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Theoretical and data-driven approaches to semiconductors and dielectrics: from prediction to experiment

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ABSTRACT

Computational approaches using theoretical calculations and data scientific methods have become increasingly important in materials science and technology, with the development of relevant methodologies and algorithms, the availability of large materials data, and the enhancement of computer performance. As reviewed herein, we have developed computational methods for the design and prediction of inorganic materials with a particular focus on the exploration of semiconductors and dielectrics. High-throughput first-principles calculations are used to systematically and accurately predict the local atomic and electronic structures of polarons, point defects, surfaces, and interfaces, as well as bulk fundamental properties. Machine learning techniques are utilized to efficiently predict various material properties, construct phase diagrams, and search for materials satisfying target properties. These computational approaches have elucidated the mechanisms behind material functionalities and explored promising materials in combination with synthesis, characterization, and device fabrication. Examples include the development of ternary nitride semiconductors for potential optoelectronic and photovoltaic applications, the exploration of phosphide semiconductors and the optimization of heterointerfaces toward the improvement of phosphide-based photovoltaic cells, and the discovery of ferroelectricity in layered perovskite oxides and the theoretical understanding of its origin, all of which demonstrate the effectiveness of our computer-aided materials research.

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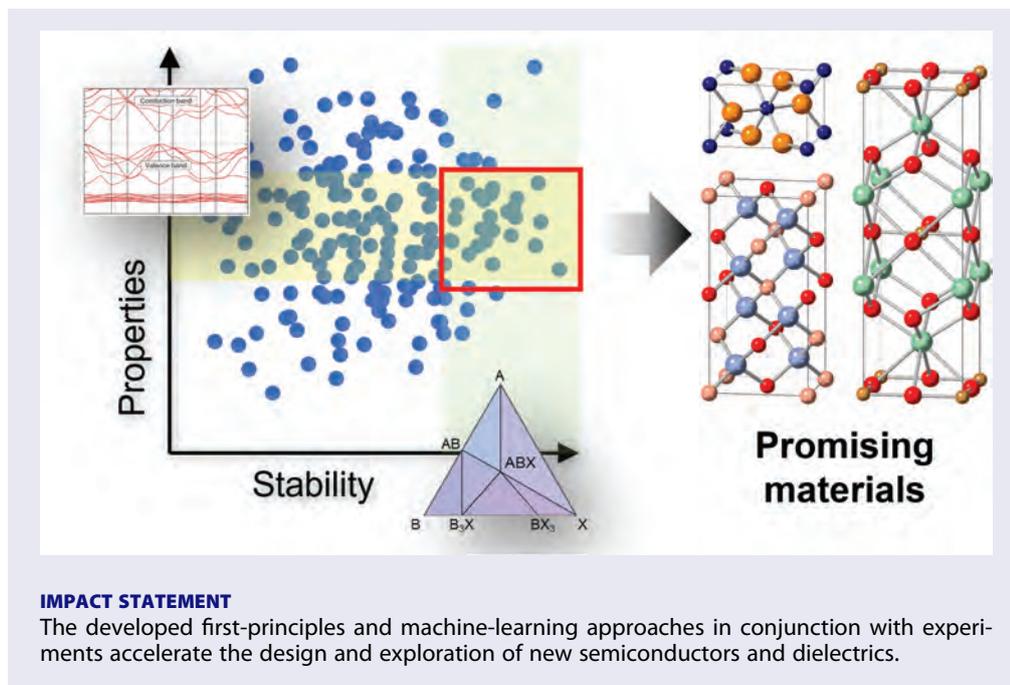
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1. Introduction

Inorganic semiconductors and dielectrics are essential components of electronic and optoelectronic devices in modern society. To meet the demands for further improvement of the material and device performance, not only the optimization of device structures but also the exploration of novel materials with superb properties is needed, particularly for innovating new applications [1–8]. However, the search for as-yet-unknown or as-yet-unused materials with desired properties and sufficient stability is generally challenging. It typically requires huge efforts and costs for a series of experiments from synthesis, crystal- and micro-structure characterization, and property measurements to device fabrication and optimization.

Computational approaches using theoretical calculations and data scientific methods can be helpful to accelerate such materials research if they allow us to make reliable predictions. With the development of methodologies and algorithms and the enhancement of computer performance, theoretical calculations such as first-principles calculations are applicable to a wider variety of material issues and, therefore, play increasingly important roles nowadays [9–13]. Data scientific methods such as machine learning are being spread in the field of materials science and technology, as larger data becomes available [14–16]. Several computational databases are already open to the public and widely used in materials research. Examples include the Material Project [17], AFLOW Distributed Materials Property Repository [18], Open Quantum Materials Database (OQMD) [19], Electronic Structure Project (ESP) [20], Open Materials Database [21], Materials Cloud [22], and

Novel Materials Discovery (NOMAD) Repository [23]. Efforts to construct open experimental databases for inorganic materials have also been made [24,25].

To deploy such theoretical calculations, machine learning, and their combinations in a broader variety of materials studies, further development of methodologies and algorithms is desired. In particular, high-quality and large datasets are key to the effective use of machine learning. First-principles calculations are able to generate data on diverse material properties, including the atomic and electronic structures and related properties of perfect crystals, amorphous phases, point defects, dislocations, surfaces, grain boundaries, and heterointerfaces. However, the prediction accuracy and the computational efficiency of first-principles calculations are usually in a trade-off relationship. This issue particularly matters when we deal with imperfect crystals that require the use of large simulation models and the consideration of various configurations. It is essential to develop accurate but efficient computational schemes to overcome this situation. Moreover, the development of new materials should be accelerated if reliable computational screening of a tremendous number of candidate materials is doable and combined with experiments appropriately.

In this article, we review our combined computational and experimental approaches to inorganic semiconductors and dielectrics. In Sec. 2, we describe the first-principles calculation methods that allow for accurate and efficient prediction of various material properties, not only bulk electronic structures and relevant fundamental properties but also the formation energies, geometries, and electronic structures of

polarons, point defects including native defects and dopants, surfaces, and interfaces. As a novel machine learning approach, we emphasize an efficient construction of phase diagrams with the aid of active learning. The applications of these computational methods to various materials research issues are also shown, where first-principles calculations, machine learning techniques, and/or experiments are combined to accelerate the prediction of material properties, the exploration of materials with desired properties, and the suggestion of synthesis conditions. In Sec. 3, we discuss the design and exploration of new semiconductors and dielectrics by combining experimental and computational approaches. This includes the development of nitride and phosphide semiconductors for optoelectronic and photovoltaic applications, followed by phosphide-based photovoltaic cell fabrication, and the discovery of new layered perovskite oxide ferroelectrics. The origins of their functionalities have been elucidated with the support of electronic structure and chemical bonding analyses using the first-principles calculations. We then summarize our work and provide an outlook on computer-aided materials research in Sec. 4.

2. Development and application of computational methods for predicting the structures, properties, and phase diagrams of semiconductors and dielectrics

2.1. First-principles approaches to polarons, point defects, surfaces, and interfaces

We have developed computational methods based on first-principles calculations to predict the formation energies and the local atomic and electronic structures of polarons [26–28], point defects [28–39], dislocations [40], surfaces [41–49], and interfaces [40,41,44,46,50,51], as well as bulk fundamental properties and stability [41,44,52–58], accurately and/or efficiently. The first-principles calculations are conducted using the projector augmented-wave method [59] as implemented in the Vienna *Ab initio* Simulation Package [60,61]. Appropriate corrections have been applied to overcome the limitations in the description of the electronic structure [26,34,44,62] and the restriction in the simulation cell size [30,63]. The construction of simulation models and calculation processes are automated to enable high-throughput calculations [34,35,42,45,55]. Such approaches are briefly described below with an emphasis on the case of surfaces.

Inorganic semiconductors and dielectrics, which are our main targets, have finite band gaps, as well as insulators. It is well known that standard exchange-correlation functionals such as the local (spin) density approximation (L(S)DA) [64–66]

and the generalized gradient approximation (GGA) [67–70] to density functional theory [71] tend to underestimate band gaps of many systems. In addition, localized states in solids are not well described by these approximations [13,52,72,73]. This drawback can be concisely remedied using Hubbard U corrections, so-called L(S)DA+ U and GGA+ U , when the localized states mainly originate from particular atomic orbitals [72,73]. To better describe the overall electronic structures of semiconductors, dielectrics, and insulators, hybrid functionals that partially incorporate the non-local Fock exchange are often used nowadays [74–77]. However, the results depend on the form and parameters in hybrid functionals such as the mixing amount of the Fock exchange, the optimal value of which by nature depends on the electronic structures of the systems of interest because of the system-dependent screening behavior [74,78,79]. Dielectric-dependent hybrid functionals are often effective in overcoming this issue, a concise form of which uses the Fock exchange mixing value determined from the electronic contributions to the dielectric constants calculated for materials of interest [78,80,81]. This approach has been reported to well reproduce the band gaps of prototypical semiconductors and insulators [78,80,81].

Hybrid functional calculations are computationally demanding, particularly when dealing with polarons, dislocations, point defects, surfaces, and interfaces using supercells. If reasonably applicable, we conduct dielectric-dependent hybrid functional calculations non-self-consistently on top of the GGA or GGA+ U . For prototypical materials, we have shown that this approach dramatically accelerates electronic structure calculations with sufficient accuracy preserved [34,44,45,49,82]. Examples of band gap predictions for prototypical semiconductors and insulators are shown in Figure 1 [44].

The GW [83] and related approximations based on many-body perturbation theory enable us to obtain quasi-particle band structures. For prototypical systems with relatively simple band structures, the band gaps are well reproduced using standard GW calculations on top of the GGA, as shown in Figure 1. However, the accuracy depends significantly on the level of the approximations for other properties such as localized semi-core states, ionization potentials (IPs), and electronic affinities (EAs) [41,52], as well as the case of materials with more complicated band structures [84]. Calculations using a high level of the GW approximation and beyond are rather computationally demanding and, therefore, not suited to high-throughput data generation. In our study that aims at data-driven approaches, we have occasionally used GW calculations to check the results of GGA(+ U) and hybrid functional calculations.

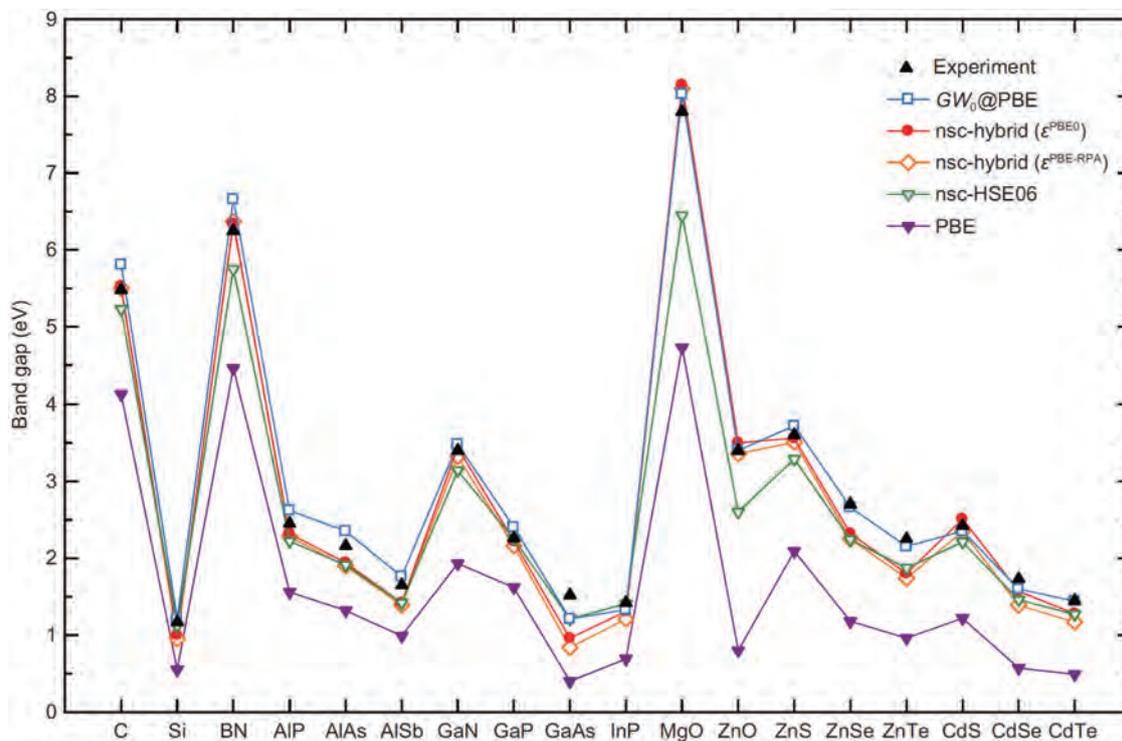


Figure 1. Band gaps of prototypical semiconductors and insulators from first-principles calculations using various approximations, including non-self-consistent (nsc) dielectric-dependent hybrid functional approaches, compared with reported experimental values. Adapted with permission from ref. [44]. Copyright 2017 American physical society.

The non-self-consistent hybrid functional approach allows us to make band-edge corrections to the GGA(+U) results by utilizing the fact that electrostatic potentials do not change from the input GGA (+U) values. This feature is advantageous in the accelerated evaluation of the formation energies and electronic levels of polarons and point defects if the polaron and defect states are reasonably described using the GGA(+U) [34,85–88]; otherwise, we use self-consistent hybrid functional calculations even though they are computationally rather demanding.

Surface band-edge positions with respect to the vacuum level, namely IPs and EAs, and interfacial band offsets are also efficiently evaluated using the non-self-consistent hybrid functional approach. In typical evaluation procedures for the surface and interface band positions, the results for surface or interface and bulk models are combined through the electrostatic alignment [13]. This is schematically shown in Figure 2 [49]. Here, standard self-consistent hybrid functional calculations of surfaces and interfaces are computationally expensive in many cases. Overall computational costs are dramatically reduced if the bulk electronic structures are obtained using non-self-consistent hybrid functional calculations on top of the GGA(+U) and electrostatically aligned with the surface or interface results treated using the GGA (+U) [44].

In combination with an algorithm for automatic non-polar surface model generation [42], we have

conducted first-principles calculations of in total about 3000 surface models for non-metallic binary

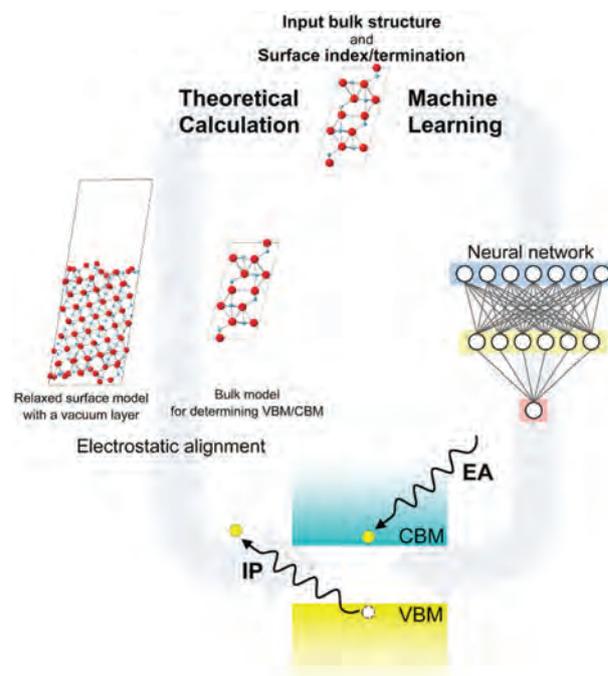


Figure 2. Schematic showing the computational procedures for the evaluation of IPs and EAs of non-metallic solids. Two approaches using theoretical (first principles) calculations and machine learning (neural network) are illustrated. Reprinted from ref. [49] under a creative commons attribution 4.0 international license.

and ternary oxides using the aforementioned non-self-consistent dielectric-dependent hybrid functional approach [49]. The resultant IPs and EAs for 2195 binary oxide surfaces are shown in Figure 3 [49]. The upper panel includes the results for various polymorphs and surface orientations, and the IPs and EAs are widely spread even within the common cation species. Still, we can see some chemical trends related to the group and period of the constituent element (cation) in the periodic table. The lower panel shows a comparison with reported experimental values. The IPs and EAs of solids, by nature, depend on the surface atomic structures and, therefore, the surface orientation, termination plane, reconstruction, adsorption, and contamination, all of which affect the surface dipole contributions [89–93]. Ideally speaking, identical surfaces should be considered for the experiments and theoretical calculations. However, available experimental values are limited for oxides, especially for well-characterized surfaces at the atomistic level. Nevertheless, reasonable agreement is found between theory and experiment overall. Our previous study has found that the non-self-consistent dielectric-dependent hybrid functional calculations can reproduce

experimental values better in the case of prototypical semiconductors with surface atomic structures reported [44], for which a direct comparison between theory and experiment is more straightforward and meaningful. In addition, highly accurate prediction of the IPs and EAs requires a rather sophisticated approximation, for instance, the GWT^1 approach based on many-body perturbation theory [41,52]. Given that such calculations are computationally demanding even for systems with small unit cells, we believe that the non-self-consistent dielectric-dependent hybrid functional approach is a well-balanced scheme suited for high-throughput studies.

