First-Principles Investigation of the Li–Fe–F Phase Diagram and Equilibrium and Nonequilibrium Conversion Reactions of Iron Fluorides with Lithium

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We have used density functional theory (DFT) to investigate the ternary phase diagram of the Li–Fe–F system and the reactions of Li with iron fluorides. Several novel compounds, not previously identified in the Li–Fe–F system, are predicted to be stable. Electrochemical voltage profiles, derived from the evolution of the Li chemical potential in the calculated phase diagram, are in reasonable agreement with experimental trends. The effect of particle size on the Fe that precipitates when Li2FeF6 reacts with Li is also investigated. We find that when 1 nm Fe particles form, the potential for this reaction is considerably reduced from its bulk value and relate this to the experimental observations. Furthermore, we formulate a model for the significant hysteresis that is observed in the lithiation and delithiation of Fe3F. Nonequilibrium paths derived by assuming much faster diffusion of Li than Fe are in reasonable agreement with experimental profiles. Our kinetic model predicts that the iron fluoride reaction follows a different path through the phase diagram during conversion (discharge) and reconversion (charge), which results in the voltage profile hysteresis observed during experiment. The proposed kinetic model also explains why upon extraction of Li from a 3/1 mixture of LiF and Fe a rutile FeF2-like structure can form, even when iron should be oxidized to Fe3+ by extraction of three Li+ per Fe.

I. Introduction

The lithium ion (Li-ion) battery has become the premier technology for portable power because of its ability to meet the ever-increasing power and energy density demands of modern electronic devices.1 The principle by which Li-ion chemistry moves electrons through an external circuit relies upon Li+ insertion into one electrode material while extracting Li+ from the opposing electrode material.2,3 Commonly referred to as the “rocking chair” mechanism, this process requires the electrodes to be composed of host materials that remain structurally and electrochemically stable during repeated charge/discharge cycles. This requirement has practically restricted the reversible reaction to at most a single electron transfer per formula unit, in turn limiting the energy density that can be achieved with current anodic (e.g., graphite) and cathodic (e.g., LiCoO2, LiNiO2, LiFePO4) insertion materials.

An alternative to Li+ insertion reactions is the displacement or “conversion” reaction in which the active electrode material, MXy, is consumed by Li+ and reduced to the metal, M0, and a corresponding lithium compound, Liy/X:

\[
M^{x+}X_y + yLi^+ \rightarrow M^0 + yLi_yX
\]

In eq 1, M represents the metal cation and X represents the anion. These reactions can make use of all energetically favorable valence states of the metal cation, enabling a large theoretical energy density. Reversible conversion reactions have been demonstrated with a variety of materials including metal oxides,4,5 metal nitrides,6–8 metal sulfides,9,10 and metal fluorides.11–15 Because the conversion reaction potential is directly proportional to the strength of the bond ionicity only metal fluorides have a high enough potential to be used as Li-ion cathodes. Although the very ionic nature of the


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metal–fluorine bond produces a high reaction potential, the corresponding wide band gap results in electronically insulating behavior, which has inhibited the use of metal fluoride electrodes until recently. A variety of metal fluorides (e.g., those based on Fe, Co, Ni, Cu, Bi) exhibited significant electrochemical activity when prepared as a nanocomposite consisting of small particles (on the order of ~1–20 nm) embedded in a conductive matrix (e.g., carbon, V₂O₅, MoS₂). Use of fine nanoscale metal fluorides minimizes the length of the ion diffusion path while the conductive matrix serves to enhance and maintain interparticle electronic and ionic conductivity during prolonged cycling.

One well-studied metal fluoride that typifies conversion reaction behavior is iron fluoride. It has been demonstrated that carbon metal fluoride nanocomposites containing 85/15 wt % FeF₃/C exhibit a reversible capacity of 600 mAh/g (FeF₃ theoretical capacity = 712 mAh/g) between 4.5 and 1.5 V versus Li/Li⁺. This performance is roughly 4 times the specific capacity and nearly 3 times the gravimetric energy density of LiCoO₂. Investigation of the FeF₃ conversion mechanism by electrochemical and structural analysis has lead other authors to ascertain that initial Li⁺ insertion occurs at ~3.3 V versus Li/Li⁺ according to the reaction

\[ \text{Fe}^{3+}\text{F}_3 + \text{Li}^+ \rightarrow \text{LiFe}^{2+}\text{F}_3 \]  

