Manganese-based layered oxides for electrochemical energy storage: a review of degradation mechanisms and engineering strategies at the atomic level

Shuo Sun,†a Jin Li,‡a Cuixia Xu,b Teng Zhai,*a and Hui Xia*a

The ever-increasing demand for high-energy-density electrochemical energy storage has been driving research on the electrochemical degradation mechanisms of high-energy cathodes, among which manganese-based layered oxide (MLO) cathodes have attracted high attention thanks to their low cost and eco-friendliness. More importantly, MLO materials with large and tunable interlayer spacing are ideal candidates for the insertion of (monovalent, divalent, trivalent) alkaline ions, such as Li+, Na+, K+, Zn2+, Mg2+, and Al3+, enabling impressive electrochemical performance. Nevertheless, the local MnO6 octahedron distortion induced by the Jahn–Teller (J–T) effect can lead to irreversible phase transformation, dissolution/disproportionation reactions, interfacial degradation arising from Mn2+, and crack formation, which significantly impact the electrochemical stability of MLO materials. Hence, in this review, we discuss the various degradation processes caused by J–T distortion in MLO cathodes at the atomic level. Advances in the atomic-level structure and property optimizations of MLO materials and in-depth structure–function–property correlations are also systematically reviewed. Finally, we provide our perspectives on the future development of MLO materials. The integration of high-performance MLO cathodes in energy storage devices has great potential to address growing global energy demands.

1. Introduction

Triggered by the overuse of traditional nonrenewable energy supplies, energy crisis and climate change have become serious global issues.1,2 In this regard, it is of particular importance to transit from fossil fuels towards clean and sustainable energy sources (especially wind and solar energy). Thus, developing low-cost, environmentally friendly, and highly efficient energy storage techniques is crucial for storing and utilizing these

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Intermittent sustainable energies for the following utilization, which can address the energy crisis, reduce CO₂ emissions, and contribute to achieving a more sustainable society. Particular interest has been directed towards electrochemical energy storage systems, such as supercapacitors, Li-ion batteries, Na-ion batteries, Zn-ion batteries, and Al-based batteries, which are regarded as the most promising technologies to meet the skyrocketing energy consumption demands of consumer.

Fig. 1  (a) Schematic illustrations of the balance between the electrochemical performance and sustainability in MLO materials. (b) MLO materials with large and tunable layered spacing enabling the insertion of various valence ions. (c) Volumetric capacity and gravimetric capacity for the various metal anodes.

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electronics, electric vehicles, and large-scale energy storage devices.\textsuperscript{7–10} Electrochemical energy storage involves the conversion of chemical energy into electrical energy via ion shift between the cathode side and the anode side. Therefore, designing high-performance electrodes is crucial for advancing energy storage solutions. In the endeavors to improve energy storage technologies, the high-performance cathode material is the main bottleneck.

Among the various cathode candidates, transition metal (TM) layered oxides provide many advantages thanks to their high capacities, simple structures, and ease of synthesis.\textsuperscript{11–13} Although different TMs, including Ni, Co, and Mn, have been utilized in layered cathode materials, only Mn-based layered cathode materials realize the ideal balance between sustainability and performance (Fig. 1a). Mn has the highest crust abundance among Ni, Co, and Mn (84 ppm, 25 ppm, and 950 ppm, respectively)\textsuperscript{14} and thus the lowest costs. Additionally, elemental Mn is nontoxic and eco-friendly, and the risks of metal pollution with Mn use are lower than those of Ni and Co.\textsuperscript{8,15,16} In terms of performance, owing to the high redox activity of Mn\textsuperscript{2+/3+}/Mn\textsuperscript{4+} and the corresponding high redox potential, as well as the relatively low atomic weight, Mn-based layered oxides (MLOs) deliver much higher energy densities than other cathode materials.\textsuperscript{6,14} Meanwhile, unlike Ni-based layered oxides, MLOs possess high thermal stability and are thus safer.\textsuperscript{17} Furthermore, the large and tunable interlayer spacing of MLO materials guarantees fast ionic diffusion kinetics, providing an ideal structure for (monovalent, divalent, trivalent) alkaline ions (e.g., Li\textsuperscript{+}, Na\textsuperscript{+}, K\textsuperscript{+}, Zn\textsuperscript{2+}, Mg\textsuperscript{2+}, Ca\textsuperscript{2+}, Al\textsuperscript{3+}) insertion, representing a star family as cathode materials in various electrochemical energy storage devices for a long period (Fig. 1b).\textsuperscript{18–20} Apart from the commonly used metal anodes, such as Li, Na, and K, metal anode materials like Zn, Mg, and Al, which boast the merits of earth abundance, safety, good machinability, and ultrahigh volumetric energy density, are also attractive candidates for energy storage. Therefore, the development of MLO cathodes is beneficial to develop advanced energy storage devices and satisfy the future demands of large-scale energy storage.

The typical MLO materials are built by alternately layering TM–oxygen and alkaline metal ions. The TM is coordinated in an octahedral geometry by oxygen, and the alkaline metal ions are inserted in prismatic (P), tetrahedral (T), and octahedral (O) configurations, corresponding to the P2/P3, T2, and O2/O3 phases, respectively (Fig. 2a), in which the numbers 2 and 3 represent the repeat period of the TM–O layers.\textsuperscript{21–22} In addition,
a special kind of MLO material, called birnessite, comprises a mixed-valent oxide surrounded by six oxygen atoms to constitute a basic unit $[\text{MnO}_6]$ octahedron with cations of various valences ($\text{e.g.}$, $\text{K}^+$, $\text{Na}^+$, $\text{Zn}^{2+}$) and crystal water filling the interlayers.\textsuperscript{23,24} Birnessite, with 2D interlayer space for ionic diffusion, is a hydrous Mn oxide. Layered oxides were first applied in energy storage by J. B. Goodenough in 1980;\textsuperscript{25} substantial efforts have since advanced the development of high-performance MLO materials for various energy storage purposes (Fig. 2b).\textsuperscript{26–30} Although remarkable progress has been made, the applications of MLO materials are still limited by irreversible phase transition, Mn dissolution/disproportionation reactions, interfacial degradation arising from Mn$^{2+}$, and crack formation (Fig. 3), which lead to rapid electrochemical degradation. It is worth noting that the root of these detrimental issues is local MnO$_6$ octahedron distortion caused by the Jahn–Teller (J–T) effect. The J–T theorem was proposed by H.A. Jahn and E. Teller in 1937;\textsuperscript{32} due to this effect, a nonlinear molecular system suffers from instability in the degenerate electron state. Mn has several oxidation states (+2, +3, +4, +7), and electrochemical redox reactions usually happen between the +4 and +2 oxidation states. The ligand field of Mn$^{3+}$ is shown at the center of Fig. 3. The Mn ions are coordinated by six oxide ligands and are symmetrically arranged along the Cartesian axes. The electric field of the octahedrally arranged ligands separates the d orbitals of Mn into doubly degenerate $e_g$ ($d_z^2$ and $d_x^2−y^2$) and triply degenerate $t_{2g}$ ($d_{xy}$, $d_{xz}$, and $d_{yz}$) sets. The $e_g$ orbital set of Mn$^{4+}$ is empty, thus possessing a cubically symmetrical electron density. This feature confers stability to Mn$^{4+}$-derived layered materials. However, when Mn$^{3+}$ is coordinated with oxygen, like in LiMnO$_2$, the 3d$^4$ configuration of Mn$^{3+}$ exhibits an asymmetrical electron density distribution and a high-spin state due to the single-electron occupancy in the $e_g$ level.\textsuperscript{33–35} In this case, the two axial Mn–O bonds are elongated and the four equatorial bonds are shortened to relax this unfavorable state, i.e., the J–T distortion, transforming MnO$_6$ from the cubic configuration into the tetragonal configuration. As such, this structural evolution at the atomic level detrimentally impacts the physicochemical properties of MLO materials.

