Strained crystalline nanomechanical resonators with quality factors above 10 billion

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In strained mechanical resonators, the concurrence of tensile stress and geometric nonlinearity dramatically reduces dissipation. This phenomenon, called dissipation dilution, is employed in mirror suspensions of gravitational-wave interferometers and at the nanoscale, where soft clamping and strain engineering have allowed extremely high quality factors. However, these techniques have so far been applied only to amorphous materials, specifically Si₃N₄. Crystalline materials exhibit substantially lower intrinsic damping at cryogenic temperatures. Applying dissipation dilution engineering to strained crystalline materials could, therefore, enable extremely low loss nanomechanical resonators, as they combine low internal friction, high intrinsic strain and high yield strength. This potential has not yet been fully exploited. Here we demonstrate that single-crystal strained silicon—a material developed for high-mobility transistors—can be used to realize mechanical resonators with ultralow dissipation. We fabricate strained silicon nanostrings with high aspect ratios supporting megahertz mechanical modes with quality factors exceeding 10⁸ at 7 K, a tenfold improvement over values reported in Si₃N₄. We estimate a thermal-noise-limited force sensitivity of (5 ± 2) × 10⁻²⁶ N Hz⁻¹/₂ at 7 K—approaching that of carbon nanotubes—and a heating rate of only 60 quanta per second. The low mass and high quality factors of our nanomechanical resonators make them particularly promising for quantum sensing and transduction.

Crystalline materials feature exceptionally low acoustic (mechanical) losses, as observed in early resonant-bar gravitational-wave detectors, and used for decades in stable-frequency quartz oscillators. The low internal dissipation of crystalline resonators, by virtue of the fluctuation–dissipation theorem, leads to lower thermomechanical noise, which motivated the transition from amorphous to crystalline materials as substrates for mirrors in interferometric gravitational-wave detectors. Furthermore, the development of crystalline mirror coatings, substrates and spacers has improved the frequency stability of reference cavities and lasers. Crystalline nanomechanical resonators have been employed in bolometric, inertial and magnetic force sensors, as well as for cavity optomechanical systems such as silicon optomechanical crystals and Fabry–Pérot cavities with a movable mirror.

Amorphous or glassy materials, irrespective of their precise composition, exhibit universal properties due to the presence of two-level systems (TLSs), such as high dissipation at low temperatures. TLSs can couple to the strain field of acoustic vibrations, giving rise to acoustic absorption. In crystalline materials, TLSs only form due to defects, and energy exchanges with TLS ensembles are drastically reduced. Through the control of spurious damping channels, exceptionally low vibrational damping has been demonstrated in single-crystal resonators, with quality factor \(Q\) of \(~10^9\) at low temperatures in single-crystal silicon optomechanical cavities.

Despite the high internal friction of amorphous materials, dissipation dilution has enabled extremely high \(Q\)s for flexural modes of strained nanomechanical resonators. Dissipation dilution is characterized by the factor \(D_{op}\), which relates the \(Q\) of a resonator mode to the intrinsic \(Q\) of the material as \(Q = D_{op} \times Q_{int}\). Here \(Q_{int} = 1/\tan(\phi) \approx 1/\phi\) where \(\phi \ll 1\) is the phase delay between the strain and stress, that is, the material loss angle. A number of methods to engineer \(D_{op}\) have been developed in recent years, with a majority of implementations reported in amorphous Si₃N₄—a material that can be deposited with large tensile stress. Despite the relatively low \(Q_{int}\) of Si₃N₄, \(Q\)s exceeding \(10^8\) were demonstrated, corresponding to a dilution factor \((D_{op})\) of up to \(10^4\). The combination of dissipation dilution and crystalline materials could enable even lower acoustic damping. In nanomechanical resonators, \(Q_{int}\) is known to be lower than in macroscopic devices, primarily due to pronounced surface effects such as friction in surface oxides, adventitious adsorbed layers or reconstructed surfaces. Despite this, the use of crystalline materials can be highly advantageous: the temperature dependence of intrinsic mechanical dissipation processes often makes the coupling of phonons to TLSs the dominant loss mechanism at cryogenic temperatures. Several demonstrations of dissipation dilution in single-crystal materials, such as GaAs and SiC, have been reported but could not attain lower dissipation than Si₃N₄ devices.