This portion of the reaction has been found to take place readily and be fully reversible, whereas the subsequent step, observed at ~2 V versus Li/Li⁺, is thought to correspond to the reaction

\[ \text{LiFe}^{2+}\text{F}_3 + 2\text{Li}^+ \rightarrow \text{Fe}^0 + 3\text{LiF} \]  

This segment of the overall reaction occurs with a significant voltage drop, and the amount of capacity obtained is extremely rate and temperature-sensitive. Observations such as these lead to speculation that conversion reactions are kinetically, rather than thermodynamically, hindered. Reconversion of the Fe and LiF nanocomposite to iron fluoride takes place with significant polarization, resulting in considerable hysteresis of the voltage profile, a characteristic common to many conversion reactions. Reconversion of LiF and Fe is also quite intriguing because a capacity equivalent to three Li⁺ per Fe is obtained although an “FeF₂-like” structure is produced instead of the initial FeF₃ structure. In addition, features of the electrochemical profile for an FeF₃/C nanocomposite conversion/reconversion process appear quite similar to those obtained from charge/discharge cycles of the FeF₃/C nanocomposite (particularly cycles subsequent to the first). Since it is difficult to reconcile how three Li⁺ can be extracted per Fe, when Fe is only oxidized to +2 in FeF₂, mechanisms involving interfacial or surface charge storage have been proposed. At the present time, the details of this process are poorly understood primarily because of the analytical limitations of current experimental instrumentation encountered when examining such fine nanostructures.

The work presented herein demonstrates the application of first-principles calculations to the study of conversion reactions by specifically investigating the iron fluoride conversion mechanism. In recent years, first-principles calculations have been used to accurately predict and explain a variety of properties pertinent to Li-ion insertion materials such as average insertion potential(s), voltage profile, phase stability, and lithium and electron mobilities. In this paper, we use first-principles methods to investigate the stable and metastable phases in the Li–Fe–F ternary system and predict plausible reaction paths. We also explain the apparent paradox how three Li⁺ can be extracted from the discharged state while forming an FeF₂-like structure upon reconversion. Finally, we also propose a model for the significant hysteresis observed in the voltage profile of iron fluoride, and for conversion reactions in general.

II. Computational Methodology and Crystal Structures

The first-principles calculations were conducted within the formalism of density functional theory (DFT) and the generalized gradient approximation (GGA) of the exchange-correlation function as formulated by Perdew, Burke, and Ernzerhof. Pseudopotentials generated by the projector-augmented wave (PAW) method were utilized as implemented by the Vienna ab Initio Simulation Package (VASP). The pseudopotentials utilized the valence state 1s²2s²2p⁴ for Li, 3p³3d⁴s¹ for Fe, and 2s²2p⁵ for F. Planar convergence to less than 5 meV/atom was achieved by using an energy cutoff of 550 eV, higher than the default value of 400 eV for fluorine, and Brillouin zone integration was performed on a 4 x 4 x 4 grid for all metal fluorides and a 12 x 12 x 12 grid for Li and Fe. Minimization of the total energy was realized with a...
Magnetic (AFM) ground states below the Néel temperature (T_N), 363 and 79 K, respectively. The AFM state is well below room temperature (RT) for FeF_2, and recent findings indicate that T_N drops to 39 K for FeF_3 as the crystallite size is reduced to amorphous nature, so the FM configuration may not result in significant error. It should also be noted that the spin-magnetic moment obtained with FM configuration (e.g., FeF_2 = 3.70 μ_B/Fe^{2+}) is in good agreement with both AFM GGA calculations by other authors (e.g., FeF_2 = 3.51 μ_B/Fe^{2+}) and the experimental value (e.g., FeF_2 = 3.75 μ_B/Fe^{2+}). The total energy of F, used to derive the ternary phase diagram defined by Li, Fe, and F at 0 K, was determined from the experimental reaction enthalpy obtained for Mg + F_2 → MgF_2

and the calculated energies for Mg and MgF_2. This enables the total energy of fluorine to be determined by fitting the total energies of an insulator (i.e., MgF_2) and a nontransition metal (i.e., Mg) to the experimental reaction enthalpy and minimizes ab initio errors arising from the metal to anion charge transfer.

