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We demonstrate that ultraviolet Raman spectroscopy is an effective technique to measure the transition temperature, T_c , accurately in ferroelectric ultrathin films and superlattices. We show that one-unit-cell thick BaTiO_3 layers in $\text{BaTiO}_3/\text{SrTiO}_3$ superlattices are not only ferroelectric (T_c as high as 250 K), but also polarize the quantum paraelectric SrTiO_3 layers adjacent to them. T_c was tuned by ~ 500 K by varying the thicknesses of BaTiO_3 and SrTiO_3 layers, revealing the essential roles of electrical and mechanical boundary conditions for nanoscale ferroelectricity.

Ferroelectricity at the nanoscale has emerged as fertile ground for new physical phenomena and devices (1–3). Shrinking dimensions demand characterization techniques capable of probing the properties of ferroelectrics in, for example, ultrathin films and superlattices. In particular, it is difficult to measure the ferroelectric phase transition temperature T_c in such systems, and the T_c information is largely missing or inaccurate in reports of ferroelectricity in nanoscale ultrathin films and superlattices (3–5). One fundamental property of ferroelectrics that changes qualitatively during the phase transition is the dynamics of lattice vibrations (6). Thus, its temperature dependence allows determination of T_c . While lattice dynamics in ferroelectric films (7,8) and superlattices (9) from 150 nm to 2 μm in thickness have been investigated previously, such studies are very difficult for films thinner than ~ 100 nm. We report the use of ultraviolet (UV) Raman spectroscopy on $\text{BaTiO}_3/\text{SrTiO}_3$ superlattices with total thicknesses down to 24 nm, which enabled us to measure T_c of the BaTiO_3 layers in the superlattices. We found that the BaTiO_3 layers are ferroelectric even when their thickness is only one unit cell (0.4 nm), and that they can induce polarization in the adjacent paraelectric SrTiO_3 layers that are much thicker. By varying the thickness of both the BaTiO_3 and SrTiO_3 layers, T_c was tuned from 250 K below to 235 K above the bulk value of BaTiO_3 (403 K). The result shows that under favorable electrical and mechanical boundary conditions ferroelectricity is robust in nanoscale

systems.

Conventional visible Raman spectroscopy works poorly for thin films of ferroelectrics and other wide-bandgap materials because the visible photon energy is much smaller than the bandgap (10). Consequently, the absorption is extremely weak and the penetration depth is large, allowing light to travel through the film into the substrate, which generates overwhelming signals in the Raman spectra. For UV excitation, the photon energy is above the bandgaps of ferroelectrics, leading to a much stronger absorption and a shorter penetration depth, preventing light from entering the substrate. UV excitation near the bandgap also leads to strong resonance enhancement of Raman signals. This is demonstrated by Fig. 1 where Raman spectra of a BaTiO₃/SrTiO₃ superlattice measured with visible (514.5 nm) and UV (351.1 nm) excitations are shown. The substrate features dominate the 514.5 nm spectrum while they are greatly reduced in the UV spectrum, in which peaks of superlattice phonons are clearly observed.

UV Raman spectroscopy has not been widely used for ferroelectric films because of technical difficulties such as lower throughput efficiency, insufficient dispersion, and higher stray light level of UV Raman spectrometers compared to those operating in the visible range. Recently, room temperature measurement of SrTiO₃ films using 325 nm excitation has been reported (11). The recent progress in UV Raman instrumentation has made measurement of ferroelectric films possible. In our experiment, a triple monochromator was used to provide high resolution and effective stray light reduction (12). Powerful laser sources and optimized optical paths were used to improve the throughput. With these setups, we have measured Raman scattering in BaTiO₃/SrTiO₃ superlattices as thin as 24 nm, and in (Ba_{0.5}Sr_{0.5})TiO₃ films 10 nm in thickness.

The BaTiO₃/SrTiO₃ superlattices are denoted by (BTO_{*n*}/STO_{*m*}) × number of periods, where *n* and *m* refer to the thickness, in unit cells, of the BaTiO₃ and SrTiO₃ layers, respectively. They were all grown on (001) SrTiO₃ substrates. Details of the sample preparation by reactive molecular-beam epitaxy (13) and structural characterization are presented in Supporting Online

Material (12).

Curve 3 in Fig. 1 is typical of the UV Raman spectra of BaTiO₃/SrTiO₃ superlattices below T_c , exhibiting strong first-order (single-phonon) peaks as labeled in the figure. Weak second-order (two-phonon) features from the SrTiO₃ substrate can be seen between 600 and 700 cm⁻¹ and as a background in the range 200-500 cm⁻¹. The phonon mode assignment is made by comparison with the spectra of SrTiO₃ and BaTiO₃ single crystals (12) and with the help of first-principles calculations. The lines at about 290 cm⁻¹ have similar positions and shapes to the TO₂ modes of A_1 symmetry of the tetragonal-phase BaTiO₃ (14, 15), thus they are assigned to the BaTiO₃ layers. The line at about 180 cm⁻¹ corresponds closely to the TO₂ line in the electric-field-induced Raman spectrum of SrTiO₃ crystals (16). It is not from the SrTiO₃ substrate because the first-order Raman lines are symmetry-forbidden in bulk SrTiO₃ (17). Although the TO₁ mode of A_1 symmetry of BaTiO₃ is at about the same position (177 cm⁻¹), it has markedly different relative intensity and shape (14) from the 180-cm⁻¹ line. Therefore, we attribute this line to the TO₂ phonon in the SrTiO₃ layers. The LO₃ and TO₄ modes involve both SrTiO₃ and BaTiO₃ layers and extend through the superlattice. A doublet of folded longitudinal acoustic (LA) phonon due to the superlattice periodicity (18) is also observed. The two triangles indicate the predicted first doublet frequencies by an elastic continuum model (19). The observation of the LA phonon folding suggests that these superlattices possess the requisite structural quality for acoustic Bragg mirrors and cavities used for coherent phonon generation (20, 21).

