycle number.<sup>123</sup> SEM images of (F) octahedron-shaped and (G) polyhedron-shaped single-crystalline NCM811 particles, with their schematic inserted.<sup>124</sup> (H) The rate capability of polycrystalline NCM811.<sup>124</sup> (I) High-resolution TEM microstructure of polyhedron-shaped NCM811 after 100 cycles at a 6C rate.<sup>124</sup>

observed by SEM in Fig. 10A. Electron backscatter diffraction (EBSD) of the cross-sectional particles in Fig. 10B indicates that as-synthesized NCM622 indeed consisted of single crystalline particles. EDS mapping in Fig. 10C and D shows a uniform distribution of Ni, Co, and Mn in each particle. In Fig. 10E, the single crystal NCM622 cathode exhibited an excellent rate capability of 135 mA h g<sup>-1</sup> at 20C, which is 75% of the capacity at 0.1C, while the polycrystalline NCM622 retained 80 mA h g<sup>-1</sup>, 45% of its 0.1C capacity.

However, the morphology of single crystalline NCM particles is often non-spherical, making the cathode calendering process difficult. Additionally, the single crystals tend to have a larger particle size that leads to a longer solid-state Li diffusion length, and thus sluggish Li kinetics, than the polycrystalline. Note that achieving a high Ni fraction in an NCM single crystal is non-trivial due to the low synthesis temperature insufficient to promote grain growth.<sup>122</sup> By hydrothermal synthesis, Lu et al.<sup>124</sup> obtained octahedron- (Fig. 10F) and polyhedron-shaped (Fig. 10G) single-crystal NCM811 particles with a controlled surface orientation and 1 µm average particle size. Fig. 10H demonstrates the rate capability of the two NCM811 particles and compares them with polycrystalline NCM811. While all NCM811 cathodes deliver a similar charge capacity of  $\sim$ 200 mA h g<sup>-1</sup> at 0.1C, the two single-crystal NCM811 outperforms the polycrystalline at higher rates. Fig. 10I shows that polyhedron-shaped NCM811 does not develop surface phases after 100 cycles at 6C. The authors claimed that enhanced stability of single-crystal NCM811 at fast charge is attributable to a good rate capability. Density-functional theory (DFT) calculations further confirmed that (104) facet-dominated polyhedron-shaped particles have higher thermodynamic stability. The surface reconstruction is more likely to occur on the (012) facets compared with (104) facets, which is associated with the segregation of Ni to the surface and driven by the local distortions due to the mixing of Li ions and Ni ions.<sup>125</sup>

#### 4.2 Engineering particle morphology

The surface structure of Ni-based layered oxide particles is one of the factors that dictate Li kinetics.<sup>126,127</sup> Co-precipitation and subsequent solid-state reactions is the most common method to prepare Ni-based layered oxides,<sup>128</sup> and a wide variety of high-Ni NCM can be obtained by precisely controlprecursor ling the stoichiometry and calcination parameters.<sup>129,130</sup> To achieve novel particle morphology that is unachievable by the conventional co-precipitation method for the Ni-based cathode, Su et al.<sup>131</sup> employed two-step (secondary) co-precipitation that can lead to core-shell-structured NCM, as illustrated in Fig. 11A. The core structure was generated through the first co-precipitation process and then the active-plane-exposing shell was anchored on the core via the



**Fig. 11** (A) Schematic of the secondary coprecipitation method.<sup>131</sup> (B) Cross-sectional SEM images of synthesized cathode materials along with schematic diagrams.<sup>131</sup> (C) Schematic of assembling process of NCM622 cathode materials.<sup>132</sup> (D) TEM image of one-dimensional NCM622 cathode material. Inset is an associated SEM image.<sup>132</sup> (E) HRTEM image corresponding to the region taken from the pink dashed box in (D).<sup>132</sup> (F) Schematic of plasma treatment.<sup>133</sup> (G) The rate capability of plasma-treated NCA and bare NCA cathode.<sup>133</sup>

secondary co-precipitation. A cross-sectional SEM image and the associated schematic in Fig. 11B show this core–shell structure. The core particle was constructed by tetragonal-shape primary particles that can have high packing density while the surface shell exhibited a loose sheet-like structure with increasing surface area that can promote Li transport and charge transfer. This core–shell NCM811 delivered 160 mA h  $g^{-1}$  at 10C.

Lai *et al.*<sup>132</sup> employed ultrasound-triggered cation chelation to produce one-dimensional (1D) NCM622. For an aqueous solution of precursors, the ultrasonic wave can remove the hydration layer surrounding the metal ions (step 1 in Fig. 11C) and promote a chelating reaction between the metal ions and ethylenediaminetetraacetic acid (EDTA) (step 2). The chelates can be separated from the EDTA (step 3) and assembled into fiber-like structures by continuous ultrasonication (step 4), as shown in Fig. 11D. The obtained 1D NCM622 has the layered structure, as identified by HRTEM in Fig. 11E. This fibrous morphology provides continuous, percolated Li diffusion pathways on a nanometer scale, leading to high rate capability (122 mA h g<sup>-1</sup> at 10C).

Yuan et al.<sup>133</sup> reported improved cyclability for the LiNi<sub>0.8</sub>Co<sub>0.15</sub>Al<sub>0.05</sub>O<sub>2</sub> (NCA) cathode by tailoring the surface characteristics of the precursors using nondestructive plasma treatment with a mixed gas of O<sub>2</sub>/Ar, as illustrated in Fig. 11F. The plasma treatment can expose pre-existing microvoids between primary precursor particles and form a layer of NiOOH with an ordered crystal structure on the surface, ensuring good layeredness of the entire particle. This plasma-treated precursor was mixed with Li and calcined at 750 °C for 12 h under an O2 atmosphere to obtain modified NCA. Fig. 11G shows the rate capability of plasma-treated NCA (PTNCA), outperforming conventional NCA especially at high current density, which delivered  $\sim$ 170 mA h g<sup>-1</sup> at 5C. The improved rate capability can be plausibly explained by the effect of plasma treatment that creates the continuous layered structure throughout the particle, from bulk to surface, facilitating Li diffusion.

#### 4.3 Cation doping

Doping has been considered an effective way to enhance the electrochemical properties and chemical stability of the

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Table 2 Summary of doping elements reported as promoting the rate performance of LIBs

Dopant	Function	Cathode material	Test parameter/ cell structure	Electrolyte	Mass loading	Rate capability	Cycle retention	Ref.
Mg	Improve the structural stability	NCM90	2.7-4.3 V/NCM90 Li	1 M LiPF <sub>6</sub> in EC : EMC = 3 : 7 $(v/v)$ + 2 wt% VC	12.0 ± 0.5 mg cm <sup>-2</sup>	~165 mA h $g^{-1}$ (3C)	80 (350 cycles under 0.33C)	140
Nb	Improve the structural stability	NCM811	2.8-4.6 V/NCM811 Li	1 M LiPF <sub>6</sub> in EC : DMC = 1 : 1 $(v/v)$	15.0 mg cm <sup>-2</sup>	~160 mA h $g^{-1}$ (2C)	92 (60 cycles under 0.33C)	141
Fe	Improve the structural stability and decrease the particle size	LiNiO <sub>2</sub>	3.0–4.5 V/LiNiO <sub>2</sub>  Li	1 M $LiPF_6$ in EC : DMC = 1 : 1 (v/v)	_	~150 mA h $g^{-1}$ (5C)	96 (60 cycles under 1C)	142
Cr	Improve the structural stability and decrease the particle size	LiNiO <sub>2</sub>	3.0–4.5 V/LiNiO <sub>2</sub>  Li	1 M LiClO <sub>4</sub> in EC : DMC = $1:1 (v/v)$	_	~135 mA h $g^{-1}$ (10C)	95 (50 cycles under 1C)	143
Sn	Improve the structural stability	NCM622	2.5-4.3 V/NCM622 Li	$1 \text{ M LiPF}_6 \text{ in EC : DMC } =$ 1:1 (v/v)	—	~120 mA h g <sup>-1</sup> (5C)	90 (150 cycles under 1C)	144
Ga	Improve the structural stability and Li ion transition kinetics	NCM811	2.8–4.3 V/NCM811 Li	1 M $LiPF_6$ in EC : DMC : EMC = 1 : 1 : 1 (v/v/v)	2.21 mg cm <sup>-2</sup>	$\sim 120 \text{ mA h}$ $g^{-1} (10C)$	90 (100 cycles under 1C)	145
Zr	Improve the structural stability and Li ion transition kinetics	NCM811	2.8-4.3 V/NCM811 Li	1 M LiPF <sub>6</sub> in EC : EMC = $3:7 (v/v)$	$4 \text{ mg cm}^{-2}$	~135 mA h $g^{-1}$ (2C)	85 (50 cycles under 0.2C)	146
Ti	Improve the Li ion transition kinetics	NCM622	2.8-4.3 V/NCM622 Li	1 M LiPF <sub>6</sub> in EC : DMC : EMC = 1 : 1 : 1 (y/y/y)	_	~130 mA h $g^{-1}$ (5C)	_	147
Sb	Improve the structural stability	NC89	2.7-4.3 V/NC89 Li	1.2 M LiPF <sub>6</sub> in EC : EMC = 3 : 7 $(v/v)$ + 2 wt% VC	_	~195 mA h g <sup>-1</sup> (2C)	84 (1000 cycles under 1C)	148
Мо	Improve the structural stability	NM90	2.7–4.3 V/NM90 Li	1.2 M LiPF <sub>6</sub> in EC : EMC = 3 : 7 $(v/v)$ + 2 wt% VC	$4-5 \text{ mg} \text{ cm}^{-2}$	~180 mA h $g^{-1}$ (3C)	86 (1000 cycles under 1C)	149

cathode.<sup>134–136</sup> Table 2 summarizes the list of reported cation dopants. Intuitively, electrochemically inert elements, such as  $Mg^{2+}$ ,  $Al^{3+}$ , and  $Zr^{4+}$ , as well as  $Mn^{4+}$  and  $Ti^{4+}$ , that do not participate in the redox reaction can stabilize delithiated structures and thus are favorable in general for battery materials design.<sup>137–139</sup>