Here we implement strained silicon (sSi) mechanical resonators with ultralow dissipation; sSi was developed as a material for microelectronic devices, and has been used to improve the carrier mobility in MOSFETs and to induce a Pockels coefficient for optical modulation in silicon photonics, but nanomechanical applications of sSi remain unexplored thus far. We show that soft-clamped sSi resonators achieve \(Q\)s of up to \((1.3 ± 0.2) \times 10^8\) at around 1.4 MHz and temperatures of approximately 7 K. To the best of our knowledge, this is the highest \(Q\) reported for a mechanical oscillator at liquid-helium temperatures, exceeded only at millikelvin temperatures in single-crystal silicon optomechanical cavities.

Fabrication of sSi mechanical resonators

In this work, we use strained silicon-on-insulator (sSOI) substrates, where the sSi layer is heteroepitaxially grown and bonded to a...
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Fig. 1 | Dissipation in strained crystalline mechanical resonators.

a. Localized flexural mode of an sSi nanostring, simulated with finite element methods. The insets depict the crystalline lattice of silicon in the string bulk system and TLSs localized at the interfaces. b. Scanning transmission electron micrograph of the cross-section of a processed sSOI sample. The inset shows the sSi crystalline lattice, with crystal directions indicated in the figure. c. Photograph of a chip with an array of sSi strings. d. Diagram of dissipation dilution, with boundaries of coloured regions indicating the \( Q \) versus frequency limits for resonators with 12 nm thickness and lengths below 6 mm. The opaque regions correspond to the estimated \( Q_{\text{int}} \) of Si\(_{3}\)N\(_4\) whereas semi-transparent regions are drawn for the estimated \( Q_{\text{int}} \) of Si at 7 K.

carrier wafer\(^{16}\). As in SOI technology, the sSi layer is conveniently separated from the silicon substrate by a buried oxide layer, facilitating the fabrication of suspended nanomechanical resonators. The average initial thicknesses of the sSi and buried oxide films in our samples are about 14 and 145 nm, respectively. A transmission electron microscopy (TEM) cross-sectional image of the film stack is shown in Fig. 1b.

We developed a fabrication method to suspend high-aspect-ratio phononic crystal (PnC) nanostrings\(^{26}\) (Fig. 2a). Millimetre-scale nanostrings require a large clearance (>10 \( \mu \)m) from the substrate to be reliably suspended without stiction and collapse. An added complication is the chemical identity of the substrate and the device layer, offering no chemical selectivity for undercut steps. To circumvent these issues, our process includes the deposition of several encapsulation layers and a sequence of dry and wet etch steps for suspending the nanostrings (Fig. 2b–e). The fabrication process is detailed in Methods and Extended Data Fig. 1.

We assess the strain in the sSi film after fabrication by imaging a thin cross-section of the chip, taken from the PnC nanostring pad, with a transmission electron microscope. An interferometric electron optics technique, namely, dark-field electron holography\(^{40}\), is employed to construct a strain map with nanometre resolution (Fig. 2f; Methods and Extended Data Fig. 2 provide further details). We measure an average biaxial strain of (0.85 ± 0.06)% (stress, (1.53 ± 0.11) GPa; 95% confidence interval on the mean), close to the supplier specification of 1.3 GPa. The strain is significantly higher than the ~0.4% value of stoichiometric Si\(_{3}\)N\(_4\) on silicon\(^{26}\) due to a lower (direction-averaged) Young’s modulus.

