Comparative Investigation of 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.2}$Mn$_{0.3}$O$_2$ Cathode Materials Synthesized by Using Different Lithium Sources

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Lithium-rich manganese-based cathode materials has been attracted enormous interests as one of the most promising candidates of cathode materials for next-generation lithium ion batteries because of its high theoretic capacity and low cost. In this study, 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.2}$Mn$_{0.3}$O$_2$ materials are synthesized through a solid-state reaction by using different lithium sources, and the synthesis process and the reaction mechanism are investigated in detail. The morphology, structure, and electrochemical performances of the material synthesized by using LiOH·H$_2$O, Li$_2$CO$_3$, and CH$_3$COOLi·2H$_2$O have been analyzed by using Thermo gravimetric analysis (TGA), X-ray diffraction (XRD), Scanning electron microscope (SEM), X-ray photoelectron spectroscopy (XPS), and electrochemical measurements. The 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.2}$Mn$_{0.3}$O$_2$ material prepared by using LiOH·H$_2$O displays uniform morphology with nanoparticle and stable layer structure so that it suppresses the first cycle irreversible reaction and structure transfer, and it delivers the best electrochemical performance. The results indicate that LiOH·H$_2$O is the best choice for the synthesis of the 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.2}$Mn$_{0.3}$O$_2$ material.

Keywords: lithium-ion battery, lithium-rich manganese-based cathode materials, lithium sources, solid reaction, electrochemical performances

INTRODUCTION

Rechargeable Li-ion batteries (LIBs) play a dominant role in energy storage devices of portable electronic devices and electric vehicles (EVs) for its excellent safety, long cycle life, and high energy density (Armand and Tarascon, 2008; Goodenough and Park, 2013; Zhang Q. et al., 2015, 2016; Li et al., 2017a,b; Zhang et al., 2018). However, the specific energy density of LIBs still cannot meet the demand of EVs owing to lower energy density of cathode material (Whittingham, 2014). At present, classic cathode materials, such as LiFePO$_4$ (Zheng et al., 2008, 2017), LiMn$_2$O$_4$ (Kim et al., 2008), LiCoO$_2$ (Kang et al., 1999), and LiNi$_{0.8}$Mn$_{0.2}$O$_2$ (Li et al., 2017) etc., offer a reversible discharge capacity less than 200mAh g$^{-1}$. Recently, lithium-rich manganese-based cathode materials has been attracted enormous interests as a promising cathode material for next-generation LIBs because of high discharge capacity (more than 250mAh g$^{-1}$) and low cost (Ohzuku et al., 2011).

Lithium-rich manganese-based cathode materials contain double component: one phase of Li$_2$MnO$_3$ with C2/m space group and the other phase of LiMO$_2$ with R-3m space group. Because
Mn$^{4+}$ in Li$_2$MnO$_3$ phase cannot be oxidized any more, it possesses electrochemically inert. However, Li$^+$ and O$_2$ can be extracted from the TM (transition metal) layer and the lattice, respectively, which indicates that the Li$_2$MnO$_3$ phase is activated by initial charging process and forms an irreversible loss of Li$_2$O (Yabuuchi et al., 2011). It is demonstrated that the high capacity of the materials originates from the oxygen escape from Li$_2$MnO$_3$ phase at high voltage (Yabuuchi et al., 2011). In addition, there are so many fatal disadvantages in lithium-rich manganese-based cathode materials, such as severe voltage fading during cycling (Zheng J. et al., 2015; Zhang T. et al., 2016) poor rate performance (Fan et al., 2015; Rozier and Tarascon, 2015; Zhang K. et al., 2015), large initial irreversible capacity, and low initial coulombic efficiency (Bai et al., 2015). Presently, many methods like comprising doping (Dianat et al., 2013; Wang et al., 2013; Li et al., 2014; Zhang H. et al., 2014), coating (Shi et al., 2012, 2013; Gu et al., 2013; Zhang et al., 2013; Zhou et al., 2017; Liu et al., 2018), nano crystallization (Wang et al., 2010), and morphology control (Yang et al., 2013; Remith and Kalaiselvi, 2014) have been proposed to promote the electrochemical performance of the materials.