In attempts to prevent J–T distortion, various strategies have been developed to optimize structural stability at the atomic level during the charging/discharging process.\textsuperscript{7,14,36–40} These strategies can be divided into four categories (Fig. 3): (i) elemental doping, (ii) interlayered structure manipulation, (iii) vacancy introduction, and (iv) novel structural design. Although previous reviews have discussed Mn-based and/or layered cathode materials, the relationships between the structural evolution of MLO materials at the atomic level and capacity fade during the ion deintercalation/intercalation processes have not been systematically reviewed.\textsuperscript{12,41–43} In particular, a comprehensive review of the atomic-level mechanisms and strategies that remedy the electrochemical degradation of MLO cathodes in
various valence ion-based electrochemical energy storage devices is needed.

In this review, we discuss the relationships between the intrinsic structure, electrochemical behaviors, and degradation mechanisms from the perspective of the atomic level and summarize the effective strategies for structural degradation during ion deintercalation/intercalation. We begin by discussing the different electrochemical behaviors resulting from alkaline ion insertion into MLO cathodes, from monovalent ions to trivalent ions. Moreover, we describe the corresponding electrochemical degradation caused by irreversible phase transition, Mn dissolution/disproportionation reactions, interfacial degradation arising from Mn$^{3+}$, and crack formation in various energy storage devices, as well as the relationships between these degradation mechanisms and J-T distortion. Finally, we outline the recent achievements in overcoming the aforementioned bottlenecks and provide perspectives on the future development of high-performance MLO materials.

2. Fundamental understanding of alkaline ion insertion in MLO materials

Energy storage devices based on multivalent metal ions with high crust abundances, such as Ca, Mg, Zn, and Al, have the potential to satisfy the future demands of large-scale energy storage. Therefore, as an alternative to the development of conventional monovalent metal energy storage (e.g., Li batteries, Na batteries), divalent and trivalent ion intercalation is of particular importance. Nevertheless, unlike monovalent ions, the intercalation of divalent/trivalent ions into electrode materials is difficult due to the strong electrostatic interactions between the electrode material atoms and the highly charged intercalation ions. Diffusion channels that allow ion mobility are crucial to enable multivalent ion insertion. Notably, the large and tunable interlayer spacing of MLOs can accommodate structural deformations, enabling the reversible insertion of both monovalent ions and multivalent ions into MLO materials. Layered MLO structures possess 2D diffusion channels, providing a favorable structure for ionic diffusion and guaranteeing fast kinetics. The MLO materials with different structures used in various energy storage systems are summarized in Table 1, in which MLO materials exhibit superior performance due to their large and tunable interlayered spacing. In the next sections, we discuss the insertion of monovalent, divalent, and trivalent ions in MLO cathodes.

<table>
<thead>
<tr>
<th>Energy storage type</th>
<th>Material</th>
<th>Crystal structure</th>
<th>Interlayer space (Å)</th>
<th>Electrolyte type</th>
<th>Rate performance</th>
<th>Ref.</th>
</tr>
</thead>
<tbody>
<tr>
<td>Supercapacitor</td>
<td>δ-MnO$_2$</td>
<td>Birnessite</td>
<td>7.2</td>
<td>Aqueous</td>
<td>250 F g$^{-1}$ at 0.2 A g$^{-1}$</td>
<td>19</td>
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<td></td>
<td>LiMnO$_2$</td>
<td>O3</td>
<td>5.4</td>
<td>Organic</td>
<td>286 mA h g$^{-1}$ at 1C</td>
<td>44</td>
</tr>
<tr>
<td></td>
<td>Li$<em>{2/3}$[Ni$</em>{1/3}$Mn$_{2/3}$]O$_2$</td>
<td>T2</td>
<td>5.0</td>
<td>Organic</td>
<td>183 mA h g$^{-1}$ at 12C</td>
<td>45</td>
</tr>
<tr>
<td></td>
<td>Li$_2$MnO$_3$</td>
<td>O2</td>
<td>4.7</td>
<td>Organic</td>
<td>180 mA h g$^{-1}$ at 0.1C</td>
<td>46</td>
</tr>
<tr>
<td></td>
<td>Na$_2$MnO$<em>3$-$δ$-(OH)$</em>{2y}$ Birnessite</td>
<td>7.2</td>
<td>Aqueous</td>
<td>211 mA h g$^{-1}$ at 0.2C</td>
<td>47</td>
<td></td>
</tr>
<tr>
<td></td>
<td>Na$_3$MnO$_2$</td>
<td>O3 ($x = 0.9–1.0$)</td>
<td>5.7</td>
<td>Organic</td>
<td>156 mA h g$^{-1}$ at 50C</td>
<td>48</td>
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<tr>
<td></td>
<td>K$_x$γ-MnO$_2$</td>
<td>Birnessite</td>
<td>7.2</td>
<td>Organic</td>
<td>185 mA h g$^{-1}$ at 0.1C</td>
<td>49</td>
</tr>
<tr>
<td></td>
<td>Zn$_x$MnO$_2$</td>
<td>Birnessite</td>
<td>9.2</td>
<td>Aqueous</td>
<td>77 mA h g$^{-1}$ at 1 A g$^{-1}$</td>
<td>50</td>
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<td></td>
<td>Ca$_3$MnO$_2$</td>
<td>Birnessite</td>
<td>6.9</td>
<td>Aqueous</td>
<td>242 mA h g$^{-1}$ at 3 A g$^{-1}$</td>
<td>51</td>
</tr>
<tr>
<td></td>
<td>Mg$_3$MnO$_2$</td>
<td>Birnessite</td>
<td>6.7</td>
<td>Aqueous</td>
<td>89 mA h g$^{-1}$ at 2 A g$^{-1}$</td>
<td>52</td>
</tr>
<tr>
<td></td>
<td>Mg$_3$MnO$_2$</td>
<td>Birnessite</td>
<td>9.7</td>
<td>Aqueous</td>
<td>55 mA h g$^{-1}$ at 8 A g$^{-1}$</td>
<td>53</td>
</tr>
<tr>
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<td>Al$_2$O$_3$</td>
<td>Birnessite</td>
<td>7.0</td>
<td>Aqueous</td>
<td>59 mA h g$^{-1}$ at 0.1 A g$^{-1}$</td>
<td>54</td>
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pseudocapacitance, Mn oxidation state changes, local bonding, and structure. In the charge/discharge process, the monovalent cations in the electrolytes, such as Na⁺ and K⁺, can be intercalated/deintercalated into the layered space of δ-MnO₂. However, unlike the intercalation of other cations, the integration and confinement of structural water within the interlayers of δ-MnO₂ leads to electrochemical behaviors between faradaic and traditional capacitive non-faradaic electrical double layer intercalation (Fig. 4a). These confined water molecules in the interlayers weaken the interactions between the intercalated ions and MnO₂ layers, resulting in minimal structural change and potential dependence (manifested as rectangular CV curves). Boyd et al. performed electrochemical quartz crystal microbalance experiments (EQCM) and discovered that the alkali cation and H₂O intercalation into δ-MnO₂ simultaneously proceeded during the electrochemical process. In sharp contrast, the CV curves of δ-MnO₂ materials with no water showed well-defined peaks originating from the structural transitions. When H₂O molecules entered the interlayers, the CV curves became featureless, and the corresponding X-ray diffraction pattern showed only a small change in the interlayer spacing during intercalation/deintercalation.