Local strain can also be measured in suspended resonators by Raman spectroscopy\(^{41}\): for silicon, tensile (compressive) strain is known to decrease (increase) the Stokes scattering frequency. When collecting the Stokes signal from the sSOI stack, we observed spectra similar to those depicted in Fig. 2g, where a redshifted contribution from the sSi film can be resolved from the more intense peak at 521 cm\(^{-1}\), scattered by the unstrained substrate. We then collect Raman-scattered light from a suspended nanostring with non-uniform width and display the spatial variation of the Stokes frequency (Fig. 2h): the frequency changes along the longitudinal axis, being lower in the thin parts of the unit cell. This occurs due to the redistribution of stress in a string with non-uniform width\(^{26}\). From the Stokes frequency, we can estimate the uniaxial stress along the unit cell (Fig. 2i), taking laser heating into consideration\(^{41,42}\) (Methods and Extended Data Fig. 3). We display a band of finite width for the reconstructed stress, accounting for uncertainties in the elements of the stiffness—strain tensor and in the Stokes peak from unstrained silicon. The reconstructed stress profile agrees fairly well with an initial stress of approximately 1.5 GPa (Fig. 2i).

**Soft-clamped PnC nanostrings**

In a strained string resonator, \( D_n \) for flexural modes takes the form\(^{26}\)

\[
D_Q = \frac{1}{2\alpha_n \xi + \beta_n \omega_n^2 \lambda^2},
\]

where \( n \) is the mode order; \( \alpha_n \) and \( \beta_n \) are factors that depend on the width profile of the string and mode shape; \( \xi \) is the dimensionless mechanical frequency \( (\xi_n = \omega_n^2 = \frac{\mu \Omega_n}{\mu_s} \omega_n^2) \); \( \lambda \) is the average static strain; \( E \) is the Young’s modulus; \( \rho \) is the material density; \( h \) is the string thickness; \( L \) is the length of the string; and \( Q_n \) is the resonant frequency of mode number \( n \) in angular units. The dominant \( \lambda \) term in the denominator of equation (1) stems from the displacement-field curvature, \( u'' \), close to the clamping points of the string, whereas the \( \lambda^2 \) term originates from the curvature maxima associated with each antinode in the vibrational mode shape\(^{43}\).

We apply the technique of soft clamping\(^{22}\) to nanostrings using width corrugations, implementing a PnC and opening a bandgap around an acoustic wavelength twice as long as the width modulation period\(^{16}\) (Fig. 3a). A schematic of the width corrugation profile is shown in Fig. 3a and is implemented with width modulation \( w_{\text{mod}}/w_{\text{max}} = 2 \) and unit cell length \( l_{\text{uc}} = 108 \mu \text{m} \). The transition between wide and narrow parts of the unit cell is defined by a polynomial curve extending over 0.1\( l_{\text{uc}}\), with vanishing derivatives up to the third order at both ends. A defect that perturbs the translational symmetry is introduced by stretching the length of the narrow region in the string centre to \( l_{\text{sc}} \). The defect is surrounded on either side by seven unit cells (Fig. 2a).

We characterize the devices in an interferometric setup (Methods and Extended Data Fig. 4) and acquire thermomechanical noise spectra to find the out-of-plane (OOP) frequencies of the resonators (Fig. 3d). The resulting vibrational spectra are compiled in Fig. 3b, displaying the OOP flexural modes of soft-clamped nanostrings with \( L \approx 1.6 \text{ mm} \). To match the frequencies and numerical predictions, we assume an initial stress in the sSi layer of around 1.0–1.2 GPa, varying for different chips and fabrication runs. We speculate that this value, significantly lower than the stress...
reconstructed via TEM and Raman methods, might be attributed to the presence of a thin layer of native oxide on the exposed silicon surfaces, which lowers the effective stress in the string cross-section.

A bandgap manifests at around 2.8–3.2 MHz. One or two resonances appear in the bandgap, corresponding to localized modes; their frequencies respond much more sensitively to variations in $l_{def}$ than those of the distributed modes. Such a localized mode, implemented in a 6.0-mm-long nanosting, with 12 unit cells on each side of the defect ($l_{def} \approx 2.3l_{uc} \approx 0.5$ mm), is displayed in Fig. 3e and exhibits soft clamping, that is, a suppressed clamping-point curvature with respect to a uniform string (Fig. 3f). For the localized mode, $\alpha_0 \approx 0$ in equation (1), yielding a more favourable $D_Q$ scaling:

$$D_Q \approx \frac{1}{\pi^4 \Omega_i^2 \Omega_m l_{def}^2} \lambda^2 \propto \frac{\epsilon L^2}{h^2},$$

where $\Omega_m$ is the localized mode frequency and $\Omega_i$ is the fundamental mode frequency. Soft-clamped modes of high-aspect-ratio ($L/h$) strings exhibit enhanced $D_Q$ compared with distributed modes, far from resonance with the PnC, and their $Q$ scales inversely with thickness $h$ in the common case of surface loss-limited $Q_m$ (refs. 21,26).