Kim and Sun et al. investigated the effect of doping (Mn, Al, B, W, and Ta) on the microstructure of LiNi<sub>0.91</sub>Co<sub>0.09</sub>O<sub>2</sub>.<sup>150</sup> In Fig. 12A, it can be seen that morphology of the primary particles depends on the doping elements: large equiaxed particles with Mn and Al and fine needle-like (preferred [003] orientation) particles with B, W and Ta. DFT was employed to theoretically confirm the surface energy modified by Ta and Al, through which the surface energies of (003) and (104) facets from both Ta-doped and Al-doped Ni-rich models were calculated. The results suggest that inducing ions with a larger ionic radius (i.e., Ta) leads to an increase of the surface energy on the (104) facets rather than (003)facets, which indicates the higher stability of (003) facets in NCTa90 than in NCA90 cathode materials. Note that the equiaxed primary particles are also obtained in the pristine NCA. This needle-like morphology is expected to effectively dissipate the strain energy of the Ni-rich layered oxide cathode upon charge and discharge as a significant amount of grain boundaries associated with ultrafine grains can deflect crack propagation compared with the large equiaxed particles.<sup>151,152</sup> These authors also explored mechano-chemical stability of Ni-rich cathodes upon Nb doping.<sup>153</sup> Nb was incorporated into NCA at the lithiation process of synthesis. As the doping content increases, the aspect ratio of NCA increases, transforming the equiaxed to elongated particles,

as denoted in Fig. 12B. Specifically, Nb-doped Ni-rich primary particles are less than 400 nm wide with an average aspect ratio (particle length divided by width) of 6.3, whereas pristine Ni-rich primary particles are 200 nm -1 µm wide with an average aspect ratio of 1.5. The orientation of the layered planes is parallel to the long axis of the elongated primary particles in Fig. 12C, implying that Nbdoped NCA has a strong crystallographic texture. According to EDS line mapping across multiple particles, Nb was mostly found at the grain boundaries that likely function as Nb diffusion pathways.<sup>153</sup> Grain boundary pinning of Nb may stabilize the surface structure of NCA mechanically by suppressing lattice distortion and/or phase transformations of the layered phase. Indeed, Nb doping leads to a highcapacity retention of 93.1% (~170 mA h g<sup>-1</sup>) for 500 cycles at a 3C rate.

It should be highlighted that cation doping has been sought to stabilize Co-free Ni-rich cathodes for high-power applications of Li-ion batteries. Park *et al.* used high-valence  $Mo^{6^+}$  as a dopant to enhance the electrochemical stability of LiNi<sub>0.9</sub>Mn<sub>0.1</sub>O<sub>2</sub> (NM90) upon delithiation and its Li kinetics.<sup>149</sup> As observed by TEM and the corresponding fast Fourier transform (FFT) in Fig. 12D, LiNi<sub>0.89</sub>Mn<sub>0.1</sub>Mo<sub>0.01</sub>O<sub>2</sub> (Mo-NM90) charged at 4.4 V maintained the layered structure without signs of surface damage and intragranular fractures. To compare the rate capability, LiNi<sub>0.9</sub>Co<sub>0.05</sub>Mn<sub>0.05</sub>O<sub>2</sub> (NCM90), NM90, and Mo-NM90 were charged at two different rates alternately between 0.5C and 3C. In Fig. 12E, Mo-NM90 demonstrates higher capacity retention rates (~200 mA h g<sup>-1</sup> at 0.5C and ~174 mA h g<sup>-1</sup> at 3C) after 100 cycles at both rates than the others. Li diffusion can be 3 to 5 times faster along the



**Fig. 12** (A) Scanning TEM image of a cross-sectional cathode particle and the corresponding schematic of the primary particles (Li  $[Ni_{0.90}Co_{0.05}Mn_{0.05}]O_2$ , Li $[Ni_{0.895}Co_{0.09}B_{0.015}]O_2$ , Li $[Ni_{0.90}Co_{0.09}Al_{0.01}]O_2$ , Li $[Ni_{0.90}Co_{0.09}W_{0.01}]O_2$  and Li $[Ni_{0.90}Co_{0.09}Ta_{0.01}]O_2$ , denoted as NCM90, NCB90, NCA90, NCW90 and NCTa90, respectively).<sup>150</sup> (B) The aspect ratio of the primary particle.<sup>153</sup> (C) Plan-view TEM image of a Nb-doped NCA85 cathode and associated EDS line scans evaluating the Nb concentrations of the primary particles in the region marked with a yellow box.<sup>153</sup> (D) TEM image and SAED pattern of the primary particle marked in D.<sup>149</sup> (E) Cycling performances of half-cells featuring NCM90, NM90, and Mo-NM90 cathodes alternately charged at 0.5C for four cycles and 3C for three cycles at a time.<sup>149</sup>

grain boundaries than in bulk.<sup>149</sup> Given that Mo-NM90 exhibits fine primary particles, the improved charged state stability and high rate capability can be attributable to high grain boundary density as a result of Mo doping. However, introducing dopants may lead to the reduction of the discharge capacity such as Mg<sup>2+</sup>, Al<sup>3+</sup> and Ti<sup>4+</sup> which are considered inactive in the layered structure of Ni-based oxide cathode materials. In addition, dopants like W<sup>6+</sup> and Co<sup>3+</sup> are expensive and, therefore, may result in higher costs of battery manufacturing.<sup>154</sup> Accordingly, cost-effective dopants that improve

the overall electrochemical performance without reducing the discharge capacity need to be investigated and developed in the future.

#### 4.4 Coating

Coating has been widely used to tailor the properties of cathode particles. Coating processes can also be compatible with other cathode optimization techniques such as cation doping and particle/grain size engineering by systematic experimental design. A wide variety of materials listed in

Table 3	Summary c	of coating	materials	reported a	as promoting	the rate	performance	of LIBs
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Coating material	Cathode material	Test parameter/ cell structure	Electrolyte	Mass loading	Rate capability	Cycle retention	Ref.
LiF	NCM111	2.5–4.5 V/NCM111 Li	1 M LiPF <sub>6</sub> in EC : DMC = 1 : 1 (v/v)	_	~140 mA h $g^{-1}$ (10C)	89 (50 cycles under 1C)	155
Li <sub>3</sub> PO <sub>4</sub>	NCA	2.7-4.3 V/NCA Li	1 M LiPF <sub>6</sub> in EC : DEC = 1 : 1 (v/v)	4.5 mg cm <sup>-2</sup>	~125 mÅ h $g^{-1}$ (10C)	88 (100 cycles under 0.5C)	156
$C_6H_9O6La \cdot_x H_2O,$ $C_{16}H_{36}O_4Ti$ (Dual)	NCM811	2.7–4.3 V/NCM811 Li	$1 \text{ M LiPF}_6 \text{ in EC : DMC : EMC =}$ 1:1:1(v/v/v)	$2.3 \text{ mg cm}^{-2}$	~160 mÅ h $g^{-1}$ (10C)	90 (200 cycles under 1C)	157
TiO <sub>2</sub>	NCM622	3.0-4.5 V/NCM622 Li	1 M LiPF <sub>6</sub> in EC : DMC = 1 : 1 (v/v)	—	~145 mÅ h $g^{-1}$ (5C)	89 (50 cycles under 1C)	158
Li <sub>x</sub> TiO <sub>2</sub> -silica	NCM622	2.75–4.3 V/ NCM622 Li	1 M LiPF <sub>6</sub> in EC : DMC : EMC = 1 : 1 : 1 $(v/v/v)$	$4 \text{ mg cm}^{-2}$	~140 mA h $g^{-1}$ (5C)	91 (100 cycles under 0.5C)	159
MgO	NCM523	3.0-4.3 V/NCM523 Li	1 M LiPF <sub>6</sub> in EC : DEC = 1 : 1 (v/v)	_	$\sim 120 \text{ mA h}$ g <sup>-1</sup> (10C)	80 (180 cycles under 1C)	160
MOF-derived alumina (MDA)	NCM622	3.0-4.5 V/NCM622 Li	1 M LiPF <sub>6</sub> in EC : DMC = 1 : 1 ( $v/v$ )	2-3  mg cm <sup>-2</sup>	~150 mA h $g^{-1}$ (10C)	92 (100 cycles under 1C)	161
PEDOT	NCM811	3.0-4.3 V/NCM811 Li	1 M LiPF <sub>6</sub> in EC : DMC : EMC = 1 : 1 : 1 $(v/v/v)$	$3.6 \text{ mg cm}^{-2}$	~170 mÅ h $g^{-1}$ (7C)	80 (300 cycles under 1C)	162
Co <sub>x</sub> B	NCM811	3.0-4.4 V/NCM811 Li	1 M LiPF <sub>6</sub> in EC : DMC : EMC = 3 : 3 : 4 $(v/v/v)$ + 5 wt% FEC	$10.5 \pm 0.2 \text{ mg cm}^{-2}$	~200 mA h $g^{-1}$ (7C)	82 (200 cycles under 7C)	163

Table 3 can be used to coat primary particles (nanoscale), secondary particles (microscale), and the entire cathode (macroscale) using different methods.