Bulk crystalline BaTiO₃ is cubic and paraelectric above $T_c = 403$ K, becomes tetragonal and ferroelectric below T_c , and goes through additional transitions to orthorhombic at 278 K and rhombohedral at 183 K (22). Bulk crystalline SrTiO₃ is paraelectric at all temperatures due to quantum fluctuations (23). The temperature evolution of Raman spectra for two superlattice samples is shown in Fig. 2: **A** (BTO₂/STO₁₃) \times 20 and **B** (BTO₈/STO₄) \times 10. The shapes and positions of the BaTiO₃ lines at low temperatures are characteristic of BaTiO₃ in the tetragonal

phase (12, 14, 15), indicating that the BaTiO₃ layers are tetragonal and ferroelectric below T_c . The presence of the first-order Raman lines of SrTiO₃ shows that the SrTiO₃ layers are polar because the first-order lines are symmetry-forbidden in non-polar SrTiO₃ (17). The intensities of the first-order superlattice phonons decrease as the temperature increases and disappear at T_c . Above T_c , the spectra contain only the second-order features as expected from the symmetry selection rules. When the BaTiO₃ layers are paraelectric, the induced polarization in the SrTiO₃ layers also disappears.

By plotting the first-order Raman intensity as a function of temperature, we can determine T_c accurately as the temperature where the intensity becomes zero. For this purpose, the TO₂ and TO₄ phonon lines are the most suitable because they do not overlap with the second-order features. The results, with the phonon intensities normalized by the Bose factor $n + 1 = (1 - \exp(-\hbar\omega/kT))^{-1}$ and by the intensities at 7 K, are presented for four superlattices: (BTO₂/STO₄) \times 40 and (BTO₅/STO₄) \times 25 in Fig. 3A, and (BTO₈/STO₄) \times 10 and (BTO₈/STO₄) \times 40 (strain partially relaxed) in Fig. 3B. Both TO₂ and TO₄ phonons show similar behaviors and the dashed-dotted lines are linear fits to the average of the two modes. The linear fit corresponds to a parabolic decrease of polarization with temperature as Raman intensity is proportional to the square of atomic displacement. The intersection of a dash-dotted line with the horizontal axis is taken as the T_c of the sample.

The temperature dependence of polarization from a phase-field model calculation (24) is plotted in Figs. 3C and 3D for the same samples as in Figs. 3A and 3B. The model assumes that the BaTiO₃ and SrTiO₃ layers in the superlattices have their respective bulk elastic and thermodynamic properties. The in-plane lattice constant is commensurately constrained to the SrTiO₃ substrate except for the partially relaxed case, and the top surface is stress-free. The surface depolarization field is ignored and a short-circuit electrostatic boundary condition is employed. A computational cell of 64 nm along the two in-plane directions and one unit cell along the

growth direction was employed. The corresponding 3-dimensional time-dependent Ginzburg-Landau equations are then numerically solved using the perturbation method with semi-implicit Fourier-spectral algorithms (25). The result reveals a spontaneous polarization along the growth direction with multiple 180° domains in the BaTiO_3 layers, which induces polarization in the adjacent SrTiO_3 layers whose magnitude and distribution vary with the thickness and domain size of the BaTiO_3 layers. The spontaneous polarization in the BaTiO_3 layers becomes zero at T_c , and the predicted T_c values agree with those from the Raman data. This is remarkable considering that no fitting parameters from the Raman experiments are used in the calculations.

In Fig. 4, T_c determined by the Raman data, XRD, and the phase-field model are shown as a function of the BaTiO_3 and SrTiO_3 layer thicknesses. The XRD measurement provides an additional confirmation of the Raman results, where a change in the temperature dependence of the out-of-plane lattice constant can be taken as an indication of T_c (12). The figure shows that the BaTiO_3 layers in the superlattices are ferroelectric even when their thickness is only one unit cell, with a T_c as high as 250 K. T_c increases with increasing n as the dipole-dipole interaction in BaTiO_3 layers becomes stronger, while large m suppresses T_c by reducing the coupling between the BaTiO_3 layers. By changing the values of n and m , we were able to tune T_c from 151 K to 638 K, i.e. from 250 K below to 235 K above the bulk value of BaTiO_3 . The higher-than-bulk T_c is due to the strain in the BaTiO_3 layers, just as strain enhances T_c in single-layer ferroelectric films (26, 27). When the strain is partially relaxed in sample $(\text{BTO}_8/\text{STO}_4)\times 40$, T_c drops almost to the bulk BaTiO_3 value. While the 3-dimensional phase-field model allowing domain formation provides a good description of the Raman data, simulations assuming a single domain in the BaTiO_3 layers yield significantly lower T_c for $m = 13$, demonstrating the importance of domain formation in theoretical calculations (28).

