Large-scale ultra-fast strain engineering of CVD-grown two-dimensional materials on strain self-limited deformable nanostructures towards enhanced field effect transistors

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Abstract

Strain engineering of 2D materials is capable of tuning the electrical and optical properties of the materials without introducing additional atoms. However, there are still great challenges in realizing straining of 2D materials with CMOS compatibility. Here, a method for large-scale ultrafast strain engineering of CVD-grown 2D materials is proposed. We introduce locally non-uniform strains through the cooperative deformation of materials and metal/metal oxide core/shell nanoparticles through cold laser shock. Raman and PL spectra reveal that the tensile strain of MoS2 changes and the band gap decreases after laser shock. MD simulations are used to investigate the mechanism of the ultrafast straining of CVD-grown 2D materials. Field effect transistors of CVD MoS2 were fabricated, and the performances before and after straining of the same devices are compared. By adjusting the strain level of MoS2, the field effect mobility can be increased from 1.9 cm2V-1s-1 to 44.1 cm2V-1s-1. This is the maximum value of MoS2 FETs grown by CVD with SiO2 as dielectric. As an environment-friendly, large-scale and ultra-fast manufacturing method, laser shock provides a universal strategy for large-scale adjustment of 2D materials strain, which will help to promote the manufacturing of 2D nano electronic devices and optoelectronic devices.

Full Text

Large-Scale Ultra-Fast Strain Engineering of CVD-Grown Two-Dimensional Materials on Strain Self-Limited Deformable Nanostructures Towards Enhanced Field-Effect Transistors

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Abstract

Strain engineering of two-dimensional (2D) materials offers a powerful approach to tune their electrical and optical properties without introducing foreign atoms. However, major challenges remain in implementing strain engineering of 2D materials with CMOS compatibility. Here we propose a method for large-scale ultrafast strain engineering of chemical vapor deposition (CVD)-grown 2D materials that introduces locally non-uniform strains through the cooperative deformation of materials and metal/metal oxide core/shell nanoparticles via cold laser shock. Raman and photoluminescence (PL) spectroscopy reveal that the tensile strain in MoS_2 increases and its band gap decreases after laser shock treatment. Molecular dynamics (MD) simulations are employed to investigate the mechanism of this ultrafast straining process in CVD-grown 2D materials. Field-effect transistors (FETs) based on CVD-grown MoS_2 were fabricated, enabling direct comparison of device performance before and after straining. By controlling the strain level in MoS_2, the field-effect mobility can be enhanced from 1.9 $cm^2V^{-1}s^{-1}$ to 44.1 $cm^2V^{-1}s^{-1}$, which represents the highest value reported for CVD-grown MoS_2 FETs with SiO_2 as the dielectric. As an environmentally friendly, large-scale, and ultra-fast manufacturing method, laser shock provides a universal strategy for large-scale strain adjustment in 2D materials, promising to advance the fabrication of 2D nanoelectronic and optoelectronic devices.

Introduction

The physical properties of most materials can be widely tuned through strain engineering. Among various material systems, strain engineering of two-dimensional materials is particularly promising because individual atomic layers can inherently withstand much greater mechanical strain than their

bulk counterparts or conventional electronic materials. In 2D materials, strain alters the configuration of atomic bonds (their length, angle, and strength) and modifies the interactions between electron orbitals, thereby enabling control over material properties and providing a rich library for advanced applications.

Typical straining techniques include strain transfer from flexible or stretchable substrates, MEMS-based stretching, sphere diameter engineering, and nanoscale surface morphology regulation. Among these, only the nanoscale surface morphology regulation approach is CMOS-compatible and allows fabrication of field-effect transistors. However, current surface morphology regulation methods typically rely on rigid surface structures defined by high-resolution lithography or chemical processes, which are difficult to fabricate accurately at large scales or when multiple layers are involved. While 2D materials on these nanoscale features can be strained through chemical solution filling and vaporization, residual contaminants prove detrimental to transistor performance. Alternatively, transferred 2D crystals can be strained through weak van der Waals interactions with the substrate, and studies have shown that exfoliated samples on nanofeatures smaller than 2 nm exhibit tremendous improvements in device mobility. In this case, the feature size is comparable to the material thickness (~1 nm). However, for large-scale applications, CVD-grown 2D materials with non-uniform wrinkles (tens of nanometers) and defects are more practical. These wrinkles easily relax strains on 2 nm-scale nanofeatures, and increasing the aspect ratio of such features would fracture the 2D materials, interrupting current flow.

