Synthesis and Lithium-ion Conductivity of Sr(La$_{1-x}$Li$_{3x}$)ScO$_4$ with a K$_2$NiF$_4$ Structure

Guowei ZHAO$^{a,b,§}$, Kota SUZUKI$^{a,c,d,§}$, Masaaki HIRAYAMA$^{a,c,d,§}$, and Ryoji KANNO$^{a,c,d,*§}$

$^a$ All-Solid-State Battery Unit, Institute of Innovation Research, Tokyo Institute of Technology, 4259 Nagatsuta, Midori-ku, Yokohama 226-8502, Japan
$^b$ College of Chemistry and Chemical Engineering, Huanggang Normal University, 438000 Huanggang, Hubei Province, China
$^c$ Research Center for All-Solid-State Battery, Institute of Innovation Research, Tokyo Institute of Technology, 4259 Nagatsuta, Midori-ku, Yokohama 226-8502, Japan
$^d$ Department of Chemical Science and Engineering, School of Materials and Chemical Technology, Tokyo Institute of Technology, 4259 Nagatsuta, Midori-ku, Yokohama 226-8502, Japan

* Corresponding author: kanno@echem.titech.ac.jp

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ECSJ Active Member
G. Zhao orcid.org/0000-0002-4344-3923
K. Suzuki orcid.org/0000-0002-2473-0724
M. Hirayama orcid.org/0000-0003-4804-4208
R. Kanno orcid.org/0000-0002-0593-2515
Synthesis and lithium-ion conductivity of Sr(\(\text{La}_{1-x}\text{Li}_{3x}\))\(\text{ScO}_4\) with a \(\text{K}_2\text{NiF}_4\) structure

Guowei ZHAO,\(^{a,b,§}\) Kota SUZUKI,\(^{a,c,d,§}\) Masaaki HIRAYAMA,\(^{a,c,d,§}\) Ryoji KANNO\(^{a,c,d,*,§}\)

\(^a\) All-Solid-State Battery Unit, Institute of Innovation Research, Tokyo Institute of Technology, 4259 Nagatsuta, Midori-ku, Yokohama 226-8502, Japan

\(^b\) College of Chemistry and Chemical Engineering, Huanggang Normal University, 438000 Huanggang, Hubei Province, China

\(^c\) Research Center for All-Solid-State Battery, Institute of Innovation Research, Tokyo Institute of Technology, 4259 Nagatsuta, Midori-ku, Yokohama 226-8502, Japan

\(^d\) Department of Chemical Science and Engineering, School of Materials and Chemical Technology, Tokyo Institute of Technology, 4259 Nagatsuta, Midori-ku, Yokohama 226-8502, Japan

*Corresponding author: kanno@echem.titech.ac.jp

§ ECSJ Active Member

ORCID:

Guowei ZHAO: 0000-0002-4344-3923

Kota SUZUKI: 0000-0002-2473-0724

Masaaki HIRAYAMA: 0000-0003-4804-4208

Ryoji KANNO: 0000-0002-0593-2515
ABSTRACT

K$_2$NiF$_4$-type oxides are expected to be potential lithium-ion conductors because they have a structure similar to that of perovskites, which provide a reasonably flexible framework for accommodating defects such as charge carriers within the lattice. However, the K$_2$NiF$_4$-type structure as a framework for lithium conductors is scarcely reported. This article presents the preparation of Sr(La$_{1-x}$Li$_{3x}$)ScO$_4$ with a K$_2$NiF$_4$ structure by a solid-state reaction at a high pressure of 2 GPa, and elucidates its lithium-ion-conducting properties. Sr(La$_{1-x}$Li$_{3x}$)ScO$_4$ forms solid solutions in the $x$ range of 0.05–0.20. Its orthorhombic lattice expands with lithium doping, indicating the incorporation of lithium ions as interstitial species in the structure. The highly doped samples exhibit high ionic conductivities (e.g., $4.66 \times 10^{-6}$ S cm$^{-1}$ at 250 °C for $x = 0.15$ and $4.29 \times 10^{-2}$ S cm$^{-1}$ at 375 °C for $x = 0.2$) with an activation energy of ~100 kJ mol$^{-1}$. The samples show an abnormal increase in the ionic conductivity at ~300 °C, possibly due to the increase in the orthorhombicity of Sr(La$_{1-x}$Li$_{3x}$)ScO$_4$. As the electronic conductivities of the developed oxide materials are a few orders of magnitude lower than their total conductivities, they can be used as solid electrolytes in all-solid-state lithium batteries. The study reveals that K$_2$NiF$_4$-type oxides are attractive candidates for developing novel lithium-ion conductors.
Keywords: Lithium-ion conductor, K$_2$NiF$_4$ structure, Solid electrolyte, All-solid-state lithium battery
1. Introduction