In the above, we consider the cases where atomic relaxation is not significant at the surfaces. In reality, however, surfaces often show complicated and dramatic structural reconstructions [94–98]. As a case study, we have modeled macroscopically stoichiometric and nonpolar reconstructed (001) surfaces of $A(I)B(V)O_3$ ($NaTaO_3$, $KNbO_3$, and $KTaO_3$), $A(II)B(IV)O_3$ ($CaTiO_3$, $SrTiO_3$, $BaTiO_3$, and $BaZrO_3$), and $A(III)B(III)O_3$ ($YAlO_3$, $LaAlO_3$, and $LaGaO_3$) perovskites using a combination of first-principles calculations and an evolutionary algorithm [43,47]. Four

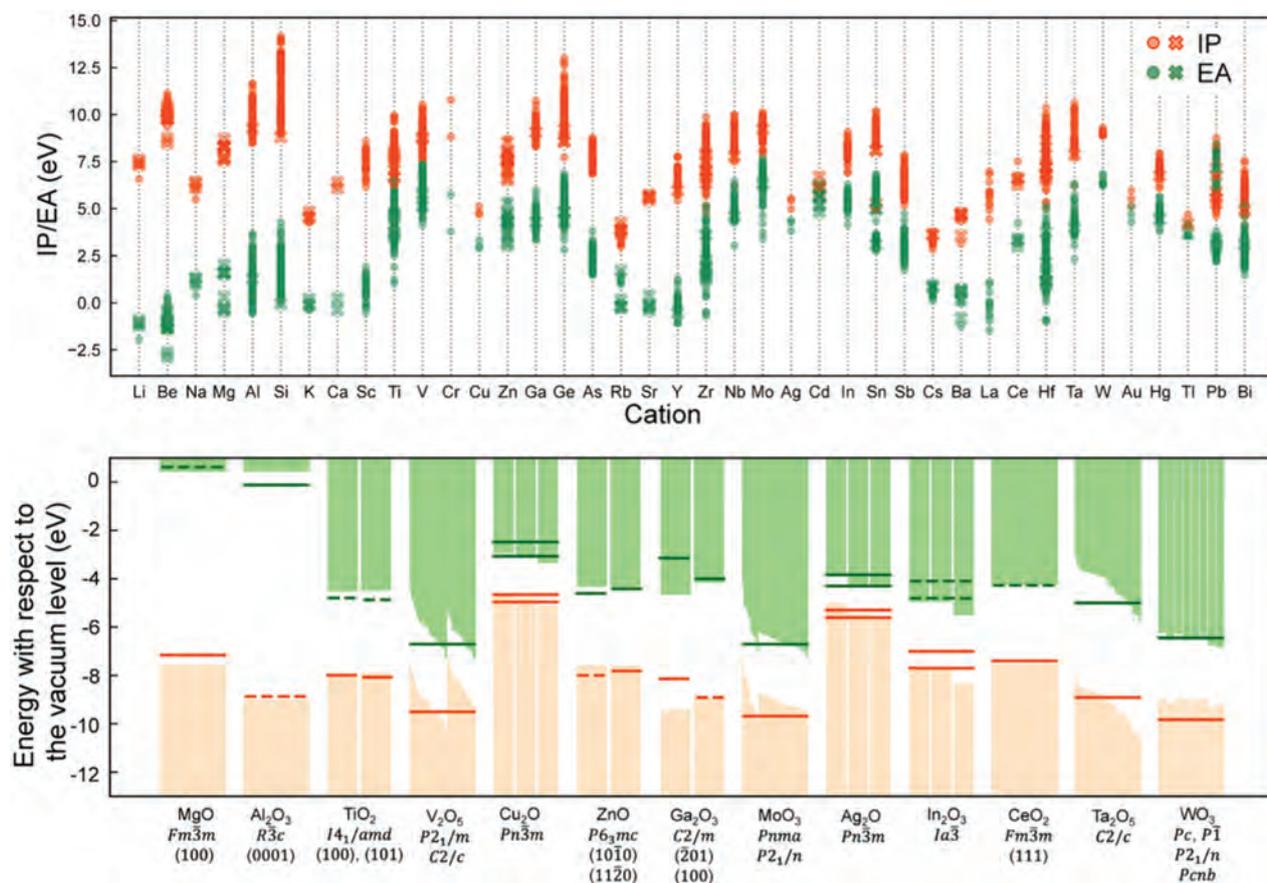


Figure 3. Theoretical IPs and EAs for 2195 nonpolar surfaces of binary oxides obtained using non-self-consistent dielectric-dependent hybrid functional calculations. A comparison with reported experimental values is given in the lower panel, where the negatives of the IPs and EAs corresponding to the valence band maxima and conduction band minima against the vacuum level, respectively, are plotted. The experimental values are indicated by horizontal solid and broken bars. The theoretical values for various polymorphs and surface orientations are included when the experimentally investigated crystal structures and/or surface orientations have not been specified. Reprinted from ref. [49] under a creative commons attribution 4.0 international license.

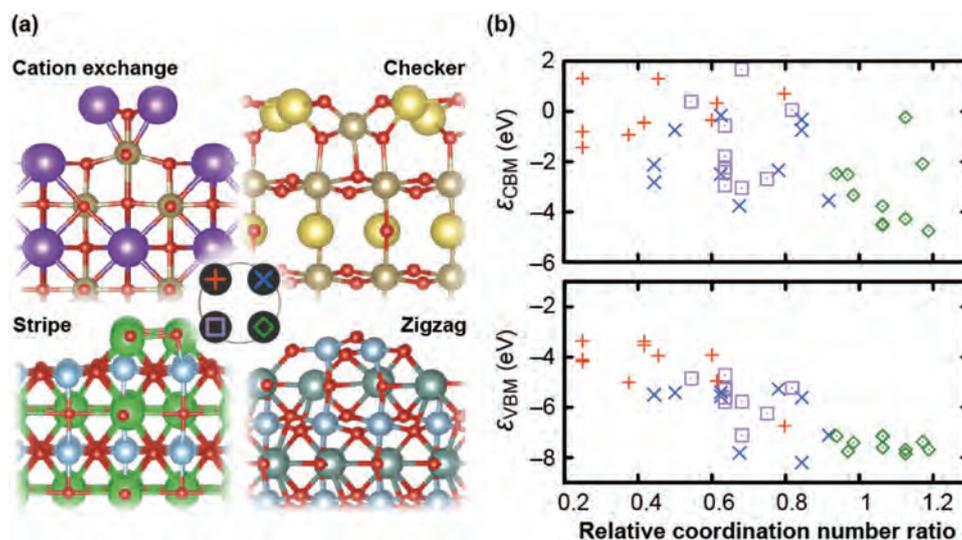


Figure 4. (a) Reconstructed structures of macroscopically stoichiometric and nonpolar (001) surfaces of $A(I)B(V)O_3$ ($NaTaO_3$, $KNbO_3$, and $KTaO_3$), $A(II)B(IV)O_3$ ($CaTiO_3$, $SrTiO_3$, $BaTiO_3$, and $BaZrO_3$), and $A(III)B(III)O_3$ ($YAlO_3$, $LaAlO_3$, and $LaGaO_3$) perovskites predicted using a combination of first-principles calculations and an evolutionary algorithm. Four types of representative reconstructed structures are shown: cation-exchange, checker, stripe, and zigzag models, which are labeled with red plus, blue cross, purple square, and green diamond marks, respectively. (b) CBM and VBM relative to the vacuum level (ϵ_{CBM} and ϵ_{VBM} , respectively) as a function of relative coordination number (see texts for details). Reprinted from ref. [47] under a creative commons attribution 4.0 international license.

types of representative reconstructed patterns were obtained with diverse atomic arrangements, as illustrated in Figure 4(a) [47]; here, the cation exchange structure has reproduced the previously predicted one for $KTaO_3$ [99]. Significant dependency of the valence band maximum (VBM) and the conduction band minimum (CBM) relative to the vacuum level (the negatives of IPs and EAs, respectively) on both composition and reconstructed structure is recognized in Figure 4(b), which is explainable in terms of the atomic configurations near the surfaces [47]: the decrease of the coordination number of cation A (B) at the surfaces leads to shallow (deep) VBMs and CBMs relative to the vacuum level. This chemical trend is clarified using *relative coordination number ratio* $\frac{n_{surf}^{A-O}/n_{bulk}^{A-O}}{n_{surf}^{B-O}/n_{bulk}^{B-O}}$, where $n_{surf}^{A(B)-O}$ and $n_{bulk}^{A(B)-O}$ are the coordination numbers of cation A (B) at the surface and the bulk, respectively.

Other computational methodologies developed include those for predicting the band positions at polar surfaces and interfaces using slab models with appropriate pseudo-hydrogen passivation [46].

2.2. Combination of first-principles calculations and machine learning to accelerate material-property prediction and materials exploration

Needless to say, machine-learning models can be effective in accelerating the prediction of various material properties, the exploration of target materials for diverse purposes, and so forth. Since the assessment of the accuracy and efficiency of the surrogate

models particularly matters here, we mainly discuss the accuracy of our property prediction models in this section, as well as the efficiency of our models for exploring materials with desired properties and assisting the construction of phase diagrams.

Once a reasonably accurate and large dataset is obtained from first-principles calculations, machine learning can be powerful if appropriate descriptors are available. We have constructed accurate and efficient prediction models using machine learning techniques for bulk fundamental properties such as band gaps and dielectric constants [55]. Moreover, point defect and surface properties can also be predicted with sufficient accuracy using high-throughput first-principles calculation datasets [34,49]. Examples include the prediction models of the IPs and EAs of binary oxide surfaces shown in Figure 5 [49]. An artificial neural network with an attention layer is constructed for each of the IP and EA cases using the smooth overlap of atom positions as structural descriptors [100]. Assuming that the atomic relaxation does not significantly alter the structures of ideally cleaved surfaces, only the information on the bulk composition and crystal structure and the surface orientation and termination plane, namely the surface structure before relaxation, is inputted, as illustrated in Figure 2. The coefficients of determination, root mean squared errors, and mean absolute errors of the test data are given in the caption of Figure 5. All the values indicate good prediction accuracy for IPs and EAs at relaxed surfaces, despite the fact that IPs and EAs are complicated surface structure-dependent electronic properties as mentioned above.

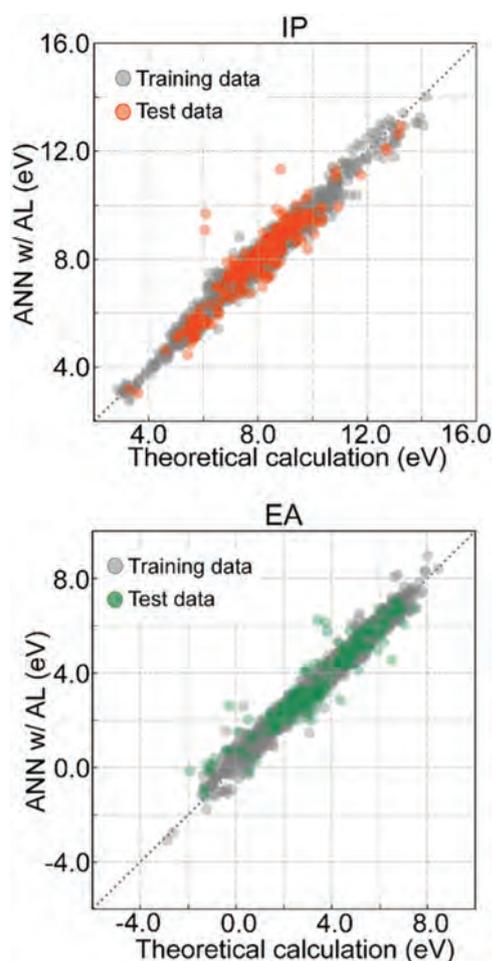


Figure 5. IPs and EAs of binary oxide surfaces by first-principles calculations versus those predicted by the artificial neural network with an attention layer (ANN w/AL). The coefficients of determination, root mean squared errors, and mean absolute errors of the test data are 0.90, 0.29 eV, and 0.21 eV for the IPs, and 0.93, 0.27 eV, and 0.19 eV for the EAs, respectively. Reprinted from ref. [49] under a creative commons attribution 4.0 international license.

A combination of first-principles calculations and machine learning is also a powerful scheme for materials exploration. Adaptive sampling or black-box optimization methods are widely used, including Bayesian optimization to search for the global maxima or minima of target material properties [101]. However, maximizing or minimizing material properties is not always required; there are also demands for exploring materials whose properties fall in target ranges. For this purpose, it is useful to set the acquisition function based on the probability that a data point achieves a target property within a specific range (PTR) [102]. We have implemented PTR acquisition functions in our material exploration code, including their extension to multi-objective ones, and conducted extensive performance tests [103]. Figure 6 shows examples of materials exploration simulations for two kinds of target conditions for the band gap and formation energy [103]. In both cases of Figure 6(a,b), target materials are quickly found when both the band

gap and formation energy are considered using the multi-objective PTR acquisition function.

It is also interesting to search for exceptional materials whose chemical compositions, structures, and/or properties do not follow typical trends, given that such materials may have exceptionally good properties. The Bound Objective-free eXploration (BLOX) method has been suggested to efficiently identify exceptional materials [104]. The effectiveness of BLOX has been demonstrated through the exploration simulations of molecules [104] and solids [105] with out-of-trend properties.

Understanding the tendency of a large number of materials would be helpful in constructing new principles for materials design and exploration. Interpreting the machine learning prediction models in terms of important features is one useful approach. Clustering is also widely used to classify materials into groups according to particular features, supporting us in overviewing large material data. Typical clustering approaches consider either target material properties or fundamental features related to chemical compositions and crystal structures. The relationships between the former and the latter are unclear, preventing us from understanding the chemical and structural origins of the material properties. In this context, we have conceived a novel clustering method that simultaneously considers target material properties and fundamental chemical and structural features, enabling us to classify materials depending on the properties of our interest [106].

2.3. Prediction of phase diagrams using machine learning techniques

We have also developed methods to accelerate the investigation of phase diagrams and to predict them. Although phase diagrams are crucial in materials research, suitable machine learning techniques to assist with their determination have not been much considered.

As mentioned above, a prominent method for optimizing material properties is Bayesian optimization in the materials science community [101]. This approach effectively suggests new materials with tailored properties. However, the direct application of Bayesian optimization to phase diagram construction is impossible. Therefore, we have introduced a new machine learning strategy, uncertainty sampling [107], which is tailored to study phase diagrams in detail. Uncertainty sampling assumes that the most uncertain points contain the most valuable information. Applied to phase diagram construction, it suggests the next experimental conditions for synthesis or measurement, which are crucial for an accurate determination. The developed method is called Phase Diagram Construction (PDC) and is outlined in Figure 7

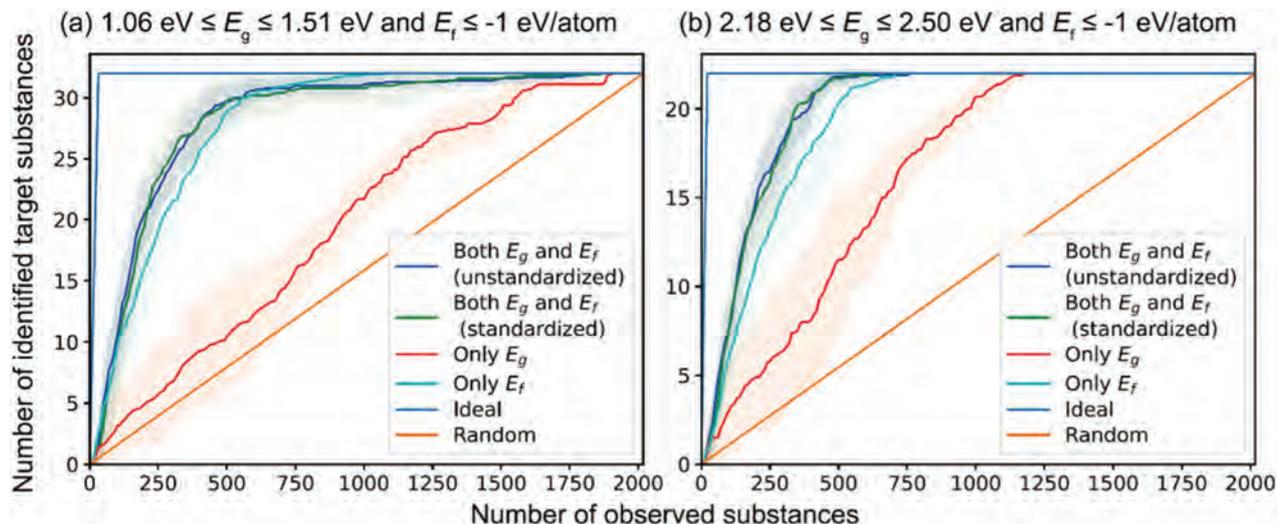


Figure 6. Performance of multi-objective adaptive sampling methods in the search for substances satisfying the respective conditions of the band gap (E_g) and formation energy (E_f) given in panels (a) and (b). Lines and shaded regions indicate the average and standard deviation of 10 simulations using different initial datasets, respectively. Reprinted from ref. [103] under a creative commons attribution 4.0 international license.