As for the P/O/T phases, such as layered LiₓMnO₂ and NaₓMnO₂, both consist of edge-sharing MnO₆ octahedra separated by an interlayer of Li⁺ or Na⁺. Generally, most AMO₂ (A = Li, Na; M = transition metal) phases have α-NaFeO₂-type structures. However, when M is Mn³⁺, the crystal structure is monoclinic (space group C2/m), not orthorhombic, due to the
strong J-T effect of high-spin Mn$^{3+}$, which distorts the local site symmetry near Mn$^{3+}$ from a regular octahedron and changes the corresponding crystal structure.\textsuperscript{55-67} The practical use of layered LiMnO$_2$ in batteries is greatly hindered by severe structural degradation and rapid capacity fading. The small difference in size between Mn$^{3+}$ and Li$^+$ makes it difficult to form the layered LiMnO$_2$ structure; Mn migration into the Li layer is often observed.\textsuperscript{68-69} Thus, an irreversible transition from the layered structure to the spinel structure usually occurs during the first deintercalation. Specifically, the formation of Li vacancies during oxidation results in covalent MnO$_2$ slab formation and an empty interlayer space with face-to-face O layers. Correspondingly, the Mn ions tend to migrate to the Li layers to homogenize the charge distribution.\textsuperscript{70} To realize this transfer process, the Mn ions must cross a tetrahedral interstitial site. Low-spin Co$^{4+}$ (d$_5$), low-spin Co$^{3+}$ (d$_6$), low-spin Ni$^{3+}$ (d$_6$), and low-spin Ni$^{4+}$ (d$_6$) have very strong stability in octahedral sites. However, high-spin Mn$^{3+}$ ions are prone to disproportionate into Mn$^{2+}$ and Mn$^{4+}$.\textsuperscript{71,72} Mn$^{3+}$ (d$_5$) is much less stable in octahedral sites than in tetrahedral sites, promoting movement and thus fast capacity fade. As such, the crystal field stability determines whether the TM ions transfer.\textsuperscript{73} Generally, stable redox reactions of monoclinic layered LiMnO$_2$ can be observed at ~3.9 V in the cathodic scan and 3.8 V in the anodic scan (Fig. 4b).\textsuperscript{46,74-77} However, the sizes of Na$^+$ and Mn ions differ, allowing NaNMnO$_2$ to form the O3-type layered structure. Meanwhile, there is a great increase in the covalency of the Mn-O bond due to the Na$^+$ and Mn$^{4+}$ ions forming the Mn$^{4+}$ compounds Li$_4$Mn$_5$O$_{12}$ and MnO$_2$. The phase diagrams in Fig. 4e facilitate the understanding of the electrochemical reaction mechanisms of Li-rich MLO materials. However, the detected voltage curves do not follow these three-phase reactions (Fig. 4f), which is mainly attributed to the slow kinetics of these reactions resulting from the long-range diffusion of Mn or O. Compared with Li, the relatively slow migration of Mn leads to the formation of a metastable Li$_{1.2}$Mn$_{0.54}$Co$_{0.13}$Ni$_{0.13}$O$_2$ compound. The oxygen-redox hypothesis also demonstrates the formation of this metastable phase where the oxygen anion valence states are $-1,56,57$.

2.2 Divalent and trivalent ion insertion in MLO materials

Compared with monovalent cations, inserting divalent and trivalent cations in MLO materials can lead to higher energy density. The layered hydrated TM oxide birnessite has attracted tremendous attention because it contains crystal water, which can not only change the interlayered spacing but also weaken the polarizing nature and accelerate the diffusion of divalent and trivalent cations. As the interlayers of birnessite contain multivalent ions and crystal water, their spacing can be expanded to nearly 1 nm, allowing different layered structures to be obtained. Herein, the insertion of divalent and trivalent cations in MLO materials is demonstrated in the context of typical Zn-ion and Al-ion battery applications, respectively.

Zn-ion batteries possess high gravimetric and volumetric capacity (820 mA h g$^{-1}$ and 5854 mA h cm$^{-3}$). In addition, the excellent reversibility of Zn plating/stripping reactions, safety, and low toxicity make Zn-ion batteries a promising energy storage device. The insertion of Zn$^{2+}$ into birnessite forms the layered $\delta$-Zn$_2$MnO$_4$. Alfaruqi et al.\textsuperscript{27} demonstrated the high capacity of birnessite-based Zn-ion batteries (>250 mA h g$^{-1}$). Despite the high capacity of layered $\delta$-Zn$_2$MnO$_4$ compared with tunnel MnO$_2$, $\delta$-Zn$_2$MnO$_4$ still has low capacity retention during electrochemical processes. Recent research demonstrated that the capacity decay is due to the phase transition from the layered structure to the spinel structure and the dissolution of Mn ions in electrolytes.\textsuperscript{88} Wang et al.\textsuperscript{89} prepared a mesoporous layered MnO$_2$ nanosphere with H$_2$O and Zn$^{2+}$ in the interlayers.
AlCl3 were used, the intercalation of Al3+ was prohibited.

J-T distortion, which elongates the two axial Mn–O bonds, induces MnO2 to transform from the cubic form into the tetragonal form. This atomic structure evolution influences the physicochemical properties of MLO materials and impacts their electrochemical performance. Generally, the degradation issues caused by J-T distortion in MLO cathodes can be divided into (i) irreversible phase transition, (ii) dissolution/disproportionation reaction, (iii) interfacial degradation arising from Mn2+, and (iv) crack formation. As a result, a comprehensive understanding of these problems is necessary for developing MLO-based energy storage devices with high structural stability and prolonged life spans.

3. Degradation mechanisms

3.1 Irreversible phase transformation

3.1.1 Layered-to-tunnel/spinel phase transition. Birnessite with layered packing composed of edge-sharing [MnO6] octahedra can convert to other types of tunnel structures during the electrochemical process. At the beginning of phase transition, crystal water migrates out of the interlayer spaces. The layered structure collapses during the oxidation of Mn2+ to Mn3+, or the disproportionation of Mn3+ ions to Mn2+ and Mn4+. As a result, the [MnO6] octahedral layers are distorted and rearranged, which reduces lattice stress and causes the layer-to-tunnel phase transition (Fig. 5a). Subsequently, the destroyed layers may interlink again to form a new phase of α-MnO2, or other tunnel phases may be formed in an analogous process. The irreversible phase transition from the layered structure to the spinel structure is considered the main reason for the severe electrochemical degradation of MLO cathodes. Specifically, the layered structures of MLO materials prefer to transform into spinel-like structures when alkali ions are removed from the interlayers. This phase transition originates from the rearrangement of Mn and alkali ions and is thus generally irreversible, causing rapid electrochemical performance decay. In addition, the high mobility of Mn2+ in LiMnO2 accelerates the transformation from the layered structure into the spinel structure during the first charge. Mn ions can move into tetrahedral sites as vacancies are generated in the Li layer. Meanwhile, the Li/Mn dumbbell configuration is generated (Fig. 5b) when the Li ions migrate to a tetrahedral site in an adjacent Li layer, which endows the Mn in tetrahedral sites with high stability and serves as a nucleus for the layered-to-spinel phase transition.

Gu et al. reported that newly formed spinel domains show random orientation within the same particle during cycling. The extraction of Li+ ions together with the loss of oxygen from Li2MnO3 induces the displacement of TM cations. This change further destroys the lattice and creates numerous nanometer-scale spinel particles, leading to rapid capacity and voltage decay. Correspondingly, Li+ ions cannot intercalate in the transformed structure again, resulting in low initial coulombic efficiency. For example, in the first oxidation process, atomic columns in the Li layers can be observed in both the surface and bulk regions. This feature implies the migration of TM ions to...
the Li layer (Fig. 5c). Meanwhile, the development of pores and cracks among the layered nanoparticles results in lattice degradation and vacancy formation upon the extraction of Li$^+$ ions, suggesting that the layered-to-spinel phase transition is a nucleation and growth mechanism. Recently, Radin et al.$^{57}$ modeled the irreversible phase transition. As shown in Fig. 5d, a typical local configuration forms upon Li$_2$MnO$_3$ charging. Mn$^{4+}$ remains in the [MnO$_6$] octahedral sites, while Mn$^{7+}$ tends to migrate to the tetrahedral sites in Li$_{1/2}$MnO$_3$. To form Li$_{1/2}$MnO$_3$, half of the Mn ions must migrate, with one quarter moving downward and one quarter moving upward (Fig. 5e). This behavior finally produces an unstable lattice structure.