We now can conclude that ferroelectricity can be very strong in one-unit-cell thick BaTiO_3 layers ($T_c \sim 250$ K for $n/m = 1/4$). The electrical boundary condition plays a critical role.

With the highly-polarizable SrTiO₃ in contact with the BaTiO₃ layers, the critical thickness is reduced to a single unit cell. Meanwhile, the mechanical boundary condition imposed by the SrTiO₃ substrate leads to strain in the BaTiO₃ layers and thus enhanced ferroelectricity. The interplay between the electrical and mechanical boundary conditions enables the tuning of T_c by nearly 500 K.

References and Notes

1. C. H. Ahn, K. M. Rabe, J.-M. Triscone, *Science* **303**, 488 (2004).
2. J. Junquera, P. Ghosez, *Nature (London)* **422**, 506 (2003).
3. D. D. Fong, *et al.*, *Science* **304**, 1650 (2004).
4. T. Tybell, C. H. Ahn, J. -M. Triscone, *Appl. Phys. Lett.* **75**, 856 (1998).
5. H. N. Lee, H. M. Christen, M. F. Chisholm, C. M. Rouleau, D. H. Lowndes, *Nature* **433**, 395 (2005).
6. M. E. Lines, A. M. Glass, *Principles and applications of ferroelectrics and related materials* (Clarendon Press, Oxford, 1977).
7. A. A. Sirenko, *et al.*, *Nature (London)* **404**, 373 (2000).
8. T. Ostapchuk, *et al.*, *Phys. Rev. B* **66**, 235406 (2002).
9. R. S. Katiyar, Y. I. Yuzyuk, R. R. Das, P. Bhattacharya, V. Gupta, *Ferroelectrics* **329**, 907 (2005).
10. K. van Benthem, C. Elsässer, R. H. French, *J. Appl. Phys.* **90**, 6156 (2001).
11. L. H. Tisinger, *et al.*, *J. Vac. Sci. Technol. B* **21**, 53 (2003).
12. See Supporting Online Material for details .
13. D. G. Schlom, *et al.*, *Mater. Sci. Eng. B* **87**, 282 (2001).
14. A. Scalabrin, A. S. Chaves, D. S. Shim, S. P. S. Porto, *Phys. Status Solidi B* **79**, 731 (1977).
15. D. A. Tenne, *et al.*, *Phys. Rev. B* **69**, 174101 (2004).

16. P. A. Fleury and J. M. Worlock, *Phys. Rev.* **174**, 613 (1968).
17. W. G. Nilsen and J. G. Skinner, *J. Chem. Phys.* **48**, 2240 (1968).
18. B. Jusserand, M. Cardona, *Light Scattering in Solids V* (Springer, Heidelberg, 1989), pp. 49-152.
19. C. Colvard, *et al.*, *Phys. Rev. B* **31**, 2080 (1985).
20. M. Trigo, A. Bruchhausen, A. Fainstein, B. Jusserand, V. Thierry-Mieg, *Phys. Rev. Lett.* **89**, 227402 (2002).
21. A. Bartels, T. Dekorsy, H. Kurz, K. Köhler, *Phys. Rev. Lett.* **82**, 1044 (1999).
22. L. E. Cross, *Ferroelectric ceramics*, N. Setter, E. L. Colla, eds. (Birkhäuser Verlag, Basel, 1993), p. 1.
23. K. A. Müller, H. Burkard, *Phys. Rev. B* **19**, 3593 (1979).
24. L.-Q. Chen, *Annu. Rev. Mater. Res.* **32**, 113 (2002).
25. L.-Q. Chen, J. Shen, *Comput. Phys. Commun.* **108**, 147 (1998).
26. K. J. Choi, *et al.*, *Science* **306**, 1005 (2004).
27. J. H. Haeni, *et al.*, *Nature (London)* **430**, 758 (2004).
28. V. A. Stephanovich, I. A. Luk'yanchuk, M. G. Karkut, *Phys. Rev. Lett.* **94**, 047601 (2005).
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P0005 (Fainstein), N00014-05-1-0559 (Eom), N00014-00-1-0261 (Rabe), and N00014-01-1-0365 (Rabe), by NASA under Grant No. NASA3-NCC1034 (Katiyar), by NSF under Grant Nos. DMR-0507146 (Schlom, Chen, Pan, Rabe, Xi), DMR-0122638 (Chen), DMR-0313764 (Eom), ECS-0210449 (Eom), and DMR-0315633 (Pan), and by a Guggenheim fellowship (Chen). The work at Los Alamos National Laboratory was supported by the Laboratory-Directed Research and Development Program under US DOE.