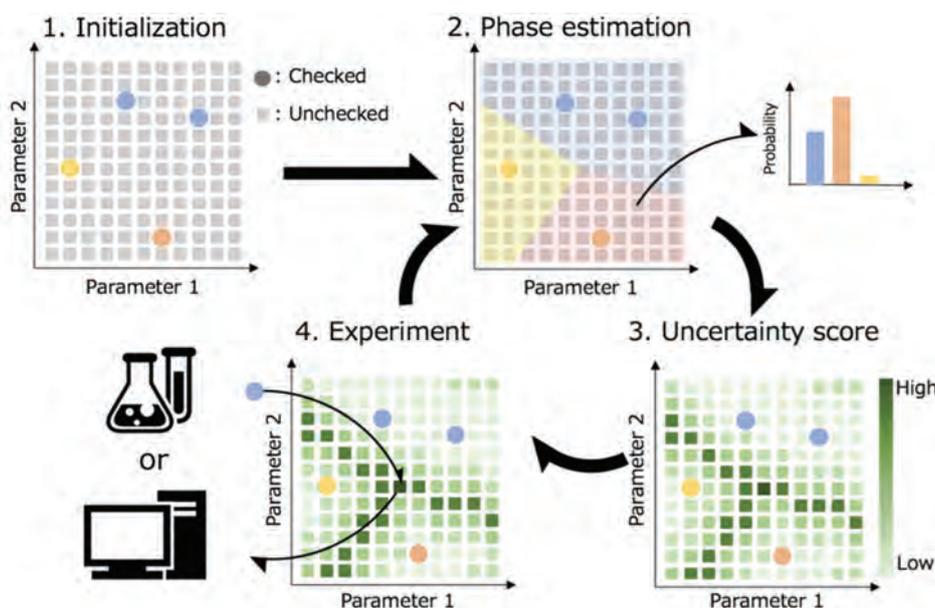


Figure 7. Procedure of the PDC algorithm to efficiently construct phase diagrams. Reprinted from ref. [107] under a creative commons attribution 4.0 international license.

[107]. In Sec. 2.3.1, the details of the PDC algorithm are introduced. In Sec. 2.3.2, the results of applications of the PDC algorithm for known and unknown phase diagrams are shown. Improvements of the PDC algorithm are discussed in Sec. 2.3.3. Finally, in Sec. 2.3.4, a complementary method to the PDC algorithm is considered, namely a machine learning method combined with CALPHAD (CALculation of PHase Diagrams) [108] databases to predict unknown phase diagrams.

2.3.1. PDC algorithm

The following is a detailed procedure of the PDC algorithm [107].

Step 1: initialization. We define the space where the phase diagram is to be constructed and discretize it. Each discretized point in the phase diagram serves as candidate experimental conditions. There are no restrictions on the dimensions of the phase diagram. Let \mathbf{x} denote the vector of discretized positions, and N candidate points $\{\mathbf{x}_i\}_{i=1,\dots,N}$ are prepared. From these, phase equilibria are identified for M initial points through experiments or simulations. These initial points can be chosen randomly or according to specific rules (e.g. selecting points that are easy to conduct experiment). Existing databases can also be utilized. Discrete

indices are assigned to the identified phase regions at this stage. For instance, if there are P different phase regions identified initially, we prepare indices $p = 1, \dots, P$ and assign these indices as labels to each data point. This process prepares the training data $D = \{\mathbf{x}_j, p_j\}_{j=1, \dots, M}$.

Step 2: phase diagram estimation. Using machine learning techniques such as label propagation (LP) or label spreading (LS), we estimate the probabilities $P(p|\mathbf{x})$ of belonging to each phase region p for all points in the phase diagram when D is the training data. LP maintains the phase information unchanged for the points where phases are already identified, whereas LS allows phase information to vary based on the surrounding environment. Utilizing these probabilities $P(p|\mathbf{x})$, we can predict the phase diagram by selecting the phase region with the highest probability at each point.

Step 3: uncertainty score. We define an uncertainty score $u(\mathbf{x})$ using the probabilities $P(p|\mathbf{x})$. We choose the next candidate \mathbf{x}^* where $u(\mathbf{x})$ is maximum using the following equation:

$$\mathbf{x}^* = \arg \max_{\mathbf{x}} u(\mathbf{x}). \quad (1)$$

In the PDC algorithm, three types of scores are available for the uncertainty score definition, namely the least confident (LC) method, margin sampling (MS) method, and entropy-based approach (EA):

$$u_{\text{LC}}(\mathbf{x}) = 1 - P(p_1|\mathbf{x}), \quad (2)$$

$$u_{\text{MS}}(\mathbf{x}) = 1 - [P(p_1|\mathbf{x}) - P(p_2|\mathbf{x})], \quad (3)$$

$$u_{\text{EA}}(\mathbf{x}) = - \sum_p P(p|\mathbf{x}) \log P(p|\mathbf{x}), \quad (4)$$

where $P(p_1|\mathbf{x})$ and $P(p_2|\mathbf{x})$ represent the highest and second highest probabilities of phase regions at the point \mathbf{x} , respectively.

Step 4: experiment. We perform experiments or simulations at the point \mathbf{x}^* selected in Step 3 where the uncertainty score is maximum, and we identify the phase region at this point. If a new phase region is discovered, the phase index is increased by one as $p = 1, \dots, P$ with $P + 1$. This process adds one more data point to the training dataset as $\{\mathbf{x}_j, p_j\}_{j=1, \dots, M, M+1}$. This dataset is utilized to return to Step 2.

This iterative process enables the construction of an accurate phase diagram with fewer iterations, leveraging uncertainty sampling to efficiently guide experimental or simulation efforts toward discovering new phases and determining phase boundaries in the phase diagram.

2.3.2. Examples of phase diagram constructions

First, to verify the efficiency of the PDC algorithm, known phase diagrams were virtually constructed. The examples include the low-pressure and high-pressure phase diagrams of water, as well as the liquidus projection of the $\text{SiO}_2\text{-Al}_2\text{O}_3\text{-MgO}$ pseudo-ternary system [107]. We found that using the PDC algorithm, the number of experiments required to determine a phase diagram can be reduced by a factor of five compared with random sampling. Figure 8 shows the results for each sampling method [107]. In this result, we utilized the LP method for phase diagram estimation and applied the LC approach for uncertainty score evaluation. With the PDC algorithm, the selected points were concentrated near phase boundaries, demonstrating efficient determination of phase boundaries. Moreover, the PDC algorithm found narrow phase regions quickly. We confirmed that even with a small number of initial data points (when many phases are still unidentified), the PDC algorithm efficiently produces phase diagrams. This illustrates that PDC is effective for investigating complex unknown phase diagrams from scratch.

Next, to verify the efficiency of the PDC algorithm for real experiments, the construction of novel phase diagrams for Zn – Sn – P film growth using molecular beam epitaxy (MBE) was performed using the PDC algorithm with LP and LC [109]. The phase diagram with varying temperatures of Sn and P evaporation sources was focused, and phase regions were identified through X-ray diffraction experiments. Starting with initial data from seven existing experimental points, the PDC algorithm suggested four additional experiments, leading to the discovery of new phase regions not identified in the initial dataset. Subsequent to this, additional seven experiments guided by PDC recommendations allowed us to construct the detail of the phase diagram [109] (see Figure 9). The phase diagram represents non-equilibrium phase regions depending on device-specific conditions, and it is difficult to predict by humans, requiring substantial effort during early-stage research. Particularly, the red regions in Figure 9 denote film deposition conditions where thin films do not grow. PDC effectively treats such conditions as distinct ‘phase regions’ enabling efficient discovery of new regions and precise determination of phase boundaries in non-equilibrium phase diagrams.

2.3.3. Improvements of PDC algorithm

2.3.3.1. Utilizing Gibbs' phase rule. For more effective construction of phase diagrams, two ingenuities are incorporated into the PDC algorithm: preparing training data in multiple-phase coexisting regions and reducing the search space based on the Gibbs' phase rule [110]. The former can increase the number of training data when one experiment is performed. In addition, the latter can reduce the number of

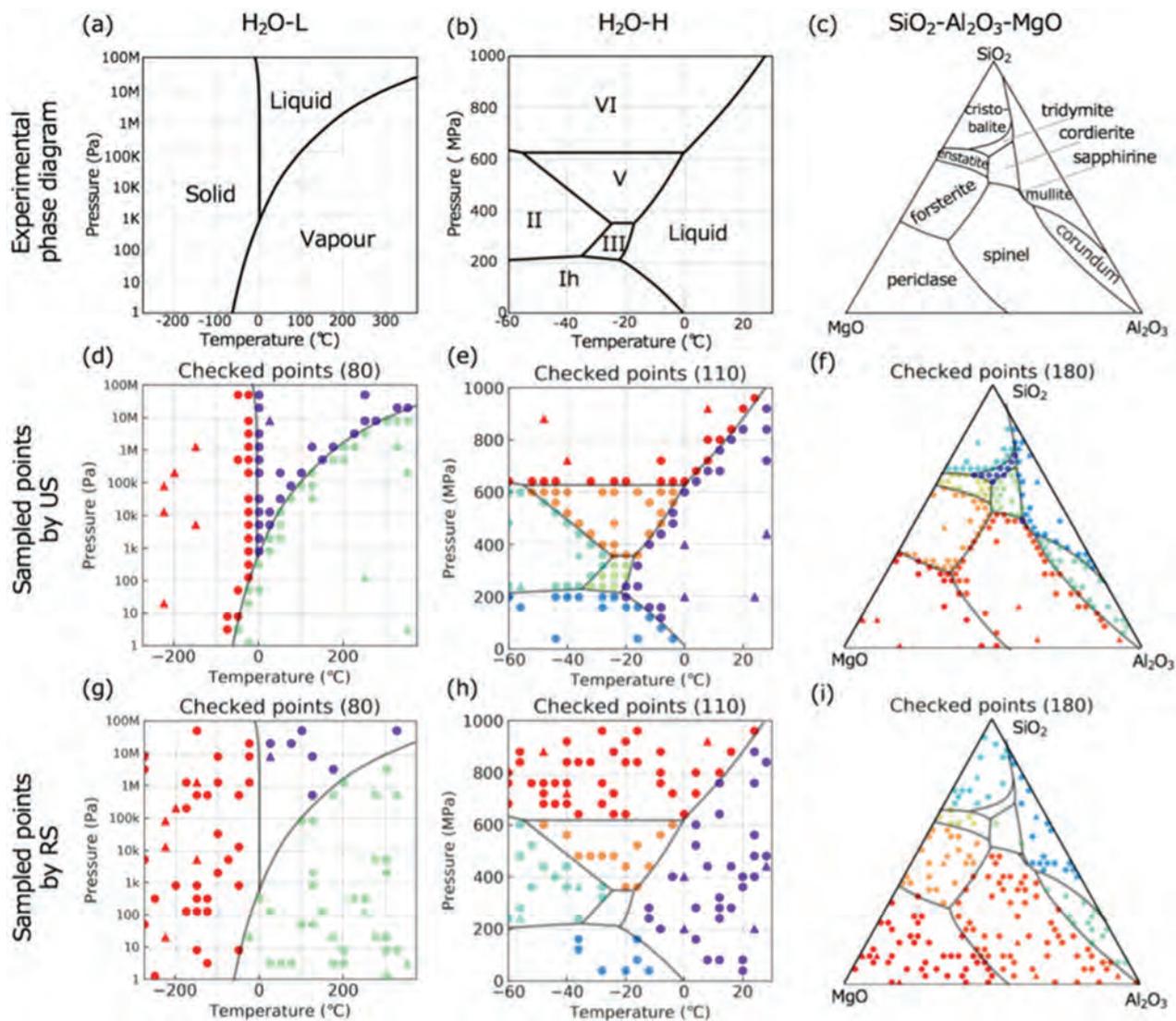


Figure 8. Sampling results for known phase diagrams: low pressure and high pressure phase diagrams of water, and the liquidus projection of the $\text{SiO}_2\text{-Al}_2\text{O}_3\text{-MgO}$ pseudo-ternary system. The top panels are the identified phase regions. The middle and bottom panels are the sampling results by the PDC algorithm and random sampling, respectively. Reprinted from ref. [107] under a creative commons attribution 4.0 international license.

candidate experimental conditions. We demonstrated the effectiveness of our strategy by constructing ternary phase diagrams of alloy systems, incorporating these ingenuities. Even with initial knowledge limited to single-component systems, our approach significantly reduces the number of experiments required to construct a phase diagram, achieving approximately a 1/8 reduction compared to random sampling. Thus, by incorporating scientific knowledge into PDC, further acceleration can be achieved.

2.3.3.2. Treatment of high-throughput batch experiments. Recent advancements in laboratory automation and robotics enable high-throughput batch experiments. To leverage this capability, multiple experimental conditions must be selected simultaneously to effectively construct a phase diagram using machine learning techniques. We proposed strategies to select multiple conditions and compared their

performance in exploring ternary isothermal sections (two-dimensional) and temperature-dependent ternary phase diagrams (three-dimensional) [111]. Our results demonstrated that even when exploring several suggestions simultaneously instead of one at a time, the overall performance does not change significantly. Therefore, we concluded that employing the PDC algorithm with multiple suggestions is suitable for high-throughput batch experiments.

2.3.3.3. Web application. Visualization is important for phase diagrams. We released the web application version of the PDC algorithm on <https://aiphad.org/>. The corresponding Python code is distributed as AIPHAD on <https://github.com/NIMS-DA/aiphad>. Using the web application, PDC can be used directly without any programming skills, and the suggestions for the next experiments, the uncertainty score mapping, and the predicted phase diagram are displayed in

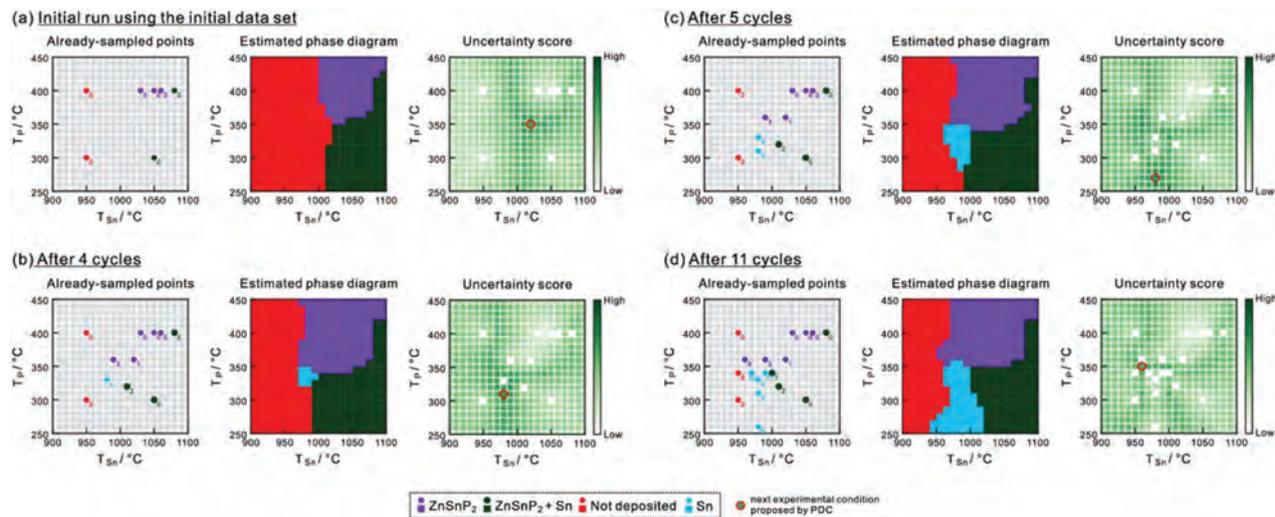


Figure 9. Procedures for constructing a phase diagram for the MBE growth of Zn–Sn–P films by the PDC algorithm. For each step, sampled points, an estimated phase diagram, and an uncertainty score are shown. Reprinted from ref. [109] under a Creative Commons attribution-NonCommercial license.

the browser. We have demonstrated that the phase diagrams in the Fe-Ti-Sn system exhibiting a Heusler phase can be efficiently constructed using this web application [112].