The conformability of 2D materials on surface features at the tens-of-nanometers scale has been extensively studied. It is generally understood that the strain value is established through a balance between strain energy and adhesion potential within a certain geometric scale. In this slow, self-balancing process, wrinkles in CVD-grown materials relax elastic strains, making them difficult to build up and control. Recently, ultrafast straining using shock pressure induced by laser ablation has been investigated. The ultra-high strain rate deformation capability prevents excessive self-relaxation of wrinkles. Previous studies have strained nanowires and 2D materials using hard, rigid SiO_2 nanofeatures defined by electron beam lithography. While this approach benefits FET device fabrication, it is too costly and easily generates geometric stress concentrators that break devices. Polymer cushion layers have been used to address this issue, but the charging effects of polymers degrade device performance. Thus, for CVD-grown 2D materials, large-scale ultrafast straining with low cost, high efficiency, and good controllability has not yet been demonstrated.

In this paper, we demonstrate strain engineering of 2D materials on SiO_2/Si substrates using strain self-limited deformable metal@metal oxide nanoparticles (M@MO NPs) through laser shock-induced cooperative deformation (LSICD). This process modulates the strain in MoS_2 through laser shock-induced cooperative deformation of metal nanoparticles and MoS_2 on rigid substrates, thereby altering its band structure. Field-effect transistors based on CVD-grown MoS_2 were fabricated, achieving the highest field-effect mobility reported

for CVD MoS_2 with low-k dielectrics. This contamination-free, ultra-fast opto-mechanical nanomanufacturing method provides a universal strategy for large-scale strain adjustment in 2D materials and is compatible with traditional semiconductor manufacturing, promising to advance the development of high-performance electronic and mechanical devices.

Results and Discussion

Laser Shock Treatment and Cooperative Deformation Mechanism

Laser shock treatment of 2D materials on rigid surface structures typically causes material fracture at stress concentration points, as shown in Fig. S1. Molecular dynamics simulation results reveal that during laser shock, 2D materials experience enormous stress at mold turning points, leading to lattice destruction, defects, or fractures (Fig. S1(a-b)). SEM images of samples confirm that 2D materials shocked on rigid substrates suffer irreparable damage under impact pressure, degrading their electrical and optical properties.

To overcome this limitation, we developed a method to produce large-area non-uniform strain in 2D materials, as illustrated in Fig. 1(a). The 2D materials were transferred to SiO_2/Si substrates pre-deposited with nanoparticles using PMMA-assisted transfer, then subjected to laser shock treatment. Due to the non-uniform size of nanoparticles on the substrate surface, the transferred 2D materials were supported like a tent, creating local strain. As reported by Garaj et al., 2D materials transferred to rough substrates can produce strain, but this depends entirely on substrate roughness and cannot be effectively controlled. Here we demonstrate that strain adjustment in 2D materials can be achieved through high strain rate deformation induced by laser shock. Specifically, when a high-energy-density nanosecond laser beam is focused on a graphite surface, plasma is generated, and the recoil momentum of this plasma produces a pulse pressure. The gap between the 2D materials and SiO_2/Si substrates decreases in an ultrafast manner, which eliminates strain relaxation due to wrinkles and increases local strain during LSICD (Fig. 1(b)). In this method, nanoparticles can be formed by sputtering or direct spin coating without cumbersome lithography, making it scalable for large-area and roll-to-roll manufacturing. The cooperative deformation of metal nanoparticles also prevents defects or fractures caused by stress exceeding the strain limit of 2D materials. Laser processing offers additional advantages including high production efficiency, reliable quality, economic benefits, and easy integration with existing semiconductor manufacturing processes.