We experimentally characterize $D_Q$ enhancement resulting from soft clamping in the 6.0 mm nanostring. The measurements are performed at the base temperature of our cryostat, namely, $T \approx 7$ K. The PnC localizes a mechanical mode at $\Omega_i/2\pi = 1.46$ MHz (Fig. 3c). We evaluated the damping of OOP modes by resonantly exciting them and recording ringdown traces on interruption of the drive (Fig. 3d). Since the laser beam increases the sample temperature and influences the reconstructed damping rates (Methods), we perform gated detection by periodically blocking the probe laser with a shutter (Extended Data Fig. 5). The repetition rate of the gates is much slower than thermal relaxation, which occurs at timescales below 1 s.

The measured $Q$s are shown in Fig. 3g: modes out of the vibrational bandgap display a weak dependence on frequency due to the $\lambda^{-1}$ dependence (equation (1)) for hard-clamped modes. Mode $n=26$ (blue circle), localized to the defect region, exhibits $Q=(1.3 \pm 0.2) \times 10^{10}$, more than two orders of magnitude higher than the $Q$ of the distributed modes. The uncertainty corresponds to a 95% confidence interval, estimated from seven separate gated ringdowns. The observed $Q$ enhancement from soft clamping is too large to be explained only by the suppression of intrinsic loss produced by the bending curvature around the clamping points. To reliably model the experimental data, we have to assume further dissipation in the clamping regions, which can be due to phonon radiation to the substrate or mechanical losses in the SiO$_2$ layer underneath the clamping points. We fit a model with two free parameters (Fig. 3g, open circles): the first related to the intrinsic...
dissipation of Si ($Q_m$) and the second representing dissipation in the boundary regions of the string (mostly affecting the out-of-bandgap modes; Supplementary Information). We obtain $Q_m = (8 \pm 3) \times 10^4$ for silicon at 7K, consistent with previous observations and surface-dominated mechanical losses.

**Temperature dependence of dissipation**

We then characterize the $Q$s of several modes of the 6.0 mm nanostring as the sample-holder temperature is tuned with a resistive heater. We probe two modes well below the lower bandgap edge ($Q_s/2\pi \approx 500$ and 900kHz) and the localized mode at $Q_s/2\pi \approx 1.4$ MHz (Fig. 3g). The results are shown in Fig. 4. We initially increase the sample temperature (measured with a thin-film resistive thermometer connected to the sample plate) from 7 to 20K (Fig. 4a, open symbols), without substantial variations in all the $Q$s. We then heat the sample to 300K and gradually decrease the temperature (Fig. 4a, closed symbols). Although the $Q$ of the localized mode around room temperature is partially limited by gas damping (Supplementary Information), we believe this to be negligible at low temperatures due to the lower pressures attained in cryogenic conditions ($<10^{-4}$ mbar).