Fig. 1. Room temperature Raman spectra of (1) a bare SrTiO₃ substrate (black curve), (2) a (BTO₅/STO₄)×25 superlattice ($T_c = 530$ K, blue curve) measured with visible excitation (514.5 nm), and (3) the same superlattice measured with 351.1 nm UV excitation (red curve). The dashed black line shows the bare SrTiO₃ substrate spectrum measured with 351.1 nm UV excitation. Triangles show the calculated frequencies of the first folded LA doublet.

Fig. 2. Temperature evolution of UV Raman spectra of superlattices **A** (BTO₂/STO₁₃)×20 and **B** (BTO₈/STO₄)×10. The arrows mark the SrTiO₃-like TO₂ mode at 180 cm⁻¹ and the TO₄ mode at about 530 cm⁻¹, whose intensities decrease as the temperature increases and disappear at T_c .

Fig. 3. Temperature dependencies of normalized Raman intensities of TO₂ (solid triangles) and TO₄ (open triangles) phonons for **A** (BTO₂/STO₄)×40 and (BTO₅/STO₄)×25 and **B** (BTO₈/STO₄)×10 and (BTO₈/STO₄)×40. Sample (BTO₈/STO₄)×40 is partially relaxed while the other three samples are commensurate with the SrTiO₃ substrate. The dash-dotted lines are fits to a linear temperature dependence. **C** and **D**: The 3-dimensional phase-field model calculations of polarization as a function of temperature in the same superlattice samples. Polarization is given as a fraction of the polarization of bulk BaTiO₃, $P_0 = 0.26$ C/m².

Fig. 4. Dependence of T_c on n and m in superlattices BTO _{n} /STO _{m} . Blue symbols are for $m = 4$ and red symbols for $m = 13$. Open triangles are from temperature-dependent XRD measurement. Circles with lines are from the 3-dimensional phase-field model calculations. The black dash-dotted line shows the T_c in bulk BaTiO₃.