Generally speaking, the compositions of this kind of material are varied. According to the chemical formula of lithium-rich manganese-based cathode materials, it can be simply written as $x$Li$_2$MnO$_3$·($I$-$x$)LiMnO$_2$ ($M$ = Ni, Co, Mn, Ti, Fe, etc.; Ohzuku et al., 2011; Yabuuchi et al., 2011; Bai et al., 2015; Fan et al., 2015; Rozier and Tarascon, 2015; Zheng J. et al., 2015; Zhang T. et al., 2016). Common and representative components of lithium-rich manganese-based materials are 0.5Li$_2$MnO$_3$·0.5LiNiO$_3$·0.5Mn$_2$O$_3$ and 0.5Li$_2$MnO$_3$·0.5LiNi$_{1/3}$Co$_{1/3}$Mn$_{1/3}$O$_2$ (Shi et al., 2012, 2013; Dianat et al., 2013; Gu et al., 2013; Wang et al., 2013; Yang et al., 2013; Zhang et al., 2013; Zhang H. et al., 2014; Li et al., 2014; Zhou et al., 2017).

To the best of our knowledge, the presence and content of nickel that improves the cathode capacity (Sun et al., 2005). Zheng Z. et al. (2015) discussed that the roles and the functions of nickel in electrochemical cycling of lithium-rich Mn-based cathode materials. Yang et al. (2017) reported that the Ni substitution at 2c sites not only enhances oxygen stability and delays oxygen loss from the lattice but also suppresses the cation mixing that induces the undesired phase transition. Gao et al. (2017) demonstrated that Li-rich material Li$_{1.2}$(Ni$_{0.25}$Co$_{0.25}$Mn$_{0.5}$)O$_{2.8}$ was prepared by a novel core-shell structure, in which Ni element acts as stabilizing ions to inhibit the Jahn-Teller effect of active Mn$^{3+}$. Based on the above considerations, lithium-rich manganese-based material 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.25}$Mn$_{0.3}$O$_2$ is designed for the first time.

In fact, the cathode materials with different morphology and electrochemical performance can be achieved by using different lithium sources (Zhang B. et al., 2014; Cao et al., 2017). In this work, we; try to synthesize 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.25}$Mn$_{0.3}$O$_2$ materials with nanosize particles by using different lithium sources (LiOH·H$_2$O, Li$_2$CO$_3$ and CH$_3$COO$Li$·2H$_2$O). The solid state reaction mechanism is investigated and the effects of lithium sources on the morphology, structure and electrochemical performance are clarified.

**EXPERIMENTAL**

**Material Preparation**

The 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.25}$Mn$_{0.3}$O$_2$ cathode material was prepared via solid-state reaction. Analytical grade chemicals NiC$_2$H$_4$O$_4$·H$_2$O (AR, 99.9%), CoC$_2$H$_4$O$_4$·4H$_2$O (AR, 99.5%), MnC$_6$H$_8$O$_4$·H$_2$O (AR, 99%), and different lithium sources, LiOH-H$_2$O, Li$_2$CO$_3$, and CH$_3$COO$Li$·2H$_2$O (excess 3.33% molar ratio, AR 95%, AR 98%, AR 99%, respectively), with a stoichiometric amount were mixed thoroughly and ball milled (200 rpm) for 1 h. With an amount of ethanol added, the materials were ball milled (200 rpm) continually for 3 h and dried in an oven at 80°C for 12 h to obtain a uniform mixed precursor. Then the dried precursor was ball milled for 0.5 h and sintered at 900°C in air atmosphere to prepare the targeted 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.25}$Mn$_{0.3}$O$_2$ material. The rate of heating was retained at 5°C min$^{-1}$. The 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.25}$Mn$_{0.3}$O$_2$ compounds synthesized by using LiOH-H$_2$O, Li$_2$CO$_3$, and CH$_3$COO$Li$·2H$_2$O as lithium sources are marked as Sample A, Sample B, and Sample C, respectively.

**Sample Characterization**

The crystalline structure of 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.25}$Mn$_{0.3}$O$_2$ was tested by X-ray diffraction (XRD, Rigaku D/maxb) with Cu Kα radiation ($\lambda = 1.54056\AA$) in the range of 10°-80° with the speed of 5° min$^{-1}$. The morphology was investigated with scanning electron microscopy (SEM, Philips, FEI Quanta 200 FEG) and transmission electron microscopy (TEM, TECNAI G2 F20, FEI). The sample was examined by Thermo gravimetric/Differential Scanning calorimeter (TG/DSC, SDT Q600) under the air from ambient temperature to 1,000°C at 10°C min$^{-1}$. X-ray photoelectron spectroscopy (XPS, VG Multilab 2000) was used to observe the chemical valence of the TMs (transition elements Ni, Co, Mn, O) of the sample.