2.3.4. Phase diagram prediction by combining machine learning and the CALPHAD method

CALPHAD is the most widely used method to model and calculate phase diagrams. Combining machine learning methods and CALPHAD databases is an important future development. As a first step, we attempted to estimate unknown phase diagrams from known phase diagrams obtained by CALPHAD calculations using machine learning [113]. We focused on predicting the number of coexisting phases in each of the ten 800 K isothermal ternary sections within the Al-Cu-Mg-Si-Zn system based on the other nine sections. Using the random forest classifier, we achieved an average prediction accuracy of 84% across all ten sections, with a standard deviation of 11%. This indicates that the machine learning model can accurately predict whether a section contains single-phase, two-phase, or three-phase regions most of the time. Here, to improve prediction accuracy, we introduced new descriptors derived from the thermodynamic properties of the elements and extrapolations using the CALPHAD method. This approach is of general interest to efficiently develop new materials by predicting phase equilibria, compounds, and solid and liquid solutions in systems where phase diagrams are initially unknown. Besides, it can also be used for the initialization of PDC to select the first experiments where three-phase regions that provide a lot of information are likely to be found. This work shows that by leveraging machine learning and CALPHAD data, we can accelerate the discovery and design processes in materials science.

Phase equilibria with the liquid phase are important for the development of all kinds of materials, but currently they cannot be predicted from high-throughput experiments or simulations. Therefore, as a second step, we focused on the prediction of the liquidus [114]. We proposed a general framework for predicting binary liquidus from the properties of the pure elements and thermodynamic properties calculated using Miedema’s semi-empirical model [115]. Our framework combines three machine learning models trained and evaluated on liquidus data collected from 466 CALPHAD assessments of binary phase diagrams [114] (see Figure 10). The first model predicts the presence of liquid miscibility gaps with high accuracy (95.3%). The second and third models predict the onset temperature of solidification and the critical temperature of liquidus miscibility gaps, respectively. Using our framework, we successfully predicted the liquidus temperatures in 1563 binary systems not included in our CALPHAD dataset, many of which were previously unknown. An important feature of our models is their ability to predict the presence of eutectics. This is significant because eutectic alloys have many applications such as brazing, coolants in fast-neutron nuclear reactors, or latent thermal energy storage.

3. Experimental approaches toward the understanding, design, and exploration of new semiconductors and dielectrics

3.1. Nitride semiconductors

3.1.1. Earth-abundant nitride, Ca(Mg_{1-x}Zn_x)₂N₂ (x = 0–1): polycrystalline bulk and epitaxial film

Recent progress in computational materials science and materials informatics is remarkably fast and

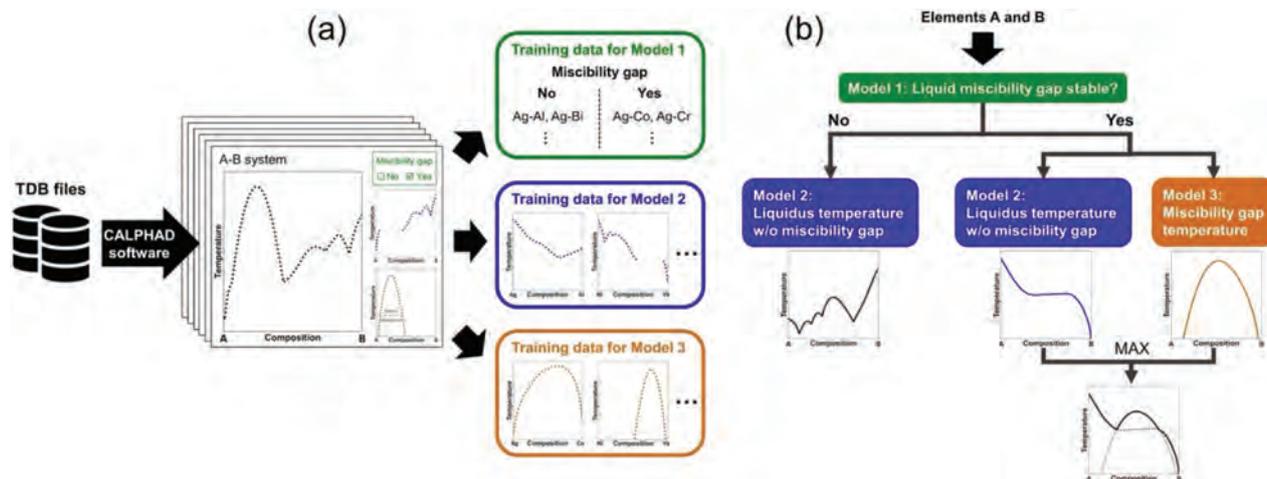


Figure 10. Framework for predicting binary liquidus. (a) Training data of 466 binary systems for three models. (b) Three machine learning models and procedures for predicting binary liquidus from any two elements. Reprinted from ref. [114] under a creative commons attribution 4.0 international license.

leads to important predictions and subsequent experimental verification of novel (i.e. previously unknown) functional complex materials [13,116–124]. An earth-abundant ternary nitride, CaZn_2N_2 , was proposed through computational screening using first-principles calculations as a promising optoelectronic functional nitride semiconductor with a band gap of 1.8 eV because of its small electron and hole effective masses and direct transition-type energy band structure [117]. It was also predicted that the band gap could be widely tuned from the ultraviolet region to the red – infrared one by simple element-substitutions such as Mg or Cd substitution at the Zn site and Sr substitution at the Ca site [117,122], suggesting that this nitride would be a promising host candidate of a light-emitting semiconductor and/or a photovoltaic semiconductor devices with appropriately tuned band gaps.

Based on the prediction, polycrystalline samples of $\text{Ca}(\text{Mg}_{1-x}\text{Zn}_x)_2\text{N}_2$ ($x = 0-1$) were successfully synthesized via a solid-state reaction using binary nitride precursors at ambient pressure or an extremely high pressure (5 GPa) [125]. It is experimentally validated that the optical band gap can be continuously tuned from 3.2 eV to 1.8 eV (i.e. the ultraviolet-to-red region) and the corresponding band-to-band photoluminescence (PL) is observed even at room temperature [125] (Figure 11(a)). In the entire x range, the observed band gaps are close to those predicted by first-principles calculations [117,125] (Figure 11(b)). Additionally, heteroepitaxial growth of a CaZn_2N_2 (i.e. $x = 1$) thin film was demonstrated by MBE with precise optimization of growth conditions [126] (Figure 11(c,d)). Three key factors for heteroepitaxy were found: (i) the precise tuning of the individual flux rate of Knudsen cells for Ca and Zn, (ii) the use of GaN template layer on c -plane sapphire as a substrate, and (iii) the application of MBE with an active nitrogen radical source because other attempts at physical vapor deposition and thermal annealing

processes did not produce pure CaZn_2N_2 films. Since the GaN template layer exhibits non-negligible n -type conduction, reliable carrier transport measurement was difficult. Thus, the electronic transport properties of CaZn_2N_2 films deposited on a yttria-stabilized zirconia (YSZ) single-crystal and a silica glass were evaluated by Hall-effect measurements at room temperature. It was clarified that the dominant carrier types were p - (holes) and n -type (electrons) for the films on YSZ and silica glass, respectively. This ambipolar carrier-doped state was unintentionally formed, while would be an advantageous dopability to fabricate a pn homojunction. The estimated carrier densities of both the films were as low as 10^{13} cm^{-3} order, which implies that CaZn_2N_2 has potential for controlling carriers over a wide range from intrinsic (i.e. close to insulating), semiconducting, through to degenerate states by impurity doping. Despite the theoretically predicted small effective masses [117], the carrier mobilities were as low as $0.3 \text{ cm}^2/(\text{V}\cdot\text{s})$ for holes (YSZ) and $4.3 \text{ cm}^2/(\text{V}\cdot\text{s})$ for electrons (silica glass) due probably to the poor crystallinity of the films. These studies on polycrystalline bulks and epitaxial films experimentally clarified that the earth-abundant nitride, $\text{Ca}(\text{Mg}_{1-x}\text{Zn}_x)_2\text{N}_2$ ($x = 0-1$), can be a promising semiconductor material that plays an emitting layer in a wide-range wavelength-tunable light-emitting diode or a light absorber layer in a photovoltaic cell. However, to achieve superior carrier transport, different nitridation processes must be essential to overcome a dilemma; i.e, high nitrogen chemical potential requires low substrate temperature, but high crystallinity usually does high substrate temperature.

Similar to the discovery of CaZn_2N_2 by computational screening [117], the exploration of new functional ternary nitrides has been actively performed by considering thermodynamic stability/metastability and predicting functionality [118,120,127]. Even though nitrides particularly remain unexplored

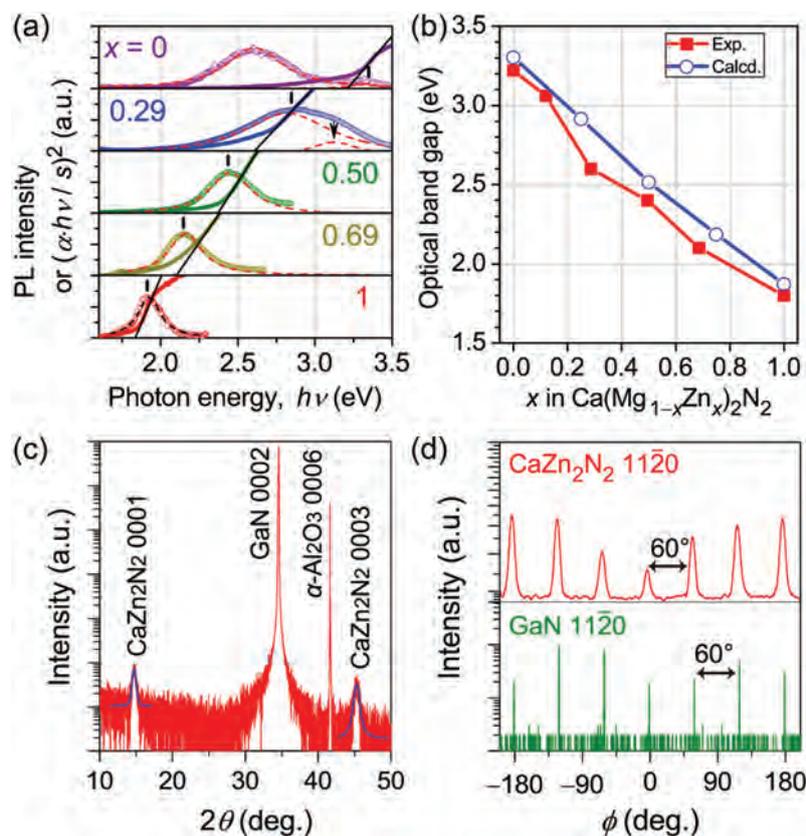


Figure 11. Optical and structural characteristics of $\text{Ca}(\text{Mg}_{1-x}\text{Zn}_x)_2\text{N}_2$. (a) Photoluminescence (PL) and absorption $(ah\nu/s)^2$ spectra of polycrystalline samples at room temperature. (b) Optical band gaps (Exp.) estimated from (a). The calculated ones (Calcd.) are shown for comparison. (c, d) X-ray diffraction profiles of an epitaxial CaZn_2N_2 (i.e. $x = 1$) thin film with ~ 50 nm thickness, optimally grown by MBE, on a GaN template layer on *c*-plane sapphire substrate. (c) Out-of-plane 2θ scan. (d) In-plane ϕ scan. Adapted with permission from refs. [125,126]. Copyright 2019 American chemical society.

mainly due to difficulty in synthesis, it has recently become to overcome the difficulty owing to advancing various synthetic techniques [128]. A representative demonstration would be a successful synthesis of new ternary perovskite-type nitrides and II-IV- N_2 type ones. Although nitride perovskites, excluding TaThN_3 [129] and anti-perovskite-type nitrides [130], had not been reported due to the synthesis difficulty, first-principles studies have proposed unreported but thermodynamically stable perovskite-type nitrides as well as their electronic structures [131]. After that, among the predicted ternary nitrides, LaReN_3 (in 2021) [132] and LaWN_3 (in 2021–2023) [133–136] have been experimentally synthesized. Recently, novel ternary II-IV- N_2 compounds such as AeTmN_2 ($\text{Ae} = \text{Ca}, \text{Sr}, \text{Ba}$; $\text{Tm} = \text{Ti}, \text{Zr}, \text{Hf}$) [137,138] have also been successfully synthesized. Therefore, a large research platform on ternary nitrides may be about to expand more actively and rapidly.

3.1.2. Thin-film growth of perovskite-type LaWN_3 : toward ferroelectric semiconductor applications

Ferroelectric photovoltaic cells are expected to offer highly efficient electric power generation owing to their extremely high operating voltages [3]. The ferroelectric photovoltaic effect has been examined mainly

in perovskite-type oxides such as BiFeO_3 . However, the conversion efficiency of ferroelectric photovoltaic cells based on the related oxides remains low ($\sim 8\%$) [139]. One of the origins seems to be its low sunlight absorption efficiency originating from the wide band gap of ~ 2.7 eV for BiFeO_3 [140].

Under such a circumstance, a perovskite-type nitride, LaWN_3 , has been expected to have potential as a rare and new ferroelectric semiconductor because it is predicted to exhibit ferroelectricity with a high spontaneous polarization of ~ 60 $\mu\text{C}/\text{cm}^2$ [141] and have a narrower band gap of approximately 1.6–1.8 eV according to first-principles calculations [134,142]. These characteristics may make it more suitable for applications as the light-absorbing layer of ferroelectric photovoltaic cells than the perovskite-type oxides. However, the ferroelectric properties of LaWN_3 have yet to be experimentally confirmed: only a piezoelectric response was reported for a sample synthesized [133]. In addition, the LaWN_3 samples reported so far are crystallographically non-oriented polycrystalline [133,134], which hinders the predicted ferroelectric polarization. Thus, it was strongly required to fabricate a highly oriented epitaxial film, especially the *c*-axis, which is a polar direction of LaWN_3 with a rhombohedral $R3c$ space group.

LaWN₃ epitaxial films with ~65 nm thickness were grown on *c*-plane α-Al₂O₃ substrates with the multi-cathode rf magnetron sputtering guns and nitrogen gas [135]. A substrate temperature (*T_s*) higher than 1000°C and optimization of rf power densities of the La and W cathodes were essential for heteroepitaxial growth (Figure 12(a)). The heteroepitaxial relationship between the film and the substrate is LaWN₃ [0001] || α-Al₂O₃ [0001] for out-of-plane and LaWN₃ [1010] || α-Al₂O₃ [1120] for the in-plane; i.e. *c*-axis of LaWN₃ is oriented perpendicular to the *c*-plane of α-Al₂O₃ substrate and the in-plane domain of LaWN₃ is rotated by 30° with respect to that of α-Al₂O₃ (Figure 12(b)). Electronic transport measurements for the obtained epitaxial film revealed that the film exhibited no temperature dependence of *n*-type electrical conductivity (σ) (i.e. a heavily electron-doped degenerate semiconductor) (Figure 12(c)). Because the carrier density was as high as $2 \times 10^{22} \text{ cm}^{-3}$ and the corresponding sample resistance was too low (<1 kΩ), the predicted ferroelectricity was not observed as also reported in Ref. [133]. Optical measurements indicated that the high-density carriers induced free carrier absorption over near-infrared wavelengths and widened the optical band gap to 1.83 eV owing to the band-filling effect (Figure 12(d)). Moreover, the film exhibited

superconductivity at 0.7 K (inset of Figure 12(c)). The degenerate state and superconductivity are attributed to the high-density electron carriers that resulted from the unavoidable off-stoichiometry (excess of W and/or N deficiency) to achieve heteroepitaxial growth.