Importantly, as the base temperature of the cryostat is reached again, we do not recover the initially observed $Q$s: mechanical dissipation is increased by a factor of about three. This observation corresponds to gradual sample degradation over several days (the time required to complete the temperature-dependent measurements), which also explains the difference in dissipation of the measurements displayed with the star and open-circle symbols in Fig. 4a, where the latter points were acquired approximately four days later. This degradation was reversible: we could reproduce $Q > 10^8$ multiple times after heating the chip to room temperature and cooling it back to 7K. On the other hand, the mechanical frequencies drifted irreversibly to lower values, hinting to stress relaxation or to the growth of native oxide. This reversible increase in mechanical dissipation could be due to the slow processes of condensation on the sample of the residual gases in the cold cryostat or to gas damping induced by the rapid variation in the hydrogen partial pressure at around 10K (ref. 44).
The Q values of all the modes have a minimum (the dissipation is peaked) at around 175–180 K and increase rapidly (but not monotonically) for lower temperatures. A closer examination of the relevant temperature range (Fig. 4b) shows that the temperature of the dissipation peak varies with the mode frequency, suggesting the presence of thermally activated defects\(^{21}\), with activation energy of around 0.2 eV, extracted with a fit to an Arrhenius thermal relaxation model\(^{21}\) (Fig. 4b, solid lines). The Q degradation is smaller for the localized mode, which could indicate a weaker coupling to the defect ensemble. A secondary dissipation peak is visible for all the modes at around 120 K. The absence of a dissipation peak at around 175–180 K, where the linear expansion coefficient of silicon approaches zero, excludes thermoelastic damping as a relevant dissipation mechanism. The high yield stress of crystalline materials is particularly attractive for implementing strain-engineered resonators\(^{26,47}\), and the low intrinsic loss of silicon may be combined with recently proposed searches for dark-matter particles\(^{48}\). The built-in mechanical isolation of the soft-clamped modes in this work eliminates many spurious loss mechanisms and provides access to the intrinsic material damping of silicon. The highest observed Q is due to a combination of ultrahigh dilution factor (\(D_{\text{D}}^{} = 2 \times 10^8\)) and a suppressed intrinsic loss of single-crystal silicon in a cryogenic environment.

The temperature scaling of dissipation could be extremely favourable at millikelvin temperatures, owing to low densities of TLSs. In light of recent demonstrations in unstrained optomechanical crystals of Si\(_3\)N\(_4\) at liquid-helium temperatures\(^{25,29}\). The high yield stress of crystalline materials is particularly attractive for implementing strain-engineered resonators\(^{26,47}\), and the low intrinsic loss of silicon may be combined with other mode-shape engineering techniques\(^{24,25,27}\) to decrease damping even further. In addition, surface passivation techniques could be employed to diminish the effect of native oxide formation on dissipation\(^ {46}\). At millikelvin temperatures, thermal transport in strings with low dimensionality might be particularly unconventional, and laser heating could be probed by examining the mechanical properties of the strings. The exceptionally high Q may also lend itself to force sensing, including recently proposed searches for dark-matter particles\(^ {48}\).

### Online content

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Importantly, the separate patterning of Al2O3 allows fine control of the geometry of encapsulate the sSi strings in a protective oxide layer and then use deep RIE and step to strip the resist layers.

Methods
Detailed fabrication process. Nanofabrication of the sSi mechanical resonators starts from 300-mm-format sSOI wafers purchased from Soitec. In this fabrication technology, sSi is obtained by heteroepitaxial growth on Si1-xGex; the mismatch between the crystal lattice constants can be tuned by changing the Ge concentration x, leading to typical biaxial stress levels in excess of 1 GPa. The epitaxial sSi film is bonded to a silicon carrier wafer capped by an oxide layer, and separated from its original substrate by a sequence of ion implantation, thermal treatment and selective etching of the Si1-xGex buffer layer. Our wafers consist of an ~800-μm-thick Si substrate, with 145 nm SiO2 and 14 nm (nominal thickness) sSi thin films on the front side.

The wafers are resized via laser cutting to a 100 mm diameter, suitable for handling and processing in our clean room. After thoroughly stripping the wafer of the photoresist protective film (with a combination of room-temperature acetone and O2 plasma), we proceed to pattern the sSi layer.