Mechanical Properties and Nanoparticle Selection

The mechanical properties of the materials are crucial for successful strain engineering. We simulated the morphological changes of 2D materials suspended on rigid nanoparticles versus different metal nanoparticles under laser shock using MD simulations (Fig. 2). The results show that Si and Ti nanoparticles

undergo minimal deformation during laser shock. As the 2D materials impact these hard nanoparticles, the films experience enormous stress at the nanoparticle tops, resulting in cracks. With continued movement of the Al plate, these cracks expand further, significantly affecting the mechanical and photoelectric properties of the 2D materials. In contrast, metal nanoparticles with lower yield strengths, such as Au, Ag, and Al, can undergo cooperative deformation with 2D materials under laser shock, reducing defect generation.

MD simulations of the entire surface and strain distributions are shown in Fig. S2. As the Al plate moves downward, the metal nanoparticles gradually deform and the MoS_2 films approach the substrate, wrapping around the nanoparticles (Fig. S2a(-)). When the Al plate collides with the substrate and rebounds, the gap between MoS_2 films and the substrate temporarily increases due to interactions between the Al plate and MoS_2 and stress waves propagating along the plane (larger than in reality due to total reflection at the flat surface in the simulation) after the MoS_2 films collide with the substrate at high speed (Fig. S2a( )). After the Al plate departs, the in-plane stress waves continue adjusting the gap between MoS_2 films and the substrate. As the stress wave energy gradually dissipates, the nanoparticles become wrapped through interactions between the nanoparticles, substrate, and film (Fig. S2a( )).

Deformation Mechanism and Strain Distribution

The deformation process is schematically depicted in Fig. 3(a). When monolayer MoS_2 films are transferred onto nanoparticles, they are supported like a tent, bonding only at the nanoparticle tops, with wrinkles relaxed at this stage. Under laser shock pressure, MoS_2 quickly attaches to the flat plane without time for relaxation from other areas. After impact, the van der Waals force between MoS_2 and nanoparticles maintains this state, and the contact area between the three-dimensional nanoparticle surface and MoS_2 increases, achieving greater strain (Fig. 3(b)). This mechanism is confirmed by SEM images of the NPs/MoS_2 structure before and after LSICD (Fig. 3(d-e)). The maximum atomic strain in monolayer MoS_2 after LSICD reaches 10%. As the MoS_2 films approach the substrate, the strained region expands and strain energy increases sharply. When the films contact the substrate, maximum atomic strain appears at the nanoparticle tops and gradually decreases along both sides from the center, while flat areas in full contact with the substrate exhibit essentially no strain.

Spectroscopic Characterization of Strain

Tensile strain softens the lattice spring constant, which influences crystal phonon modes. Raman spectroscopy was used to study these phonon modes and characterize strain in MoS_2 before and after laser shock. As shown in Fig. 4, the Raman spectrum of MoS_2 exhibits two prominent characteristic vibration modes: $E^1{}_2g$ at 385.4 cm^{-1} and A_1g at 405.8 cm^{-1}, corresponding to in-plane and out-of-plane vibration modes, respectively. After transferring monolayer

MoS_2 to M@MO NPs substrates, the $E^1{}_2g$ peak red-shifts by 0.6 cm^{-1} relative to as-grown MoS_2. Following shock treatment with different laser energies, the $E^1{}_2g$ peak first red-shifts then blue-shifts, with a maximum red shift of 3.3 cm^{-1}. The $E^1{}_2g$ mode is highly sensitive to uniaxial tensile strain because strain-induced crystal symmetry breaking in transition metal dichalcogenides softens the doubly degenerate in-plane vibration. Thus, the $E^1{}_2g$ shift demonstrates that MoS_2 experiences strain when transferred to M@MO NPs substrates, and this strain increases after LSICD. Compared with previous literature reports, M@MO NPs/MoS_2 produces approximately 0.2% tensile strain, and the average maximum strain in MoS_2 after laser shock is about 1.2%.