**Electrochemical Measurements**

Electrochemical measurements of 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.25}$Mn$_{0.3}$O$_2$ were tested by CR2025 coin-type cells. The positive electrode was operated as slurry by 80% active material (0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.25}$Mn$_{0.3}$O$_2$), 10% acetylene black, 10% polyvinylidene fluoride (PVDF), and N-methylpyrrolidone (NMP) solvent. Then the electrode slurry was cast on aluminum foil and dried at 120°C for 12 h under vacuum atmosphere. Typical active material areal loadings were about 1.2 mg cm$^{-2}$. The cells were assembled in a filled argon glove box. Lithium metal was used as the anode and the separator was a Celgard 2500. The electrolyte utilized was a 1 M LiPF$_6$ solution by mixtures of ethylene carbonate (EC) and dimethyl carbonate (DMC) with a volume ratio of 1:1. The Galvanostatic charge-discharge measurements were carried out using NEWARE CT-3008 battery testing system (Shenzhen, China) within the voltage range of 2.0–4.8 V at
room temperature. Cyclic voltammetry (CV) measurements were conducted using CHI660D Electrochemical Workstation (Shanghai Chen Hua) at 0.1 mV s⁻¹ between 2.0 and 4.8 V. Electrochemical impedance spectroscopy (EIS) of the cell was carried out using CHI660D Electrochemical Workstation (Shanghai Chen Hua) in the frequency range of 0.1 Hz−10 kHz, and the AC voltage was applied as 5 mV.

RESULTS AND DISCUSSIONS

In order to ascertain the optimum temperature for heat treatment and explore the effects of different lithium sources, TG-DSC analyses are done for the precursor in the air. As presented in Figure 1, there are three main stages for weight losses in the TG plots and several endothermic and exothermic peaks in the DSC plot. The temperature range from ambient to about 200°C, the weight loss is the release of hydration water from precursor (Deng et al., 2010). In the region from 200 to 500°C, there is a sharp exothermic peak (Figures 1A,B) or two exothermic peaks (Figure 1C) accompanied by abrupt weight loss observed in DTG/DSC curves, it should be related to the volatilization of crystallized water from $\text{MC}_2\text{H}_2\text{O}_4\cdot\text{H}_2\text{O}$ (M = Ni, Co, Mn) and the decomposition of precursor. The weight loss of precursor mainly comes from the escape of water and carbon dioxide during the reaction. As the temperature increases from 500 to 1,000°C, weight loss (Figures 1A,C) almost can’t be observed in DTG curves. However, a little exothermic peak and a small amount weight loss is observed in DSC curves (Figure 1B) when the temperature reaches 720°C, which corresponding to the melting temperature of lithium carbonate. We think that it attributes to carbon dioxide emission from lithium carbonate (Li$_2$CO$_3$).

The mass loss of sample A, B, and C is 51.44, 54.93, and 62.63%, respectively. The weight loss is from the decomposition of precursor and the release of water and carbon dioxide under heating (Cao et al., 2017). The greater the weight losses, the greater the quantity of gas releases (Cao et al., 2017). During the material formation under heat treatment, a large number of gas releases will destroy the primary particle morphology and promote it growth, resulting in aggregation. Because the mass loss of sample A is the least, the average size of particles and uniformity of the sample A should be much better than others. It will be proved in subsequent SEM images (Cao et al., 2017).

The XRD patterns for sample A synthesized at different temperatures are shown in Figure S1. It can be seen that the crystallinity of the samples increases accompanied by temperature increment from 600 to 1,000°C. As the temperature increases to 800°C, there appears a couple of peaks during 20–23° attributed to the super lattice diffraction of the monoclinic Li$_2$MnO$_3$ phase (Gao et al., 2017). With the temperature increasing to 900°C, the diffraction peaks of sample A are well indexed to a hexagonal α-NaFeO$_2$ structure (Seteni et al., 2017). Sample A displays the (006)/(012) and (018)/(110) peaks with a fine splitting, it indicates that it possesses highly ordered good crystallinity layered structure. When the temperature reaches 1,000°C, the peak at 36.5° corresponding to LiMn$_2$O$_4$ with spinel structure. It indicates that LiMn$_2$O$_4$ structure can be formed at higher temperature (Zhang B. et al., 2014). Based on the above article 159