All-solid-state lithium batteries afford a stable, reliable, and safer mode of power storage with a higher power density than those of the conventional lithium-ion batteries based on flammable organic liquid electrolytes. Therefore, lithium-ion-conducting solid materials have recently received considerable attention. The materials science community has spent decades in the search for new solid lithium-ion conductors with better ionic conductivity and electrochemical stability. Among the available materials, sulfide- and oxide-based lithium-ion conductors have attracted significant interest. In particular, the lithium superionic conductor, Li₁₀GeP₂S₁₂, exhibits an extremely high ionic conductivity of 1.2 × 10⁻² S cm⁻¹ at room temperature. However, sulfide-based materials suffer from drawbacks such as incompatibility with oxide cathode materials and instability in ambient air, which impede their application in high-performance solid-state batteries. In contrast, oxide-based lithium-ion conductors are relatively stable in air, but have relatively low ionic conductivities compared to sulfides.

Among the oxide-based lithium-ion conductors, perovskite-type Li₃La(2/3−x)TiO₃ is considered to be a promising electrolyte candidate because of its high bulk ionic conductivity (>10⁻³ S cm⁻¹ at room temperature). However, the reduction of Ti⁴⁺ to Ti³⁺ at the interface with a low-potential anode during the electrochemical process results in
undesirable electronic conduction.\textsuperscript{6, 7} Therefore, the practical application of Li$_{3x}$La$_{(2/3)-x}$TiO$_3$ as a solid electrolyte has not yet been realized. To improve the electrochemical stability of perovskite-structured lithium-ion conductors, perovskite-type materials consisting of elements with high redox tolerance have been investigated.\textsuperscript{8–10} However, a material with a bulk ionic conductivity exceeding that of the Li$_{3x}$La$_{(2/3)-x}$TiO$_3$ system remains to be discovered.

To date, many lithium-ion conductors have been developed based on the perovskite structure, and certain perovskite-related structures (e.g., K$_2$NiF$_4$-type and Ruddlesden–Popper layered perovskites) have been suggested as potential frameworks for lithium conduction. Considering the general formula of these perovskite-related materials (A$_{n+1}$B$_n$X$_{3n+1}$), certain materials with stoichiometric compositions ($n = 2, 3,$ and 6) and non-stoichiometric compositions have been reported as lithium-ion conductors.\textsuperscript{11–13} The conductivities of these materials range from $10^{-7}$ to $10^{-3}$ S cm$^{-1}$ at $\sim$300 °C. Further, LiATiO$_4$ ($A =$ La, Eu) materials with $n = 1$ have been synthesized,\textsuperscript{14} and their lithium-intercalation capabilities as anodes have been reported.\textsuperscript{15} The lithium-intercalation capability of these materials is also related to the reducibility of Ti$^{4+}$ at the $B$ site. However, promising K$_2$NiF$_4$-type candidates that can serve as solid electrolytes have not yet been reported.
In this study, we investigated the lithium-ion conductivity of SrLaScO$_4$, which consists of alternating perovskite ($ABX_3$) and rock salt ($AX$) layers along the $c$-axis. This work focuses on a Sc$^{3+}$-based framework because Sc$^{3+}$ can facilitate a relatively high redox tolerance as the $B$-site cation.$^{9,16}$ Using the SrLaScO$_4$ framework, lithium doping was conducted for substituting the La$^{3+}$ species to produce lithium-ion-conducting materials with the chemical composition of Sr(La$_{1-x}$Li$_{3x}$)ScO$_4$. Herein, we report the synthesis and ionic conductivity of this novel K$_2$NiF$_4$-type lithium-ion-conducting oxide.