FeF_3 possesses a trigonal structure with R-3c space group symmetry (S.G. no. 167), which is common to the majority of 3d metal trifluorides. The hexagonal representation of the lattice, shown in Figure 1A, is composed entirely of corner-sharing FeF_6 octahedra. It is akin to the ABX_3 cubic perovskite wherein the “A” site cation has been removed, yielding a collapsed BX_3 structure in which the M–F–M bond angle has decreased from the ideal 180° (i.e., that observed for ReO_3 structure). As a result of the “A” site cation vacancy, continuous channels form along the a and b vectors of the unit cell, which result in vacant planes parallel to the (012) plane of FeF_6 octahedra in the hexagonal setting. Because of the relatively small ionic radii for Li^+ (i.e., 0.59 Å for 4-fold coordination and 0.76 Å for 6-fold coordination) it is reasonable to expect that these channels can accommodate up to one Li^+ per FeF_3 formula unit via topotactic insertion.

The majority of 3d metal difluorides, including FeF_2, display a tetragonal structure with space group symmetry P4_2/mnm (S.G. no. 136). This structure, shown in Figure 1B, is of the rutile type and is composed of an hcp anion lattice with cations occupying half of the octahedral sites to form FeF_6 octahedra. These octahedra link in an edge-sharing manner along the [001] to form alternating chains of FeF_6 octahedra and vacant channels. The channels along the [001] can reasonably be expected to accommodate Li^+ coordinated in either tetrahedral or octahedral sites. However, Li^+ insertion into the most common form of rutile, TiO_2, has been found to be quite difficult due to strong anisotropic effects from Li–Li interaction in the crystal structure upon increasing Li^+ concentration. In addition, TiO_2 makes use of the Ti^{4+}/F^{−} couple, whereas initial Li^+ insertion into FeF_2 is restricted because Fe^{2+}

Figure 1. Panel A depicts the hexagonal representation of the FeF_3 unit cell with 6a site vacancies parallel to the (012) plane. Panel B illustrates the rutile structure of FeF_2, highlighting the vacant channels along the [001]. The unit cell is outlined in solid, black lines for each lattice.

For Li^+ insertion to occur, some Fe^{2+} would need to be reduced to either Fe^{1+} or disproportionation to Fe^{0} and Fe^{3+} would need to take place. In general, solid-state Fe^{1+} is very unlikely and oxidation states <2+ are uncommon for elements of the first transition series unless a π-acid-type ligand (e.g., CO and NO) is present, or in some organometallic compounds. If disproportionation occurs, combining some Fe^{2+} reduction to Fe^{0} with some Fe^{2+} oxidation to Fe^{3+}, then metallic iron would likely precipitate out of the iron fluoride structure. Therefore, this mechanism would not be a strict insertion reaction because it requires an exchange of iron with lithium. During this process it would be possible for Li^+ to either fill an interstitial site in the channels, occupy the site that Fe^{3+} is removed from, or occupy space that is a combination of both limits. Additional possibilities exist such as charge-compensated alovalent exchange, in which two Li^+ exchange for one Fe^{2+} of the rutile structure.

As illustrated by the above discussion of the FeF_2 structure, the variety of possibilities in which a structure can be altered during a reduction is required. For Li^+ insertion to occur, some Fe^{2+} would need to be reduced to either Fe^{1+} or disproportionation to Fe^{0} and Fe^{3+} would need to take place. In general, solid-state Fe^{1+} is very unlikely and oxidation states <2+ are uncommon for elements of the first transition series unless a π-acid-type ligand (e.g., CO and NO) is present, or in some organometallic compounds. If disproportionation occurs, combining some Fe^{2+} reduction to Fe^{0} with some Fe^{2+} oxidation to Fe^{3+}, then metallic iron would likely precipitate out of the iron fluoride structure. Therefore, this mechanism would not be a strict insertion reaction because it requires an exchange of iron with lithium. During this process it would be possible for Li^+ to either fill an interstitial site in the channels, occupy the site that Fe^{3+} is removed from, or occupy space that is a combination of both limits. Additional possibilities exist such as charge-compensated alovalent exchange, in which two Li^+ exchange for one Fe^{2+} of the rutile structure.

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conversion reaction makes the search for thermodynamically stable lithiated structures significantly more difficult than predicting structures based strictly upon Li$^+$ insertion into interstitial sites. Table 1 illustrates the various approaches undertaken in this paper to determine stable, lithiated phases of iron fluoride. More than 100 compounds have been computed in order to investigate the probability of not only Li$^+$ insertion into FeF$_3$ (i.e., filling interstitial space) and direct FeF$_2$ conversion (i.e., exchanging iron for lithium) but also the possibility of reaction intermediates resulting from both partial lithium insertion and conversion. Additionally, LiF supercells containing minimal amounts of substituted iron were constructed in an effort to find the intermediates one might expect upon initial reconversion. A variety of compositions based on spinel and olivine structures have also been examined because of their close structural relation to rocksalt (e.g., LiF) and because consultation of structure–field maps for ternary compounds$^{50}$ indicates that these may be favorable structures based on ionic radii of lithium and iron.