Electron-beam resist ZEP520A at 50% diluted is employed, spun at a thickness of roughly 150 nm. During the electron-beam exposure, the patterns are divided into a subregion (surrounding the nanostraining edges and discretized with a fine grid) and a bulk region (including larger patches of the patterns with no common borders with the mechanical resonators). Proximity effect correction is performed to adjust the local exposure dose point by point. With this procedure, we can achieve line roughness of ≤10 nm, while a reasonable duration for the exposure job is maintained (line-edge roughness is measured in a scanning electron microscope (over an ~5×5 μm² field of view) after dry etching the sSi layer; here we quote a 3σ confidence interval). The nanobeams are patterned along the (110) directions (with a perimeter approximately 2π-dominated of the precision of the reference wafer cut), where the stress relaxation on release is negligible due to a minimal Poisson’s ratio of ν=0.06 (Young’s modulus, E=169 GPa)⁴⁴. Using ZEP520A as a mask, we pattern the sSi layer with inductively coupled plasma reactive-ion etching (RIE) using SF6 + C4F8 gases. Only a few seconds of exposure to plasma is needed to etch the film, landing on the SiO2 underneath. To increase reproducibility, we, therefore, pre-condition the inductively coupled plasma chamber by exposing a test Si wafer to the same chemistry before etching the actual devices. The resist is stripped with a sequence of hot N-methyl-2-pyrrolidone and O2 plasma; this procedure is carefully repeated after each etch step to strip the resist layers.

The goal of the process at this stage is to create sufficient clearance between the sSi layer and substrate to suspend millimetre-scale devices. To achieve this, we encapsulate the sSi strings in a protective oxide layer and then use deep RIE and an isotropic undercut (in a sequence similar to the method described elsewhere⁴⁷) with high oxide chemical selectivity. However, an extra constraint arises: when such a multilayer beam is suspended, it will buckle and very commonly collapse or break, due to the high compressive stress intrinsic to thermal and deposition oxides (Extended Data Fig. 1a). The solution is to ensure that the encapsulating layer has tensile stress and is sufficiently thick to compensate the compressive stress in the buried oxide underneath (the presence of the sSi layer can be ignored due to its negligible thickness; Extended Data Fig. 1b).

In practice, multiple dielectric films are deposited. First, roughly 10 nm SiO2 is deposited through atomic layer deposition, covering the patterned surface. Then, hydrogen-rich SiNx is grown through plasma-enhanced chemical vapour deposition. With 40 W RF power generating the plasma, the film exhibits strong tensile stress, stable in time, of approximately +330 MPa (characterized through the Stoney wafer-bending method⁴⁸), consisting of electron-beam lithography, pattern transfer via dry etching and HF-vapour release, exploiting the thin sacrificial layer of our substrates.

Estimating stress with electron holography. A thin cross-sectional slice (~100 nm thickness) is carved from the support pads of the nanobeams, using a focused-ion-beam specimen preparation technique. The slice is observed in a TEM instrument (Fig. 1b). An electronic interference method, namely, dark-field electron holography⁴⁰, is employed to quantitatively map the strain in the crystalline films (Si substrate and bonded sSOI layer), with nanometre-scale resolution (Fig. 2). A magnified illustration, centred on the sSi film, is reproduced in Extended Data Fig. 2a. From the signal-to-noise ratio of the strain distribution in the sSi film (Extended Data Fig. 2b), we calculate the 95% confidence interval on the average biaxial strain as (0.85±0.06)% as quoted in the main text.

To acquire the dark-field electron holography strain map, the impinging electron wave is diffracted by the crystal of the specimen, set in the two-beam condition to enhance the intensity of the (220)-diffraacted beam that is selected with an objective aperture placed in the back focal plane of the objective lens of the TEM instrument. Then, an electron biprism is used to interfere the reference wave emerging from the unstrained Si substrate (the dark-field image of the substrate) with the wave emerging from the measurement area (the dark-field image of the sSOI) to generate a hologram. Fourier processing is used to retrieve the geometric phase encoded in the interference fringes of the hologram, and the in-plane strain map is calculated from the gradient of the geometric phase, being proportional to the displacement field of the (220) silicon lattice planes⁴⁹.

Electron holography experiments were performed with a double aberration-corrected FEI Titan Temis operated at 300 kV. Stacks of holograms of 5 s exposure time were recorded with a FEI Ceta camera to improve the signal-to-noise ratio of the strain maps⁵⁰. The biprism bias was 200 V, giving a hologram carrier frequency of 0.69 nm⁻¹. The numerical aperture used for the Fourier processing limits the spatial resolution of the strain map to 2.75 nm.