Li *et al.* employed atomic layer deposition (ALD) to directly apply a thin layer of LiTaO<sub>3</sub> coating on the surface of the NCM111 cathode.<sup>164</sup> Fig. 13A compares the surface texture of the NCM particles before and after the ALD coating process and shows a uniform and conformal ALD layer. The EDS result in Fig. 13B clearly indicates that a Ta-rich coating (~5 nm) is formed on the particle. Fig. 13C shows the rate capability tests of ALD-coated NCM cathodes. Optimized ALD-coated NCM111 delivers high capacity, cyclability, and high rate capability. At 700 mA g<sup>-1</sup>, NCM111 coated with five ALD layers (NMC-5 in Fig. 13C) achieves a high discharge capacity of 125 mA h g<sup>-1</sup>, 77% of the capacity obtained at 100 mA g<sup>-1</sup>. While a thin coating is unlikely to impede Li diffusion across the cathode– electrolyte interface, a thick ALD coating leads to poor rate capability for the NCM cathode (*e.g.*, NMC-10 in Fig. 13C).

Oxidative chemical vapor deposition (oCVD) that leverages deep penetration of vaporized phases into porous media is another effective method to conformally coat NCM particles. As illustrated in Fig. 13D, Xu et al. used oCVD to apply a poly (3,4-ethylenedioxythiophene) (PEDOT) coating on primary and secondary NCM particles simultaneously.165 At high-voltage charging in 4.3-4.6 V, the in situ XRD in Fig. 13E shows continuous peak shifting but no peak splitting, indicating the integrity of the layered phase as a result of the PEDOT coating, likely dissipating the strain energy associated with the volume change of the H3 phase. Zhang et al. demonstrated that PEDOT coating by oCVD can be directly applicable to cathodes on a larger scale (Fig. 13F).<sup>162</sup> NCM811 cathodes with PEDOT coating outperform pristine NCM 811 for high rate capability and cyclability. In Fig. 13G, the discharge capacities of PEDOTcoated and pristine NCM811 obtained at 7C are 170 and 150 mA h g<sup>-1</sup>, respectively. In a subsequent 0.1C discharge,

coated NCM recovers the full initial capacity, while pristine NCM shows substantial capacity loss, indicative of structural and chemical damage in the particles.

Similarly, Cho and coworkers developed a coating-plus-infusion method to passivate both primary and secondary particles.<sup>163</sup> They demonstrated that as-synthesized Co<sub>x</sub>B can completely coat the secondary particles and subsequently infiltrate into interfaces between the primary particles (Fig. 13H). The cycling stability of NCM811 was dramatically improved by this infused microstructure at high discharge rates (up to 1540 mA  $g^{-1}$ ). It should be pointed out that these tests were performed at 45 °C, which generally shows fast Li kinetics unless the structure is unstable. The rate performance of Co<sub>x</sub>Bcoated NCM811 and pristine NCM811 at 25 °C is shown in Fig. 13I. The Co<sub>x</sub>B-coated NCM811 cathode delivers a specific capacity  $\sim 175$  mA h g<sup>-1</sup> at 10C compared with  $\sim 140$  mA h g<sup>-1</sup> for pristine NCM811, indicating that CoxB coating-and-infusion can enhance the rate capability and cycling stability of NCM. To reveal the functionality of the amorphous Co<sub>x</sub>B coating layer on the electrochemical performance of the NCM811 cathode material, DFT calculations were conducted to unveil that the strength of Co-O and B-O bonds is stronger than Ni-O bonds. The results indicate the strong covalent B-O and Co-O bonds stabilized the interface O, which enhanced the electronic-structure stability, and hence the rate capability and cycling stability. The coating layers achieved by novel methods such as ALD and oCVD are conformal and the thickness controllable. However the overall cost leveraging the ALD and oCVD techniques may increase and the precursor materials of these methods are limited, depending on the reaction mechanisms. More economical strategies such as hydrothermal, solid state, and sol-gel processes are available on large-scale electrodes with an easier selection of coating materials. However, the coating layer by these conventional approaches is less uniform and conformal (or low step cover-



**Fig. 13** (A) Typical morphologies of pristine and ALD LiTaO<sub>3</sub>-modified NCM111 and (B) linear EDX scan of ALD-modified NCM.<sup>164</sup> (C) The rate capability of various ALD-modified NCM cathodes.<sup>164</sup> (D) An illustration of the structural stability of both the primary particle and secondary particle coating *via* oCVD after long-term cycling.<sup>165</sup> (E) Selected *in situ* high-energy XRD patterns from oCVD PEDOT-coated cathodes in the high-voltage (4.3–4.6 V) region during the charge process.<sup>165</sup> (F) Schematic of the vapor-phase synthetic process of a Ni-rich cathode modified with highly conformal oCVD PEDOT.<sup>162</sup> (G) The rate capability of various kinds of PEDOT coated NCM cathode.<sup>162</sup> (H) Schematic coating-plus-infusion microstructure in which Co<sub>x</sub>B uniformly coats the surface of NCM secondary particles and infuses grain boundaries between the NCM primary particles.<sup>163</sup> (I) 7C cycling tests in the range of 3.0–4.4 V *versus* Li/Li<sup>+</sup> at 45 °C.<sup>163</sup>

age) for Ni-based layered oxide particles, compared with those by novel techniques.<sup>154</sup> The uncoated regions due to the nonuniformity of the conventional approaches will be directly exposed to the electrolyte and result in performance fading due to electrolyte oxidation. Therefore, affordable coating methods realizing high coverage and uniformity are required.

#### 4.5 Electrolyte additives

The localized inhomogeneity at the particle and electrode levels induced by fast-charging can lead to partial overdelithiation in Ni-rich NCM (section 3.1). Highly oxidized Ni ions ( $Ni^{4+}$ ) are unstable by nature and can be readily reduced to Ni<sup>3+</sup> and Ni<sup>2+</sup> by oxidizing the surrounding oxygen (*i.e.*, oxygen gas evolution). If Ni<sup>4+</sup> oxidizes electrolyte molecules, electrolyte decomposition products known as cathode–electrolyte interphases (CEIs) that increase cell impedance can form. Functional additives to the electrolyte can address problems associated with undesirable Ni reduction by increasing the anodic stability of the electrolyte as well as changing the composition of the CEIs. Indeed, designing uniform cathode–electrolyte interphases (CEI) that have a self-limiting thickness to provide robust particle passivation while permitting facile Li diffusion is key to developing next-generation cathodes for high-voltage, fast-charging Li-ion batteries. Table 4 summar-

Table 4 Summary of electrolyte additives reported as promoting the rate performance of LIBs

Additive	Cathode material	Test parameter/ cell structure	Electrolyte	Mass loading	Rate capability	Cycle retention	Ref.
[4,4'-Bi(1,3,2-dioxathiolane)]2,2'- dioxide	NCM532	3.0–4.6 V/ NCM532 Li	1 M LiPF <sub>6</sub> in EC : EMC = $3:7 (v/v)$	$22.5 \text{ mg cm}^{-2}$	~160 mA h $g^{-1}$ (10C)	92 (100 cycles under 0.5C)	166
<i>p</i> -Toluenesulfonyl isocyanate	NCM532	3.0–4.6 V/ NCM532 Li	1 M $LiPF_6$ in EC : EMC = 3 : 7 (v/v)	$1.8 \text{ mg cm}^{-2}$	~120 mÅ h $g^{-1}$ (10C)	86 (100 cycles under 1C)	167
Qisophorone diisocyanate	NCM532	2.5–4.6 V/ NCM532 Li	1 M LiPF <sub>6</sub> in PC : DMC = $1:1 (v/v)$	—	~150 mÅ h $g^{-1}$ (5C)	83 (200 cycles under 1C)	168
Lithium difluorophosphate	NCM532	2.8–4.6 V/ NCM532  graphite	1 M $LiPF_6$ in EC : DEC = 1 : 3 (wt/wt)	31.38 mg cm <sup>-2</sup>	~120 mA h $g^{-1}$ (5C)	62 (160 cycles under 0.5C)	169
Tris(trimethylsilyl)phosphite	NCM622	2.8-4.45 V NCM622  MCMB	1 M LiPF <sub>6</sub> in EC : DMC : EMC = 1 : 1 : 1 (v/v/v)	11.7 mg cm <sup>-2</sup>	~130 mA h $g^{-1}$ (5C)	79 (300 cycles under 1C)	170
Diphenyldimethoxysilane	NCM622	2.8–4.3 V/ NCM622 Li	1 M LiPF <sub>6</sub> in EC: DMC: EMC = 1:1:1 (v/v/v)	1.52 mg cm <sup>-2</sup>	~143 mA h $g^{-1}$ (10C)	99 (200 cycles under 2C)	171
3-Isocyanatopropyltriethoxysilane	NCM622	2.5–4.5 V/ NCM622 Li	1 M LiPF <sub>6</sub> in EC: DMC: EMC = 1:1:1 (y/y/y)	2.76–3.46 mg cm <sup>-2</sup>	$\sim 120 \text{ mA h}$ g <sup>-1</sup> (10C)	73 (159 cycles under 0.3C)	172
Succinic anhydride	NCM811	2.8–4.3 V/ NCM811  graphite	1 M LiPF <sub>6</sub> in EC : EMC = $3:7 (v/v)$	10 mg cm <sup>-2</sup>	~150 mA h $g^{-1}$ (5C)	94 (400 cycles under 1C)	173
<i>N</i> -allyl- <i>N</i> , <i>N</i> -bis(trimethylsilyl) amine	NCA85	3.0–4.2 V/ NCA85 Li	1 M LiPF <sub>6</sub> in EC : DEC : EMC = $3 : 2 : 5$ (v/v/v)	3.68 mg cm <sup>-2</sup>	~100 mA h $g^{-1}$ (7C)	86 (300 cycles under 1C)	174
1,2,4–1 <i>H</i> -triazole	NCM90	3.0–4.3 V/ NCM90 Li	1 M LiPF <sub>6</sub> in EC : DEC = $3:7$ (v/v)	$3-4 \text{ mg cm}^{-2}$	~175 mA h g <sup>-1</sup> (10C)	87 (150 cycles under 1C)	175
LiDFOB	LiNiO <sub>2</sub>	2.7–4.4 V/ LiNiO <sub>2</sub>  Li	1 M LiP $F_6$ in FEC: FEMC: HFE = 2:6:2 (wt/wt/wt)	$3 \text{ mg cm}^{-2}$	~120 mA h $g^{-1}$ (5C)	80 (400 cycles under 0.5C)	176

izes electrolyte additives used to improve the electrochemical properties of Ni-rich layered oxide cathodes.