### III. Li–Fe–F Ternary Phase Diagram and the Equilibrium Reaction Path

The ternary Li–Fe–F phase diagram (plotted with LiF on one vertex, rather than Li, to focus on the relevant area) is presented in Figure 2. This plot has been constructed using the structure with lowest energy at each composition. The compositions marked by filled circles are part of the convex lowest energy hull, indicating that they are overall stable and have lower energy than any linear combination of other structures that add up to the same composition. In a ternary phase diagram at zero K, the phase regions are triangles indicating coexistence of three phases. At the compositions marked with an “×,” inside a stable three-phase triangle, one or more structures were calculated but found to have higher energy than a mixture of the three compounds that define the triangle. Although such compounds are therefore thermodynamically unstable, they can be metastable if diffusion of some components is not fast enough to form the stable phases or if there are nucleation limitations for some of the compounds. Given that the chemical potential of all species is constant in each three-phase equilibrium region the Li chemical potential and voltage can be calculated as function of composition. For instance Figure 2 shows that any composition within the tie-triangle formed by LiF, Fe, and Li$_{1/2}$FeF$_3$ will possess a potential of 2.91 V versus Li metal under equilibrium conditions. All of the other three-phase regions have higher Li potentials. A potential for the three-phase equilibria connecting to pure F (shaded in gray) is not given as we did not systematically search for stable structures in this composition space where the valence state of Fe would be very large.

An expanded view of the phase diagram area containing the relevant stoichiometries is presented in Figure 3, while the crystallographic information for the optimized structures of the lithiated iron fluorides is presented in Table 2. The experimentally observed structures, rocksalt, rutile, and distorted perovskite, were utilized for the binaries LiF, FeF$_3$, and FeF$_2$, respectively. The only experimentally known ternary compound within the Li–Fe–F system to have been directly synthesized is a trirutile of composition Li$_{1/2}$FeF$_3$; however, another experimental effort indicated that electro-chemical formation of Li$_{1/2}$FeF$_3$ (via Li$^+$ insertion into FeF$_3$) is also possible when $x \leq 1$. Results of our study do indicate that Li$^+$ insertion into FeF$_3$ forms thermodynamically stable structures, but only up to about $x = 0.25$ as

\begin{table}[h]
\centering
\begin{tabular}{|c|c|}
\hline
structural basis for calculation & no. of compounds calculated \\
\hline
perovskite-related Li$_4$FeF$_6$ & $\sim$40 \\
Fe/Li exchange in FeF$_3$ & $\sim$10 \\
FeF$_2$ partial insertion/exchange & $\sim$10 \\
spinel- and olivine-type Li$_2$FeF$_4$ compositions & $\sim$15 \\
LiF substituted with Fe & $\sim$15 \\
\hline
\end{tabular}
\caption{A Significant Number of Computations Have Been Conducted in Order to Survey the Great Variety of Theoretically Possible Structures That May Exist as Reaction Intermediates of the Li–Fe–F Ternary System}
\end{table}


indicated by the color coding in Figure 3. In addition, we observe from the calculated crystal structures that it is significantly more energetically favorable to shift Li\(^+\) off the “A” site of the FeF\(_3\) perovskite enabling it to occupy roughly half the interstitial space of each “A” site. In doing so the Li\(^+\) becomes octahedrally coordinated and shares a single face with an adjacent Fe-centered octahedra. At composition Li\(_{1/2}\)Fe\(_3\) the trirutile structure yields a slightly more favorable energy (~25 meV/atom) than the best structure we could find that has Li\(^+\) inserted topotactically into perovskite FeF\(_3\). This energy difference is small enough that it could be overcome by entropy effects at finite temperature or by minor polarization of the Li potential. Additional, thermodynamically stable, rutile structures (i.e., Li\(_{1/4}\)Fe\(_{3/4}\)F\(_3\), Li\(_{1/2}\)Fe\(_{1/2}\)F\(_3\), and LiFe\(_{1/2}\)F\(_3\)) were generated by substituting iron for lithium in a one-for-one manner in FeF\(_2\), Li\(_{1/4}\)Fe\(_{5/4}\)F\(_3\), Li\(_{3/4}\)Fe\(_{1/4}\)F\(_3\), and Li\(_{3/2}\)Fe\(_{1/2}\)F\(_3\), respectively. LiFe\(_{1/2}\)F\(_3\) displays the R3\(_H\) (no. 146) space group of prototype Ni\(_3\)TeO\(_6\). This compound can be related to corundum, Al\(_2\)O\(_3\), by replacing Al with Li and Fe ordered in the 3\(_a\) site. It can also be related to ilmenite, FeTiO\(_3\), because of the similar cation ordering. However, it is an iron-deficient ilmenite because the Li/Fe ratio is 3:1, rather than the 1:1 Fe/Ti ratio observed in the prototype. Although structurally different from Li\(_{1/2}\)Fe\(_{1/2}\)F\(_3\), the inverse spinel, Li\(_{15/8}\)Fe\(_{3/8}\)F\(_3\), is also iron-deficient with respect to the standard A\(_2\)B\(_x\) composition of spinel. The structure of Li\(_{15/8}\)Fe\(_{3/8}\)F\(_3\) possesses the symmetry of the space group P4\(_3\)2 (no. 212) with the prototype LiFe\(_2\)O\(_3\). More detailed discussion of the importance of these structures and their relation to the electrochemical reaction is presented in section V.