2. Experimental

Reagent-grade Li$_2$O (Kojundo Chemical Laboratory Co. Ltd., Sakado, Saitama Pre., Japan, ≥99 % purity), SrO (Alfa Aesar, Thermo Fisher Scientific, Waltham, MA, USA, ≥99.5 % purity), La$_2$O$_3$ (Kanto Chemical Co., Inc., Tokyo, Japan, ≥99.99 % purity), and Sc$_2$O$_3$ (Alfa Aesar, Thermo Fisher Scientific, Waltham, MA, USA, ≥ 99.9 % purity) were used as precursors for preparing Sr(La$_{1-x}$Li$_{3x}$)ScO$_4$ by a high-pressure solid-state reaction. Typically, the precursor mixtures at stoichiometric molar ratios were ground and enclosed in platinum capsules. Then, they were reacted under a high pressure of 2 GPa by heating at 1000–1200 °C for 30 min. The reaction was quenched by cooling the mixture to room temperature. This high-pressure synthetic route ensured the incorporation of an accurate amount of Li, apart from accelerating the reaction.
The obtained products were crushed to powders, and then characterized by powder X-ray diffraction (XRD; Rigaku, Smart Lab., Tokyo, Japan) using Cu Kα1 radiation (\(\lambda = 1.5418\) Å, 45 kV, 200 mA). The diffraction data were collected at room temperature over the 2θ range of 10–70° with a step width of 0.01°. The synchrotron diffraction data were collected at 25 °C using a Debye–Scherrer camera with an imaging plate at the BL19B2 beamline of SPring-8 (1-1-1, Kouto, Sayo-cho, Sayo-gun, Hyogo 679-5198, Japan). The powder samples for XRD measurements were sealed in Lindermann glass capillaries (~0.3 mm). The incident-beam wavelength was calibrated using NIST SRM Ceria 640b CeO₂, and was fixed at 0.5 Å. The synchrotron XRD data were refined by the Rietveld method using the RIETAN-FP package.\(^{17}\)

The conductivity of the sample was determined via alternating current (AC) impedance spectroscopy in the frequency range of \(10^{-1}\) to \(10^7\) Hz using a frequency response analyzer (Solartron Analytical, Solartron 1260, Leicester, England). A disk-shaped sample (diameter ~3 mm; thickness ~1 mm) was prepared for conductivity measurements; the sample surfaces were polished with emery paper. Gold plates were used as blocking electrodes on both sides of the sample. The measurements were conducted under Ar flow between 25 and 400 °C with an applied voltage of 10–100 mV. The electronic conductivity of the sample was evaluated by direct current (DC) polarization.
measurement; the voltage, which was controlled using a potentiostat electrochemical interface (Solartron Analytical, Solartron 1287, Leicester, England), was varied from 0.1 to 1.8 V.

Differential thermal analysis (DTA) was performed using a differential thermal analyzer (Rigaku, Thermo Plus TG 8120, Tokyo, Japan) in the temperature range of 27 to 500 °C under a dry Ar atmosphere with a heating and cooling rate of 5 °C/min.

3. Results and discussion

3.1. Phase identification and crystal structure

The lithium-doped SrLaScO$_4$-layered perovskite oxides with the K$_2$NiF$_4$ structure were prepared by a high-pressure solid-state method. Figure 1 shows the XRD patterns of the obtained oxides, Sr(La$_{1-x}$Li$_x$)ScO$_4$ ($x = 0.05, 0.10, 0.15,$ and $0.20$). The main peaks in all the patterns could be indexed to the orthorhombic K$_2$NiF$_4$ lattice with the space group of Abma (No. 64). This result indicated the formation of a K$_2$NiF$_4$-type phase of the obtained oxides. Weak diffraction peaks originating from impurities (e.g., LaScO$_3$) were also observed at lower angles, especially for samples with higher $x$ values.