### 3.1.2 P/O phase transition

Unlike Li$_2$MnO$_3$, Na$_x$MnO$_2$ does not easily transform from the layered phase into spinel structures.$^{48}$ This difference in structural evolution is attributed to the fact that Na$^+$ does not favorably occupy the tetrahedral site in NaMnO$_2$ because it is larger than Li$^+$. Therefore, with Na$^+$, the transformation from layered to spinel structures is unfavorable, and the layered Na$_x$MnO$_2$ is highly metastable. Nevertheless, a series of phase transformations can occur in the P2-type and O3-type Na$_x$MnO$_2$, changing the stacking sequences during the electrochemical processes. If some Na$^+$ ions are removed, P2-type Na$_x$MnO$_2$ transforms to O2-type Na$_x$MnO$_2$, deriving from the gliding (π/3 rotation) of some MnO$_6$ octahedra sheets. This behavior contracts the crystal structure and decreases the interlayer distance (Fig. 5f).$^{90}$

As the large Na$^+$ ions stabilize the prismatic sites in the P2 phase, MnO$_2$ slabs tend to form octahedral sites, enabling the phase change from P2 to O2.$^{90}$ For the O3-type phase, when a sufficient amount of Na$^+$ is removed from the interlayers, MnO$_2$ sheet gliding induces the formation of broad prismatic sites. The stacking of oxygen transits to AB, BC, CA, known as the P3 phase. In other words, the extraction of Na$^+$ from both P2 and O3 phases generally causes the gliding of MnO$_2$ sheets, promoting the phase transition (layer-to-layer). Na$_x$MnO$_2$ has a more complicated phase than its Li analogs due to the large size of Na$^+$ coupled with the ordering arrangement of Na$^+$ and vacancies. Because the MnO$_2$ sheet gliding combined with the change in oxygen stacking brings severe structural degradation, cracks are often observed in Na$_x$MnO$_2$ during cycling. Therefore, reducing or suppressing these phase transformations is critical for improving the electrochemical performance.

### 3.2 Dissolution/disproportionation reaction

The J–T effect modulates the Mn–O bond lengths, and the strength of this distortion is related to the covalency of the Mn–
O bonds. In contrast to the shorter axial Mn–O bonds, longer axial Mn–O bonds with weaker overlap have a higher negative charge (Fig. 6a). From a Lewis acid–base perspective, the axial oxygen atoms enhance the Lewis base strength and thus have high reactivity with acid. Generally, the electrolytes contain substantial H⁺, and these ions exclusively attack the axial oxygens that possess higher negative charges, resulting in severe Mn dissolution. For example, in Li-ion batteries, the LiPF₆ salt in the electrolyte is prone to hydrolysis, and this process induces the formation of a strong Lewis acid, HF. The H⁺ in HF interacts with the elongated axial oxygens, generating H₂O through oxide protonation. Meanwhile, an electron must transfer from the Mn to the axial oxide orbital; this charge transfer derives from the (dₓ²−z²) orbital of Mn³⁺. Correspondingly, the protonation process along with a metal-to-ligand electron transfer increases the oxidation state of the Mn ions. Unfortunately, the formed Mn⁴⁺ ions are a strong oxidizer that can be directly converted to the J-T-stabilized free state (Mn²⁺)

Fig. 6  (a) The six equivalent Mn–O bonds (left) in MLOs are distorted and transformed into the four shorter equatorial bonds and two longer axial bonds (right). (b) H⁺ in the electrolyte attacks the axial oxide and transfers electrons from the manganese to oxygen. Reproduced with permission. Copyright 2020 American Association for the Advancement of Science. (c and d) Decomposition reaction of EC in electrolytes with/without interacting with metallic Mn, Mn²⁺, and Li⁺ before and after one-electron reduction. (e) The catalytic effect of Mn²⁺ on the decomposition of electrolytes. Reproduced with permission. Copyright 2019 Springer Nature.
via a disproportionation reaction (the two-electron reduction pathway), and Mn$^{2+}$ ions are easily dissolved into the electrolyte (Fig. 6b).

### 3.3 Interfacial degradation arising from Mn$^{2+}$

Electrochemical degradation mechanisms originate from not only altering the surface structure of the MLO materials by the dissolution/disproportionation reaction, but also deteriorating the anode interface by the impact of Mn$^{2+}$. The electric field force and/or concentration gradient drives the Mn ions in the electrolyte to the anode side where Mn ions tend to change the interface compositions by decomposing the electrolytes and/or depositing on the anode, resulting in interfacial degradation and electrochemical degradation.

Specifically, the lower operation potential of conventional anodes like Li metal and graphite compared with the redox potential of Mn/Mn$^{2+}$ results in the electroplating of soluble Mn$^{2+}$ in the electrolyte on the anode. Simulations have indicated that the deposited Mn on the anode has high electrical conductivity or electrocatalytic properties, accelerating the decomposition of the electrolyte.$^{112-114}$ The reconstructed SEI gradually covers the anode and affects the ionic diffusion kinetics. Recently, Wang et al.$^{115}$ demonstrated that overlooking the effect of the solvation sheath on the Mn$^{2+}$ ions is the main reason for the inaccurate conclusions that Mn is the final product of the anode. They proposed that the solvation sheaths around Mn$^{2+}$ act as molecular-sized reaction vessels, in which the central Mn$^{2+}$ activates its solvation member and guides the reaction process. This process explains why Mn atoms are rarely detected on anodes.$^{115,116}$ If the solvation sheath begins to lose solvating members, Mn$^{2+}$ ions can convert to Mn atoms, explaining the existence of Mn atoms on the surface of the anode. It was found that the Mn atoms (Fig. 6c) substantially decrease the activation energy of EC decomposition (from TS1 247.20 to TS3 33.69 kJ mol$^{-1}$). In particular, when coordinated with Mn$^{2+}$, the EC decomposition activation energy (TS7) is less than that of neat EC (TS1), which is only 16.57 kJ mol$^{-1}$. After the EC anion receives an electron, the decomposition of EC (TS10) becomes much easier (Fig. 6d). With Mn$^{2+}$, the reduced EC decomposition activation energy (TS21: 9 kJ mol$^{-1}$) is higher than that of reduced EC (TS10), suggesting that the EC anion moves away from Mn$^{2+}$ after receiving one electron. Nevertheless, the generated coordination vacancies around Mn$^{2+}$ become occupied by other incoming EC molecules. Overall, the Mn$^{2+}$ solvation sheath serves as a catalytic center on the electrode surface or in the electrolytes, which continuously accelerates the reduction of carbonate molecules, resulting in severe interfacial degradation, and cell performance fading (Fig. 6e).

### 3.4 Crack formation

It has been proven that the ion intercalation/deintercalation process causes the expansion and contraction of the structure, generally resulting in a high strain on the materials that gradually induces cracking and disintegration.$^{117,118}$ Moreover, irreversible phase transition and Mn dissolution have been demonstrated to cause crack formation and severe capacity degradation.$^{119}$ Therefore, the observed macroscale cracks originate from the evolution of the atomic-scale structure.$^{120}$ By deeply analysing the atomic structure change, we can elucidate the crack formation process. Recently, Sharifi-Asl et al.$^{118}$ used aberration-corrected scanning transmission electron microscopy to demonstrate the correlation between the atomic structure evolution during the electrochemical cycling of LRMO materials and the formation of cracks. They observed a spinel-type grain boundary derived from the migration of TM ions to the Li-octahedral sites (Fig. 7a). Strain analysis revealed that the spinel-type grain boundary had a high strain, which may serve as crack initiation points in the cycled LRMO (Fig. 7b). Furthermore, a nanocrack was observed initiating on the grain boundary, which could develop into a large-scale crack in the cathode during further cycling. Apart from the irreversible phase transition from the layered structure to the spinel or rocksalt phase, a disordered phase was also observed (Fig. 7c). The energy-dispersive spectroscopy map shows Ni enrichment and Mn depletion in the disordered area (Fig. 7d). The depletion of Mn is attributed to the dissolution of Mn at the high-energy grain boundaries, which further accelerates the diffusion of Ni toward these Mn vacancies.