Sun and coworkers reported that succinic anhydride (SA) addition to the carbonate electrolyte enhances the rate capability of Li/NCM811 cells.<sup>173</sup> A DFT calculation was conducted to compare the highest occupied molecular orbital (HOMO) energies of SA molecules and molecules of the electrolyte solvent. Through Gauss fitting, the values of the HOMO energies of SA, EC, and EMC were determined to be about -6.45 eV, -8.50 eV, and -8.20 eV, respectively, indicating that the SA molecules are oxidized preferentially at the cathode during the charging process. They found that SA induces uniform CEIs that prevent internal cracking from the inside of secondary particles (Fig. 14A). Electrochemical impedance spectroscopy (EIS) in Fig. 14B shows that impedance, as a lump sum of interphase and charge transfer resistance contributions, for the cell with SA after 100 cycles is much smaller than for the cell without it. This result suggests that SA-related CEIs can passivate and stabilize the high-voltage structure of NCM811 particles at fast charge (~150 mA h  $g^{-1}$  at 5C). For a Ni-rich  $(Ni \ge 0.9)$  layered oxide cathode,  $LiNi_{0.9}Co_{0.05}Mn_{0.05}O_2$ (NCM90), Zou et al. used 1,2,4-1H-triazole (HTZ) as the electrolyte additive that can reduce solution resistance and tailor CEI compositions.175 Similarly, the HOMO/lowest unoccupied molecular orbital (LUMO) values of EC, DEC, and HTZ were compared through DFT calculations. The higher HOMO value and lower LUMO value of HTZ suggest that HTZ would be preferably oxidized at the cathode and reduced at the anode compared with the EC and DEC molecules in electrolyte, generating a CEI and SEI layer on the cathode and anode respectively. Fig. 14C plots the specific capacity of the NCM90 cathodes with the baseline electrolyte and 0.3% HTZ-containing electrolyte as a function of the discharge rates for multiple cycles. By introducing HTZ to the electrolyte, the NCM90 cathode demonstrates improved rate capability (~175 mA h g<sup>-1</sup> at 10C) and capacity retention (87% after 150 cycles). This result corroborates with the SEM and TEM images in Fig. 14D and E, which show clean NCM90 particles with uniform CEI after cycling with the HTZ-added electrolyte.

In addition to oxidation stability against the cathode, the judicious selection of additives can enhance the electrolyte's reduction stability against the anode. Lu and coworkers reported the effect of lithium tetrafluoroborate (LiBF<sub>4</sub>) and lithium nitrate (LiNO<sub>3</sub>) on electrochemical stability of a carbonated-based electrolyte.<sup>177</sup> These dual-salt additives can form stable CEIs on the primary particles of NCM811, as observed by HRTEM in Fig. 14F, and possibly form stable solid–electrolyte interphases (SEIs) on the anode surface (Fig. 14G), promoting Li kinetics across the entire cell and cycling stability. In Fig. 14H, conventional ethylene carbonate (EC)/diethyl carbonate (DEC) does not sustain the high voltage cutoff (4.4 V) of a Li/NCM811 half-cell at rates higher than 0.3C. In contrast, the cell with fluoroethylene carbonate (FEC)/ethyl methyl carbonate (EMC) shows respectable rate capability even without





uniform CE

в

Fig. 14 (A) Schematic illustration of the protection mechanism of CEI formed by SA.<sup>173</sup> (B) Nyquist plots of Li/NCM811 cells after 100 cycles.<sup>173</sup> (C) Rate performance of the HTZ-additive and baseline cells at 30 °C<sup>.175</sup> (D) SEM and (E) TEM images of NCM90 electrodes after 150 cycles.<sup>175</sup> (F) HRTEM images of NCM811 cycled in electrolyte with dual additives after 100 cycles.<sup>177</sup> (G) Schematic of dual-salt-additives working on both cathode and anode.<sup>177</sup> (H) Charging at an increasing C rate but discharging at a constant 0.5C rate.<sup>177</sup>

additives. Fluorides are favorable interphase compositions due to their high reduction and oxidation stability,<sup>178,179</sup> and thus the fluorination of CEIs and SEIs by decomposing FEC likely enables reversible Li intercalation at high rates. While LiBF4 addition to the FEC/EMC electrolyte alone does not improve Li intercalation kinetics much, the dual-salt additives lead to a substantial increase in high rate capacity, showing 185.6 mA h  $g^{-1}$  at 5C (~10 mA cm<sup>-2</sup>). It is likely that electrically conductive nitrides may form in the interphases, reducing the charge transfer kinetics at high rates. Noteworthy, the electrolyte additive (e.g., salt-type) may be reduced and decomposed at the anode side in full cells due to the low LUMO level of these additives and will not form a suitable CEI on the cathode.

Besides, some types of electrolyte additive are detrimental to the anode material (e.g., Si-based), which should be utilized with other compounds to build a F-donating SEI protection layer for the Si anode.180

#### 4.6 Electrode architectures

While a conventional cathode construction that mixes active NCM particles uniformly with activated carbon and a binder is non-directional at a macroscopic scale, the electric field developed in the cathode upon charge and discharge is directional at a microscopic scale. Thus, the SoC for individual NCM particles (i.e., different Li concentrations) may be different from each other, especially if the charge carrier and electron conduction pathways are poorly percolated, developing local inhomogeneity in the cathode upon charge and discharge. This kinetic limitation known as polarization increases as the charging rate increases (*i.e.*, fast-charging). Therefore, the fast charge performance of NCM cathodes can be improved by tailoring the electrode architectures.

To develop thick electrodes for high energy density, Wood et al. investigated how the particle size distribution of a Nibased layered oxide cathode and the graphite anode affects specific capacity as a function of rates.<sup>181</sup> Fig. 15A shows five different electrode configurations: (1) single-sized small ( $6 \mu m$ ) particles, (2) random mixing of small and large (12 µm) particles, (3) layer-by-layer where large particles are near the current collector, (4) layer-by-layer where small particles are near the current collector, and (5) single-sized large particles. In Fig. 15B, although the rate capability is poor for all cases due to the thick electrode construction, cathode configuration (3) provides the highest 2C capacity of 100 mA h  $g^{-1}$  in combination with anode configuration (4), compared with the others. This result can suggest that the tortuosity of the porous electrode and the particle size distribution play a critical role in Li kinetics, while clear understanding is lacking. Hu et al. argued that cathode calendering can physically enhance Li transport and electron conduction in the cathode.<sup>182</sup> As shown in Fig. 15C, Li ions can migrate inside the cathode through the porosity where a large free volume for electrolyte penetration exists. In this case the cathode is charge transfer-limited as electronic conduction is not percolated. Calendering provides a high packing density, leading to an effective percolation of both Li and electron pathways. Note that over-calendering may lead to particle cracking and/or reduce electrolyte penetration that decreases the number of accessible Li at fast charge. These exercises suggest that a balance between high energy

density (packing fraction) and high power (tortuosity) can be achieved by engineering the electrode configurations.

# 5. Summary and outlook

Rapidly increasing EV production demands next-generation Liion batteries. Ni-based layered oxides are the strongest cathode system for future EV technology due to the potential for high capacity at a reasonable cost. Many car makers have implemented the Ni-rich cathode in battery packs for their EVs. However, pushing to the limit of the Ni content in the layered oxide cathodes faces challenges due to electrochemical degradation especially at fast-charging and high voltage.

This review discusses the degradation mechanisms of Nibased cathodes at fast charge in various length scales. Chemical and mechanical inhomogeneities developed in charged Ni-rich layered oxides are critical to determining the rate capability and cyclability of the cathode. The electrode level inhomogeneity is attributed to the non-uniform SoC in individual particles, leading to microcracks in particles and local delamination of the electrode. The single particle-level inhomogeneity results in the different degrees of delithiation. Local overcharge leads to oxygen loss from layered oxides, inducing electrolyte oxidation, irreversible phase transformations (from layered to spinel and to rock-salt phase), and transition metal dissolution, all of which lead to sudden cell failure. Of note, the fast charging process will accelerate the oxygen loss due to large polarization.