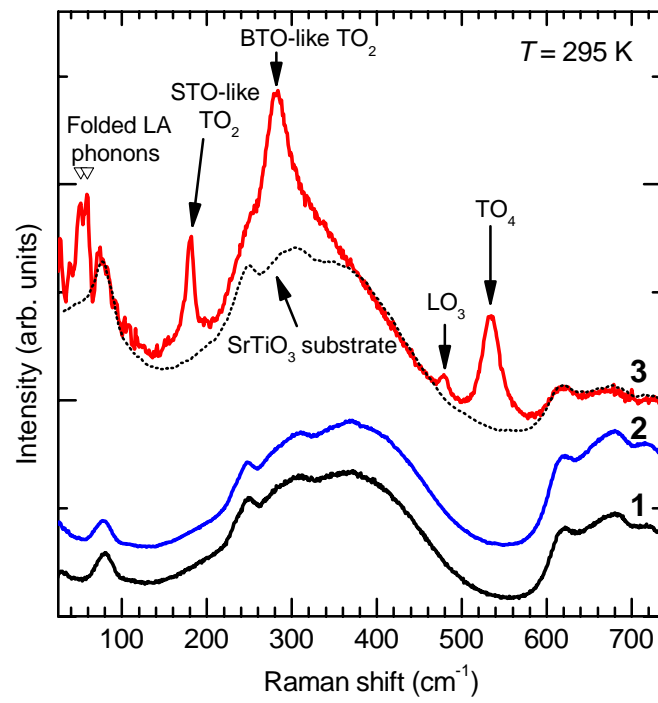


Figure 1

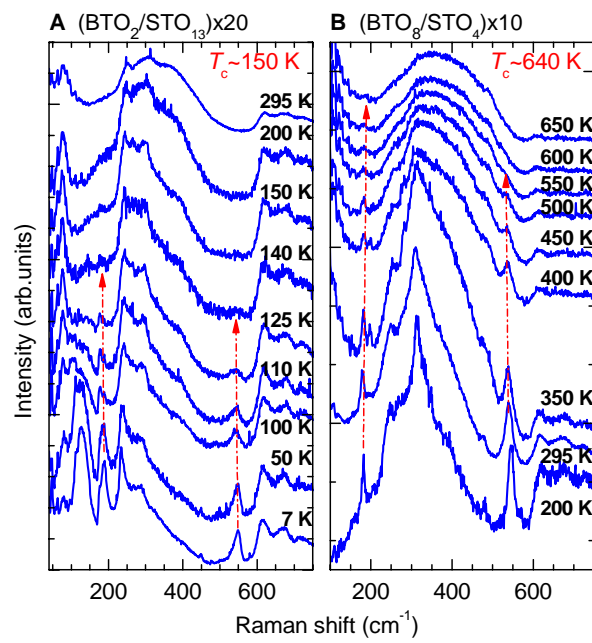


Figure 2

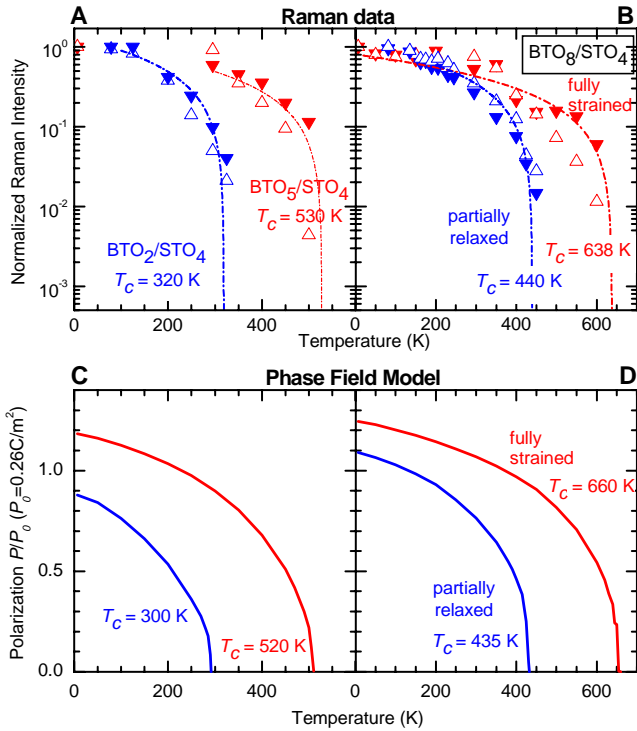


Figure 3

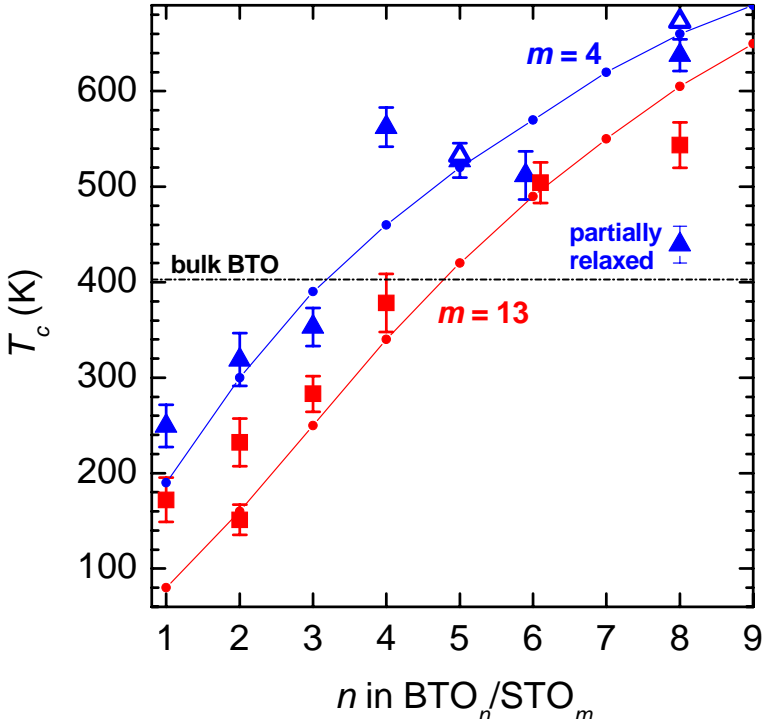


Figure 4

Supporting Online Material

Materials and Methods

1. Preparation and structural characterization of $\text{BaTiO}_3/\text{SrTiO}_3$ superlattices

The $\text{BaTiO}_3/\text{SrTiO}_3$ superlattice samples were grown by reactive molecular-beam epitaxy (1). The substrate is (001) SrTiO_3 terminated at the TiO_2 layer, with BaO being the first layer deposited. The n and m values are controlled using reflection high-energy electron diffraction (RHEED) oscillations during growth, and confirmed by x-ray diffraction (XRD) and in some samples by high resolution transmission electron microscopy (HRTEM) (2). XRD shows excellent epitaxy and crystallinity in the superlattice samples, with all superlattice peaks present in the $\theta - 2\theta$ scan. HRTEM images show atomically sharp $\text{BaTiO}_3/\text{SrTiO}_3$ interfaces and accurate periodicity (2). According to XRD, all samples are commensurate to the SrTiO_3 substrate ($a = 0.3905$ nm), meaning that the SrTiO_3 layers are strain free and the BaTiO_3 layers are under 2.2% compressive biaxial strain ($a = 0.3992$ nm in bulk BaTiO_3 at room temperature), except for a 200 nm thick superlattice $(\text{BTO}_8/\text{STO}_4) \times 40$ in which the strain is partially relaxed ($a = 0.3946$ nm in-plane). In this case, the SrTiO_3 layers are under biaxial tensile strain and the BaTiO_3 layers are under biaxial compressive strain. Some samples were annealed in oxygen (760 Torr, 550°C , 5 hours) to reduce the population of oxygen vacancies, and we found no observable effect of the annealing on the Raman spectra.

As an example of the structural characterization of the superlattice samples in this work, the four-circle x-ray diffraction (XRD) scans (Figs. S1–S3) and a high resolution transmission electron microscopy (HRTEM) image for the superlattice $(\text{BTO}_5/\text{STO}_4) \times 25$ (Fig. S4) are presented here.