The strain-induced band gap reduction is consistent with existing theoretical and experimental reports, confirming effective strain transfer to MoS_2 samples. The Raman and PL intensities of MoS_2 transferred to M@MO NPs are significantly enhanced (Fig. S6), primarily due to the strong plasmonic effect of Ag nanoparticles that amplifies the local electric field. We note that the observed strain and plasmonic enhancement factors represent averages over a 1 m sample region (limited by the laser spot size), while the maximum local strain and plasmonic enhancement could be substantially larger.

Electrical Performance of Strained MoS_2 FETs

Monolayer MoS_2 FETs and MoS_2 FETs transferred to M@MO NPs (NPs/MoS_2 FETs) were fabricated to investigate strain effects on electrical properties. To preserve the lattice and intrinsic properties of 2D materials, we used mechanically laminated and transferred metal electrodes on M@MO NPs/MoS_2. Fig. 5 shows the transfer and output characteristic curves for MoS_2 FETs, NPs/MoS_2 FETs, and M@MO NPs/MoS_2 FETs after LSICD (LSICD NPs/MoS_2 FETs). The output current of pristine MoS_2 FETs is 3.8 A, showing nonlinear I-V characteristics and Schottky contact behavior. In contrast, MoS_2 transferred to M@MO NPs exhibits linear behavior, indicating ohmic contact at room temperature. This contact type change may be related to film strain, as recent studies have shown that tensile strain can reduce the contact energy barrier between MoS_2 and Au. The maximum output currents of NPs/MoS_2 FETs and LSICD1 NPs/MoS_2 FETs are 96 A and 134 A, respectively—about 25 times and 35 times higher than that of pristine MoS_2 FETs. When MoS_2 is transferred to M@MO NPs, the electrode-MoS_2 contact changes from Schottky to ohmic, attributable to the contact barrier reduction caused by strain.

Fig. 5d shows the transfer characteristic curves of MoS_2 FETs and NPs/MoS_2 FETs with and without LSICD under a source-drain bias of 2.5 V. The on-state current density of pristine MoS_2 FETs is 0.07 A/ m. Similar to the output performance improvement, the on-state current density of NPs/MoS_2 FETs increases to 1.89 A/ m—more than an order of magnitude higher than pristine MoS_2 FETs. Remarkably, the on-state current density of the laser-treated sample (LSICD1 NPs/MoS_2 FETs) reaches 2.73 A/ m, 1.4 times higher than the sample without laser shock treatment. To further compare device characteris-

tics, we calculated the field-effect mobility (_{FE}) using the standard equation. In this equation, the maximum transconductance is extracted from the transfer characteristic curve, L (20 m)/W (80 m) represents the channel aspect ratio, and C_{ox} is the capacitance of the 300 nm silica gate oxide. The calculated field-effect mobility is 1.9 $cm^2V^{-1}s^{-1}$ for pristine MoS_2 FETs and 36.1 $cm^2V^{-1}s^{-1}$ for NPs/MoS_2 FETs—a nearly 20-fold increase. The _{FE} of LSICD1 NPs/MoS_2 FETs further increases to 44.1 $cm^2V^{-1}s^{-1}$, which is the maximum value reported for CVD-grown MoS_2 FETs with SiO_2 as the dielectric.

We also calculated the carrier concentration and resistivity of MoS_2 using the standard formulas, where V_{th} is the threshold voltage, V_G = 16 V, and e is the electronic charge. The carrier concentration of monolayer MoS_2 and Ag NPs/MoS_2 is 3.22$×10^{12}$ cm^{-2} and 3.69$×10^{12}$ cm^{-2}, respectively. After laser shock treatment, the carrier concentration increases to 3.93$×10^{12}$ cm^{-2}. Although _{FE} and carrier concentration decrease slightly for samples treated with higher laser shock energy (LSICD3 NPs/MoS_2), their values remain higher than directly transferred samples overall, and all four sample types have carrier concentrations in the same order of magnitude.