Estimating stress from Raman shifts. Micro-Raman spectroscopy has long been a powerful technique to characterize the local stress state of crystalline samples, owing to the high spatial resolution achievable with confocal microscopy and the high resolving power of diffraction gratings, compared with typical stress-induced Raman shifts.⁵¹,⁵²

Silicon is arguably the best-studied material in this respect. The frequencies of long-wavelength optical phonons are perturbed by strain. In the absence of strain, the modes are degenerate at wavenumber k=π/2R cm⁻¹ (ωs=2π×15.6 THz); the introduction of strain modifies the crystal stiffness tensor and perturbs the mode frequencies, breaking their degeneracy⁵³. This influence is generally approximated as a linear dependence; by symmetries of the silicon lattice, only three independent elements appear in the stiffness–strain tensor, namely, p, q and r. These are usually determined experimentally, and different values have been reported in the literature.⁵⁴ We report here the numerical values from another study⁵⁵:

\[
p = -1.30 \times 10^6 \text{ cm}^{-1} \text{ atm}^{-1} \\
q = -1.90 \times 10^6 \text{ cm}^{-1} \text{ atm}^{-1} \\
r = -0.66 \times 10^6 \text{ cm}^{-1} \text{ atm}^{-1}
\]
The perturbed normal mode frequencies are obtained by finding the eigenvalues $\lambda$ of the matrix

$$
\begin{bmatrix}
\epsilon_{11} + q (\epsilon_{22} - \epsilon_{33}) & 2r_{e12} & 2r_{e13} \\
2r_{e12} & \epsilon_{22} + q (\epsilon_{31} - \epsilon_{33}) & 2r_{e23} \\
2r_{e13} & 2r_{e23} & \epsilon_{33} + q (\epsilon_{11} + \epsilon_{22})
\end{bmatrix},
$$

(5)

from which the phonon wavenumbers can be computed as $\omega = \epsilon_{xx} + \lambda / 2 \epsilon_{xx}$. Phonon polarizations are found as the corresponding eigenvectors. Raman-scattering selection rules impose that only scattering by longitudinal phonons (polarized along $z$) can be measured, on reflection on a (100) surface.

A particular strain state is assumed to simplify the eigenvalue equation and a single strain or stress component is extracted from the measurement of $\Delta \omega$. Importantly, the matrix shown in equation (5) is expressed in the reference system of 100]. Therefore, the strain tensor must be rotated to the same axes.

We report here the resulting linear relations between stress and Raman shift in the relevant cases: uniaxial stress directed along [110] or [100]: $\Delta \omega / \sigma \approx -2.0 \text{ cm}^{-1} \text{GPa}^{-1}$ and biaxial stress $\sigma_{xx} = \sigma_{yy} \approx -4.1 \text{ cm}^{-1} \text{GPa}^{-1}$. Uniaxial stress directed along [110] or [100]: $\Delta \omega / \sigma \approx -2.0 \text{ cm}^{-1} \text{GPa}^{-1}$, Biaxial stress $\sigma_{xx} = \sigma_{yy} \approx -4.1 \text{ cm}^{-1} \text{GPa}^{-1}$, confirming the general observation that tensile (compressive) stress leads to a redshift (blueshift) in the Raman-scattering peak.

In our confocal microscopy setup, we excite Raman scattering by focusing a 488 nm laser with a typical output power of 10–50 mW on the sample (located at room temperature and atmospheric pressure), we detect the Stokes peak in reflection and separate it from the laser line, using a diffraction grating with a high resolving power. The intensity and wavenumber of the Stokes peak are recorded by a charge-coupled device sensor. When the laser beam is focused on the suspended string, the spectrum consists of a single Lorentzian, since the recessed substrate is beyond the depth of focus of the microscope. On the other hand, when the laser is focused on a region where sSi is not undercut, two contributions can be distinguished: an intense Lorentzian line at $\nu = \nu_0$ collected from the unrestrained substrate and a redshifted Lorentzian peak from the biaxially stressed sSi layer. By extracting $\Delta \omega = \nu - \nu_0$ from the fit, the stress magnitude can be inferred as described above. We measure a Stokes shift from the unrestrained substrate of $\nu_0 = 521 \text{ cm}^{-1}$, at a small impinging optical power.