Reports on successful synthesis of LaWN₃ samples are still scarce because extremely high external pressure or precise control of film-growth conditions is necessary for polycrystalline powders [134,136] or for thin films [133,135], respectively. However, fundamental researches on LaWN₃ have been gradually progressing. A recent research on neutron diffraction reveals that the nitrogen vacancy stabilizes an orthorhombic structure, not rhombohedral *R3c*, with a polar symmetry (*Pna2₁*) that possesses a unique atomic polarization along the *c*-axis [136]. Additionally, a recent research on thin films clarifies that the electrical resistivity of LaWN₃ is quite sensitive to La/(W+La) in the films (i.e. cation ratio) [143]. Thus, different growth processes that realize a higher nitrogen chemical potential will be required to achieve a stoichiometric chemical composition for the realization of lower carrier density and observation of the predicted ferroelectricity. Because this perspective is almost the same as the case of CaZn₂N₂ in the former section, the

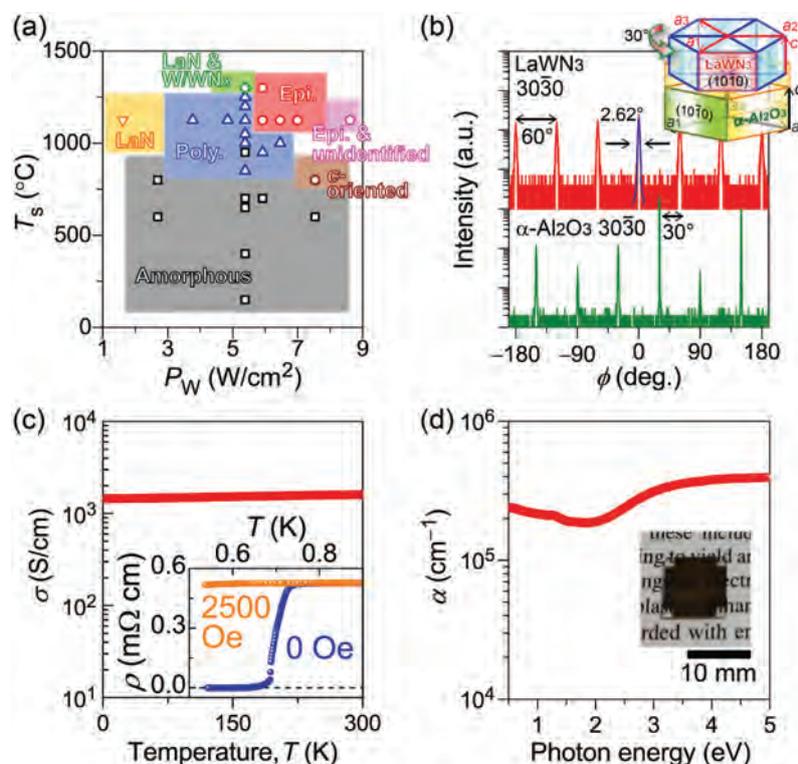


Figure 12. Heteroepitaxial growth and optoelectronic properties of perovskite-type LaWN₃ thin films. (a) Optimization of film-growth conditions: *T_s* and *P_w* denote substrate temperature and rf power density of a tungsten cathode, respectively. The optimum region for epitaxial growth is indicated by red. (b) Epitaxial relationship of the optimally grown film. (c) Temperature dependence of the electrical conductivity (σ) of the epitaxial film. The inset shows resistivity ($\rho = 1/\sigma$) in an extremely low temperature region of ≤ 0.9 K under external magnetic fields. (d) The optical absorption coefficient (α) spectrum of the epitaxial film. The inset is a sample picture. Adapted with permission from ref. [135]. Copyright 2023 American chemical society.

development of a new active nitrogen source apparatus alongside a new concept would be a critical issue to advance research on these nitride semiconductors.

3.2. Phosphide semiconductors for photovoltaic applications

Recently, some novel semiconductors for photovoltaics have been investigated with the aid of theoretical calculations and/or machine learning. We focused on ternary zinc phosphides because they have suitable electronic structures and phosphorus vacancies are less likely to form mid-gap states considering the behavior reported for the binary zinc phosphide [144]. In particular, $ZnSnP_2$ (ZTP) with a chalcopyrite structure is an emerging absorber in view of its light-absorption and electric characteristics. To achieve higher energy conversion efficiency in ZTP solar cells, the suppression of the recombination of photo-generated carriers is a key issue. Point defects in bulk regions and heterointerfaces in devices can be sites of carrier recombination. Therefore, we explored ways to understand and improve the carrier recombination properties through a combination of experimental and theoretical studies as described below. In addition, new materials for photovoltaic applications were investigated.

3.2.1. Evaluation and control of native point defects in phosphide semiconductors

Identifying deep and shallow levels in ZTP is the first step to understand the carrier recombination properties in absorber materials. Deep level transient spectroscopy (DLTS) is well known to be a technique to detect trap levels in semiconductors and has also been applied to recent state-of-the-art devices. A discussion based on the formation energies and energy levels of point defects derived from first-principles calculations is fruitful for conventional materials of thin-film solar cells and even for emerging absorber materials. Actually, it was clearly shown for some absorber semiconductors that

emissions observed in PL spectra consist of some recombination processes related to shallow and deep levels. In this study, ZTP bulk crystals obtained by a conventional solution growth were characterized by DLTS and photoluminescence (PL) for discussion on levels related to point defects in conjunction with the results of first-principles calculations by Kumagai et al. [145].

A conventional DLTS measurement and theoretical calculations revealed two shallow trap levels coming from antisite defects and a deep trap state [146], but detailed properties were still unknown. Therefore, advanced DLTS and minority carrier transient spectroscopy (MCTS) using the correlation function method were carried out in cooperation with Ceramicforum Co. Ltd. [147]. Figure 13(a) shows the Arrhenius plot for the time constant t of hole or electron emission obtained from DLTS and MCTS spectra, and five (H1–H5) and two (E1, E2) traps were evaluated for hole and electron, respectively [147]. The capture cross section and the activation energy corresponding to a trap level can be evaluated by analyzing the plot based on the Shockley – Read – Hall model, and the trap density was obtained from the capacitance measurements. The trap density is in the range of 10^{14} – 10^{15} cm^{-3} in all traps, whereas the capture cross section of trap E1 is 4.0×10^{-11} cm^{-2} which is 2–5 orders of magnitude larger than others, resulting in the much shorter time constant. This means that trap E1 might have a dominant contribution to the capture of electrons from the CBM. The theoretical calculations suggest that trap E1 originates from the Sn antisite defect on the Zn site, Sn_{Zn} in the viewpoints of the activation energy and thermodynamic transition level.

Next, we attempted to suppress the formation of Sn_{Zn} antisites by controlling the chemical potentials during crystal growth. According to the theoretical calculations, the formation energy of Sn_{Zn} is higher under the conditions of higher chemical potential of Zn. Conventionally, ZTP crystals were prepared with

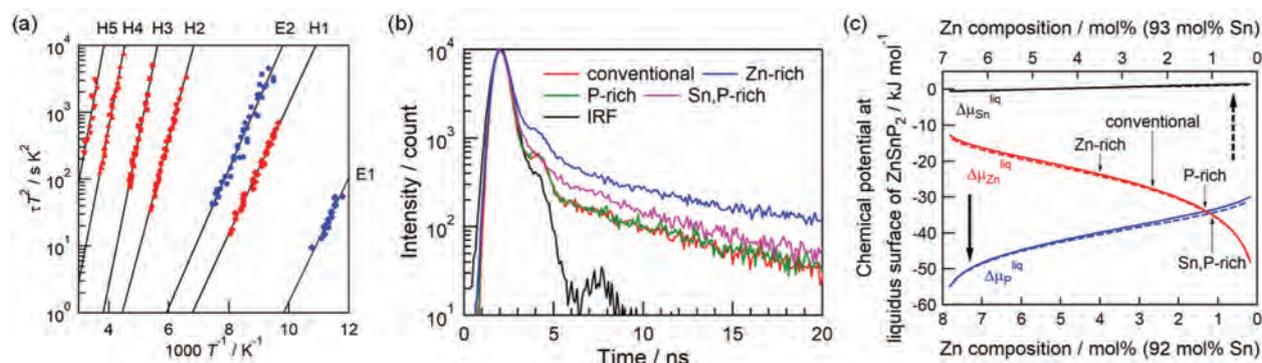


Figure 13. (a) Arrhenius plot of time constant from DLTS and MCTS. (b) TRPL spectra of ZTP crystals. (c) Chemical potentials of Zn, Sn, and P in equilibrium of ZTP and liquid solution. Reproduced from ref. [147] with the permission of AIP publishing.

the composition of the Sn-ZTP pseudo-binary system. We prepared crystals with various initial compositions such as Zn-rich, P-rich, and Sn,P-rich in addition to the conventional composition. Time-resolved photoluminescence (TRPL) spectra of ZTP crystals shown in Figure 13(b) revealed that the TRPL lifetime is longer for crystals in the Zn-rich condition, which is consistent with the decrease in the concentration of Sn_{Zn} [147]. However, the actual defect formation behavior depends on the chemical potentials of constituent elements, and then they were evaluated using a sub-regular solution model under the condition in equilibrium ZTP solid and liquid solution, in other words, precipitation from liquid. Figure 13(c) clarifies that the chemical potential of Zn in the Zn-rich condition is higher than those in other conditions and suggests the suppression of the Sn antisite formation [147].

As seen in the above, a fusion of experiments and calculations is effective in the discussion of defect formation based on chemical potentials. It is desired that a more quantitative discussion will be developed in the future.

3.2.2. Interfaces for phosphide solar cells

In thin-film solar cells with compound semiconductors, there are some heterointerfaces because of the multilayer structure with various materials. As mentioned above, interfaces can be carrier recombination sites and the effect appears in the resistance. Since ZTP is an emerging absorber, every interface should be examined and controlled in addition to the *pn* junction. We first focused on the interface between the ZTP absorber and the back electrode.

In solar cells consisting of sulfide or selenide absorbers, Mo is conventionally used as a back electrode to obtain ohmic contact. However, our previous work clarified that the series resistance of ZTP solar cells could be reduced by applying Cu to the electrode rather than Mo [148]. The Cu/ZTP interfacial structure and its impact on carrier transport behavior are discussed here.

Annealing experiments of Cu-deposited ZTP crystals revealed the formation of a copper phosphide, Cu_3P , at the interface, leading to the reduction of the resistance. In this case, selected area electron diffraction (SAED) suggested Cu_3P epitaxially formed on ZTP crystals with the relationship of $\text{Cu}_3\text{P}[1100](0001)//\text{ZTP}[110](112)$. Additionally, it was revealed that the IP of ZTP and the work function of Cu_3P are comparable by photoelectron yield spectroscopy (PYS). These results indicate that the heterostructure with lattice-matched and band-aligned is required for the suppression of carrier recombination at interfaces. This interface structure was applied to solar cells. In particular, the insertion of Cu_3P before annealing enhanced the solar cell performance, as shown in Figure 14, and consequently, the best energy conversion efficiency of 3.87% was achieved [149].

Next, turning to the *pn* junction, an *n*-CdS/*p*-ZTP structure was adopted in the solar cell with the best efficiency. CdS is a conventional *n*-type layer for compound thin-film solar cells, but the large CBM offset at the *n*-CdS/*p*-ZTP interface was suggested by X-ray photoelectron spectroscopy (XPS) [150]. Considering the discussion on the Cu_3P /ZTP interface, CdSnP_2 (CTP) is a candidate as an *n*-type partner with ZTP. The crystal structure of CTP is also chalcopyrite and the first-principles calculations suggest a smaller CBM offset in the *n*-CTP/*p*-ZTP junction compared to CdS/ZTP [151]. Samples including a CTP/ZTP interface were prepared by phosphidation of Cd-Sn precursor films on ZTP crystals [152,153] and the current-voltage curve was measured. The epitaxial relationship between the deposited CTP and ZTP crystals is clearly confirmed from the scanning transmission electron microscope (STEM) dark-field (DF) image and SAED patterns in Figure 15 and a good rectification was obtained in the *n*-CTP/*p*-ZTP junction [153]. A higher performance of ZTP solar cells is expected by adopting this junction. Our work means that the research assisted by theoretical calculations is also effective in the interface design in devices.

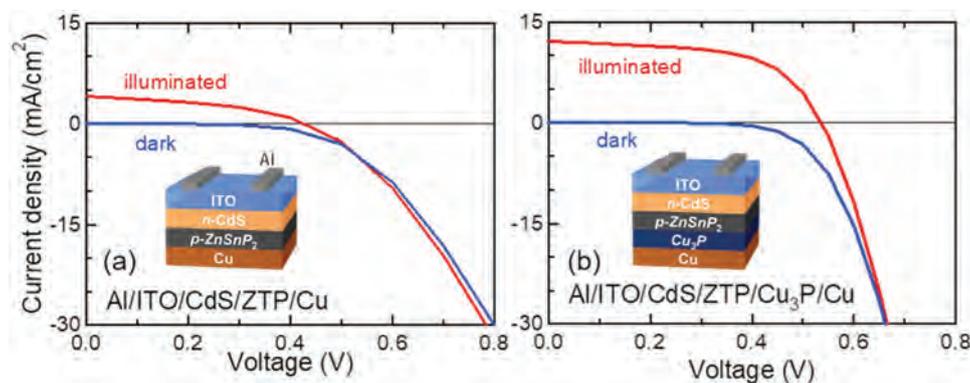


Figure 14. Current density-voltage characteristics of ZTP solar cells (a) without and (b) with insertion of Cu_3P . Created based on ref. [149]. Copyright 2020 Elsevier B.V. All rights reserved.

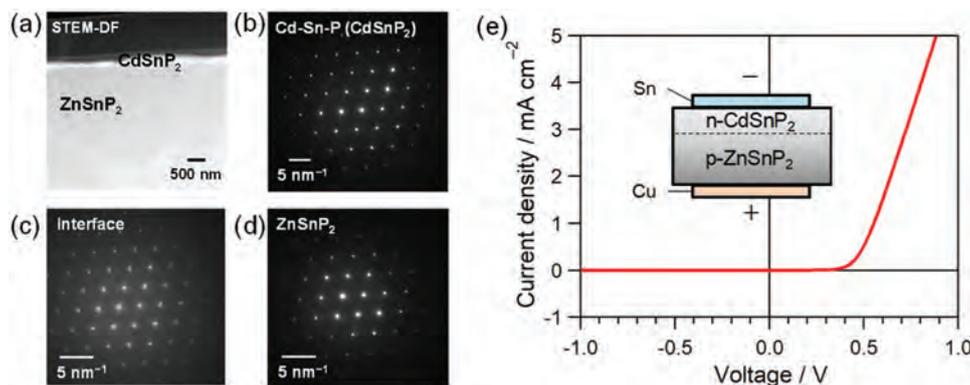


Figure 15. (a) STEM-DF image of CTP/ZTP interface and corresponding SAED patterns of (b) CTP, (c) interface, and (d) ZTP regions. (e) Current density-voltage curve for *n*-CTP/*p*-ZTP junction. Reproduced from ref. [153] with permission. Copyright IOP publishing. All rights reserved.

3.2.3. Novel materials for photovoltaic applications

We also tried to develop novel materials with the assistance of theoretical calculations. Among ternary compounds containing Zn and P, compounds including alkaline earth elements were suggested for potential photovoltaic applications from the viewpoints of semiconductor properties such as the band gap and effective mass.