Strategies to improve the rate capability and cyclability of Ni-rich layered oxide cathodes include, but are not limited to, synthesis optimization (modified co-precipitation), single crystal growth, engineering of electrolyte additives, cation



**Fig. 15** (A) Thick electrode designs with different active material particle sizes and configurations.<sup>181</sup> (B) Rate performance comparison for singlelayer pouch cells made with all 25 cathode/anode design combinations.<sup>181</sup> (C) Sketch of cathode calendering from high thickness (h1) to low thicknesses (h2 and h3, h2 > h3).<sup>182</sup>

doping, protective surface coating, and cathode architecture design. To summarize recent progress, the development of Nibased layered oxide cathodes can be envisioned as below:

(1) Tailoring the size, shape, and combination of primary particles of Ni-based layered oxides to reduce local overcharging;

(2) Functional coating, doping or designing zero-strain compositions to enhance the electrochemical and chemical stability of Ni-based layered oxides by suppressing oxygen loss;

(3) Electrode construction with engineered thickness and tortuosity to promote rate capability and energy density.

It should be emphasized that addressing complex problems present in modern cathodes by one approach is highly unlikely as electrochemical and physicochemical properties are strongly coupled. A holistic approach leveraging multiple strategies is therefore required to enable fast charging Ni-rich layered oxides for emerging EV applications.

# Conflicts of interest

There are no conflicts to declare.

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#### References

- 1 A. L. Stephen Nalley, International Energy Outlook 2021, https://www.eia.gov/outlooks/ieo/, (accessed October 6th, 2021).
- 2 T. U. S. E. P. Agency, Inventory of U.S. Greenhouse Gas Emissions and Sinks: 1990–2019, https://www.epa.gov/ ghgemissions/inventory-us-greenhouse-gas-emissions-andsinks-1990-2019, (accessed April, 2021).
- 3 G. Motors, GM Corporate Newsroom, https://media.gm. com/media/us/en/gm/home.detail.ht.
- 4 F. M. Company, Ford Media Center, https://media.ford. com/content/fordmedia/fna/us/en.
- 5 H. Shimotake, *Progress in Batteries and Solar Cells*, JEC Press, 1990, vol. 9.
- 6 M. Armand and J. M. Tarascon, *Nature*, 2008, **451**, 652–657.
- 7 L. Lu, X. Han, J. Li, J. Hua and M. Ouyang, *J. Power* Sources, 2013, **226**, 272–288.
- 8 J. B. Goodenough and K.-S. Park, J. Am. Chem. Soc., 2013, 135, 1167–1176.
- 9 BloombergNEF, Electric Vehicle Outlook 2021, https:// bnef.turtl.co/story/evo-2021/page/1?teaser.

- 10 M. S. Whittingham, Chem. Rev., 2004, 104, 4271-4302.
- 11 J. Zhang, H. Zhong, C. Zheng, Y. Xia, C. Liang, H. Huang, Y. Gan, X. Tao and W. Zhang, *J. Power Sources*, 2018, **391**, 73–79.
- 12 Y. Zhu, S. He, J. Ding, G. Zhao and F. Lian, *Nano Res.*, 2022, DOI: 10.1007/s12274-022-4852-y.
- 13 W. Li, E. M. Erickson and A. Manthiram, *Nat. Energy*, 2020, 5, 26–34.
- 14 W. Liu, P. Oh, X. Liu, M.-J. Lee, W. Cho, S. Chae, Y. Kim and J. Cho, Angew. Chem., Int. Ed., 2015, 54, 4440– 4457.
- 15 Y.-K. Sun, S.-T. Myung, B.-C. Park, J. Prakash, I. Belharouak and K. Amine, *Nat. Mater.*, 2009, **8**, 320–324.
- 16 P. Morrissey, P. Weldon and M. O'Mahony, *Energy Policy*, 2016, **89**, 257–270.
- A. Meintz, J. Zhang, R. Vijayagopal, C. Kreutzer, S. Ahmed,
  I. Bloom, A. Burnham, R. B. Carlson, F. Dias, E. J. Dufek,
  J. Francfort, K. Hardy, A. N. Jansen, M. Keyser, A. Markel,
  C. Michelbacher, M. Mohanpurkar, A. Pesaran,
  D. Scoffield, M. Shirk, T. Stephens and T. Tanim, *J. Power* Sources, 2017, 367, 216–227.
- 18 F. R. C. NACS, The US Petroleum Industry Statistics, https://www.nacsonline.com/YourBusiness/FuelsCenter/Basics/ Articles/Pages/The-US-Petroleum-Industry-Statistics-Definitions. aspx#.WEStQHrN7hE, (accessed December 4, 2016).
- 19 M. Li, M. Feng, D. Luo and Z. Chen, *Cell Rep. Phys. Sci.*, 2020, 1, 100212.
- 20 D. Howell, T. Duong, P. Faguy and B. Cunningham, US Department of Energy Vehicle Battery R&D: Progress Update, https://www.hydrogen.energy.gov/pdfs/htac\_nov2011\_ howell.pdf, 2011.
- 21 A. Manthiram, J. C. Knight, S.-T. Myung, S.-M. Oh and Y.-K. Sun, *Adv. Energy Mater.*, 2016, 6, 1501010.
- 22 K. M. Shaju, G. V. Subba Rao and B. V. R. Chowdari, *Electrochim. Acta*, 2002, **48**, 145–151.
- 23 I. Belharouak, W. Lu, D. Vissers and K. Amine, *Electrochem. Commun.*, 2006, **8**, 329–335.
- 24 J. Kim, H. Lee, H. Cha, M. Yoon, M. Park and J. Cho, *Adv. Energy Mater.*, 2018, 8, 1702028.
- 25 L. Ma, M. Nie, J. Xia and J. Dahn, J. Power Sources, 2016, 327, 145–150.
- 26 T. Ohzuku, A. Ueda and M. Nagayama, *J. Electrochem. Soc.*, 1993, **140**, 1862.
- 27 X. Zeng, G.-L. Xu, Y. Li, X. Luo, F. Maglia, C. Bauer, S. F. Lux, O. Paschos, S.-J. Kim, P. Lamp, J. Lu, K. Amine and Z. Chen, ACS Appl. Mater. Interfaces, 2016, 8, 3446– 3451.
- 28 L. Zou, W. Zhao, Z. Liu, H. Jia, J. Zheng, G. Wang, Y. Yang, J.-G. Zhang and C. Wang, *ACS Energy Lett.*, 2018, 3, 2433– 2440.
- 29 H. Li, N. Zhang, J. Li and J. R. Dahn, J. Electrochem. Soc., 2018, 165, A2985–A2993.
- 30 S. S. Zhang, InfoMat, 2020, 2, 942-949.
- 31 A. Tomaszewska, Z. Chu, X. Feng, S. O'Kane, X. Liu, J. Chen, C. Ji, E. Endler, R. Li, L. Liu, Y. Li, S. Zheng, S. Vetterlein, M. Gao, J. Du, M. Parkes, M. Ouyang,

M. Marinescu, G. Offer and B. Wu, *eTransportation*, 2019, 1, 100011.

- 32 S. S. Zhang, ChemElectroChem, 2020, 7, 3569-3577.
- 33 S. S. Zhang, *Energy Technol.*, 2022, 2200370.
- 34 A. Meintz, J. Zhang, R. Vijayagopal, C. Kreutzer, S. Ahmed, I. Bloom, A. Burnham, R. B. Carlson, F. Dias, E. J. Dufek, J. Francfort, K. Hardy, A. N. Jansen, M. Keyser, A. Markel, C. Michelbacher, M. Mohanpurkar, A. Pesaran, D. Scoffield, M. Shirk, T. Stephens and T. Tanim, *J. Power Sources*, 2017, 367, 216–227.
- 35 D. Howell, S. Boyd, B. Cunningham, S. Gillard and L. Slezak, *Enabling Fast Charging: A Technology Gap* Assessment, https://www.energy.gov/sites/default/files/2017/10/ f38/XFC%20Technology%20Gap%20Assessment%20Report\_ FINAL\_10202017.pdf, 2017.
- 36 M. Motors, i-MiEV charging features, https://www.mitsubishicars.com/imiev/features/charging, (accessed April, 2017).
- 37 H. Boeriu, BMW and Nissan are joining forces to offer public DC Fast charging, https://www.bmwblog.com/2015/ 12/21/bmw-and-nissan-are-joining-forces-to-offer-public-dcfast-charging/, (accessed December 21, 2015).
- 38 B. USA, BMWi3: Overview, https://www.bmwusa.com/ vehicles/bmwi/i3.html, (accessed April, 2017).
- 39 Chevrolet, 2017 BOLT EV owner's manual, https://my.chevrolet.com/content/dam/gmownercenter/gmna/dynamic/manuals/ 2017/Chevrolet/BOLT%20EV/Owner's%20Manual.pdf, (accessed April, 2017).
- 40 Tesla, Supercharger, https://www.tesla.com/supercharger, (accessed November 30, 2016).
- 41 Tesla, Charging, https://www.tesla.com/charging, (accessed April, 2017).
- 42 S. Ahmed, I. Bloom, A. N. Jansen, T. Tanim, E. J. Dufek, A. Pesaran, A. Burnham, R. B. Carlson, F. Dias, K. Hardy, M. Keyser, C. Kreuzer, A. Markel, A. Meintz, C. Michelbacher, M. Mohanpurkar, P. A. Nelson, D. C. Robertson, D. Scoffield, M. Shirk, T. Stephens, R. Vijayagopal and J. Zhang, *J. Power Sources*, 2017, 367, 250–262.
- 43 J. Dahn, E. Fuller, M. Obrovac and U. Von Sacken, *Solid State Ionics*, 1994, **69**, 265–270.
- 44 H. Arai, S. Okada, Y. Sakurai and J. I. Yamaki, J. Electrochem. Soc., 1997, 144, 3117.
- 45 R. Kostecki and F. McLarnon, *Electrochem. Solid-State Lett.*, 2004, 7, A380.
- 46 J. Dahn, U. von Sacken and C. Michal, *Solid State Ionics*, 1990, 44, 87–97.
- 47 I. Belharouak, W. Lu, D. Vissers and K. Amine, *Electrochem. Commun.*, 2006, **8**, 329–335.
- 48 C. Delmas, J. Peres, A. Rougier, A. Demourgues, F. Weill,
  A. Chadwick, M. Broussely, F. Perton, P. Biensan and
  P. Willmann, *J. Power Sources*, 1997, 68, 120–125.
- 49 D. Caurant, N. Baffier, B. Garcia and J. Pereira-Ramos, Solid State Ionics, 1996, **91**, 45–54.
- 50 A. Rougier, I. Saadoune, P. Gravereau, P. Willmann and C. Delmasa, *Solid State Ionics*, 1996, **90**, 83–90.