![Figure 1](image-url) | TG-DSC curves of the precursor by using different lithium sources. (A) LiOH·H$_2$O, (B) Li$_2$CO$_3$, (C) CH$_3$COOLi·2H$_2$O.
are well separation, illustrating that the samples possess good crystallinity and fine layered structure. Nevertheless, there are a group of minor peaks at 36.5° and 44° of sample C corresponding to LiMn$_2$O$_4$ (Fd-3m) of formed under the high temperature. Lattice parameters of 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.2}$Mn$_{0.3}$O$_2$ are listed in Table 1. The c/a ratios of samples are more than 4.9, it indicates that the material possesses layered characteristics (Zhang M. et al., 2016). The $I_{(003)}/I_{(104)}$ ratios of materials are much larger than 1.2, which indicates that the samples have low cationic mixing (Deng et al., 2010).

The SEM images of 0.5Li$_2$MnO$_3$·0.5LiNi$_{0.5}$Co$_{0.2}$Mn$_{0.3}$O$_2$ layered materials obtained before and after heat treatment are shown in Figure S2. The images of all samples possess morphology of similar aggregation, but sample A and B (Figures S2B,E) synthesized by LiOH·H$_2$O and Li$_2$CO$_3$ have higher homogeneity. The sample C (Figures S2H,I) forms an irregular aggregation of primary particles. As shown in Figures S2B,E, sample A synthesized by LiOH·H$_2$O shows much smaller primary particle sizes compared with other samples. The smaller particle size can shorten the diffusion distance of lithium ions and improve the electrochemical performance of the materials (Cao et al., 2017).

To confirm SEM results, TEM analysis and the Fast Fourier Transform (FFT) are performed, as shown in Figure 3. Sample
A composed of 200–300 nm primary particles and the distance of the lattice fringes of particles is calculated to be 0.204 nm, matching well with the d_{(104)} planes, which attributed to layer structure R-3m (Yang et al., 2016). Figures 3C,D show that sample B composed of 300–400 nm primary particles and the lattice spacing are 0.273 and 0.368 nm, corresponding to the planes d_{(111)} and d_{(-111)} of Li_{2}MnO_{3} phase (C2/m) (Luo et al., 2014). As illustrated in Figures 3E,F, Sample C has severe aggregation although it possesses small primary particles from some regions owing to the destruction of released gas. The distance of the lattice fringes of this material is calculated to be 0.47 nm, corresponding to the d_{(003)} planes of layer structure (Yang et al., 2016).

In order to determine the chemical valence of major elements, the X-ray photoelectron spectra of sample A, B, and C is analyzed and shown in Figure 4. In contrast with the samples, the O 1s, Ni 2p, Co 2p, and Mn 2p peaks have no obvious chemical shift. In Figure 4A, the peak of O 1s located at 529.54 eV can be indexed to the O^{2−} in the lattice of the samples. The Ni 2p_{3/2} XPS spectra with binding energy of 854.98 eV, corresponding to the Ni 2p_{3/2} peaks of Ni^{2+} and Ni^{3+} located at 854.0 ± 0.2 and 856.0 eV, respectively (Zhang M. et al., 2016). As shown in Figure 4C, the Co 2p_{3/2} binding energy peaks of the samples centers at 780.20 eV, which is agreed with the binding energy of Co^{3+} in LiCoO_{2} (Seteni et al., 2017). The Mn 2p_{3/2} peaks is 641.94 eV, which is consistent with the value of Mn^{4+} (Lou et al., 2017). So the chemical valences of O, Ni, Co, and Mn are−2, +2/+3, +3, and +4, respectively, which indicate that lithium sources cause no effect on valence states.

Figure 5A shows the initial charge/discharge curves of the sample A, B, and C at C/10 within the voltage window of 2.0 and

| TABLE 1 | Lattice parameters of 0.5Li_{2}MnO_{3}·0.5LiNi_{0.5}Mn_{0.3}Co_{0.2}O_{2} synthesized by different lithium sources. |
|---|---|---|---|---|
| Lithium sources | a/Å | c/Å | c/a | I_{003}/I_{104} |
| Sample A | LiOH·H_{2}O | 2.8643 | 14.2943 | 4.9905 | 1.5456 |
| Sample B | Li_{2}CO_{3} | 2.8610 | 14.2963 | 4.9970 | 1.3055 |
| Sample C | CH_{3}COOLi·2H_{2}O | 2.8671 | 14.3283 | 4.9975 | 1.4124 |

FIGURE 3 | TEM images of (A) Sample A, (C) Sample B, (E) Sample C and HRTEM images of (B) Sample A, (D) Sample B, (F) Sample C.