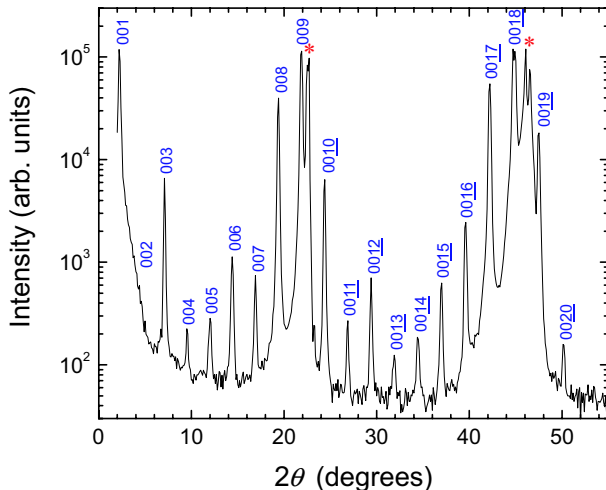


Figure S1. An x-ray diffraction $\theta - 2\theta$ scan of the $(\text{BTO}_5/\text{STO}_4) \times 25$ superlattice using $\text{Cu } K_\alpha$ radiation. Substrate peaks are marked by asterisks (*). All superlattice peaks are present for $2\theta < 55^\circ$, indicating atomically sharp interfaces between BaTiO_3 and SrTiO_3 layers and accurate periodicity. An out-of-plane lattice parameter, $c = 36.3 \pm 0.1$ Å, was obtained from a Nelson-Riley analysis (3), which is consistent with 5 unit cells of BaTiO_3 and 4 unit cells of SrTiO_3 in a period.

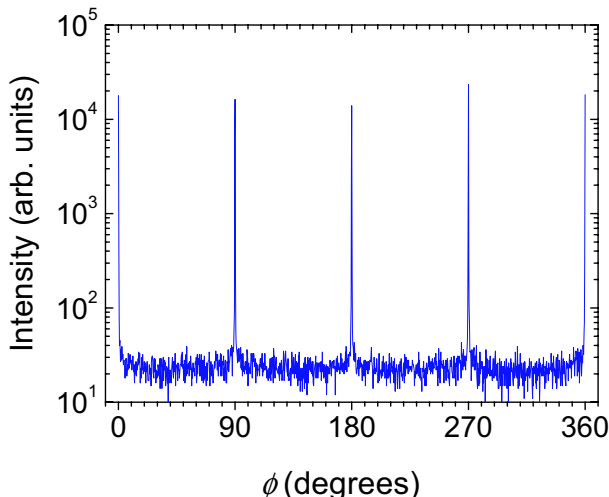


Figure S2. An x-ray diffraction ϕ scan of the 108 peak taken at $\chi = 40.7^\circ$ of the $(\text{BTO}_5/\text{STO}_4) \times 25$ superlattice. In this scan $\phi = 0^\circ$ is aligned parallel to the $[100]$ in-plane direction of the substrate and $\chi = 90^\circ$ aligns the diffraction vector to be perpendicular to the plane of the substrate. It shows that the superlattice is epitaxial with the expected ($[100]$ superlattice \parallel $[100]$ substrate) in-plane alignment with the substrate.

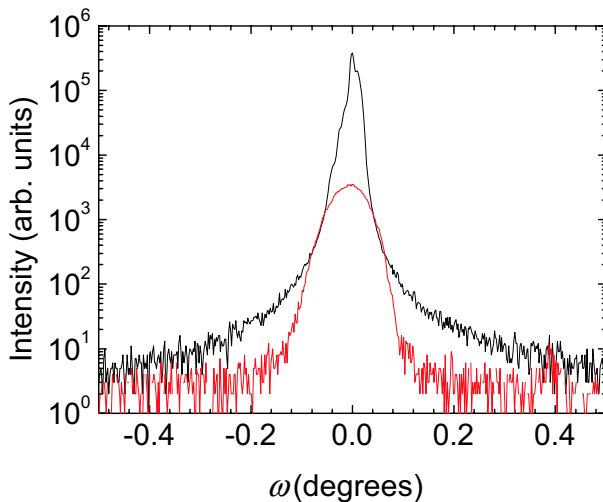


Figure S3. Rocking curves of the $(\text{BTO}_5/\text{STO}_4) \times 25$ superlattice 0018 peak (red) of the $(\text{BTO}_5/\text{STO}_4) \times 25$ superlattice and the underlying SrTiO_3 substrate 002 peak (black). The full width at half maximum (FWHM) is 0.034° for the superlattice peak as compared to 0.013° for the substrate peak. The sharp rocking curve indicates the high structural perfection of the superlattice.