The theoretical voltage profile for reaction of FeF\(_3\) with Li can be deduced from the phase diagram and the Li potentials in each three-phase triangle. Discharging FeF\(_3\) by adding Li causes the overall composition point to move along the line between FeF\(_3\) and Li. The equilibrium reaction path is highlighted in the expanded phase diagram as shown in Figure 5B, and the voltage profile is shown in Figure 5A. The calculations indicate that Li\(^+\) insertion into perovskite FeF\(_3\) is the most favorable initial reaction mechanism, up to at least Li\(_{1/4}\)FeF\(_3\), after which a transformation to the trirutile at Li\(_{1/2}\)FeF\(_3\) may occur, though further Li insertion is very

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**Table 2. Crystallographic Information of the Ternary Phases Predicted for the Li—Fe—F System and the Defect Trirutile FeF\(_3\) Predicted as the Reconversion Product**

<table>
<thead>
<tr>
<th>compd</th>
<th>space group</th>
<th>lattice parameters</th>
<th>prototype</th>
<th>atomic positions</th>
<th>occupancy</th>
</tr>
</thead>
<tbody>
<tr>
<td>Li(_{1/4})FeF(_3)</td>
<td>derived from R-3cR (no. 167)</td>
<td>(a = 5.531)  (\alpha = 56.2)  (V = 108.88)</td>
<td>FeF(_3)</td>
<td>Li 2a</td>
<td>0.796, 0.786, 0.795</td>
</tr>
<tr>
<td>Li(<em>{1/4})Fe(</em>{3/4})F(_3)</td>
<td>derived from P4(<em>2)j(</em>{2}mm) (no. 136)</td>
<td>(a = 4.820)  (\alpha = 90.0)  (V = 218.88)</td>
<td>ZnSb(_2)O(_6)</td>
<td>Fe 2a, Li 2a</td>
<td>0, 0, 0</td>
</tr>
<tr>
<td>Li(_{1/2})FeF(_3)</td>
<td>P4(<em>2)j(</em>{2}mm) (no. 136)</td>
<td>(a = 4.756)  (\alpha = 90.0)  (V = 211.23)</td>
<td>ZnSb(_2)O(_6)</td>
<td>Fe 2a, Li 2a</td>
<td>0, 0, 0</td>
</tr>
<tr>
<td>Li(<em>{1/2})Fe(</em>{1/2})F(_3)</td>
<td>R3(_H) (no. 146)</td>
<td>(a = 5.102)  (\alpha = 90.0)  (V = 314.30)</td>
<td>Ni(_3)TeO(_6)</td>
<td>Fe 2a, Li 3a</td>
<td>0, 0, 0.146</td>
</tr>
<tr>
<td>Li(<em>{15/8})Fe(</em>{3/8})F(_3)</td>
<td>P4(_3)2 (no. 212)</td>
<td>(a = 8.415)  (\alpha = 90.0)  (V = 595.86)</td>
<td>LiFe(_2)O(_3)</td>
<td>Fe 4b, Li 3c, Li 3d, Li 3e, Li 3f</td>
<td>0.625, 0.625, 0.625</td>
</tr>
</tbody>
</table>

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competitive since the Li-inserted FeF$_3$ at the same composition is only 25 meV/atom higher in energy. Given the kinetic ease of Li insertion, compared to the rearrangements needed to form trirutile, it seems reasonable that in experiments Li$^{+}$ insertion is still observed at $x = 0.5$ in Li$_x$FeF$_3$. The calculated equilibrium profile indicates that further lithiation causes the intermediate phase, Li$_{1/2}$FeF$_3$, to disproportionate into Fe and LiF, rather than undergo continued Li$^{+}$ insertion.