FIGURE 4 | X-ray photoelectron spectra of sample A, B, and C [(A) O, (B) Ni, (C) Co, (D) Mn].
4.8 V. During initial charge, two plateaus from 3.8 to 4.4 V and from 4.4 to 4.6 V are observed for all samples. The plateau from 3.8 to 4.4 V can be accorded with the oxidation of $\text{Ni}^{2+}/\text{Ni}^{3+}$ to $\text{Ni}^{4+}$ and $\text{Co}^{3+}$ to $\text{Co}^{4+}$, which is in good agreement with the reversible extraction of $\text{Li}^+$ from LiMO$_2$ ($M = \text{Ni, Co, Mn}$) phase (Seteni et al., 2017). Furthermore, the later plateau is attributed to the $\text{Li}^+$ and $\text{O}^{2-}$ irreversible extraction as $\text{Li}_2\text{O}$ from the inert Li$_2$MnO$_3$ phase, resulting in high irreversible capacity (Yang et al., 2016). When the voltage is from 3.8 to 4.4 V during the first charging cycle, the charging mechanism is written as (Lou et al., 2017),

$$0.5\text{Li}_2\text{MnO}_3 \cdot 0.5\text{LiNi}_{0.5}\text{Co}_{0.2}\text{Mn}_{0.3}\text{O}_2 \xrightarrow{\text{charge}} 0.5\text{Li}_2\text{MnO}_3 \cdot 0.5\text{Ni}_{0.5}\text{Co}_{0.2}\text{Mn}_{0.3}\text{O}_2 + 0.5\text{Li}^+ + 0.5\text{e}^-$$

When the voltage is from 4.4 to 4.6 V, the charging mechanism is as follows (Lou et al., 2017),

$$0.5\text{Li}_2\text{MnO}_3 \cdot 0.5\text{Ni}_{0.5}\text{Co}_{0.2}\text{Mn}_{0.3}\text{O}_2 \xrightarrow{\text{charge}} 0.5\text{MnO}_2 \cdot 0.5\text{Ni}_{0.5}\text{Co}_{0.2}\text{Mn}_{0.3}\text{O}_2 + 0.5\text{Li}_2\text{O}$$

The initial charge/discharge capacity and the coulombic efficiency of the sample A, B, and C are listed in Table 2. Obviously, the discharge capacity of the sample A is much higher than the others. Figure 5B shows rate performance of the Sample A, B, and C at different rates. It can be seen that the discharge capacities of all samples decrease as the rates increased due to the poor conductivity of the material and inert of Li$_2$MnO$_3$ (Seteni et al., 2017). However, the sample A shows much higher rate property because of smaller primary particle, which can shorten the $\text{Li}^+$ ions diffusion pathway. The midpoint voltage decay of the samples during cycling at rate of $C/10$ is shown in Figures 5C, D. The discharge capacities of sample A, B, and C reaches 231.8, 157.7, and 121.0 mA$\text{h}^{-1}$ after 30 cycles at $C/10$ rate, with capacity retentions of 88.4, 63.8, and 88.6%, respectively. Figure 5D indicates that sample A holds the most stable voltage from 3.71 to 3.43 V during cycling. However, a sudden drop appears for the midpoint voltage of Sample B after 20 cycles. The results indicate sample A keeps the most stable voltage during cycling.
Cyclic voltammetry curves of sample A, B, and C for the initial three cycles during the voltage range of 2.0–4.8 V at the scan rate of 0.1 mV s\(^{-1}\) are shown in Figure 6. The CV curves are similar. There are two oxidation peaks and two reduction peaks in the initial cycle. The oxidation peak at about 4.0 V is the oxidation of Ni\(^{2+}/Ni^{3+}\) to Ni\(^{4+}\) and Co\(^{3+}\) to Co\(^{4+}\) with the reversible extraction of Li\(^+\) from LiMO\(_2\) (M = Ni, Co, Mn) phase (Xiao et al., 2017). The oxidation peak at around 4.6 V corresponds to the Li\(^+\) and O\(^{2-}\) irreversible extraction from the Li\(_2\)MnO\(_3\) phase. In the following reduction process, the peak at about 3.7 V is the reduction of Ni\(^{4+}\) to Ni\(^{2+}/Ni^{3+}\) and Co\(^{4+}\) to Co\(^{3+}\), and the peak at around 3.2 V related to Li\(^+\) insertion into layered MnO\(_2\) (Seteni et al., 2017). Besides, the peaks of sample A and B (Figures 6A, B) are higher than sample C, indicated that sample A and B have more steady structure. It is also worth mentioning that, there are two peaks at about 2.9 V and 2.5 V for the sample C (Figure 6C), corresponding to oxidation peak and reduction peak, respectively, related to a small amount of spinel structure (Cao et al., 2017).