- 51 E. Rossen, C. Jones and J. Dahn, *Solid State Ionics*, 1992, 57, 311–318.
- 52 Y. Nitta, K. Okamura, K. Haraguchi, S. Kobayashi and A. Ohata, *J. Power Sources*, 1995, **54**, 511–515.
- 53 T. Ohzuku, A. Ueda and M. Kouguchi, *J. Electrochem. Soc.*, 1995, **142**, 4033.
- 54 M. Guilmard, A. Rougier, M. Grüne, L. Croguennec and C. Delmas, *J. Power Sources*, 2003, **115**, 305–314.
- 55 Z. Liu, H. Zhen, Y. Kim and C. Liang, *J. Power Sources*, 2011, **196**, 10201–10206.
- 56 E. A. Olivetti, G. Ceder, G. G. Gaustad and X. Fu, *Joule*, 2017, 1, 229–243.
- 57 D. Abraham, R. Twesten, M. Balasubramanian, J. Kropf, D. Fischer, J. McBreen, I. Petrov and K. Amine, *J. Electrochem. Soc.*, 2003, 150, A1450.
- 58 E. McCalla and J. Dahn, Solid State Ionics, 2013, 242, 1–9.
- 59 S.-P. Lin, K.-Z. Fung, Y.-M. Hon and M.-H. Hon, J. Solid State Chem., 2002, 167, 97–106.
- 60 H.-J. Noh, S. Youn, C. S. Yoon and Y.-K. Sun, J. Power Sources, 2013, 233, 121–130.
- 61 X. Wang, Y. L. Ding, Y. P. Deng and Z. Chen, *Adv. Energy Mater.*, 2020, **10**, 1903864.
- 62 A. Chen, K. Wang, J. Li, Q. Mao, Z. Xiao, D. Zhu, G. Wang,
  P. Liao, J. He, Y. You and Y. Xia, *Front. Energy Res.*, 2020,
  8, 593009.
- 63 S.-T. Myung, F. Maglia, K.-J. Park, C. S. Yoon, P. Lamp, S.-J. Kim and Y.-K. Sun, ACS Energy Lett., 2017, 2, 196–223.
- 64 P. Yan, J. Zheng, T. Chen, L. Luo, Y. Jiang, K. Wang, M. Sui, J.-G. Zhang, S. Zhang and C. Wang, *Nat. Commun.*, 2018, 9, 1–8.
- 65 H.-H. Sun and A. Manthiram, *Chem. Mater.*, 2017, 29, 8486–8493.
- 66 J. Zhang, C. Zheng, L. Li, Y. Xia, H. Huang, Y. Gan, C. Liang, X. He, X. Tao and W. Zhang, *Adv. Energy Mater.*, 2020, **10**, 1903311.
- 67 H.-H. Ryu, K.-J. Park, C. S. Yoon and Y.-K. Sun, *Chem. Mater.*, 2018, **30**, 1155–1163.
- 68 S. S. Zhang, J. Energy Chem., 2020, 41, 135-141.
- 69 W. Li, J. Reimers and J. Dahn, Solid State Ionics, 1993, 67, 123–130.
- 70 R. Jung, M. Metzger, F. Maglia, C. Stinner and H. A. Gasteiger, J. Electrochem. Soc., 2017, 164, A1361.
- 71 C. S. Yoon, D.-W. Jun, S.-T. Myung and Y.-K. Sun, ACS Energy Lett., 2017, 2, 1150–1155.
- 72 F. Kong, C. Liang, L. Wang, Y. Zheng, S. Perananthan, R. C. Longo, J. P. Ferraris, M. Kim and K. Cho, *Adv. Energy Mater.*, 2019, 9, 1802586.
- 73 Y. Wei, J. Zheng, S. Cui, X. Song, Y. Su, W. Deng, Z. Wu, X. Wang, W. Wang and M. Rao, *J. Am. Chem. Soc.*, 2015, 137, 8364–8367.
- 74 S. Cui, Y. Wei, T. Liu, W. Deng, Z. Hu, Y. Su, H. Li, M. Li,
  H. Guo, Y. Duan, W. Wang, M. Rao, J. Zheng, X. Wang and
  F. Pan, *Adv. Energy Mater.*, 2016, 6, 1501309.
- 75 K. Kang, Y. S. Meng, J. Breger, C. P. Grey and G. Ceder, *Science*, 2006, **311**, 977–980.

- 76 X. He, H. Sun, X. Ding and K. Zhao, J. Phys. Chem. C, 2021, 125, 10284–10294.
- 77 Y. Nomura, K. Yamamoto, Y. Yamagishi and E. Igaki, ACS Nano, 2021, 15, 19806–19814.
- 78 G. T. Park, D. R. Yoon, U. H. Kim, B. Namkoong, J. Lee, M. M. Wang, A. C. Lee, X. W. Gu, W. C. Chueh, C. S. Yoon and Y. K. Sun, *Energy Environ. Sci.*, 2021, 14, 6616–6626.
- 79 C. Hong, Q. Leng, J. Zhu, S. Zheng, H. He, Y. Li, R. Liu, J. Wan and Y. Yang, *J. Mater. Chem. A*, 2020, 8, 8540–8547.
- 80 J. C. Kim, D.-H. Seo, H. Chen and G. Ceder, *Adv. Energy Mater.*, 2015, 5, 1401916.
- 81 J. Lee, J. K. Papp, R. J. Clément, S. Sallis, D.-H. Kwon, T. Shi, W. Yang, B. D. McCloskey and G. Ceder, *Nat. Commun.*, 2017, 8, 981.
- 82 K. Ku, S. B. Son, J. Gim, J. Park, Y. J. Liang, A. Stark, E. Lee and J. Libera, *J. Mater. Chem. A*, 2021, **10**, 288– 295.
- 83 T. Y. Li, X. Z. Yuan, L. Zhang, D. T. Song, K. Y. Shi and C. Bock, *Electrochem. Energy Rev.*, 2020, 3, 43–80.
- 84 C. Y. Mao, R. E. Ruther, J. L. Li, Z. J. Du and I. Belharouak, *Electrochem. Commun.*, 2018, 97, 37–41.
- 85 E. Trevisanello, R. Ruess, G. Conforto, F. H. Richter and J. Janek, *Adv. Energy Mater.*, 2021, **11**, 2003400.
- 86 Z. Z. Yang, H. Charalambous, Y. L. Lin, S. E. Trask, L. Yu, J. G. Wen, A. Jansen, Y. F. Tsai, K. M. Wiaderek, Y. Ren and I. Bloom, *J. Power Sources*, 2022, **521**, 230961.
- 87 X. Y. Wu, B. H. Song, P. H. Chien, S. M. Everett, K. J. Zhao, J. Liu and Z. J. Du, *Adv. Sci.*, 2021, 8, 2102318.
- 88 Y. Yang, R. Xu, K. Zhang, S.-J. Lee, L. Mu, P. Liu, C. K. Waters, S. Spence, Z. Xu, C. Wei, D. J. Kautz, Q. Yuan, Y. Dong, Y.-S. Yu, X. Xiao, H.-K. Lee, P. Pianetta, P. Cloetens, J.-S. Lee, K. Zhao, F. Lin and Y. Liu, *Adv. Energy Mater.*, 2019, **9**, 1900674.
- 89 J. Li, N. Sharma, Z. Jiang, Y. Yang, F. Monaco, Z. Xu, D. Hou, D. Ratner, P. Pianetta, P. Cloetens, F. Lin, K. Zhao and Y. Liu, *Science*, 2022, 376, 517–521.
- 90 A. Mistry, F. L. E. Usseglio-Viretta, A. Colclasure, K. Smith and P. P. Mukherjee, *J. Electrochem. Soc.*, 2020, 167, 090542.
- 91 J. Park, H. Zhao, S. D. Kang, K. Lim, C.-C. Chen, Y.-S. Yu,
  R. D. Braatz, D. A. Shapiro, J. Hong, M. F. Toney,
  M. Z. Bazant and W. C. Chueh, *Nat. Mater.*, 2021, 20, 991–999.
- 92 W. E. Gent, Y. Li, S. Ahn, J. Lim, Y. Liu, A. M. Wise, C. B. Gopal, D. N. Mueller, R. Davis, J. N. Weker, J.-H. Park, S.-K. Doo and W. C. Chueh, *Adv. Mater.*, 2016, 28, 6631–6638.
- 93 Z. Xu, Z. Jiang, C. Kuai, R. Xu, C. Qin, Y. Zhang, M. M. Rahman, C. Wei, D. Nordlund, C.-J. Sun, X. Xiao, X.-W. Du, K. Zhao, P. Yan, Y. Liu and F. Lin, *Nat. Commun.*, 2020, **11**, 83.
- 94 S. Xia, L. Mu, Z. Xu, J. Wang, C. Wei, L. Liu, P. Pianetta, K. Zhao, X. Yu, F. Lin and Y. Liu, *Nano Energy*, 2018, 53, 753–762.
- 95 P. Yan, J. Zheng, M. Gu, J. Xiao, J.-G. Zhang and C.-M. Wang, *Nat. Commun.*, 2017, **8**, 14101.