For *X*-Zn-P compounds (*X*: group II elements), we investigated $MgZn_2P_2$ formed at the Mg/Zn_3P_2 junctions in Zn_3P_2 -based solar cells. The detailed nature of this device had been controversial for a long period, but we have recently revealed that this is a heterojunction of semiconductors between Zn_3P_2 and $Mg(Mg_xZn_{1-x})_2P_2$ formed through the reaction at the Mg/Zn_3P_2 interface [154]. The lattice mismatch between $Mg(Mg_xZn_{1-x})_2P_2$ and Zn_3P_2 is 0.5% at most and, thus, is favorable for carrier transport. $Mg(Mg_xZn_{1-x})_2P_2$ can be regarded as a solid solution of $MgZn_2P_2$. Therefore, our work provided the preparation of $MgZn_2P_2$ although there was no report on its synthesis. Theoretical calculations have suggested

that the stable crystal structure of $Mg(Mg_xZn_{1-x})_2P_2$ is the trigonal $CaAl_2Si_2$ -type shown in Figure 16(a). Inspired by this compound, we focused on a series of compounds, $CaZn_2P_2$ (CZP), $SrZn_2P_2$ (SZP), and $BaZn_2P_2$ (BZP), which are classified as Zintl compounds and still in an undeveloped group. A synthesis route using Sn solution as a solvent was adopted to suppress the reactivity of group II elements. Figure 16(b) shows XRD profiles of samples after the heat treatment to obtain compounds [155]. The trigonal $CaAl_2Si_2$ -type phases were identified in the profiles of CZP and SZP although they include diffractions from Sn and secondary phases, while the profile for BZP was assigned to a tetragonal $ThCr_2Si_2$ -type structure. The unidentified phase was observed in the CZP-Sn sample, but it might be the products by the decomposition of calcium phosphides and not affect the later evaluation of optoelectronic properties. The UV-vis diffuse reflectance analyses evaluated the indirect (1.7–1.85 eV) and direct band gaps (1.9–2.05 eV) for CZP and SZP, whereas it was difficult to evaluate for BZP because its band

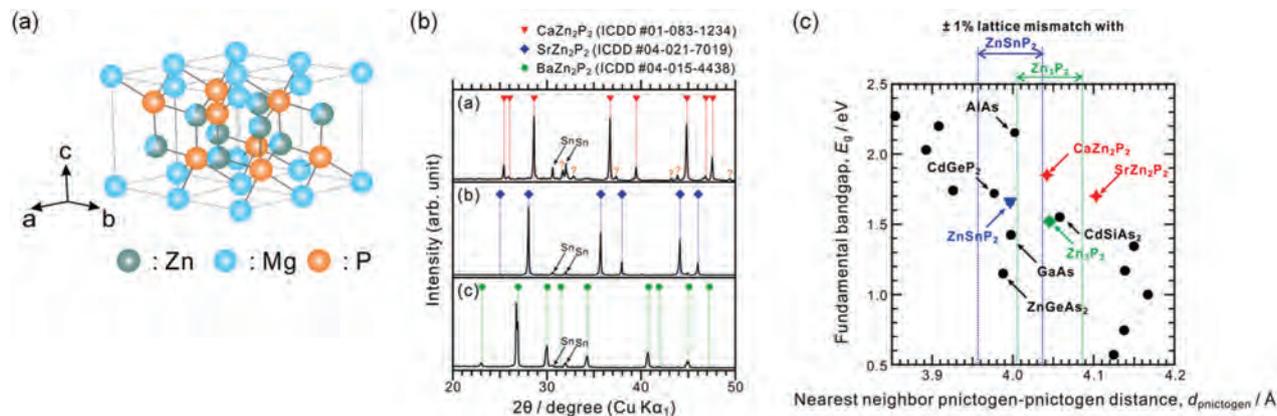


Figure 16. (a) Structure of $Mg(Mg_xZn_{1-x})_2P_2$ (trigonal $CaAl_2Si_2$ -type). (b) XRD profiles for $CaZn_2P_2$, $SrZn_2P_2$, and $BaZn_2P_2$. (c) Relationship between the nearest-neighbor pnictogen–pnictogen distance and fundamental band gap for pnictide semiconductors. Reprinted with permission from ref. [154]. Copyright 2018 American chemical society and from ref. [155] under a creative commons attribution 4.0 international license.

gap may be much narrower. The theoretical calculations also confirmed the crystal structures and the trend that their direct band gaps are slightly larger than the indirect band gaps.

The evaluated direct band gaps for CZP and SZP are somewhat too large for application to absorbers, but they can be applied to buffer layers like Mg (Mg_xZn_{1-x})₂P₂ in Zn₃P₂ solar cells. PYS measurements provided that the IPs for CZP and SZP are similar to those of ZTP and Zn₃P₂, leading to their small CBM offset within 0.2 eV. On the other hand, Figure 16(c) summarizes the relationship between fundamental band gap and nearest neighbor pnictogen – pnictogen distance, which is an index for lattice-matching, for various pnictide semiconductors. The dotted lines indicate the range of 1% lattice mismatching with ZTP and Zn₃P₂, and especially it is found that CZP is well-matched with phosphide absorbers. Consequently, CTP would be an appropriate partner material in Zn₃P₂- and ZTP-based solar cells from the viewpoints of lattice and band matching.

In previous works, the calculation-assisted developments of materials for photovoltaic applications are focused mostly on absorbers. However, lack of investigation of *n*-type buffer materials also limits the development of novel solar cells. This work indeed suggests that the investigation of such materials would be required.

3.3. Layered perovskite dielectrics

Ferroelectric oxides are fascinating targets for the exploration of functional materials due to their diverse properties, which make them valuable for various applications such as non-volatile memories and high-performance piezoceramics. Traditionally, ferroelectric oxides have been developed in perovskite-type compounds with the composition ABO₃, exemplified by BaTiO₃ and Pb(Zr,Ti)O₃ (PZT). Extensive experimental and theoretical studies have established several guiding principles for designing ferroelectric perovskite-type compounds. These principles include the second-order Jahn-Teller (SOJT) effect [156–162], the tolerance factor [163], and the stereochemical activity of lone-pair electrons [164,165].

The field of exploration for ferroelectric materials is expanding beyond perovskite-type compounds to include non-perovskite types such as fluorite-type [166–169], layered-silicate-type [170–172], and aluminate-sodalite-type compounds [173–175]. Moreover, the concept of hybrid improper ferroelectricity (HIF) has recently inspired renewed interest in layered-perovskite-type compounds [176–180]. HIF can be understood as ferroelectricity induced by a special combination of lattice instabilities at finite wave vectors [177]. Specifically, in layered perovskite-type compounds, the rotation of corner-shared oxygen octahedra contributes to the emergence of

ferroelectricity [179,181–184]. This contradicts the conventional understanding that non-polar lattice vibrations do not result in spontaneous polarization when they freeze. This discovery has underscored the significance of octahedral rotation in the design of ferroelectric materials within layered-perovskite-type compounds.

There are three typical structures in layered perovskite-type compounds: Aurivillius-type, Dion-Jacobson-type (DJ-type), and Ruddlesden-Popper-type (RP-type) phases. Li₂SrNb₂O₇ crystallizes in a similar structure but slightly different from the RP-type. As shown in Figure 17, an ‘*n* = 2’ RP-type structure of Ca₃Mn₂O₇ consists of perovskite-type building blocks and rocksalt-type CaO layers, which stack along the [001] direction. In the pseudo-RP-type structure of Li₂SrNb₂O₇, however, Li₂O layers are inserted instead of rocksalt-type layers, reducing the distance between perovskite-type building blocks. This section presents our finding that the close competition between the SOJT effect and the octahedral rotation in Li₂SrNb₂O₇ creates a fertile ground for designing various dielectric properties, including ferroelectricity and also antiferroelectricity [185–188].

Figure 18(a) presents the temperature dependence of complex dielectric permittivity for a Li₂SrNb₂O₇ polycrystalline sample over a temperature range from 4.2 K to 300 K [185]. The top and bottom panels show the real (ϵ') and imaginary (ϵ'') parts, respectively. As shown in the figure, ϵ' increases upon cooling from 300 K, reaching a maximum value of approximately 210 around 220 K. Upon further cooling, it decreases,

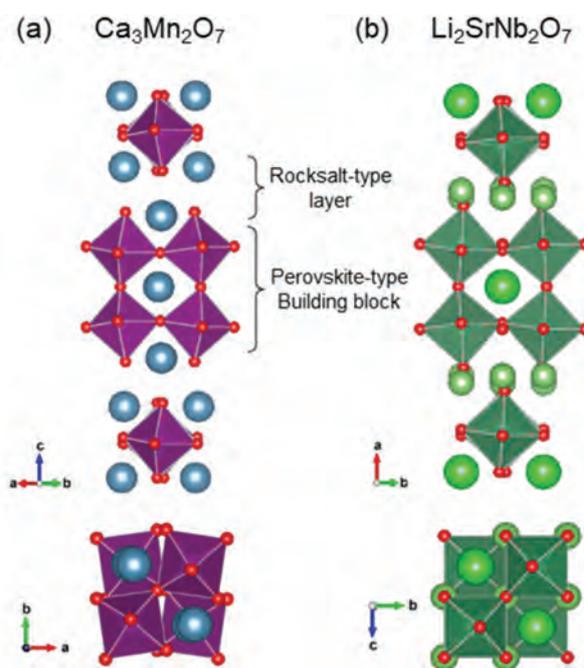


Figure 17. Crystal structures of (a) ruddlesden-popper and (b) pseudo-ruddlesden-popper phases. Adapted with permission from ref. [185]. Copyright 2019 American chemical society.

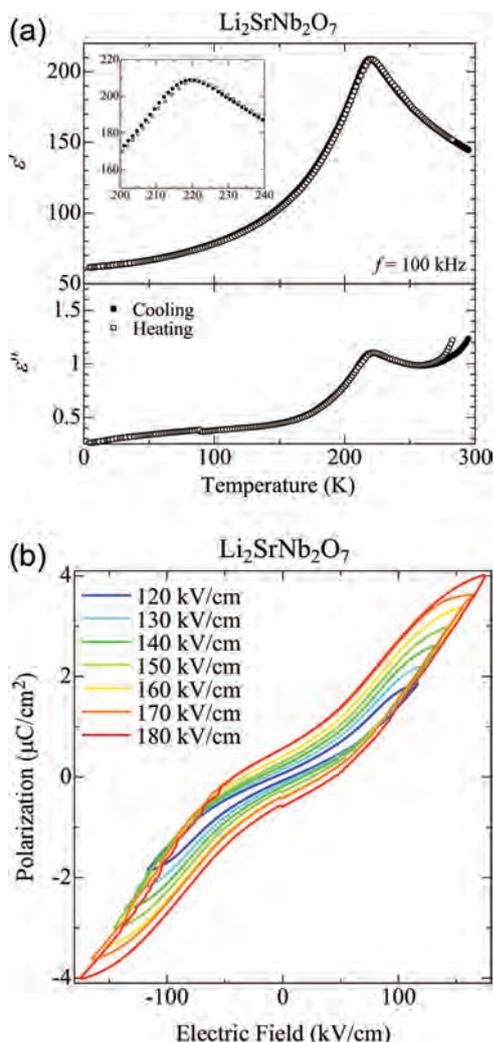


Figure 18. (a) Temperature dependences of the real and imaginary parts of the complex (relative) permittivity of $\text{Li}_2\text{SrNb}_2\text{O}_7$, measured at a frequency of 100 kHz. (b) p - E hysteresis loops of $\text{Li}_2\text{SrNb}_2\text{O}_7$, measured at a frequency of 10 hz at 80 K. Adapted with permission from ref. [185]. Copyright 2019 American chemical society.

exhibiting a cusp-like anomaly that indicates a phase transition. A corresponding anomaly in ϵ'' is also observed around 220 K. The inset in the top panel, which magnifies the temperature region where the anomaly occurs, shows slight differences between the results measured during cooling and heating, suggesting a weak first-order phase transition.

Figure 18(b) plots the electric-field-induced polarization of $\text{Li}_2\text{SrNb}_2\text{O}_7$ under various maximum applied electric fields [185]. When an electric field of 120 kV is applied, a slim but clear double hysteresis loop appears, demonstrating the antiferroelectric property of $\text{Li}_2\text{SrNb}_2\text{O}_7$. Notably, a weak polarization remains at zero electric field and gradually increases as the applied electric field rises. Since the polarization measurement was performed at 80 K, where the sample exhibits sufficiently high insulating properties, the influence of

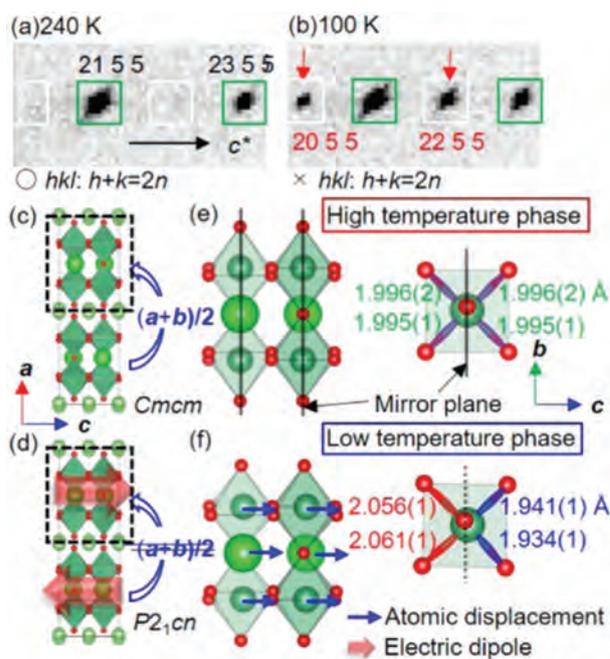


Figure 19. X-ray diffraction patterns of $\text{Li}_2\text{SrNb}_2\text{O}_7$ observed at (a) 240 K and (b) 100 K. The crystal structures in the high- and low-temperature phases are illustrated in panels (c) and (e) and panels (d) and (f), respectively. Adapted with permission from ref. [185]. Copyright 2019 American chemical society.

leakage current can be ruled out. This result suggests that a ferroelectric component coexists in $\text{Li}_2\text{SrNb}_2\text{O}_7$.

Synchrotron X-ray diffraction measurements of $\text{Li}_2\text{SrNb}_2\text{O}_7$ have clarified appearance of superlattice reflections in the low-temperature region (Figure 19(a, b)), indicating that the structural phase transition between $Cmcm$ and $P2_1cn$ symmetries occurs around 220 K during cooling [185]. The polar structure of $P2_1cn$ in the low-temperature phase is consistent with the presence of remanent polarization at zero electric field. As presented in Figure 19(c-f), structural analyses reveal that the $P2_1cn$ structure of $\text{Li}_2\text{SrNb}_2\text{O}_7$ consists of in-plane antipolar displacements and out-of-plane polar displacements of Nb atoms along the [001] and [100] directions, respectively, indicating the coexistence of antiferroelectricity and ferroelectricity in the system [185]. In the high-temperature $Cmcm$ phase, the Sr and Nb atoms are located on a mirror plane normal to the [001] direction. In the low-temperature $P2_1cn$ phase, on the other hand, the mirror plane disappears due to the relative displacements of the Sr and Nb atoms along the [001] direction, inducing a local electric dipole moment within the perovskite-type block. While all local dipoles are parallel within the perovskite-type building block, they align antiparallel between adjacent building blocks, forming an antipolar configuration along the [001] direction. Furthermore, the local dipoles slightly tilt to the [100] direction, which cooperatively generates the spontaneous polarization.

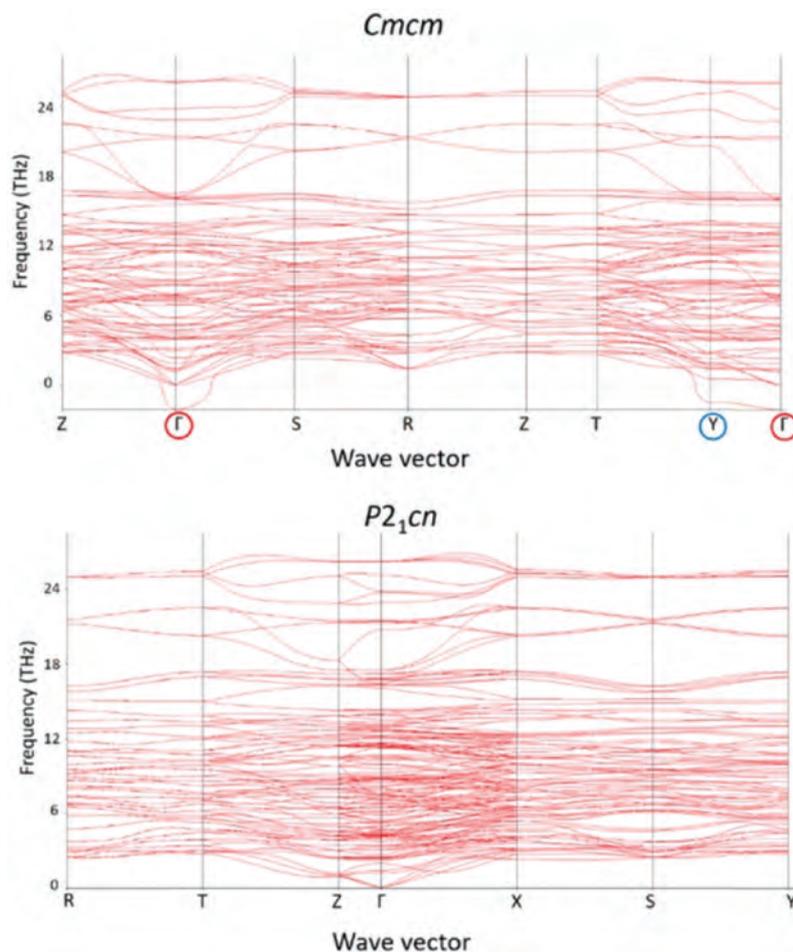


Figure 20. Calculated phonon bands of $\text{Li}_2\text{SrNb}_2\text{O}_7$ in the paraelectric $Cmcm$ phase (top) and the weak-ferroelectric $P2_1cn$ phase (bottom). The reciprocal points Γ and Y with softmodes are indicated in red and blue circles, respectively. Adapted with permission from ref. [185]. Copyright 2019 American chemical society.