IV. Deviation between Calculated and Experimental Potential

In a complex multiphase conversion system such as the FeF$_3$–Li system, there are many reasons why calculated equilibrium potentials may differ from experimentally observed charge and discharge potentials. Polarization, deviation from the equilibrium path, or nano effects, and intrinsic errors in our first-principles energy method (GGA) can all play a role in the difference between the calculated and experimental potential displayed in Figure 5. We will discuss several of these in turn and propose a model for the nonequilibrium reaction path.

The conversion step from Li$_x$FeF$_3$ to LiF and Fe involves the reduction of ionic Fe to metallic Fe. It is well-established that due to its spurious self-interaction the GGA prediction error grows as materials are compared in which the electron states are very different.$^{19}$ For this reason GGA does not capture well the energy difference between the localized d-states for ionic iron in an oxide or fluoride and the delocalized states in metallic iron.$^{42,52}$ For the purpose of our phase diagram calculations this problem can be solved by simply determining the energy of metallic Fe from the experimentally measured reaction enthalpy obtained for FeF$_2$ + 2Li $\rightarrow$ Fe + 2LiF (5)

and the calculated energies for Li, FeF$_2$, and LiF. This enables the total energy of bulk iron to be determined by fitting the total energies of two insulators (i.e., FeF$_2$ and LiF) and a nontransition metal (i.e., Li) to the experimental reaction enthalpy. This approach yields a cohesive energy of $-4.3447$ eV for bulk, metallic iron, only $\sim 65$ meV lower than that obtained from experiment$^{53}$ (i.e., $-4.28$ eV), and ensures that the energy difference between oxidized Fe and metallic Fe is well-represented in the calculations. Utilizing the fitted energy of Fe to calculate the reaction potential reduces the potential of conversion from 2.91 to 2.62 V (Figure 5A).

All calculations in Figure 5 are for bulk compounds, whereas reasonable Li reaction rates can only be obtained for nanoparticles. As Fe precipitates out from the fluoride

\begin{align*}
\text{FeF}_2 + 2\text{Li} & \rightarrow \text{Fe} + 2\text{LiF} \\
\end{align*}

(5)

Figure 5. Calculated equilibrium potential profile obtained with the GGA computed energy of bulk Fe (green, solid line) is depicted in panel A. The decrease in potential, between $x = 0.5$ and $x = 3$ in Li$_x$FeF$_3$, obtained by fitting the Fe energy to experiment is shown in panel A as a red, dashed line. Panel A also contains the experimental voltage profile for the first cycle of an FeF$_3$/C nanocomposite (black, dotted line). The data was collected under constant current conditions at a rate of C/200. It was provided to us by Professor Glenn Amatucci of Rutgers, The State University of New Jersey. The equilibrium reaction path associated with the calculated equilibrium potential profile is highlighted (green, dashed arrows) on the phase diagram in panel B.

matrix it is likely to form very small clusters. It is well-known that the cohesive energy of very small particles is significantly reduced from the bulk value.\textsuperscript{54,55} Although a full investigation of the Li–Fe–F phase space at nanoscale dimensions is outside the scope of the work presented in this paper we calculated the energy of a 1 nm spherical particle of bcc iron to assess the nanosize effect. The total energy was obtained in a manner similar to other nanoscale calculations presented in the literature\textsuperscript{55} and internally consistent with the bulk calculations in this work (e.g., full relaxation of atomic positions, planewave cutoff of 550 eV, etc., except that this computation utilized the $\Gamma$-point $k$-space integration only). The calculated cohesive energy of the 1 nm Fe particle is $-4.0545$ eV, a significant change from the $-5.0713$ eV obtained with GGA computation of bulk Fe. To remain consistent by accounting for the inherent error within the GGA method, the nanoparticle energy was also corrected by the same amount as the bulk energy in the previous section. This yields a cohesive energy of $-3.3279$ eV for fitted nanoscale Fe, a value that corresponds well with the trend expected for embedded nanoparticles within this size range.\textsuperscript{54}