### 4. Remedies for electrochemical degradation in MLO cathodes

Numerous endeavors have been devoted to developing strategies at the atomic level to improve MLO-based energy storage performance. These strategies can be classified into the following categories: (i) elemental doping; (ii) interlayered structure manipulation; (iii) vacancy introduction; (iv) novel structural design. In this section, various strategies, synthesis methods, improved mechanisms, and electrochemical properties of typical MLO materials have been discussed and summarized, as shown in Table 2.

#### 4.1 Elemental doping

Incorporating different foreign ions into MLO materials is a common practice for optimizing their intrinsic properties. Elemental doping, including anion and cation doping, has been demonstrated to significantly boost the structural stability of MLO materials, which not only changes the electronic structure but also suppresses irreversible phase transformations during ion intercalation/deintercalation. Various dopants have been proposed, including cationic dopants (Cr, Ta, and Al) and anionic dopants (F and Br).$^{124,125}$

Inferior structural stability and electronic conductivity have been demonstrated to limit the electrochemical performance of δ-MnO$_2$. Zhao et al.$^{124}$ proposed that incorporating Cr$^{3+}$ into δ-MnO$_2$ could boost its rate capability and cycling stability within the potential window of 0–1.2 V (vs. Ag/AgCl). δ-MnO$_2$ without Cr doping undergoes severe J-T distortion, which increases the average Mn–O bond length and accelerates the migration and dissolution of Mn. Numerous dissolved Mn ions in the electrolyte generate cracks and pulverize the MnO$_2$ nanosheet fragments to a moderate extent (Fig. 8a). In addition, the
dissolved Mn ions form a precipitate in the electrolyte, impacting the electrochemical performance of the whole cell. Meanwhile, buckled Mn–O slabs are formed along with the J–T distortion, and these slabs dissociate and rearrange, leading to the irreversible phase transition from a layered structure to a 2/2C2 tunnel structure. Correspondingly, only 69.1% of the capacity of the pristine δ-MnO2 electrode is retained over 10 000 cycles at the current density of 10 A g−1. The elongation of the Mn–O bond induced by the J–T distortion is effectively suppressed in Cr3+-doped δ-MnO2 (Fig. 8b), implying that the Mn ion migration barrier greatly increases. The reinforced birnessite shows significantly improved structural stability (82.6% capacity retention over 30 000 cycles) in comparison with the pristine δ-MnO2, demonstrating the excellent effect of Cr doping in δ-MnO2 on suppressing the J–T effect and stabilizing the layered structure. Moreover, owing to the improved effect of Cr on the electronic conductivity of δ-MnO2, the Cr-doped δ-MnO2 showed a much higher capacity of 250 F g−1 than pristine δ-MnO2 with 178 F g−1 at the current density of 0.2 A g−1 with a high mass loading of ∼5 mg cm−2.

To further validate the role of elemental doping, Zhang et al.44 modeled the Zn-doped layered Na[Li0.25Mn0.75]O2 material to explore its effect on thermodynamics and the electronic structure. Zn doping enables the oxidation of the Mn3+ center and thus increases the Mn4+ content of the centers in the Zn-doped materials (Fig. 8d). As a result, the materials are more structurally stable and undergo less J–T distortion, leading to negligible capacity decay of the Zn-doped layered Na[Li0.25Mn0.75]O2 electrode (99% capacity retention over 200 cycles), which is much higher than that of the pristine Na[Li0.25Mn0.75]O2 electrode (41% capacity retention over 200 cycles). From the perspective of thermodynamics, the formation energies of mixing enthalpy (ΔHmix) have an important influence on phase stability. The calculated thermodynamic energy values can confirm the structural stability of materials according to charge carrier content. Fig. 8e shows that Na[Li0.25Mn0.75]O2 comprises five ground states, whereas Zn-doped Na[Li0.25Mn0.75]O2 has seven ground states. Specifically, in the first region (x = 0–0.25), Zn doping can decrease ΔHmix, indicating a reduced volumetric strain (Zn-doped structure: εv = −0.027;
Table 2: The synthesis methods and electrochemical performance of different strategies

<table>
<thead>
<tr>
<th>Material</th>
<th>Strategy</th>
<th>Synthesis method</th>
<th>Improved mechanism</th>
<th>Capacity</th>
<th>Capacity retention</th>
<th>Ref.</th>
</tr>
</thead>
<tbody>
<tr>
<td>δ-MnO₂</td>
<td>Cr doping</td>
<td>Hydrothermal</td>
<td>Suppressing the elongation of the Mn–O bond</td>
<td>250 F g⁻¹ at 0.2 A g⁻¹</td>
<td>69.1% (10 000 cycles) vs. 82.6% (30 000 cycles) at 10 A g⁻¹</td>
<td>124</td>
</tr>
<tr>
<td>Na₀.₆₅Mn₀.₇₅Ni₀.₂₅O₂</td>
<td>F doping</td>
<td>Solid-state reaction</td>
<td>Inhibiting P₃–O₁ phase transition</td>
<td>160 mA h g⁻¹ at 0.1C</td>
<td>75.1% vs. 83.1% (100 cycles) at 0.2C</td>
<td>125</td>
</tr>
<tr>
<td>Na₀.₈₃₂[(Li₀.₂₅Mn₀.₇₅)O₂</td>
<td>Zn doping</td>
<td>Sol-gel method</td>
<td>Inhibiting P₂–P₂’ phase transition</td>
<td>241 mA h g⁻¹ at 0.2C</td>
<td>41.3% vs. 96.9% (200 cycles) at 0.2C</td>
<td>34</td>
</tr>
<tr>
<td>Na₀.₆₅Mn₀.₇₅Ni₀.₂₅O₂</td>
<td>F/B co-doping</td>
<td>Solid-state reaction</td>
<td>Suppressing P₃–O₁ phase transition</td>
<td>164 mA h g⁻¹ at 0.1C</td>
<td>86% vs. 75% (100 cycles) at 0.2C</td>
<td>125</td>
</tr>
<tr>
<td>Na₂₃MnO₂</td>
<td>Na⁺ preintercalation</td>
<td>Hydrothermal</td>
<td>Stabilizing the interlayered structure</td>
<td>366 F g⁻¹ at 1A g⁻¹</td>
<td>96% (10 000 cycles) at 4 A g⁻¹</td>
<td>126</td>
</tr>
<tr>
<td>δ-MnO₂</td>
<td>Polyaniline intercalation</td>
<td>Chemical oxidation and reduction</td>
<td>Strengthening the layered structure and eliminating layered–spinel phase transition</td>
<td>280 mA h g⁻¹ at 0.2 A g⁻¹</td>
<td>89.2% (5000 cycles) at 2 A g⁻¹</td>
<td>101</td>
</tr>
<tr>
<td>Na₀.₆₅Mn₀.₅Co₀.₅Fe₀.₅O₂</td>
<td>Fe⁺⁺⁺ preintercalation</td>
<td>Solid-state reaction</td>
<td>Restraining the potential slab sliding</td>
<td>160 mA h g⁻¹ at 0.2C</td>
<td>68.2% vs. 89.6% (100 cycles) at 10C</td>
<td>127</td>
</tr>
<tr>
<td>Na₂₀[(Zn₁₉Mn₇₉)₁₅O₂</td>
<td>Mn vacancy</td>
<td>Pechini method</td>
<td>Shortening the Mn–O bond length</td>
<td>204 mA h g⁻¹ at 0.02 A g⁻¹</td>
<td>43.4% vs. 25.6% (100 cycles) at 0.02 A g⁻¹</td>
<td>128</td>
</tr>
<tr>
<td>Li₁₂₂Na₂₃Mn₀.₅₆Ni₀.₁₆Co₀.₀₉O₂</td>
<td>O vacancy</td>
<td>Solid-state reaction</td>
<td>Modulating the local Mn coordination environments</td>
<td>250 mA h g⁻¹ at 0.1C</td>
<td>84% vs. 94% (100 cycles) at 0.5C</td>
<td>40</td>
</tr>
<tr>
<td>Li₁₂₂[Mn₀.₆₆Co₀.₁₇Ni₀.₁₇]₀.₈O₂</td>
<td>Mixed vacancies of O and TM</td>
<td>Solid-state reaction</td>
<td>Prohibiting the O₃–P₃ phase transition</td>
<td>220 mA h g⁻¹ at 0.02 A g⁻¹</td>
<td>86% (250 cycles) at 0.125 A g⁻¹</td>
<td>129</td>
</tr>
</tbody>
</table>
| NaMnO₂₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋₋˓→  