The prepared samples were also subjected to synchrotron XRD, and the obtained data were refined by the Rietveld method. A single-phase model based on K$_2$NiF$_4$-structured
SrLaScO$_4$ (space group: Abma) was used as the initial structure for the structural refinement; the model used Sr at 8$f$, La at 8$f$, Sc at 4$a$, O1 at 8$e$, and O2 at 8$f$ sites.\textsuperscript{18} The 2$\theta$ regions with the peaks of impurities were excluded from the refinement.

In a previously reported study,\textsuperscript{18} the crystal structure of LaSrScO$_4$, i.e., the original non-doped composition, was analyzed under the condition that each occupancy was fixed to a nominal value, over a wide temperature range of 300–1273 K. Accordingly, the occupancies of Sc (4$c$), O1 (8$e$), and O2 (8$f$) were fixed as 1.0 in the refinement process employed in this study.

For the refinement of the crystal structure, the Li location was fixed at A sites (8$f$), and the occupancy of the 8$f$ site was fixed as $g$(Sr) = 0.5, $g$(La) = (1.0–$x$)/2, and $g$(Li) = 0.5$x$ based on nominal compositions, while maintaining the total site occupancy at 1.0.

Therefore, a fraction of the doped lithium ions was not considered in the refinement process. As the X-ray scattering factor of lithium is weak, a precise structural analysis, including the details of lithium-ion locations, is difficult. For this reason, the interstitial lithium was not considered in the structural analysis. Additionally, we attempted to refine the lithium occupancy and found that the lithium-ion occupancy refinement has no severe effects on the parameters $R_{wp}$, $S$, etc., revealing the validity of the analysis. The Rietveld refinement patterns and refined structural parameters are shown in Figure S1 and Table
S1, respectively, of the supporting information. The crystal structure based on the refinement results is presented in Figure 2. In the crystal structure, the corner-shared ScO$_6$ units form a perovskite layer in the $ab$ plane, and an AO rock-salt layer resides between the perovskite layers. The perovskite and rock-salt layers are alternately stacked in the $c$-axis direction to form a K$_2$NiF$_4$-type structure.

Based on the refinement results, we determined that the single-phase model provides a good fit with lower $R_{wp}$ values ($R_{wp} < 6.7$) and $S$ values ($S < 2.9$). Figure 3 shows the composition-dependence of the lattice parameters determined by the Rietveld analysis. All the parameters of the doped samples are found to be larger than those of the undoped SrLaScO$_4$ lattice ($x = 0$). This result indicates that the volume of the orthorhombic lattice expanded with the doping of Li ($x = 0.05$ to 0.20). Although the composition did not exhibit an ideal linear relationship with the lattice parameters, a trend of lattice expansion following lithium doping, that is, an increase in the volume with an increase in $x$, was confirmed. These changes indicate the formation of a solid solution through lithium doping for $x$ ranging from 0.05 to 0.20. As lithium doping can be viewed as the replacement of one La ion with three Li ions (La$^{3+}$→3Li$^+$), the observed increase in the lattice volume ($V$) could be attributed to the introduction of lithium ions as interstitials in the SrLaScO$_4$ lattice. According to the ionic radius of Li$^+$ (0.76 Å; coordination number: 6)
and La$^{3+}$ (1.03 Å; coordination number: 6), lattice contraction is expected; however, the two additional Li$^+$ ions that occupy the interstitial sites may contribute to lattice expansion. A larger increase in the $c$-axis length (+0.024 Å), compared to those of the other two axes ($a$-axis: +0.0066 Å and $b$-axis: +0.0103 Å), also supports the presence of interstitial lithium species in the rock-salt layer between the perovskite layers. In the case of oxide-ion conductors (A$_2$BO$_{4+δ}$), excess O in the rock-salt layer contributes to high conductivities; therefore, the interstitial lithium species in the rock-salt layer can be expected to improve the lithium-ion conductivity of the obtained material.

Because the obtained samples contained impurities, the compositions and structures could not be analyzed precisely. Therefore, in the near future, neutron diffraction measurements and data analysis will be conducted to reveal the interstitial lithium sites and the precise compositions.