- 96 R. Ruess, S. Schweidler, H. Hemmelmann, G. Conforto, A. Bielefeld, D. A. Weber, J. Sann, M. T. Elm and J. Janek, J. Electrochem. Soc., 2020, 167, 100532.
- 97 R. Xu and K. Zhao, *J. Mech. Phys. Solids*, 2018, **121**, 258–280.
- 98 Z. Jiang, J. Li, Y. Yang, L. Mu, C. Wei, X. Yu, P. Pianetta, K. Zhao, P. Cloetens, F. Lin and Y. Liu, *Nat. Commun.*, 2020, **11**, 2310.
- 99 H. Zhang, H. Liu, L. F. Piper, M. S. Whittingham and G. Zhou, *Chem. Rev.*, 2022, **122**, 5641–5681.
- 100 F. La Mantia, F. Rosciano, N. Tran and P. Novák, J. Electrochem. Soc., 2009, 156, A823.
- 101 F. La Mantia, F. Rosciano, N. Tran and P. Novák, *J. Appl. Electrochem.*, 2008, **38**, 893–896.
- 102 J. Wandt, A. T. Freiberg, A. Ogrodnik and H. A. Gasteiger, *Mater. Today*, 2018, **21**, 825–833.
- 103 A. O. Kondrakov, H. Geßwein, K. Galdina, L. de Biasi, V. Meded, E. O. Filatova, G. Schumacher, W. Wenzel, P. Hartmann, T. Brezesinski and J. Janek, *J. Phys. Chem. C*, 2017, **121**, 24381–24388.
- 104 R. A. House, U. Maitra, L. Jin, J. G. Lozano, J. W. Somerville, N. H. Rees, A. J. Naylor, L. C. Duda, F. Massel, A. V. Chadwick, S. Ramos, D. M. Pickup, D. E. McNally, X. Lu, T. Schmitt, M. R. Roberts and P. G. Bruce, *Chem. Mater.*, 2019, **31**, 3293–3300.
- 105 C. Wang, L. Han, R. Zhang, H. Cheng, L. Mu, K. Kisslinger, P. Zou, Y. Ren, P. Cao, F. Lin and H. L. Xin, *Matter*, 2021, 4, 2013–2026.
- 106 L. Mu, R. Lin, R. Xu, L. Han, S. Xia, D. Sokaras, J. D. Steiner, T.-C. Weng, D. Nordlund, M. M. Doeff, Y. Liu, K. Zhao, H. L. Xin and F. Lin, *Nano Lett.*, 2018, 18, 3241–3249.
- 107 C. Liang, F. Kong, R. C. Longo, S. Kc, J.-S. Kim, S. Jeon, S. Choi and K. Cho, *J. Phys. Chem. C*, 2016, **120**, 6383– 6393.
- 108 R. Jung, M. Metzger, F. Maglia, C. Stinner and H. A. Gasteiger, *J. Electrochem. Soc.*, 2017, **164**, A1361– A1377.
- 109 L. M. Morgan, M. M. Islam, H. Yang, K. O'Regan, A. N. Patel, A. Ghosh, E. Kendrick, M. Marinescu, G. J. Offer, B. J. Morgan, M. S. Islam, J. Edge and A. Walsh, ACS Energy Lett., 2022, 7, 108–122.
- 110 R. Sahore, D. C. O'Hanlon, A. Tornheim, C.-W. Lee, J. C. Garcia, H. Iddir, M. Balasubramanian and I. Bloom, *J. Electrochem. Soc.*, 2020, 167, 020513.
- 111 J. Wandt, A. Freiberg, R. Thomas, Y. Gorlin, A. Siebel, R. Jung, H. A. Gasteiger and M. Tromp, *J. Mater. Chem. A*, 2016, 4, 18300–18305.
- 112 J. A. Gilbert, I. A. Shkrob and D. P. Abraham, *J. Electrochem. Soc.*, 2017, **164**, A389–A399.
- 113 D.-S. Ko, J.-H. Park, S. Park, Y. N. Ham, S. J. Ahn, J.-H. Park, H. N. Han, E. Lee, W. S. Jeon and C. Jung, *Nano Energy*, 2019, **56**, 434–442.
- 114 R. Jung, F. Linsenmann, R. Thomas, J. Wandt, S. Solchenbach, F. Maglia, C. Stinner, M. Tromp and H. A. Gasteiger, *J. Electrochem. Soc.*, 2019, 166, A378–A389.

- 115 Y. Liu, Y. Zhu and Y. Cui, Nat. Energy, 2019, 4, 540-550.
- 116 S. S. Zhang, Energy Storage Mater., 2020, 24, 247-254.
- 117 F. Yang, Y. Liu, S. K. Martha, Z. Wu, J. C. Andrews, G. E. Ice, P. Pianetta and J. Nanda, *Nano Lett.*, 2014, 14, 4334-4341.
- 118 F. Lin, I. M. Markus, D. Nordlund, T.-C. Weng, M. D. Asta, H. L. Xin and M. M. Doeff, *Nat. Commun.*, 2014, **5**, 3529.
- 119 D. Mohanty, K. Dahlberg, D. M. King, L. A. David, A. S. Sefat, D. L. Wood, C. Daniel, S. Dhar, V. Mahajan, M. Lee and F. Albano, *Sci. Rep.*, 2016, 6, 26532.
- 120 J. Li, H. Li, W. Stone, R. Weber, S. Hy and J. R. Dahn, *J. Electrochem. Soc.*, 2017, **164**, A3529–A3537.
- 121 H. Li, J. Li, X. Ma and J. R. Dahn, *J. Electrochem. Soc.*, 2018, **165**, A1038–A1045.
- 122 J. Zhu and G. Chen, J. Mater. Chem. A, 2019, 7, 5463–5474.
- 123 G. Qian, Y. Zhang, L. Li, R. Zhang, J. Xu, Z. Cheng, S. Xie, H. Wang, Q. Rao, Y. He, Y. Shen, L. Chen, M. Tang and Z.-F. Ma, *Energy Storage Mater.*, 2020, 27, 140–149.
- 124 Y. Lu, T. Zhu, E. McShane, B. D. McCloskey and G. Chen, *Small*, 2022, **18**, 2105833.
- 125 J. C. Garcia, J. Bareño, G. Chen, J. R. Croy and H. Iddir, *Phys. Chem. Chem. Phys.*, 2020, **22**, 24490–24497.
- 126 S. Zhou, T. Mei, X. Wang and Y. Qian, *Nanoscale*, 2018, **10**, 17435–17455.
- 127 J. Tian, Y. Su, F. Wu, S. Xu, F. Chen, R. Chen, Q. Li, J. Li, F. Sun and S. Chen, ACS Appl. Mater. Interfaces, 2016, 8, 582–587.
- 128 A. van Bommel and J. R. Dahn, *Chem. Mater.*, 2009, 21, 1500–1503.
- 129 X. Zhang, W. J. Jiang, A. Mauger, Q. Lu, F. Gendron and C. M. Julien, *J. Power Sources*, 2010, **195**, 1292–1301.
- 130 N. Zhang, J. Li, H. Li, A. Liu, Q. Huang, L. Ma, Y. Li and J. R. Dahn, *Chem. Mater.*, 2018, **30**, 8852–8860.
- 131 Y. Su, G. Chen, L. Chen, Y. Lu, Q. Zhang, Z. Lv, C. Li, L. Li, N. Liu, G. Tan, L. Bao, S. Chen and F. Wu, ACS Appl. Mater. Interfaces, 2019, 11, 36697–36704.
- 132 Y. Lai, Z. Li, W. Zhao, X. Cheng, S. Xu, X. Yu and Y. Liu, *Nano Res.*, 2020, **13**, 3347–3357.
- 133 K. Yuan, N. Li, R. Ning, C. Shen, N. Hu, M. Bai, K. Zhang,
  Z. Tian, L. Shao, Z. Hu, X. Xu, T. Yu and K. Xie, *Nano Energy*, 2020, 78, 105239.
- 134 T. Weigel, F. Schipper, E. M. Erickson, F. A. Susai,B. Markovsky and D. Aurbach, *ACS Energy Lett.*, 2019, 4, 508–516.
- 135 F. Schipper, M. Dixit, D. Kovacheva, M. Talianker,
  O. Haik, J. Grinblat, E. M. Erickson, C. Ghanty,
  D. T. Major, B. Markovsky and D. Aurbach, *J. Mater. Chem. A*, 2016, 4, 16073–16084.
- 136 K.-J. Park, H.-G. Jung, L.-Y. Kuo, P. Kaghazchi, C. S. Yoon and Y.-K. Sun, *Adv. Energy Mater.*, 2018, **8**, 1801202.
- 137 J. C. Kim, D.-H. Kwon, J. H. Yang, H. Kim, S.-H. Bo,
  L. Wu, H. Kim, D.-H. Seo, T. Shi, J. Wang, Y. Zhu and
  G. Ceder, *Adv. Energy Mater.*, 2020, **10**, 2001151.
- 138 P. Vassilaras, S. T. Dacek, H. Kim, T. T. Fister, S. Kim, G. Ceder and J. C. Kim, *J. Electrochem. Soc.*, 2017, 164, A3484.