| Spinel-layered LiMnO₂ | Heterostructure construction | Electrochemical conversion | Modulating interfacial orbital ordering | 286 mA h g⁻¹ at 1C | 90.4% (2000 cycles) at 12C | 44   |
| Li₁₂₂Mn₀.₃₃Nb₁.₃₃O₂F | Disordered structure | Ball-milling | Replacing the anionic reaction with a Mn reaction | 320 mA h g⁻¹ at 0.01 A g⁻¹ | 90% (25 cycles) at 0.02 A g⁻¹ | 130  |
| Li₁₂₂Ni₀.₃₃Mn₀.₆₆O₂ | Surface structure design | Chemical vapor deposition | Charge compensation | 238 mA h g⁻¹ at 0.5C | 99% (100 cycles) at 1C | 131  |
undoped structure: \( \varepsilon_V = -0.04 \). In the second region \( x = 0.25 - 0.50 \), Zn-doped Na\([Li_{0.25}Mn_{0.75}]O_2\) is illustrated by the red circles located on the tie line, which directly confirm that desodiation occurs in the same phase. In undoped Na\([Li_{0.25}Mn_{0.75}]O_2\), a pseudo-ground state appears, suggesting the formation of a new phase during the desodiation process. In the third region \( x = 0.50 - 0.88 \), Zn-doped Na\([Li_{0.25}Mn_{0.75}]O_2\) has one pseudo-ground state as marked by the grey circle \( x = 0.63 \),
indicating that phase dissociation may occur within the narrow region from $x = 0.5$ to $x = 0.75$. The extra ground states in Zn-doped Na$_{[Li_{0.25}Mn_{0.75}]}$O$_2$ are beneficial for suppressing the inherent phase transition that occurs in Na$_{[Li_{0.25}Mn_{0.75}]}$O$_2$.\cite{122} In contrast to its doped counterpart, two pseudo-ground states of undoped Na$_{[Li_{0.25}Mn_{0.75}]}$O$_2$ are shown as grey circles above the tie line (at $x = 0.63$ and $x = 0.75$, respectively), reflecting the phase separation to this extent during desodiation.

In summary, elemental doping can restrain the elongated bonds and stabilize the metastable structure to inhibit J–T distortion by changing the atomic structures around Mn ions, giving rise to a more stable structure. Additionally, the average valence of Mn ions may be tuned to above +3.5, which is beneficial to suppress the J–T effect, such as the introduction of the low-valence ions. Although elemental doping has achieved tremendous progress, some concerns should be noted. Presently, annealing at high temperatures is the most popular method to realize doping, but this process involves high energy cost and has low efficiency. In addition, it is difficult to precisely control the doping content and doping sites. Hence, it is highly desirable to explore the powerful and facile method of accurate doping with low energy costs.

Fig. 9 (a) Schematic illustration of the preintercalation process during the electrochemical oxidation. (b) The ion-preintercalation effect on widening the potential windows of layered Na$_0.5$MnO$_2$. Reproduced with permission.\cite{126} Copyright 2017 Wiley-VCH. (c) A schematic of the expanded structure by the intercalation of polyaniline (PANI) into MnO$_2$ nanolayers. Reproduced with permission.\cite{101} Copyright 2018 Springer Nature. (d) The typical in situ XRD patterns of the Na$_{0.67}$Mn$_{0.5}$Co$_{0.5}$Fe$_x$O$_{2}$ electrode. (e) The variation of cell parameters $c$ (green points), $V$ (blue points), and $a/b$ (pink points). (f) $V$ variation of Na$_{0.67}$Mn$_{0.5}$Co$_{0.5}$Fe$_x$O$_2$ and several typical sodium-storage cathode materials during the charge/discharge process. Reproduced with permission.\cite{127} Copyright 2021 Wiley-VCH.
4.2 Interlayered structure manipulation

Most MLO materials undergo the degradation of diffusion channels after long-term cycling processes. Therefore, the stabilization/enlargement of diffusion channels has emerged as a promising strategy based on improving the stabilization of the interlayer to accommodate J-T distortion. One of the characteristics of preintercalation in the interlayer hosts of MLO materials is the expansion of the interlayer spacing. The appropriate intercalated ions in the interlayer can serve as pillars to effectively expand and support this diffusion channel.24,126 Xia and co-authors proposed a hydrothermal synthesis to preintercalate Na+ into birnessite and prepared a high Na content birnessite [Na0.5MnO2] (Fig. 9a). It is known that preintercalating the Na+ protects the layered structure from collapsing. Meanwhile, the enhanced Na content in birnessite also extends the potential window from 0–1.0 V to 0–1.3 vs. Ag/AgCl (Fig. 9b). In addition, owing to the high diffusion kinetics of Na+ in large interlayers, the Na0.5MnO2 electrode could achieve a high rate performance and specific capacitance (366 F g⁻¹ at 1 A g⁻¹; 231 F g⁻¹ at 16 A g⁻¹). A 2.6 V aqueous energy storage system and large energy density of 81 Wh kg⁻¹ serve as further evidence that preintercalating Na+ ions into the interlayer can boost its energy storage. In addition to the pre-intercalation of ions into the diffusion channel, pre-intercalation of molecules can also effectively expand the interlayer space. Huang et al.ⅠⅢ prepared a polyaniline-intercalated layered MnO2 via an aqueous/organic interfacial reaction. In detail, the chemical reduction of MnO2 and the oxidation polymerization of aniline occur at the same time at the interface between the inorganic phase (KMnO4 aqueous solution) and the organic phase (CCl4-containing aniline monomer), inducing the layer-by-layer assembly of polyaniline and MnO2 (Fig. 9c) with a dramatically expanded interlayer (~1.0 nm). The large interlayer spacing of PANI-intercalated MnO2 achieved the best rate capability (280 mA h g⁻¹ at the current density of 0.2 A g⁻¹; 110 mA h g⁻¹ at the current density of 23 A g⁻¹) among the reported data in this field.77,134,135 Additionally, intercalating the PANI into the interlayers significantly reinforces the layer structure, thus avoiding irreversible phase transition and the distortion of the interlayered structure, as evidenced by the outstanding long-term cycling stability over 5000 cycles in rechargeable Zn–MnO2 batteries.