2. UV Raman Measurement

In the present work, we used UV-optimized Jobin Yvon T64000 triple spectrometers with a multichannel coupled charge-device detector. The system employs a triple monochromator to provide high resolution and effective stray light reduction. Powerful laser sources and optimized optical paths were used to improve the throughput. Spectra were recorded in a backscattering geometry from 7 - 700 K using either a closed cycle cryostat (for low temperatures) or an evacuated heater

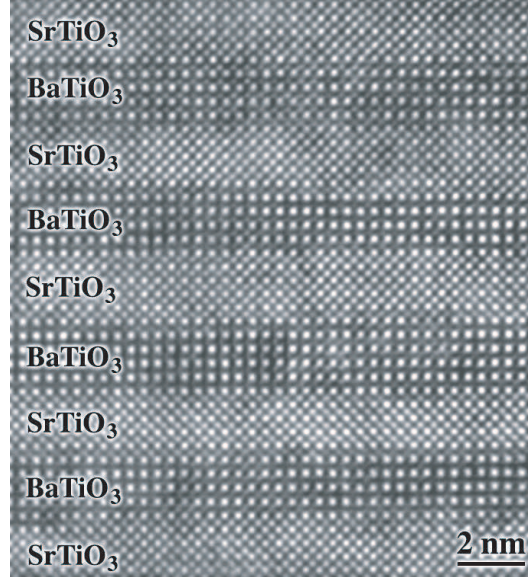


Figure S4. A cross-sectional HRTEM image of the superlattice. It shows alternating layers of 5 unit cells of BaTiO₃ and 4 unit cells of SrTiO₃, confirming the intended superlattice periodicity and the XRD result. The interfaces are abrupt and the roughness is within 1 unit cell.

stage (for above 300 K). The 351.1 nm line of an Ar⁺ laser and the 325 nm line of a He-Cd laser were used for excitation.

We used low laser power densities (about 3.5 W/cm²) to avoid possible heating of the scattering volume. To estimate the possible laser heating effect we measured the spectra of silicon at the same conditions. The Si phonon frequency has a linear dependence on temperature in the range 200 - 800 K (4,5), decreasing by 0.5 cm⁻¹ for every 20 degrees. Such a shift can be easily detected by Raman spectroscopy, which is a very sensitive technique to detect relative shifts of phonon lines (especially when measuring with a multichannel detector and keeping the same position of the monochromator). We did not observe any shifts of the Si phonon line while varying the laser power densities in the range 1 - 10 W/cm². Therefore, we can conclude that the laser heating effect for the Si sample is less than 10 degrees in the conditions we used. Although Si has higher thermal conductivity compared to BaTiO₃ and SrTiO₃, it also has about 10 times stronger absorption in the UV range used in our experiments. Hence, one can expect that the laser heating effect in the BaTiO₃/SrTiO₃ superlattices is not larger than in Si. The spectra of the (BTO₅/STO₄)×25 superlattice ($T_c \approx 530$ K) measured at room temperature with laser power density varied from 1 to 10 W/cm² did not reveal any noticeable changes in the relative intensities of phonon lines in the spectra.

3. Phonon Mode Assignment

Figure S5 shows the comparison of room temperature spectra of two superlattices, (BTO₈/STO₄)×10 and (BTO₅/STO₄)×25, ($T_c = 640$ and 530 K, respectively) with the spectra of BaTiO₃ single crystals in the three different ferroelectric phases (tetragonal, orthorhombic, and rhombohedral) (6) and the SrTiO₃ substrate spectra at 295 and 5 K.

The lines at about 290 cm⁻¹ have similar positions and shapes to the TO₂ modes of A₁ symmetry of BaTiO₃ in the tetragonal phase, thus we assign them to the BaTiO₃ layers and conclude that they are tetragonal. This is supported by the absence of the sharp peak at 310 cm⁻¹ characteristic

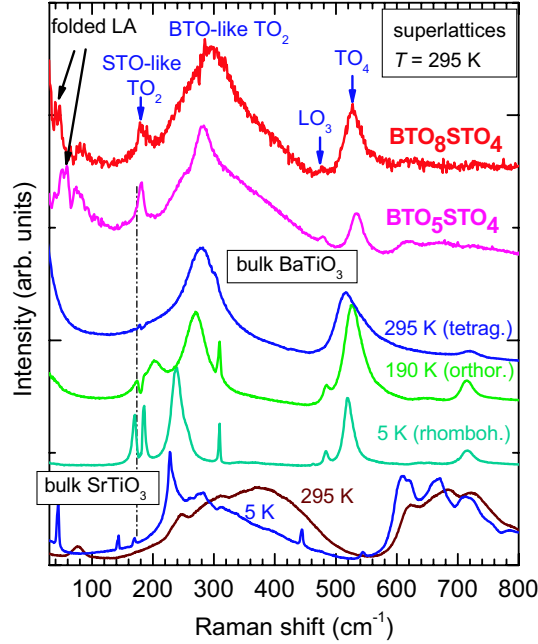


Figure S5. Raman spectra of the $(\text{BTO}_5/\text{STO}_4) \times 25$ and $(\text{BTO}_8/\text{STO}_4) \times 10$ superlattices at 295 K, measured with 351.1 nm laser line. The spectra of BaTiO_3 single crystals at 295, 190, and 100 K, corresponding to tetragonal, orthorhombic, and rhombohedral phases, respectively, and the spectra of SrTiO_3 single crystals at 295 and 5 K are shown for comparison. The bulk spectra were measured with the 514.5 nm laser line.

of the orthorhombic and rhombohedral phases of BaTiO_3 , but not pronounced in the tetragonal phase.