First-principles calculations play a key role in elucidating the driving force behind the long-range ordering of local dipoles. Figure 20 presents the phonon dispersion curves of $\text{Li}_2\text{SrNb}_2\text{O}_7$ calculated for the $Cmcm$ (top panel) and $P2_1cn$ (bottom panel) structures [185]. Data points plotted below the origin of the vertical axis in the top panel indicate an imaginary frequency, where the corresponding oscillation is unstable in the given symmetry. As shown in the top panel, there are two imaginary frequency modes at the Γ and Y points in the high-temperature $Cmcm$ phase. These modes are known as ‘soft modes’, and their freezing induces a displacive-type phase transition. In a second-order phase transition, the displacement pattern of the soft mode corresponds to the change in crystal structure through the transition. Although $\text{Li}_2\text{SrNb}_2\text{O}_7$ undergoes a first-order phase transition, as mentioned earlier, the displacement pattern of the soft mode provides insight into the mechanism of the phase transition due to its sufficiently weak first-order nature manifested by the subtle thermal hysteresis in the temperature dependence of the dielectric permittivity. Our first-principles calculations clarified that the most unstable phonon mode is at the Γ point [187]. When the Γ_2^- soft mode

freezes, a polar structure of $Cmcm2_1$ is generated, as shown in Figure 21(b), where spontaneous polarization appears along the [001] direction. However, the $Cmcm2_1$ structure is found to be incompatible with the structure determined experimentally by structural analyses. On the other hand, the freezing of the Y_2^- soft mode results in a $Pmcm$ structure. Although this $Pmcm$ structure is non-polar, it exhibits antipolar ordering of the Nb displacements along the [001] direction, which aligns with the results of the structural analyses. Therefore, it is reasonable to attribute the origin of the long-range ordering of local dipoles to freezing of the Y_2^- soft mode, despite it being slightly less imaginary than the Γ_2^- soft mode, likely due to the accuracy of the calculation. The bottom panel in Figure 20 shows the phonon dispersion curves of $\text{Li}_2\text{SrNb}_2\text{O}_7$ with the $P2_1cn$ structure [185]. There is no soft mode in the system, confirming that the $P2_1cn$ structure is the ground state of $\text{Li}_2\text{SrNb}_2\text{O}_7$, from a computational perspective as well.

The displacement pattern of the Y_2^- soft mode consists of the relative displacement of Nb and the rotation of oxygen atoms, as shown in Figure 21(d). The origin of the Nb displacement can be understood within the framework of the SOJT effect, as visualized

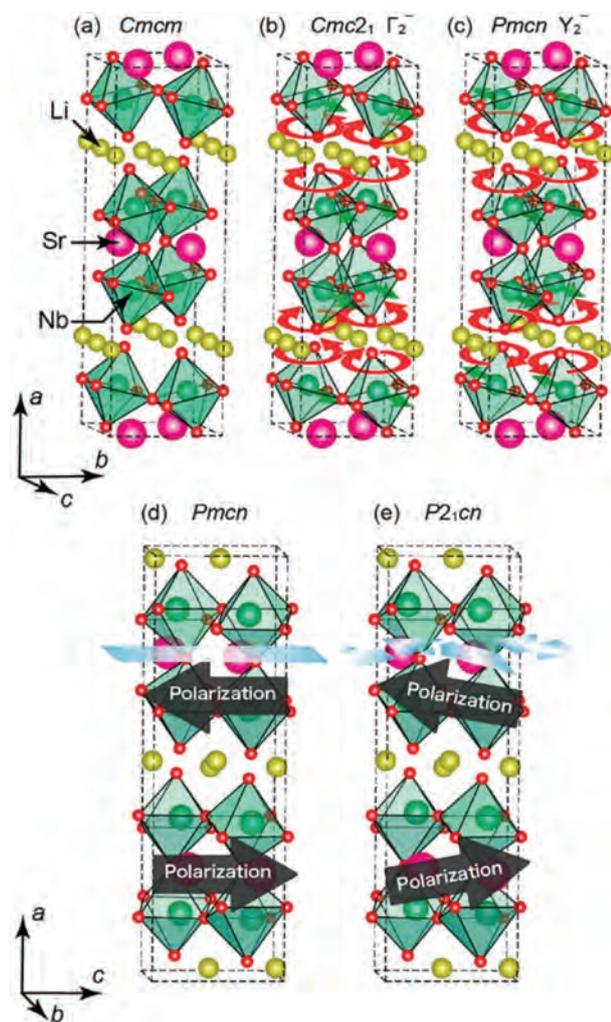


Figure 21. Crystal structures of $\text{Li}_2\text{SrNb}_2\text{O}_7$: (a) in the paraelectric $Cmcm$ phase, and (b, c) the phonon modes transforming as irreducible representations (b) Γ_2^- and (c) Y_2^- , leading to the $Cmc2_1$ and $pmcn$ phases, respectively. Schematics of (d) the antiferroelectric $pmcn$ and (e) weak ferroelectric $P2_1cn$ phases. The mirror symmetry perpendicular to the a axis is preserved in the $pmcn$ phase, whereas it is broken in the $P2_1cn$ phase. Adapted with permission from ref. [187]. Copyright 2021 American chemical society.

in Figure 22. Figure 22 shows the crystal orbital Hamilton population (COHP) [189–193] and schematics of energy-level diagrams calculated for $\text{Li}_2\text{SrNb}_2\text{O}_7$ in $Cmcm$ and $Pmcn$ symmetries [186]. The horizontal axis in panels (a) and (b) represents negative COHP ($-\text{COHP}$), with positive and negative values indicating bonding and antibonding states, respectively. As indicated by the arrow in panel (b), the bonding interaction between Nb 4d and O 2p orbitals is enhanced during the transformation from the $Cmcm$ structure to the $Pmcn$ structure due to the freezing of the Y_2^- soft mode, demonstrating that the Nb atom is attracted toward the ligand oxygen due to the SOJT effect. Panels (e–h) in Figure 22, on the other hand, present COHP and energy-level diagrams for $\text{Li}_2\text{SrTa}_2\text{O}_7$. In marked contrast to $\text{Li}_2\text{SrNb}_2\text{O}_7$, the COHPs for the $Cmcm$ and $Pmcn$ structures overlap

each other, indicating that the enhancement of the bonding interaction due to the transformation to the $Pmcn$ structure is absent. This result shows that the SOJT effect is suppressed by substituting Nb with Ta. Figure 23 shows the temperature dependence of the dielectric permittivity in $\text{Li}_2\text{Sr}(\text{Nb}_{1-x}\text{Ta}_x)_2\text{O}_7$ with varying Ta content x [186]. As shown in the figure, the cusp anomaly due to the phase transition shifts to lower temperatures as x increases, demonstrating that Ta substitution suppresses the emergence of the low-temperature structure. The variation of the phase transition temperature (T_c), plotted as a function of x in panel (b), shows that the phase transition disappears for x greater than 0.4. It would thus be reasonable to conclude that the phase transition of $\text{Li}_2\text{SrNb}_2\text{O}_7$ is driven by the SOJT effect between the Nb and ligand O atoms.

As shown in Figure 21(c), the Y_2^- soft mode includes not only the Nb displacements but also the rotational distortions of oxygen octahedra around the principal axis of the crystal. Similar to the case of $\text{Ca}_3\text{Mn}_2\text{O}_7$, the $Cmcm$ phase of $\text{Li}_2\text{SrNb}_2\text{O}_7$ originally contains the tilting distortions around the axis perpendicular to the $[100]$ axis (see Figure 21(c)). Therefore, the rotational distortions of octahedra would activate HIF through its combination with the tilting rotations, in analogy with ‘ $n = 2$ ’ RP-type oxides such as $\text{Ca}_3\text{Mn}_2\text{O}_7$. To clarify this hypothesis, phase instability was examined in $\text{Li}_2(\text{Sr}_{1-x}\text{Ca}_x)\text{Nb}_2\text{O}_7$. According to the concept of the tolerance factor for the perovskite-type structural unit, a reduction in the size of the A-site cation enhances the magnitude of the octahedral rotation. Therefore, substituting Ca for Sr provides a good test for the role of HIF in the ferroelectricity of $\text{Li}_2(\text{Sr}_{1-x}\text{Ca}_x)\text{Nb}_2\text{O}_7$, since Ca has a Shannon ionic radius of 1.00 Å in six coordination, which is smaller than 1.18 Å for Sr. Figure 24 presents the electric-field-induced polarization measured in $\text{Li}_2(\text{Sr}_{1-x}\text{Ca}_x)\text{Nb}_2\text{O}_7$ with compositions of $x = 0$ and 0.1 [187]. As shown in the figure, the remanent polarization increases with Ca substitution. This result strongly supports the hypothesis that HIF also contributes to the ferroelectricity in $\text{Li}_2\text{SrNb}_2\text{O}_7$.

The HIF also plays a role in the onset of the phase transition, in addition to the SOJT effect, by stabilizing the $P2_1cn$ structure. Figure 25 presents the effect of Ca substitution on the phase transition temperature of $\text{Li}_2(\text{Sr}_{1-x}\text{Ca}_x)\text{Nb}_2\text{O}_7$, which was examined through the temperature dependence of dielectric permittivity [187]. As shown in the figure, the cusp-like anomaly around 220 K in the composition of $x = 0$ gradually shifts to higher temperatures as the Ca content x increases. The variation of T_c is plotted in the inset as a function of x , indicating that T_c increases by more than 100 K with the substitution of 20% of Sr with Ca.

First-principles calculations provide further insight into the mechanism behind the elevation of T_c due to

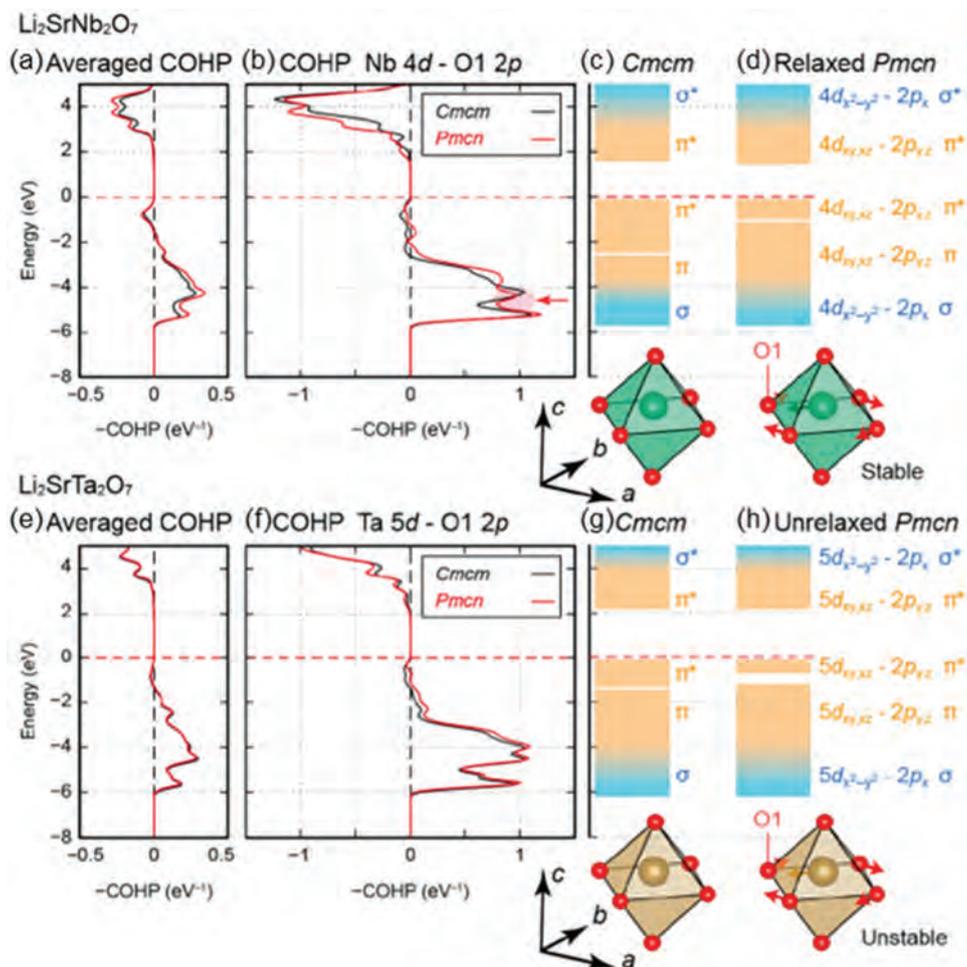


Figure 22. Averaged negative-signed COHPs in (a) $\text{Li}_2\text{SrNb}_2\text{O}_7$ and (e) $\text{Li}_2\text{SrTa}_2\text{O}_7$ for all atom pairs with bond lengths less than 3 Å. (b) Negative-signed COHPs between Nb 4d and O 2p states for $\text{Li}_2\text{SrNb}_2\text{O}_7$ and (f) those between Ta 5d and O 2p states for $\text{Li}_2\text{SrTa}_2\text{O}_7$ in the *cmm* and *pmcn* phases. Schematic energy level diagrams for the (c) *cmm* and (d) *pmcn* phases of $\text{Li}_2\text{SrNb}_2\text{O}_7$ as well as the (g) *cmm* and (h) *pmcn* phases of $\text{Li}_2\text{SrTa}_2\text{O}_7$ are described. Adapted with permission from ref. [186]. Copyright 2020 American chemical society.

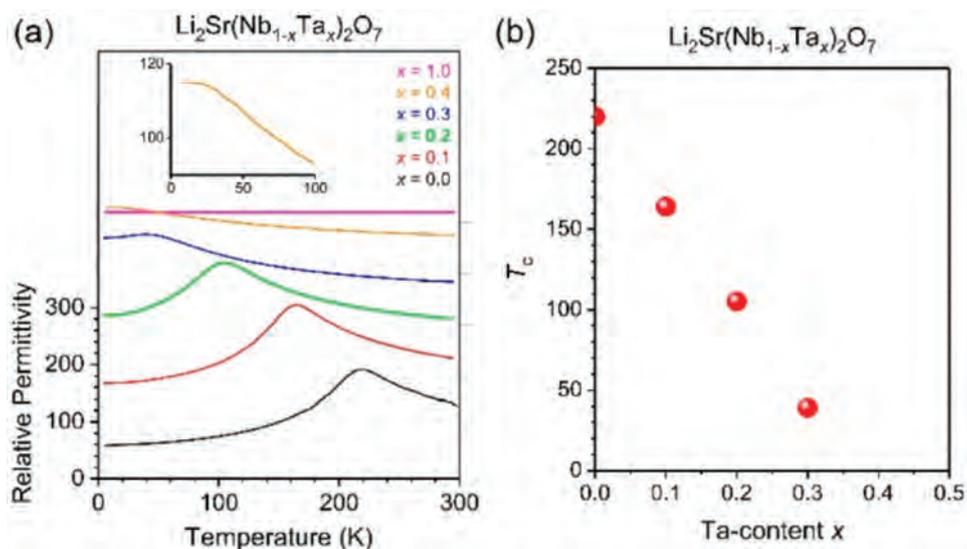


Figure 23. (a) Temperature dependences of relative dielectric permittivity in $\text{Li}_2\text{Sr}(\text{Nb}_{1-x}\text{Ta}_x)_2\text{O}_7$ with various x , which were measured at 100 kHz. The inset indicates a magnified view focused on the low-temperature range in the case of $x = 0.4$. (b) Ta-content dependence of ferroelectric phase transition temperatures in $\text{Li}_2\text{Sr}(\text{Nb}_{1-x}\text{Ta}_x)_2\text{O}_7$. Adapted with permission from ref. [186]. Copyright 2020 American chemical society.