Aside from enlarging the interlayer spacing, some intercalated ions may have a pinning effect to suppress phase transitions and thus improve the structural stability. In Na-ion batteries, Na+ ions possess a large radius and cause huge lattice changes of more than 20% during the sodiation/desodiation process, inducing a severe phase transition, resulting in more than 30% capacity fading over 100 cycles at the current density of 2C.127 Chu et al.127 reported that intercalated Fe in the Na site acts as a pinning point to suppress slab sliding. The pinning point represents a structural stabilizer that significantly reduces the volume variation to approximately 0.6% (Fig. 9d and e). This slight structural change confers outstanding structural stability and maintains a smooth channel for fast Na-ion transfer during the electrochemical processes. The ultra-low volume variation is even smaller than those of electrode materials like LiCoO2 and LiFePO4, as shown in Fig. 9f. Consequently, Na-ion batteries fabricated using this improved material can achieve high capacity retention of ~89.1% over 100 cycles at the current density of 2C. Impressively, 71% capacity can be retained over 1000 cycles at the high current density of 10C.

On the basis of the above introduction, manipulating the interlayered structure via the preintercalation can greatly enlarge the interlayer space and thus facilitate the ions to diffuse in the bulk of the MLO materials, which also enables the MLO materials to provide more active sites to accommodate cations. In addition, the support effect of preintercalated ions/molecules on the interlayered structure can improve the structural stability, guaranteeing long-term cycling stability. Nevertheless, preintercalated ions may have the electrostatic repulsion effect with the inserted ions during the electrochemical processes. Moreover, the stability of supported ions/molecules should be considered. These preintercalated species suffer from issues of extraction of ions/molecules from the host interlayers during the electrochemical processes, which is unavoidable. Such issues eliminate the advantages of manipulating interlayered structures and induce severe structural degradation during the continuous intercalation/deintercalation of cations. Excessive cation preintercalation would induce the phase transition where the structural reconstruction enables the loss of the fast ionic transport of layered pathways.109 Therefore, to avoid these disadvantages, careful screening of the preintercalated ions/molecules and optimizing their amount are necessary.

4.3 Vacancy introduction

Considering that vacancy engineering can tune the physicochemical properties of MLO materials, it is logical to anticipate that precisely designing vacancies may promote their electrochemical performance. Vacancies involving MLO materials mainly consist of anionic (nonmetal) vacancies, cationic (metal) vacancies, and mixed vacancies. MLO materials usually suffer from severe J-T distortion of the octahedral Mn2+ at low voltages and/or anionic reaction-induced structural evolution at high potentials. Incorporating the vacancies into the TM layer promotes ionic reduction/oxidation and bond length changes, which have a significant effect on the layered structure stability at both low and high potentials.136–140 To determine how vacancies improve the structural stability, Yang et al.128 investigated structural and electrochemical behavior changes by designing vacancy-containing P2-type Na2/3[Zn1/9Mn7/9]O2 (NZMO-Vac; with vacancies in the TM layer). The low-spin Mn3+ ion content increased after the introduction of vacancies at a low voltage, which was ascribed to the shortened Mn–O bond, as shown in (Fig. 10a). It is well known that high-spin Mn3+ ions are the main reason for the undesirable structural degradation and substantial capacity fading of MLO materials. Thus, increasing the content of low-spin Mn3+ is key to avoiding J-T distortion. Fig. 10b shows the coexistence of 23.6% low-spin Mn3+ and 10.9% high-spin Mn3+ ions in the discharged NZMO-Vac. Meanwhile, the oxygen atoms around the formed
Vacancies contributed much more charge compensation than those around Zn, accelerating the oxygen redox reaction and enhancing the reversibility of the anionic reaction. In addition, Mn vacancies can provide more ion intercalation sites, giving rise to reduced charge transfer resistance and improved specific capacitance. Correspondingly, NZMO-Vac delivered a much higher capacity than NZMO (204 and 187 mA h g$^{-1}$ for NZMO-Vac and NZMO, respectively) at 0.02 A g$^{-1}$ in the first cycle. Moreover, the NZMO-Vac retained 43.4% capacity after 100 cycles, which is better than the 25.6% of NZMO.

In addition to single vacancies, the introduction of mixed vacancies is also an effective way to suppress structural degradation due to the synergistic effects between anionic and cationic vacancies. In both P2-type and O3-type MLO materials, the irreversible layer gliding causes huge volume changes and significantly jeopardizes the structural stability. To solve this problem, reducing the net charge of the oxygen (i.e., creating oxygen vacancies) or enabling reversible TM migration during the charging/discharging process is critical. “Reversible” migration implies that TM can move between the TM layer and alkaline layer but does not need to return to the same site each time. Therefore, cationic vacancies can provide many more sites for TM migration. To demonstrate this concept, Xiao et al. prepared an Li- and Mn-rich layered cathode material, in which the formation of oxygen vacancies and Mn vacancies was simultaneous. The results of a combination of computational simulations and systematic characterizations demonstrated that the mixed vacancies enabled the TM to migrate to the tetrahedral sites of the Na layer during desodiation and move back during sodiation. The TM migration suppressed a complete O3–P3 phase transition by increasing the energy barriers and stabilized the performance. Fig. 10c shows the proposed energy evolution during the O3–P3 two-phase reaction by manipulating TM migrations. Such metastable states can immobilize the structure and inhibit the transition to global P stacking. Correspondingly, the Li- and Mn-rich layered cathode material delivered 86% capacity retention after 250 cycles (Fig. 10d). In addition, although the interlayer spacing decreased to 5.0 Å, the diffusion kinetics of Na$^+$ ions was not affected, which was attributed to the formation of oxygen vacancies and the distortion of the metal oxide layer.

As described above, the incorporation of Mn and/or O ion vacancies in MLO materials promotes ion-reduction/oxydation and bond length changes, which significantly increase the percentage of low-spin Mn$^{3+}$ and provide more sites for the reversible migration of Mn ions. It should be noted that the introduction of vacancies can alter the electronic structure and bonding energy in MLO materials. Too high amount of vacancies may sacrifice the conductivity of materials and structural stability. However, how to achieve precise and optimum vacancy concentration is a critical issue in the current synthesis process. Correspondingly, developing new synthesis methods to accurately control the vacancy concentration is necessary. Furthermore, advanced characterization methods to monitor the generation process of vacancies are also needed.
4.4 Novel structural design

Structural stability during the electrochemical reaction is an indispensable requirement of MLO materials. Both O3-type and P2-type Na$_x$MnO$_2$ undergo the gliding of oxygen layers during the electrochemical process and suffer from J-T distortion, leading to structural deterioration and undesirable performance. Birnessite-Na$_x$MnO$_2$·nH$_2$O possesses a much larger interlayer spacing (~7 Å) than the O2 (~5 Å) and P3 (~5 Å) phase, increasing the rate of Na$^+$ diffusion. In contrast, no oxygen layer gliding occurs in Birnessite-Na$_x$MnO$_2$·nH$_2$O during oxidation due to the pillar effect of crystal water; the corresponding layered structure can offer protection against J-T distortion and structural degradation.

Nevertheless, it should be taken into account that crystal water in the interlayer may limit Na accommodation and hinder Na$^+$ diffusion. Meanwhile, crystal water is extracted in the oxidation process, and the presence of water in the electrolyte greatly impacts its stability and accelerates decomposition.