Figure 6A shows the voltage profile when the energy of nanoscale Fe is used in the reaction energy. Both the pure first-principles voltage (GGA) and the “corrected” potential are given. Lithiation of FeF$_3$ is unchanged from the bulk phase up to $x = 0.5$ by the introduction of nanosize Fe. However, lithiation beyond $x > 0.5$ moves the system through several smaller three-phase triangles, before Fe precipitation occurs. Both the correction and the nanosized form of Fe lead to increase the energy of Fe (reduce its cohesive energy). This increase stabilizes compounds with higher Li-to-Fe ratio in the phase diagram. In particular, the lithiation path now crosses three-phase triangles connecting Li$_{1/4}$Fe$_{5/4}$F$_3$ and Li$_{3/2}$Fe$_{3/4}$F$_3$ with LiF, and FeF$_2$ with Li$_{1/4}$Fe$_{5/4}$F$_3$. Many of these reactions are nonintuitive if one is used to thinking of conversion reactions in binary systems. For example, consider the first reaction for $x = 0.5$ in which LiF and Li$_{1/4}$Fe$_{5/4}$F$_3$ form. This reaction may seem surprising at first since the Fe-to-F ratio increases in Li$_{1/4}$Fe$_{5/4}$F$_3$ from FeF$_3$. However, this reaction could rather easily occur by reaction of Li and F at the surface and diffusion of Fe from the surface of the particle toward the interior. After this reaction, the GGA nanoparticle (with uncorrected Fe energy) leads to precipitation of Fe. However, in the path using the fitted energy for Fe, the higher energy of Fe delays the precipitation and several other complex three-phase triangles cause further conversion reactions between various Li─Fe─F compounds before metallic Fe is formed. Surprisingly, under this scenario, it is even possible that FeF$_2$ forms along the reaction path. The stabilization of more intermediate phases in the phase diagram also modifies the voltage profile (Figure 6A). Following the path defined in Figure 6B (red, dashed line), the voltage profile where we use nanosized Fe with experimentally based correction agrees remarkably well with

Some of the new phases that are predicted to be stable (or metastable) along the reaction path may be particularly significant. The compound at composition Li$_{3/2}$Fe$_{3/4}$F$_3$ has an inverse spinel structure in which Li$^+$ occupies all the tetrahedral sites and half the octahedral cation sites while Fe$^{2+}$ occupies the remaining octahedral sites. Note that the prediction of stable Li$_{3/2}$Fe$_{3/4}$F$_3$ makes LiFeF$_3$ metastable (Figure 6B); however, the compound is indicated on the phase diagram as it may form if no Fe migration can occur.

V. Model for the Nonequilibrium Reaction Path

V.A. Conversion. Conversion reactions almost always display significant polarization (difference between charging and discharging voltage). To discuss this it may be useful to distinguish classical polarization including ohmic effects and diffusion polarization from changes in potential related to the fact that the system may strongly deviate from its equilibrium path in the phase diagram and go through different phases upon charge and discharge. Since the traditional polarization depends to some extent on the morphology and design of the composite electrode in which the active material is embedded, we do not treat it here and instead focus on the polarization caused by the system deviating to a nonequilibrium path through the phase diagram. The formation of different phases in charge and discharge would show up as hysteresis that is intrinsic to the active material and, as such, would be unaffected by the traditional enhancements used to improve rate such as large carbon additions or the use of thin film electrodes. Phase hysteresis is caused by the fact that, at low temperatures, structures often convert to the reaction product that is most easily reached kinetically, rather than to the thermodynamic ground state. This kinetic phase sequence is not reversible and will be distinct in charge and discharge.

One can think of the reaction with Li and FeF$_3$ (or more exactly Li$_{3/2}$Fe$_{3/4}$F$_3$ with $x \approx 1$) as an interdiffusion problem. Either Li has to diffuse into the fluoride host while Fe diffuses out and precipitates, or F has to migrate between Li and Fe. The more plausible scenario is that the cation rearrangement may take place even when the Fe diffusivity is significantly less than that of Li the reaction should be expected to continue through phase space associated with the lithiation path formed by the tie line connecting FeF$_3$ to LiFeF$_3$. Subsequently, the need to precipitate Fe$^{2+}$ from LiFeF$_3$, so as to continue accommodating Li$^+$, will push the composition of intermediates toward phase space bounded by Li$_{3/2}$Fe$_{3/4}$F$_3$, LiF, and Fe. Regardless of the exact intermediate compositions, the potentials associated with the phase space for this portion of the reaction, $x = 1$ to $x = 3$ in Li$_{3/2}$Fe$_{3/4}$F$_3$, correspond to a voltage of $\sim 2.15$ V, which is in reasonable agreement with experimental measurements.$^{11}$