Electrochemical impedance spectroscopy (EIS) can be used to investigate the electrode kinetic process of samples A, B, and C. As shown in Figure 7, The EIS plots consist of a semicircle arc and a straight line. The semicircle arc at high frequency region corresponds to the charge transfer process, and the straight line at low frequency region is the lithium diffusion process. The plots are fitted using the electric equivalent circuit model, as shown in Figure 7. The parameters of the equivalent circuit are listed in Table 3. In the equivalent circuit, \(R_s\) and \(R_{ct}\) represent the solution resistance and charge-transfer resistance, respectively (He et al., 2015). CPE is related to capacitance of the surface layer (Toprakci et al., 2013). \(Z'\) represents the Warburg impedance (Xiao et al., 2017) (\(Z'\) is the real impedance and \(Z''\) is the imaginary impedance). It is observed that the \(R_s\) and \(R_{ct}\) of sample A (5.39 \(\Omega\), 87.93 \(\Omega\)) are smaller than those of sample B (7.968 \(\Omega\), 72.08 \(\Omega\)) and sample C (27.31 \(\Omega\), 120 \(\Omega\)). The results indicate that the solution resistance and charge-transfer resistance of sample A is the smallest, which mainly because sample A prepared by using LiOH-H\(_2\)O owns the smaller primary sizes (Cao et al., 2017). Hence, sample A exhibits the best electrochemical properties.

### Table 2 | Initial charge/discharge capacity and the Coulombic efficiency of the sample A, B, and C cathodes.

| Lithium sources       | Charge capacity (mAh g\(^{-1}\)) | Discharge capacity (mAh g\(^{-1}\)) | Initial Coulombic Efficiency (%) |
|-----------------------|----------------------------------|------------------------------------|----------------------------------|
| Sample A              | LiOH·H\(_2\)O                    | 382.9                              | 255.6                            | 66.8                             |
| Sample B              | Li\(_2\)CO\(_3\)                 | 388.9                              | 238.7                            | 61.4                             |
| Sample C              | CH\(_3\)COOLi·2H\(_2\)O          | 212.1                              | 142.1                            | 67.0                             |

### Table 3 | EIS fitting values of the samples A, B, and C.

| Samples   | Sample A         | Sample B         | Sample C         |
|-----------|------------------|------------------|------------------|
| \(R_s\)  | 5.39             | 7.968            | 27.31            |
| \(R_{ct}\)| 72.08            | 87.93            | 120              |

![Figure 6](image-url) Cyclic voltammetry curves of (A) Sample A, (B) Sample B, and (C) Sample C at the scan rate of 0.1 mV s\(^{-1}\).
CONCLUSIONS

In summary, 0.5Li2MnO3·0.5LiNi0.5Co0.2Mn0.3O2 materials have been successfully synthesized by using three kinds of lithium sources, LiOH·H2O, Li2CO3, and CH3COOLi·2H2O, respectively. The effects of morphology, structure, electrochemical performances of the 0.5Li2MnO3·0.5LiNi0.5Co0.2Mn0.3O2 material prepared by using different lithium sources have been investigated. 0.5Li2MnO3·0.5LiNi0.5Co0.2Mn0.3O2 material prepared by using LiOH·H2O shows the most uniform morphology with the particle diameters of about 200–300 nm and stable layer structure. It delivers the best electrochemical performances with the initial discharge capacity reaching 255.6 mAh g⁻¹ at C/10, and the capacity retention is 88.4% after 30 cycles at C/10. LiOH·H2O is the best choice for the synthesis of 0.5Li2MnO3·0.5LiNi0.5Co0.2Mn0.3O2 material compared with Li2CO3 and CH3COOLi·2H2O.

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AUTHOR CONTRIBUTIONS

All authors listed have made a substantial, direct and intellectual contribution to the work, and approved it for publication.

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SUPPLEMENTARY MATERIAL

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Conflict of Interest Statement: The authors declare that the research was conducted in the absence of any commercial or financial relationships that could be construed as a potential conflict of interest.

The handling Editor declared a shared affiliation, though no other collaboration, with one of the authors, M-ZL.

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