- 139 J. C. Kim, D.-H. Seo and G. Ceder, *Energy Environ. Sci.*, 2015, **8**, 1790–1798.
- 140 A. Gomez-Martin, F. Reissig, L. Frankenstein, M. Heidbüchel, M. Winter, T. Placke and R. Schmuch, *Adv. Energy Mater.*, 2022, 12, 2103045.
- 141 F. Xin, H. Zhou, X. Chen, M. Zuba, N. Chernova, G. Zhou and M. S. Whittingham, ACS Appl. Mater. Interfaces, 2019, 11, 34889–34894.
- 142 P. Mohan and G. P. Kalaignan, J. Electroceram., 2013, 31, 210–217.
- 143 P. Mohan, K. A. Kumar, G. P. Kalaignan and V. S. Muralidharan, *J. Solid State Electrochem.*, 2012, **16**, 3695–3702.
- 144 M. Eilers-Rethwisch, M. Winter and F. M. Schappacher, *J. Power Sources*, 2018, **387**, 101–107.
- 145 L. Wu, X. Tang, X. Chen, Z. Rong, W. Dang, Y. Wang, X. Li, L. Huang and Y. Zhang, *J. Power Sources*, 2020, 445, 227337.
- 146 F. Schipper, H. Bouzaglo, M. Dixit, E. M. Erickson, T. Weigel, M. Talianker, J. Grinblat, L. Burstein, M. Schmidt, J. Lampert, C. Erk, B. Markovsky, D. T. Major and D. Aurbach, *Adv. Energy Mater.*, 2018, 8, 1701682.
- 147 G. Li, Z. Huang, Z. Zuo, Z. Zhang and H. Zhou, *J. Power Sources*, 2015, **281**, 69–76.
- 148 H. H. Sun, A. Dolocan, J. A. Weeks, A. Heller and C. B. Mullins, *ACS Nano*, 2020, 14, 17142–17150.
- 149 G.-T. Park, B. Namkoong, S.-B. Kim, J. Liu, C. S. Yoon and Y.-K. Sun, *Nat. Energy*, 2022, 7, 946–954.
- 150 U.-H. Kim, G.-T. Park, B.-K. Son, G. W. Nam, J. Liu, L.-Y. Kuo, P. Kaghazchi, C. S. Yoon and Y.-K. Sun, *Nat. Energy*, 2020, 5, 860–869.
- 151 H.-H. Ryu, K.-J. Park, D. R. Yoon, A. Aishova, C. S. Yoon and Y.-K. Sun, *Adv. Energy Mater.*, 2019, **9**, 1902698.
- 152 H.-H. Ryu, N.-Y. Park, J. H. Seo, Y.-S. Yu, M. Sharma, R. Mücke, P. Kaghazchi, C. S. Yoon and Y.-K. Sun, *Mater. Today*, 2020, **36**, 73–82.
- 153 U.-H. Kim, J.-H. Park, A. Aishova, R. M. Ribas, R. S. Monteiro, K. J. Griffith, C. S. Yoon and Y.-K. Sun, *Adv. Energy Mater.*, 2021, **11**, 2100884.
- 154 J. U. Choi, N. Voronina, Y.-K. Sun and S.-T. Myung, *Adv. Energy Mater.*, 2020, **10**, 2002027.
- 155 S. J. Shi, J. P. Tu, Y. Y. Tang, Y. Q. Zhang, X. Y. Liu, X. L. Wang and C. D. Gu, *J. Power Sources*, 2013, 225, 338– 346.
- 156 Z.-F. Tang, R. Wu, P.-F. Huang, Q.-S. Wang and C.-H. Chen, J. Alloys Compd., 2017, 693, 1157– 1163.
- 157 H. Yang, H.-H. Wu, M. Ge, L. Li, Y. Yuan, Q. Yao, J. Chen, L. Xia, J. Zheng, Z. Chen, J. Duan, K. Kisslinger, X. C. Zeng, W.-K. Lee, Q. Zhang and J. Lu, *Adv. Funct. Mater.*, 2019, **29**, 1808825.
- 158 Y. Chen, Y. Zhang, B. Chen, Z. Wang and C. Lu, *J. Power Sources*, 2014, **256**, 20–27.
- 159 S. Guo, B. Yuan, H. Zhao, D. Hua, Y. Shen, C. Sun, T. Chen, W. Sun, J. Wu, B. Zheng, W. Zhang, S. Li and F. Huo, *Nano Energy*, 2019, 58, 673–679.

- 160 M. R. Laskar, D. H. K. Jackson, S. Xu, R. J. Hamers, D. Morgan and T. F. Kuech, ACS Appl. Mater. Interfaces, 2017, 9, 11231–11239.
- 161 S. Li, X. Fu, J. Zhou, Y. Han, P. Qi, X. Gao, X. Feng and B. Wang, *J. Mater. Chem. A*, 2016, 4, 5823–5827.
- 162 Y. Zhang, C. S. Kim, H. W. Song, S.-J. Chang, H. Kim, J. Park, S. Hu, K. Zhao and S. Lee, *Energy Storage Mater.*, 2022, **48**, 1–11.
- 163 M. Yoon, Y. Dong, J. Hwang, J. Sung, H. Cha, K. Ahn, Y. Huang, S. J. Kang, J. Li and J. Cho, *Nat. Energy*, 2021, 6, 362–371.
- 164 X. Li, J. Liu, M. N. Banis, A. Lushington, R. Li, M. Cai and X. Sun, *Energy Environ. Sci.*, 2014, 7, 768–778.
- 165 G.-L. Xu, Q. Liu, K. K. S. Lau, Y. Liu, X. Liu, H. Gao, X. Zhou, M. Zhuang, Y. Ren, J. Li, M. Shao, M. Ouyang, F. Pan, Z. Chen, K. Amine and G. Chen, *Nat. Energy*, 2019, 4, 484–494.
- 166 S. H. Lee, S. Yoon, E.-H. Hwang, Y.-G. Kwon, Y.-G. Lee and K. Y. Cho, *J. Power Sources*, 2018, 378, 112–118.
- 167 P. Dong, D. Wang, Y. Yao, X. Li, Y. Zhang, J. Ru and T. Ren, *J. Power Sources*, 2017, 344, 111–118.
- 168 Y. Liu, D. Sun, J. Zhou, Y. Qin, D. Wang and B. Guo, ACS Appl. Mater. Interfaces, 2018, 10, 11305–11310.
- 169 C. Wang, L. Yu, W. Fan, R. Liu, J. Liu, L. Ouyang, L. Yang and M. Zhu, *J. Alloys Compd.*, 2018, 755, 1–9.
- 170 B. Deng, H. Wang, X. Li, Y. Ding, D. Sun, J. Ma, M. Chen, T. Chen, F. Gao, M. Qu and G. Peng, *Energy Technol.*, 2019, 7, 1800981.
- 171 B. Deng, H. Wang, W. Ge, X. Li, X. Yan, T. Chen, M. Qu and G. Peng, *Electrochim. Acta*, 2017, **236**, 61–71.

- 172 S. Wang, S. Chen, W. Gao, L. Liu and S. Zhang, *J. Power Sources*, 2019, **423**, 90–97.
- 173 C.-G. Shi, C.-H. Shen, X.-X. Peng, C.-X. Luo, L.-F. Shen, W.-J. Sheng, J.-J. Fan, Q. Wang, S.-J. Zhang, B.-B. Xu, J.-J. Xian, Y.-M. Wei, L. Huang, J.-T. Li and S.-G. Sun, *Nano Energy*, 2019, **65**, 104084.
- 174 Q. Zheng, L. Xing, X. Yang, X. Li, C. Ye, K. Wang, Q. Huang and W. Li, ACS Appl. Mater. Interfaces, 2018, 10, 16843–16851.
- 175 Y. Zou, K. Zhou, G. Liu, N. Xu, X. Zhang, Y. Yang, J. Zhang and J. Zheng, *ACS Appl. Mater. Interfaces*, 2021, **13**, 16427– 16436.
- 176 T. Deng, X. Fan, L. Cao, J. Chen, S. Hou, X. Ji, L. Chen, S. Li, X. Zhou, E. Hu, D. Su, X.-Q. Yang and C. Wang, *Joule*, 2019, 3, 2550–2564.
- 177 X. Y. Wang, S. Y. Li, W. D. Zhang, D. Wang, Z. Y. Shen, J. P. Zheng, H. L. L. Zhuang, Y. He and Y. Y. Lu, *Nano Energy*, 2021, **89**, 106353.
- 178 P. Bai, X. Ji, J. Zhang, W. Zhang, S. Hou, H. Su, M. Li, T. Deng, L. Cao, S. Liu, X. He, Y. Xu and C. Wang, *Angew. Chem., Int. Ed.*, 2022, **61**, e202202731.
- 179 W. D. Richards, L. J. Miara, Y. Wang, J. C. Kim and G. Ceder, *Chem. Mater.*, 2016, **28**, 266–273.
- 180 K. Kim, H. Ma, S. Park and N.-S. Choi, ACS Energy Lett., 2020, 5, 1537–1553.
- 181 M. Wood, J. Li, Z. Du, C. Daniel, A. R. Dunlop, B. J. Polzin, A. N. Jansen, G. K. Krumdick and D. L. Wood, *J. Power Sources*, 2021, 515, 230429.
- 182 J. Hu, B. Wu, X. Cao, Y. Bi, S. Chae, C. Niu, B. Xiao, J. Tao, J. Zhang and J. Xiao, *J. Power Sources*, 2020, 454, 227966.