V.B. Reconversion. If interdiffusion limitations are central to driving the conversion reaction into nonequilibrium phase space during discharge, then they are likely to have an analogous effect upon the reconversion reaction during charge. Similar to the arguments made for conversion, if the Fe diffusivity is significantly less than that of Li the reaction intermediate(s) will consistently be Fe-deficient relative to the amount of Li migrating out of the electroactive structure as Fe diffusion cannot keep up. Consequently, reconversion should promote the formation of phases containing the highest possible oxidation states of Fe as this path allows for the largest ratio of Li extraction to Fe insertion. With this principle, a reconversion reaction path and voltage profile can be estimated. We limit oxidation of Fe in the fluoride compounds to Fe$^{3+}$ because the phase space containing higher valence states corresponds to potentials of at least $\sim 5$ V versus Li/Li$^+$. Figure 7A shows (in blue) the computed nonequilibrium reconversion profile when oxidation to Fe$^{3+}$ occurs as soon as Fe is taken up in the fluoride hosts. The reaction path and compositions of the intermediate phases

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Further conversion along our proposed nonequilibrium path would cause a transition from spinel to ilmenite and then to rutile. Such transitions are also found in systems other than Li–Fe–F; this structural path is of great importance to the Fe–Ti–O system in which the spinel/ ilmenite/rutile transition corresponds to composition changes from Fe₂TiO₄ to FeTiO₃, and then to TiO₂, resulting from oxidative removal of iron.⁶⁰,⁶¹ In fact, studies examining the intergrowth of titaniferous spinel with ilmenite are of interest to geologists⁶² in part because ilmenite and rutile are the greatest, natural source for production of titanium.⁶³ Like the spinel described above, both the ilmenite and rutile structures contain only Fe⁴⁺ and are iron-deficient when compared with respective prototype compositions that would be based on pure Fe–F compounds. For example, the ilmenite composition is Li₃/₂Fe₁/₂F₃, rather than LiFeF₃, and the rutile composition is Li₃/₄Fe₃/₄F₃, rather than Fe₂O₃.

The nonequilibrium path we predict from the principle of maximum Fe oxidation shows a reasonable resemblance to the measured voltage profile in charge, though there are some deviations. The fact that the calculated initial charge voltage is somewhat too high may be related to errors in the energy of metallic Fe (which is consumed in this step) or may reflect that in the beginning of charge full oxidation to Fe⁴⁺ in the fluoride is not necessary (i.e., it may be possible to form an Fe²⁺ intermediate phase, thus yielding lower observed voltage) as there is still a lot of Fe available near the surface of the LiF particles.

An interesting aspect of our predicted nonequilibrium reconversion path is that the last stable Li–Fe–F compound before full reconversion to FeF₃ is a rutile-related compound (Li₃/₄Fe₃/₄F₃). If exchange of one Fe⁴⁺ for three Li⁺ should continue, from this compound along the easiest path, it is likely that rutile-like Li₃/₄Fe₃/₄F₃ will reconvert directly to a rutile-related FeF₃. Our first-principles computations indicate that this is quite possible by removing Li from Li₃/₄Fe₃/₄F₃ (and inserting Fe) to form a defect trirutile structure with a FeF₃ composition. The defect trirutile FeF₃ is only slightly (i.e., ~20 meV/atom) more unfavorable than the perovskite form of FeF₃. This observation supports the experimental finding that a structure related to FeF₃ forms upon completion of the charge process even though a capacity equivalent of approximately three Li⁺ has been removed.¹¹ We suggest that what actually forms is the FeF₃ defect trirutile with the stoichiometry of Li₁₅/₈Fe₃/₈F₃ can be rewritten as Li[Fe₁₂/₅Li₁₂/₅]F₂ to illustrate its correspondence to a spinel. Li occupies the tetrahedral site, while Li and Fe share the octahedral sites.

Figures 7. Panel A depicts the voltage profile for iron fluoride reconversion during charging (blue, solid line) as compared with the voltage profile for iron fluoride conversion when accounting for the energy of GGA computed nanoscale Fe (green, solid line) and the experimental profile (black, dotted line). The potentials required to form phases with Fe⁴⁺ during reconversion (red, solid line) are also presented. The corresponding reaction path is highlighted for discharge (green, dashed arrows) and charge (blue, dashed arrows) in panel B.

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