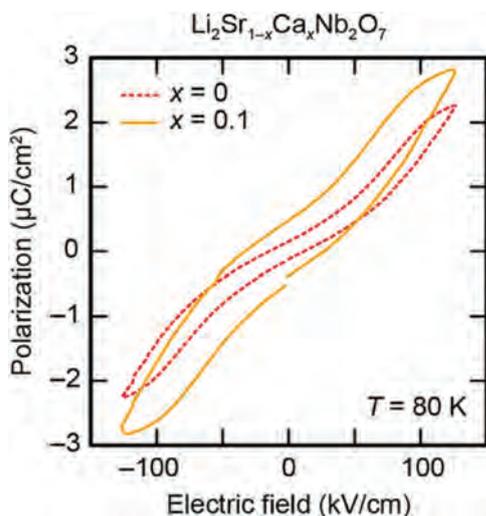


Figure 24. p - E hysteresis loops of $\text{Li}_2\text{SrNb}_2\text{O}_7$ (red dashed line) and $\text{Li}_2\text{Sr}_{0.9}\text{Ca}_{0.1}\text{Nb}_2\text{O}_7$ (orange solid line), measured at a test frequency of 10 Hz and a temperature of 80 K under an applied electric field of 130 kV/cm. Adapted with permission from ref. [187]. Copyright 2021 American chemical society.

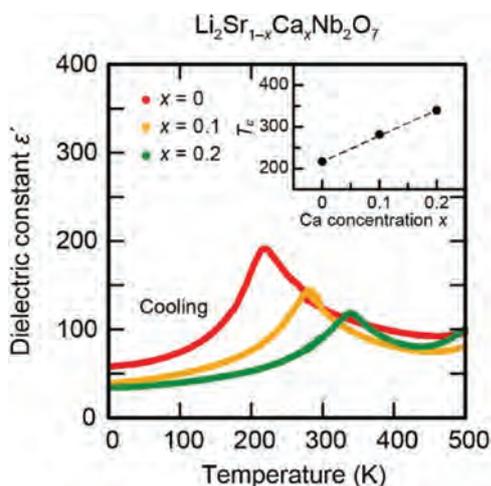


Figure 25. Temperature dependence of the relative permittivity for $\text{Li}_2\text{Sr}_{1-x}\text{Ca}_x\text{Nb}_2\text{O}_7$ measured at a frequency of 100 kHz. The inset shows the Ca-content dependence of the ferroelectric phase transition temperature T_c . Adapted with permission from ref. [187]. Copyright 2021 American chemical society.

Ca substitution. Figure 26 presents the COHP and the schematic energy level diagram of $\text{Li}_2\text{CaNb}_2\text{O}_7$, with the results for $\text{Li}_2\text{SrNb}_2\text{O}_7$ shown for comparison [187]. The calculations were performed with the $Cmcm$ and $Pm\bar{c}n$ symmetries, similar to the case of the Ta-substitution effect discussed earlier. Note that the $Pm\bar{c}n$ structure of $\text{Li}_2\text{CaNb}_2\text{O}_7$ is obtained by the freezing of the Y_2^- soft mode (Figure 21(c)) in the $Cmcm$ structure (Figure 21(a)), which was also found to be unstable in the $Cmcm$ structure of $\text{Li}_2\text{CaTa}_2\text{O}_7$ [188,194]. As shown in Figure 26, the positive - COHP is enhanced by the transformation to the $Pm\bar{c}n$ structure in $\text{Li}_2\text{CaNb}_2\text{O}_7$, demonstrating that Ca substitution encourages the instability of the Y_2^- soft mode. It should be noted that, in contrast to $\text{Li}_2\text{SrNb}_2\text{O}_7$, a σ -

bonding state is markedly enhanced in the $Pm\bar{c}n$ structure of $\text{Li}_2\text{CaNb}_2\text{O}_7$. This suggests that the octahedral rotation enhances the overlap between the Nb 4d states and the O 2p states, further motivating the displacement of Nb and leading to the increase of T_c .

The preceding discussion has clarified that both the SOJT effect and the HIF coexist in the system of $\text{Li}_2\text{SrNb}_2\text{O}_7$, where Ta-substitution for Nb and Ca-substitution for Sr are capable of tuning the SOJT effect and the HIF, respectively. This paragraph explores the co-substitution effect of Nb and Ca in $\text{Li}_2(\text{Sr}_{1-x}\text{Ca}_x)(\text{Nb}_{1-x}\text{Ta}_x)_2\text{O}_7$. Figure 27 presents a phase diagram of $\text{Li}_2(\text{Sr}_{1-x}\text{Ca}_x)(\text{Nb}_{1-x}\text{Ta}_x)_2\text{O}_7$, determined through a combination of dielectric measurements, x-ray diffraction, and second-harmonic generation measurements [188]. While the end-member composition $\text{Li}_2\text{SrNb}_2\text{O}_7$ undergoes successive phase transitions from $I4/m\bar{3}m$ to $Cm\bar{c}m$ and finally to $P2_1cn$, the counterpart $\text{Li}_2\text{CaTa}_2\text{O}_7$ exhibits an antiferroelectric $Pm\bar{c}n$ structure between the $Cm\bar{c}m$ and $P2_1cn$ phases. The ground state throughout the entire composition range is characterized by the $P2_1cn$ structure, which features in-plane antipolar and out-of-plane polar configurations of the off-center displacement of Nb within the NbO_6 octahedron.

One intriguing trend in the phase diagram is that the increase in T_c is significantly weaker compared to the case of single Ca substitution for Sr. This suggests a close competition between the suppression of the SOJT effect due to Ta substitution and the enhancement of the HIF due to Ca substitution. To examine this competition in detail, the magnitude of the SOJT effect (δl) and HIF ($||c||/|b| - 1$) has been quantified, where δl represents the displacement of the B-site cation along the c -axis, and $||c||/|b| - 1$ denotes the scale of anisotropy induced by octahedral rotation as shown in Figure 28(a,b) [188]. Figure 28(c) shows the relationship between δl and $||c||/|b| - 1$ across different compositions. It is evident from the figure that the structural instability in $\text{Li}_2\text{SrNb}_2\text{O}_7$ is primarily governed by the SOJT effect. However, the co-substitution of Ca and Nb enhances the effect of HIF, eventually surpassing the SOJT effect around the composition of $x = 0.7$. In the end-member composition of $\text{Li}_2\text{CaTa}_2\text{O}_7$, the HIF becomes dominant in the structural phase transition. The phase diagram's correspondence with the structural analysis in panel (c) strongly supports the conclusion that the close competition between the SOJT effect and the HIF provides a rich phase instability in the pseudo-RP-type structure.

Combining all the discussions, we can conclude that $\text{Li}_2\text{SrNb}_2\text{O}_7$ has two coexisting mechanisms: the SOJT effect and the HIF. The SOJT effect is the origin of the Ti displacement for the proper ferroelectrics, such as in BaTiO_3 and PbTiO_3 [196], whereas the HIF mechanism is triggered by the two octahedral rotational distortions like in $\text{Ca}_3\text{Ti}_2\text{O}_7$ and $\text{Sr}_3\text{Sn}_2\text{O}_7$

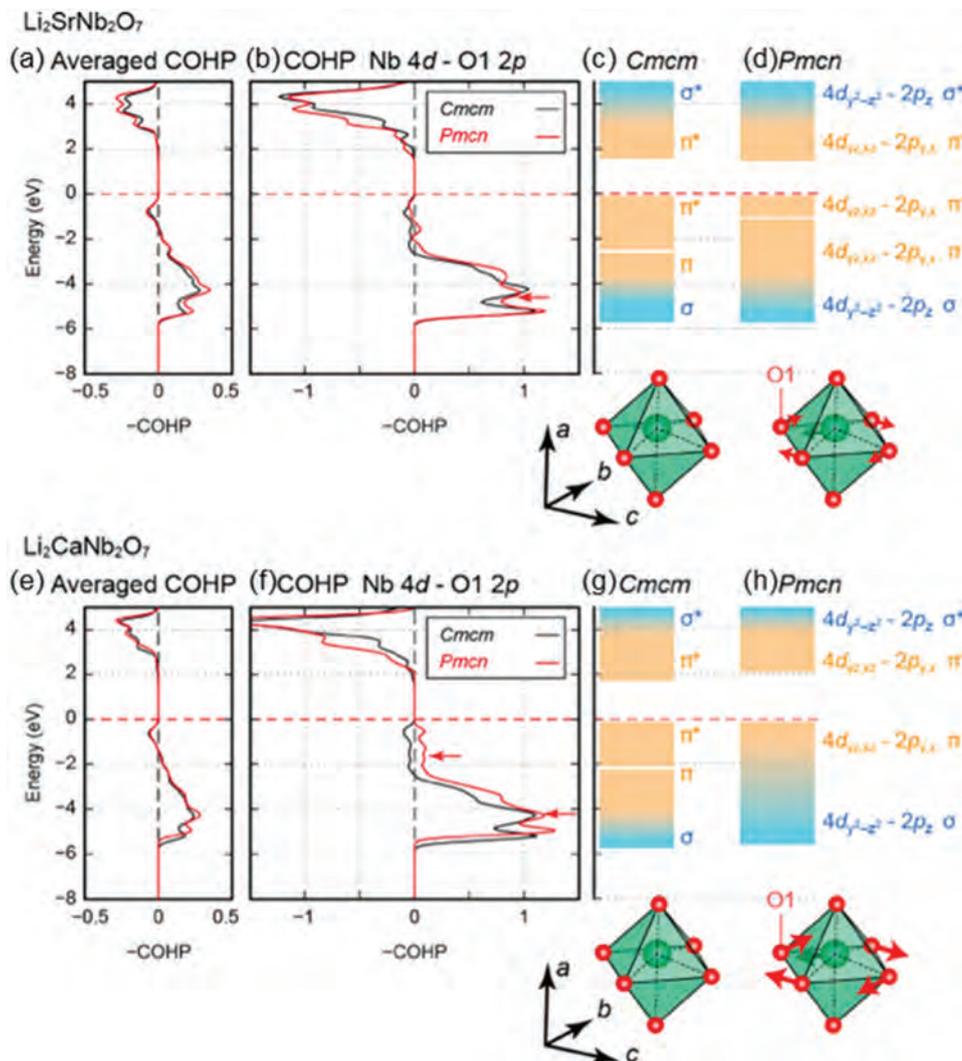


Figure 26. Averaged negative-signed COHPs in (a) $\text{Li}_2\text{SrNb}_2\text{O}_7$ and (e) $\text{Li}_2\text{CaNb}_2\text{O}_7$ for all atom pairs with bond lengths less than 3 Å. Negative-signed COHPs between Nb 4d and O 2p states for (b) $\text{Li}_2\text{SrNb}_2\text{O}_7$ and (f) $\text{Li}_2\text{CaNb}_2\text{O}_7$ in the *cmcm* and *pmcn* phases. Schematic energy level diagrams for the *cmcm* and *pmcn* phases of (c, d) $\text{Li}_2\text{SrNb}_2\text{O}_7$ and (g, h) $\text{Li}_2\text{CaNb}_2\text{O}_7$ are described. Adapted with permission from ref. [187]. Copyright 2021 American chemical society.

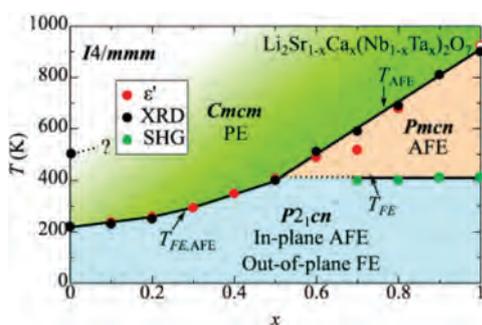


Figure 27. Phase diagram of $\text{Li}_2\text{Sr}_{1-x}\text{Ca}_x(\text{Nb}_{1-x}\text{Ta}_x)_2\text{O}_7$. The abbreviations PE, AFE, and FE denote paraelectricity, antiferroelectricity, and ferroelectricity, respectively. The open circle data for $x = 1$ is referred to in ref. [194]. Adapted with permission from ref. [188]. Copyright 2022 American physical society.

[179,183]. $\text{Li}_2\text{SrNb}_2\text{O}_7$ is a unique ferroelectric material that has two ‘switches’ to trigger the phase transition, which is confirmed by partially substituting Ta and/or Ca in $\text{Li}_2\text{SrNb}_2\text{O}_7$ from the experiments and

the calculations. It is noteworthy that these results were reproduced from the synthesis of a single crystal performed by Cheong’s group [197], although our experimental and computational results were partly in contrast to the previous reports [198].

Our study exemplifies how effective collaboration between computational modeling and experimental techniques helps the investigation of ferroelectric materials. Particularly for displacive-type ferroelectrics driven by lattice instability, we have demonstrated that first-principles calculations provide valuable guidelines for the rational design of compositions. These calculations visualize the displacement pattern of the soft mode and reveal corresponding variations in chemical bonding states, offering insights crucial for the systematic development of new materials. This integrated approach not only enhances our understanding of ferroelectric mechanisms but also facilitates targeted material design with enhanced functional properties.

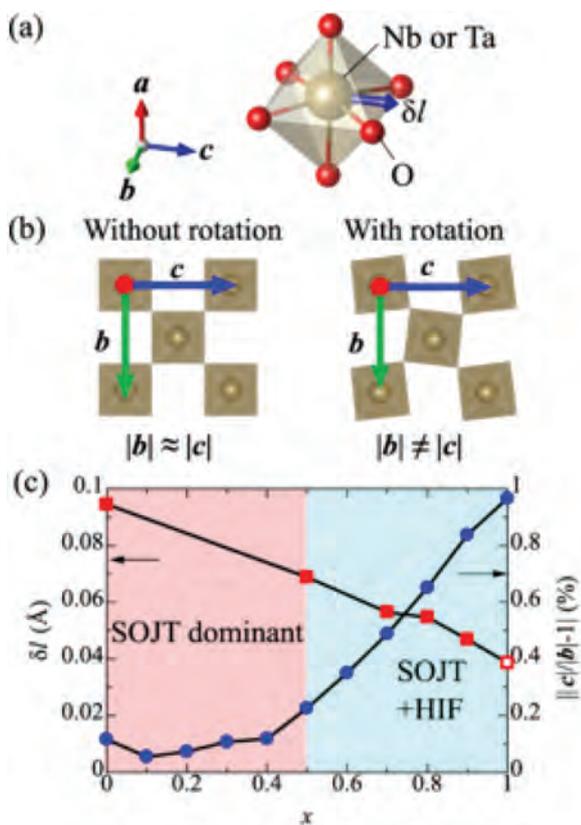


Figure 28. Schematics of (a) the displacement of B-cation in an octahedron, (b) the in-plane lattice anisotropy induced by octahedral rotation. The blue vector in (a) shows the B-cation displacement along the c axis (δl). (c) The x-dependence of δl and lattice anisotropy at 100 K. The open square data for $x=1$ is evaluated from room temperature data in tables 1 and 2 in ref. [195] (the two Ta-sites are averaged). Adapted with permission from ref. [188]. Copyright 2022 American physical society.

4. Summary and outlook

We have reviewed selected examples of our combined computational and experimental approaches to inorganic semiconductors and dielectrics, covering topics from methodology development to materials design, prediction, and exploration. High-throughput first-principles calculations of polarons, point defects, surfaces, and interfaces enable systematic and accurate predictions of their structures and properties, and the generated data are effectively used for machine learning, as well as the cases of bulk fundamental properties. The constructed machine learning models allow us to predict various material properties and search for promising materials much more quickly than the use of direct theoretical calculations. A great advantage is that a tremendous number of candidate materials can, therefore, be considered, substantially widening the materials search space.

However, we need to care about prediction accuracy associated with regression errors, as well as errors inheriting from first-principles calculations. The validation of the machine learning prediction by theoretical calculations and experiments should be important.

To show the effectiveness of the combined computational and experimental approaches in materials research, we have discussed the cases of ternary nitride semiconductors for potential optoelectronic and photovoltaic applications, phosphide-based photovoltaic cells and their constituent phosphide semiconductors, and layered perovskite oxide ferroelectrics. In addition, we have demonstrated the efficient construction of phase diagrams using active learning in conjunction with experiments. This scheme has been applied to the determination of experimental conditions for phosphide film growth.

The combination of theory, computation, and experiment is also useful for unveiling the mechanisms behind material functionalities, which is an important step toward the construction of materials design principles. This is exemplified by the case of the layered perovskite oxide ferroelectrics, where theoretical phonon and chemical bonding analyses help us elucidate the unique origin of their ferroelectricity.

Given the recent worldwide interest in materials informatics, we surely expect that the methodologies and algorithms for theoretical calculations and data scientific methods will be continuously improved and more powerful in materials research. Further development of supercomputers will allow us to generate more accurate and abundant computational materials data, with which machine learning becomes more effective and useable in a wider variety of materials studies. In addition, current trends toward open science will make more computational and experimental materials databases freely available. Effective use of such publicly open data, as well as in-house closed ones, will be a key to conducting cutting-edge data-centric research in the rapidly growing materials informatics field.

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