To address these issues, Xia et al. designed a novel monoclinic polymorph NaMnO$_{2-y}$·d(OH)$_y$ with the coexistence of the O’3 and H’3. In sharp contrast to the monoclinic, O3-, and P2-Na$_x$MnO$_2$, the monoclinic polymorph NaMnO$_{2-y}$·d(OH)$_y$ has a larger interlayer distance (~7 Å), which greatly enhances the kinetics of Na$^+$ ion diffusion. The large interlayer space is attributed to the simultaneous formation of O’3 and H’3 phases with six-coordinate octahedral sites and eight-coordinate hexahedral sites, respectively. It was found that the generated O vacancy in this material can reduce the bandgap, suggesting the enhanced electronic conductivity of NaMnO$_{2-y}$·d(OH)$_y$. Therefore, NaMnO$_{2-y}$·d(OH)$_y$ with large interlayered spacing and high electronic conductivity accelerates the electron and ion transport in the electrode, which enables the NaMnO$_{2-y}$·d(OH)$_y$ electrode to maintain a large capacity of 156 mA h g$^{-1}$ even at the current density of 50C, outperforming the previously reported data in this field. In addition, the reduced distortion of Mn–O and the small β angle in NaMnO$_{2-y}$·d(OH)$_y$ are beneficial to suppress the J-T distortion and minimize the volume change during electrochemical processes (Fig. 11a). These attributes explain why NaMnO$_{2-y}$·d(OH)$_y$ can achieve superior structural stability and excellent cycle performance. As a result,
the prepared NaMnO$_{2-x-y}$(OH)$_{2y}$ electrode achieved outstanding cycling stability (94.6% capacity retention at 10C over 1000 cycles) than the Birsnessite NaMnO$_{2-y}$(OH)$_{2y}$·0.6H$_2$O (78% capacity retention at 10C over 1000 cycles).

Designing novel interface structures is also effective in suppressing J-T distortion. Considering the orbital ordering in Mn$^{3+}$-containing MLO materials, such as LiMnO$_2$, which can generate the collinear J-T order, the strong cooperative J-T effect may cause a large volume change and promote structural degradation (Fig. 11b). To eliminate this cooperative J-T effect, Xia et al. designed a unique heterostructure of LiMnO$_2$. The heterostructure comprised layered and spinel domains with various orientations for J-T distortion, which can effectively disrupt the long-range collinear J-T ordering for either the spinel phase or the layered phase in the heterostructure. Fig. 11c compares the axial Mn–O bond elongation in this heterostructure with those in spinel and monoclinic LiMnO$_2$. The heterostructure increased the axial Mn–O bonds from 1.96 to 2.07 Å in spinel LiMnO$_2$ and 1.98 to 2.03 Å in monoclinic LiMnO$_2$, which corresponds to 5.5% and 2.5% distorted degrees, respectively. For layered LiMnO$_2$ and spinel Li$_2$Mn$_3$O$_4$, the distortion degrees were 18.0% and 15.9%, respectively, much higher than the corresponding phase in the heterostructure. Therefore, the layered–spinel heterostructure with controlled interfacial orbital ordering significantly suppressed J-T distortion. The electrochemical results demonstrated that the LiMnO$_2$ heterostructure cathode delivered high specific capability $\sim$254.3 mA h g$^{-1}$ with 97% coulombic efficiency. Furthermore, such a heterostructure achieved a high specific capacity of 183.6 mA h g$^{-1}$ at 12C. Unprecedented cycling stability with $\sim$90.4% capacity retention over 2000 cycles was achieved, demonstrating the significant effect of interfacial engineering on mitigating J-T distortion.

Cation-disordered lithium-rich MLO materials are another example of a novel structure. Lee et al. prepared a new cation-disordered (disordered rock-salt structure) material Li$_2$Mn$_{2/3}$Nb$_{1/3}$O$_2$F by manipulating the layered phase of Li$_2$MnO$_3$ (Fig. 11d). The disordered rock-salt structure, originating from a layered structure, can generate more 0 TM channels. This diffusion channel has a low migration barrier, giving rise to fast Li$^+$ diffusion kinetics. Furthermore, with cation and anion substitution, the disordered rock-salt Li-rich cathodes could deliver a high capacity (>300 mA h g$^{-1}$), mainly from the Mn$^{2+}$/Mn$^{4+}$ double electron reaction (Fig. 11e). This Mn$^{2+}$/Mn$^{4+}$ double redox effectively suppresses the problems associated with O redox by reducing the anionic redox reaction content.

In short, novel structural design shows notable effectiveness in manipulating the localized structure to suppress the J-T effect, providing stable units in localized regions and thus effectively eliminating electrochemical degradation. Nevertheless, novel structural design faces great challenges, countering which needs an in-depth understanding of the structure–property relationships in MLO materials. Additionally, accurately manipulating the structure at the atomic level is difficult, which is not suited to large-scale applications in most cases.

5. Conclusion and future perspectives

In recent decades, endeavors to build a holistic understanding of the impact of electrochemistry on the structural degradation of MLO materials have experienced a revival. Substantial progress has been made in the field of MLO materials for various energy storage technologies. All in all, developing MLO materials with high capacity, high voltage, high stability, and high rate in energy storage is our target. To reach this target, the irreversible phase transformation, Mn dissolution/disproportionation reaction, interfacial degradation arising from Mn$^{2+}$, and crack formation induced by J-T distortion must be better understood at the atomic level. From the perspective of structure–property relationships in MLO materials, this review comprehensively summarizes structure and property optimization for high-efficiency energy storage. First, elemental doping can restrain the elongated bonds and stabilize the metastable structure to inhibit J-T distortion during the electrochemical process, preventing structural degradation. Second, the interlayer structure can be manipulated by pre-intercalating ions or molecules, which serve as pillars against lattice change and endow MLO materials with a robust layered structure and/or large interlayer spacing to accommodate J-T distortion, resulting in fast diffusion kinetics and high cycling stability. Third, the incorporation of Mn and/or O ion vacancies in MLO materials promotes ion-reduction/oxidation and bond length changes, which significantly increase the percentage of low-spin Mn$^{3+}$ and provide more sites for the reversible migration of Mn ions. Finally, novel structural design shows notable effectiveness in manipulating the localized structure to suppress the J-T effect, providing stable units in localized regions and thus effectively eliminating electrochemical degradation.

Although various strategies have paved effective pathways to suppress electrochemical degradation, it is still difficult to thoroughly disentangle the issues raised above. Additional research efforts are needed to develop methods for stabilizing the crystal structure during electrochemical processes to suppress capacity degradation in the representative MLO systems. Our suggestions are as follows:

(1) Coupling various strategies to induce synergistic effects for high-performance MLO materials. Each strategy may have a limited effect on suppressing the degradation of MLO cathodes. In this case, integrating various strategies to exploit their synergistic effects may be worth exploring. For example, combining doping and interlayer manipulation may realize robust structural stability and fast diffusion kinetics.

(2) Developing novel MLO materials with stable structures. Presently, most MLO materials still have structural issues and are sensitive to air containing water vapor. It is critically necessary to innovate MLO materials with highly stable localized structures and microstructures by modulating the electronic states and tailoring the crystal domains. In addition, advanced machine learning and theoretical calculations can be applied to design and develop high-performance MLO materials.
materials. With improved computational analyses, it is possible to gain some quantitative comprehension of the effect of each adverse factor on structural stability using qualitative constraints in simulations.

(3) Realizing accurate MLO cathode characterization and comprehensively elucidating the structure–property correlations. Electrochemical energy storage devices store energy through electrochemical reactions. However, the understanding of the intrinsic electrochemical reaction mechanisms in MLO materials is still limited due to the complexity of each energy storage system and the effects of electrolytes, conductive carbon additives, and binders. In this regard, powerful characterization techniques, such as in situ transmission electron microscopy, in situ X-ray absorption spectroscopy, and electrochemical quartz crystal microbalance analysis, are indispensable to distinguish the dynamic structural evolution and electrochemical behaviors of MLO cathodes and comprehensively understand their structure–property correlations.

In conclusion, a comprehensive understanding of the electrochemical degradation mechanisms at the atomic level is essential for developing high-performance MLO cathodes, as is systematically outlined in this review. Although numerous challenges remain, the future is bright thanks to the tremendous endeavors to elucidate J-T distortion and the continual development of effective strategies, advanced characterization techniques, artificial intelligence, and theoretical calculations. We hope that the synergistic application of experiments and computational modeling will dramatically improve our understanding of the chemical degradation mechanisms and the electrochemical performance of energy storage systems based on MLO cathodes.

Conflicts of interest
The authors declare no conflict of interest.

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