Heating effects in the sSi layer influence the reconstructed stress magnitude. Silicon absorbs strongly in the pump-laser wavelength, and heating is exacerbated by the poor thermal conductance of the nanostrings. A temperature increase leads to a redshift in the phonon frequency (due to anharmonic terms in the potential energy of atomic bonds) and changes the local strain. We experimentally account for this temperature dependence by varying the pump power and extracting, point by point, the Stokes shift at vanishingly small power via linear extrapolation (Extended Data Fig. 3).

In Fig. 2i, we account for experimental uncertainties by displaying a plausible region for the reconstructed stress as a function of position. The light-green band is obtained by varying $\epsilon_{33}$ between 520.5 and 521.5 cm$^{-1}$ and $\Delta \omega / \sigma$ between $-2.0$ and $-2.3$ cm$^{-1}$ GPa$^{-1}$ (refs. 39, 40).

**Mechanical characterization setup.** We conduct the mechanical characterization measurements presented in the main text in a balanced Mach–Zehnder homodyne interferometer, in which we inject light of a 1.550 nm tunable external-cavity diode laser (Extended Data Fig. 4). We mount the samples in a low-vibration closed-box to avoid external vibrations. The setup is 185870 (Ambizione). This work was further supported by the Defense Advanced Research Projects Agency (DARPA), Defense Sciences Office (DSO) contract no. HR00111810003. All the samples were fabricated at the Center of MicroNanoTechnology (CMi) at EPFL.

**Author contributions** A.B. fabricated the devices with assistance from M.I.B. The devices were characterized by A.B., D.A.V. and N.J.E. The characterization setup was constructed by S.A.F., N.J.E. and T.J.K. The manuscript was written by A.B. and N.J.E. with assistance from the other authors. The work was supervised by N.J.E. and T.J.K.

**Competing interests** The authors declare no competing interests.

**Additional information**

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Extended Data Fig. 1 | Fabrication of sSi nanostrings. a, Suspended nanostrings without a stress-compensation layer are subject to a compressive load well beyond the critical value for out-of-plane buckling. Green - SiO₂, blue - patterned sSi + FOX16 mask. b, The encapsulation layers (dark and light green) compensate compressive stress; the suspended beam is tensioned and flat. c, Determination of the intrinsic stress in PECVD Si₃N₄ through wafer bow. The wafer height profile is measured by scanning a laser beam and recording the angular deflection of the beam reflected off the surface. The magnitude and sign of the profile curvature permit to reconstruct the biaxial stress. Red circles - wafer profile before nitride deposition, orange circles - wafer bow after nitride deposition, with the initial profile subtracted.
Extended Data Fig. 2 | Color-coded strain map in sSi, measured with dark-field electron holography.  

**a**, Portion of Fig. 2f magnified around the sSi layer.  

**b**, Distribution of reconstructed strain values in the sSi layer, centered on the average strain $\langle \epsilon \rangle = 0.85\%$. 
Extended Data Fig. 3 | Raman scans with variable laser power. Micro-Raman scans along a PnC beam unit cell are conducted, with variable laser power. A linear dependence of the Raman scattering wavevector on the power is observed; the dotted line indicates the extrapolation to zero optical power.
Extended Data Fig. 4 | Schematic of the measurement setup. ECDL: external cavity diode laser. EOM: electro-optical modulator. WDM: fiber optic wavelength division multiplexer. PZT: piezoelectric transducer. ESA: spectrum analyzer.
**Extended Data Fig. 5 | Comparison of ringdowns with continuous and gated acquisition.** The probe laser, with an impinging power $\geq 1 \text{mW}$, induces, in the continuous ringdown (green line), additional optical damping, leading to a faster amplitude decay rate than the one set by the intrinsic dissipation of the nanostring. For the gated ringdown (blue dots), the measurement duty cycle (ratio of gate duration over the time interval between successive gates) is 10%, and the gate repetition rate is 1/20 s. In both ringdowns, nonlinear damping is observed at the highest amplitudes, and exponential fits (dashed lines) are performed only within the linear decay regions. The data were acquired at $T \approx 10\text{K}$ for the 6.0 mm sample presented in the main text.