3.2. Ionic conductivity

The ionic conductivities of the samples were determined by AC impedance measurements. Figure 4 shows the typical complex impedance spectra of Sr(La$_{1-x}$Li$_{3x}$)ScO$_4$ ($x = 0.1$) at several temperatures. These spectra are composed of a
semicircle at high frequencies and a spike at lower frequencies, which correspond to the sum of the bulk and grain boundary resistances, and electrode resistance, respectively. The black curves shown in Figure 4, indicating the fitting results for a typical sample with $x = 0.1$, were evaluated using the equivalent circuit model $R_{\text{total}} \text{CPE}_{\text{total}} W_{s1}$, where $R_{\text{total}}$, $\text{CPE}_{\text{total}}$, and $W_{s1}$ represent the total (sum of the bulk and grain boundary) resistance, charge transfer resistance, constant phase element, and Warburg resistance, respectively. The spectra obtained using electrochemical impedance spectroscopy were fitted using the Zview software, and the analyzed results are presented in Table S2 and S3. The high-frequency semicircle yields a capacitance value of the order of $10^{-11}$ F, indicating that these assumptions are reasonable. According to these data, the ionic conductivity of each sample was calculated from the sum of the bulk and grain boundary resistance values, $R$, which is determined as the $x$-axis intercept of the semicircular plot. The ionic conductivities were calculated using the relationship, $\sigma = d/(R \times A)$, where $R$ is the resistance, $d$ is the thickness, and $A$ is the area of the pellet.

Figure 5(a) shows the temperature-dependence of the conductivities of the Sr(La$_{1-x}$Li$_x$)ScO$_4$ samples. The temperature-dependent conductivity data of the oxides in the lower-temperature region (<270 °C) could be fitted with the conventional Arrhenius law. In this region, the ionic conductivity of the oxide increased with an increase in $x$ from
0.05 to 0.15, whereas the sample with \( x = 0.2 \) showed a slightly decreased ionic conductivity. The activation energy \( (E_a) \) calculated from the Arrhenius plots of ion conduction decreased from 117.6 to 94.3 kJ mol\(^{-1}\) with an increase in \( x \) from 0.05 to 0.10. Furthermore, the samples with higher \( x \) values \((x > 0.1)\) also showed relatively low activation energies \((103.4 \text{ kJ mol}^{-1} \text{ at } x = 0.15 \text{ and } 99.3 \text{ kJ mol}^{-1} \text{ at } x = 0.2)\). These results indicate that the lithium doping of \( \text{SrLaScO}_4 \) results in lithium-ion-conducting properties, which can be enhanced by increasing the carrier number \((\text{i.e., } x \text{ value})\). Figure 5(b) presents the ionic conductivity of \( \text{Sr(La}_{1-x}\text{Li}_x\text{ScO}_4 \) at 250 °C as a function of the doped \( \text{Li}^+ \) content. In the lower-temperature range, the highest conductivity of \( 4.66 \times 10^{-6} \text{ S cm}^{-1} \) at 250 °C was observed at \( x = 0.15 \). The next highly doped sample with \( x = 0.2 \) exhibited a slight decrease in conductivity because of the large fraction of the impurity phase. However, its conductivity was still higher than those of the samples with lower \( \text{Li}^+ \) contents \((x = 0.05 \text{ and } 0.10)\).