The frequencies of several phonon branches in SrTiO_3 and BaTiO_3 are close to each other, and the phonons are not expected to be strongly localized within the thin SrTiO_3 and BaTiO_3 layers. According to first-principle calculations, these vibrations extend through the whole superlattice. This is the case for the LO_3 and the TO_4 modes, observed at about 478 and 530 cm^{-1} , respectively.

The line at about 180 cm^{-1} is attributed to the SrTiO_3 -like TO_2 phonons. Its position corresponds closely to the TO_2 line in the electric-field-induced Raman spectrum of SrTiO_3 crystals (7) and in the 5 K spectrum of the SrTiO_3 single crystal. The first-order Raman lines are visible in these cases because the electric field and defects break the inversion symmetry in the SrTiO_3 crystals. The 180 cm^{-1} line is not from the SrTiO_3 substrate because the first-order Raman lines are symmetry-forbidden in bulk SrTiO_3 (8). In fact, even the much stronger second-order features of the substrate at 230 and 610 cm^{-1} are barely seen in the UV Raman spectra. Although BaTiO_3 also has a feature due to the TO_1 mode of A_1 symmetry at about the same position (177 cm^{-1}), at room temperature (in the tetragonal phase) that feature is about 15-20 times weaker compared to the TO_2 and TO_4 lines, and it has a characteristic interference dip due to the coupling of the TO_1 and TO_2 modes of A_1 symmetry (9). The 180 cm^{-1} line observed in the spectra of the superlattices is of the same order of magnitude in intensity as the TO_2 and TO_4 lines of BaTiO_3 and does not have the dip characteristic of BaTiO_3 . Therefore, although we cannot absolutely rule out the contribution of the BaTiO_3 layers, we believe this line should be attributed to the phonons of the SrTiO_3 layers in the superlattices. The observation of the first-order Raman scattering by SrTiO_3 phonons indicates that the inversion symmetry is broken, and the SrTiO_3 layers in the superlattices are polar.

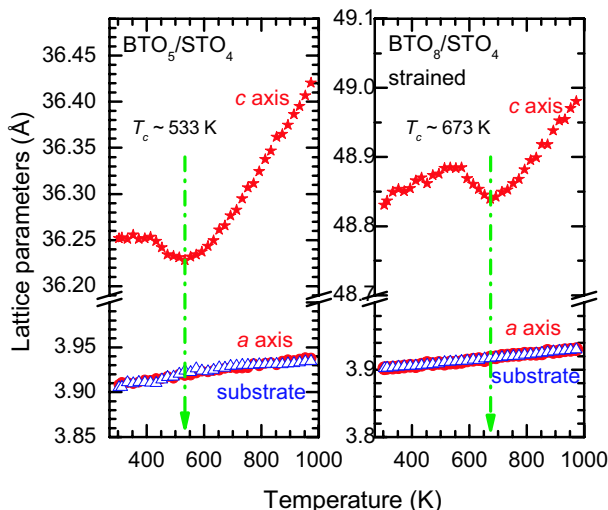


Figure S6. Temperature dependence of the lattice constants of the $(\text{BTO}_5/\text{STO}_4)\times 25$ and $(\text{BTO}_8/\text{STO}_4)\times 10$ superlattices and of the SrTiO_3 substrate, measured by x-ray diffraction..

4. Temperature-Dependent XRD Measurement

Two superlattice samples were measured by temperature-dependent XRD and the results are shown in Fig.S6. The a -axis lattice parameter of the superlattices overlap with that of the substrate, indicating that the superlattices are commensurately constrained to the SrTiO_3 substrate. The temperature dependence of the out-of-plane c -axis lattice parameter undergoes a change at $T = 533$ and 673 K, which can be taken as an indication of T_c (10–13). The T_c values from the XRD measurement provides an additional confirmation of the Raman results.

References

- S1. D. G. Schlom, *et al.*, *Mater. Sci. Eng. B* 87, 282 (2001).
- S2. A. Soukiassian, *et al.*, *unpublished*.
- S3. J. B. Nelson, D. P. Riley, *Proc. Phys. Soc. London* 57, 160 (1945).
- S4. J. Menéndez, M. Cardona, *Phys. Rev. B* 29, 2051.
- S5. G. Lang, *et al.*, *Phys. Rev. B* 59, 6182 (1999).
- S6. D. A. Tenne, *et al.*, *Phys. Rev. B* 69, 174101 (2004).
- S7. P. A. Fleury, J. M. Worlock, *Phys. Rev.* 174, 613 (1968).
- S8. W. G. Nilsen, J. G. Skinner, *J. Chem. Phys.* 48, 2240 (1968).
- S9. A. Scalabrin, A. S. Chaves, D. S. Shim, S. P. S. Porto, *Phys. Status Solidi B* 79, 731 (1977).
- S10. E. D. Specht, H.-M. Christen, D. P. Norton, L. A. Boatner, *Phys. Rev. Lett.* 80, 4317 (1998).
- S11. S. K. Streiffer, *et al.*, *Phys. Rev. Lett.* 89, 067601 (2002).
- S12. M. Sepiarsky, S. R. Phillpot, M. G. Stachiotti, R. L. Migoni, *J. Appl. Phys.* 91, 3165 (2002).
- S13. K. J. Choi, *et al.*, *Science* 306 (2004), 1005 (2004).