In the higher-temperature range above 300 °C, all the samples showed deviations from the Arrhenius law. In particular, the samples with \( x = 0.10–0.20 \) showed a significant change in conductivity at ~350 °C, indicating a phase transition of the oxide to a highly conductive high-temperature phase.\(^{20}\) In the case of non-doped \( \text{SrLaScO}_4 \), the orthorhombicity \((\text{i.e., the difference between the lattice parameters } a \text{ and } b)\) of the lattice
is known to increase during the heating, reaching a maximum near 300 °C. Further heating above 500 °C decreases the orthorhombicity, followed by a phase transition from orthorhombic (Abma) to tetragonal (I4/mmm) at temperatures exceeding 800 °C, confirmed in previous studies. Therefore, the phase transition of our lithium-doped samples occurring at ~300 °C was possibly caused by the increase in the orthorhombicity. The high-temperature XRD measurements were performed to confirm this hypothesis (Figure S2). In the XRD patterns, the peak corresponding to the 113 reflection shows a reversible peak shift to lower and higher angles, respectively, during heating and cooling processes (i.e., thermal expansion and contraction occurred during the heating and cooling processes, respectively.). By contrast, the diffraction peaks related to the $a$ and $b$ axes do not show a similar trend, and the peak position do not show a significant change during the heating and cooling processes. Furthermore, the 200 and 020 peaks are partly overlapped in the entire temperature range. Therefore, we analyzed the lattice parameters via Rietveld refinements (Figure S3). The difference between $a$ and $b$ value increased, especially, at a temperature over 200 °C, and reached a maximum value at 350 °C. Based on these results and those reported previously, we confirmed that the observed abnormal increase in the ionic conductivity in this temperature range could be related to the increase in the orthorhombicity of the crystal structure. The gradual shift of the
abnormal behavior to a lower temperature with increasing $x$ suggests that lithium doping affects the change in the orthorhombicity of the oxide structure.

The ionic conductivities at temperatures higher than 300 °C are also plotted in Figure 5(b). In the higher-temperature range, the highly doped sample ($x = 0.20$) showed conductivities comparable to those of the sample with $x = 0.15$; the highest conductivity of $4.29 \times 10^{-2} \text{ S cm}^{-1}$ was observed at 375 °C. These conductivities are one order of magnitude higher than those of the other similar $B$-site (Sc) perovskite systems; for example, the ionic conductivity of $(\text{Li}_{0.4}\text{Ce}_{0.15}\text{La}_{0.67})\text{ScO}_3$ is $1.1 \times 10^{-3} \text{ S cm}^{-1}$ at 350 °C. In addition, ionic conductivities of $>1.0 \times 10^{-3} \text{ S cm}^{-1}$ at 300 °C have been reported for Ruddlesden–Popper layered perovskites. These results suggest that the $K_2\text{NiF}_4$-type crystal structure is a good framework for developing novel ionic conductors, and that further improvements in the ionic conductivities can be achieved using this crystal structure.

Figure 6 shows the DTA curves of Sr(La$_{0.85}$Li$_{0.15}$)ScO$_4$ recorded during the heating and cooling processes, along with the synchrotron diffraction data recorded before and after the heat treatment. During the heating cycle, a clear endothermic peak appeared at 363 °C. This peak could be attributed to the phase transition related to the change in the orthorhombicity, as confirmed by the high-temperature XRD analysis.
cooling cycle, an exothermic peak was observed at 316 °C, which might have originated because of the reverse phase transition. The synchrotron diffraction patterns also did not indicate an apparent change in the crystal structure for the K$_2$NiF$_4$ phase. The reversible phase transition, confirmed by the DTA and high-temperature XRD analyses, could cause the ionic conductivity to increase abnormally at ~300 °C. Although the difference in the endothermic or exothermic temperatures in the DTA data indicated the occurrence of hysteresis, the conductivity data during heating and cooling processes showed negligible hysteresis (Figure S4). Therefore, the observed temperature difference possibly arose from the relatively high temperature scanning rate used in the DTA (5 °C/min) than that employed in the conductivity measurements.

Further, the electronic contribution to the total conductivity of Sr(La$_{0.85}$Li$_{0.45}$)ScO$_4$ was evaluated via DC polarization measurements. Figure 7 shows the steady-state current as a function of the applied voltage at 375 °C. The calculated electronic conductivity (1.1 × 10$^{-5}$ to 1.8 × 10$^{-5}$ S cm$^{-1}$ at 375 °C) could be considered negligible as it is three orders of magnitude lower than the total conductivity (3.4 × 10$^{-2}$ S cm$^{-1}$) at 375 °C, obtained by AC impedance spectroscopy. Therefore, the total conductivity of Sr(La$_{0.85}$Li$_{0.45}$)ScO$_4$ is mainly due to ion diffusion; that is, the obtained sample was a pure ionic conductor. Although the lithium doping in SrLaScO$_4$ could result in the introduction of O vacancies
and/or reduction of Sc$^{3+}$, causing electronic conductivity, no such phenomena were observed.

4. Conclusions

Lithium-ion-containing oxides, Sr(La$_{1-x}$Li$_3x$)ScO$_4$ with a K$_2$NiF$_4$ (A$_2$BO$_4$) structure, were prepared using a high-pressure solid-state reaction, and their lithium-ion conductivities and crystal structures were investigated. Despite the presence of small amounts of impurities, a K$_2$NiF$_4$-type solid solution was found to be the dominant phase formed at the examined compositions. The Sr(La$_{1-x}$Li$_3x$)ScO$_4$ system was found to exhibit ionic conductivity exclusively with an abnormal increase in the conductivity at ~300 °C, caused by phase transition. The generation of lithium interstitials and lattice expansion could be the main factors that contribute to the ionic conductivity of this oxide system. The highest ionic conductivity was obtained at a relatively large $x$ value; for example, an ionic conductivity of $1.5 \times 10^{-2}$ S cm$^{-1}$ at 350 °C was achieved for $x = 0.15$. In the low-temperature range (<270 °C), the conductivity was thermally activated ($\sigma = \sigma_0 e^{-E_a/kT}$) with an activation energy, $E_a$, of ~100 kJ mol$^{-1}$. The obtained materials could be used as templates to further improve the ionic conductivities of other materials to realize high lithium-ion conductivities at low temperatures for practical applications. The findings of this study are expected to contribute to the development of superior lithium-based solid
electrolytes for all-solid-state lithium batteries, and the ongoing material research is expected to expand the diversity of materials that can be applied in batteries. Among the structural frameworks for lithium conductors, the K$_2$NiF$_4$ structure can be one of the important candidates for further material research.

Supporting Information

Supporting Information is available on the website at DOI: xxxxxxxxxxxxxx.

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Declaration of interest

None

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Figure captions

Figure 1 X-ray diffraction patterns of Sr(La\textsubscript{1−x}Li\textsubscript{3x})ScO\textsubscript{4} (x = 0.05, 0.1, 0.15, 0.20). Reference data of SrLaScO\textsubscript{4} (ICSD: 172917) is plotted at the bottom.

Figure 2 Crystal structure of Sr(La\textsubscript{1−x}Li\textsubscript{3x})ScO\textsubscript{4} (x = 0.15), obtained using the Rietveld refinement technique. The black spheres represent the A sites, where Sr, La, and Li co-exist, whereas the gray spheres denote O atoms. The gray octahedra represents ScO\textsubscript{6}.

Figure 3 Lattice parameters of Sr(La\textsubscript{1−x}Li\textsubscript{3x})ScO\textsubscript{4} determined by the Rietveld analysis of the XRD data. Reference data of SrLaScO\textsubscript{4} (ICSD: 172917) is also plotted for x = 0.

Figure 4 Typical impedance plots for the representative Sr(La\textsubscript{1−x}Li\textsubscript{3x})ScO\textsubscript{4} (x = 0.1) sample (a) and a magnified plot (b).

Figure 5 Temperature-dependence of the conductivities of Sr(La\textsubscript{1−x}Li\textsubscript{3x})ScO\textsubscript{4} (x = 0.05, 0.1, 0.15, and 0.20) (a) and the composition-dependence of the conductivities at 250 °C (b).

Figure 6 (a) DTA curves of Sr(La\textsubscript{0.85}Li\textsubscript{0.45})ScO\textsubscript{4} recorded in a dry Ar atmosphere and (b)
synchrotron XRD patterns of Sr(La$_{0.85}$Li$_{0.15}$)ScO$_4$ before and after the DTA analysis.

Figure 7 Steady-state currents as a function of the applied voltage at 375 °C in a dry Ar atmosphere for Sr(La$_{0.85}$Li$_{0.15}$)ScO$_4$. 
Figure 1

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