Theoretical Analysis via DFT Calculations

Density functional theory (DFT) calculations were performed using the Vienna Ab initio Simulation Package to elucidate the mobility enhancement in MoS_2 FETs caused by LSICD. Fig. S14 shows the band structure of MoS_2 under biaxial tensile strain of 1-3% and different bending heights (defined in the Supporting Information, with corresponding structures shown in Fig. S15). The band gap of MoS_2 decreases under both biaxial strain and bending-induced strain, corresponding to the red shift observed in PL spectra. To estimate changes in carrier mobility, we calculated the effective masses by fitting the band structure near the valence band maximum (VBM) and conduction band minimum (CBM) using a quadratic equation. The electron and hole effective masses of MoS_2 under stretched and bent conditions are shown in Fig. S15. Under 3% biaxial strain and a bending height of 4.96 Å, the electron effective mass is 0.41 m_0 and 0.42 m_0, respectively—both lower than the 0.48 m_0 value in the strain-free state, consistent with the FET measurement results.

Conclusion

In summary, this work proposes a strategy to modulate strain through cooperative deformation of 2D materials and metal nanoparticles induced by laser shock. When 2D materials are transferred to M@MO NPs and deformed by laser shock, the ultra-high strain rate deformation of both materials prevents film defects and cracks while building up elastic strains. Using MoS_2 as a prototypical 2D material, we verified the effectiveness of this method. Raman and PL spectroscopy confirmed increased tensile strain and decreased band gap in MoS_2 after LSICD. Due to the reduction in electron effective mass caused by

strain, the field-effect mobility of transistors increases to 44.1 $cm^2V^{-1}s^{-1}$—23 times higher than pristine MoS_2 FETs and the maximum value reported for CVD-grown MoS_2 FETs with SiO_2 dielectric. As a clean processing method, laser shock does not unintentionally dope MoS_2. This pollution-free, large-scale, ultra-fast opto-mechanical nanomanufacturing method provides a universal strategy for large-scale strain adjustment in 2D materials and is compatible with traditional semiconductor manufacturing, offering significant potential for advancing two-dimensional nanoelectronic devices and expanding the application of 2D materials in semiconductor devices.

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Supporting Information

Simulation Methods

Molecular Dynamics (MD) Simulation MD simulations were performed using the LAMMPS code. In the simulation, metal nanoparticles and the Al piston were assumed to be face-centered cubic (FCC) single crystals with lattice constants of 4.09 Å and 3.9860 Å, respectively. A shock wave was generated by applying an initial velocity to the Al piston. The diameter of metal nanoparticles was 5 nm. Mixed pair styles were used in calculations: the embedded atom method (EAM) potential described metal nanoparticles and the Al piston, the Stillinger-Weber (SW) potential described interactions between atoms in monolayer MoS_2, and 12-6 Lennard-Jones potential described interactions between Ag, Au, Cu, Si and MoS_2. MD simulation was carried out using a microcanonical ensemble (NVE) at ~300 K, with metal nanoparticle positions fixed in the XY plane using fix recenter codes. The simulation time step was 1 fs, and Ovito software was used for structure visualization.

Density Functional Theory (DFT) Calculations DFT calculations were performed via the Vienna Ab initio Simulation Package (VASP). The projector augmented wave (PAW) method described the effective potential between ionic cores and electrons. The generalized gradient approximation (GGA) in the Perdew-Burke-Ernzerhof (PBE) parametrization was used for the exchange-correlation functional. Band structure was calculated along symmetry points Γ, X, S, Y, Γ. The k-point sampling of the first Brillouin zone used a 2$×8×1Γ−*centeredgrid.Theenergycutoffwas*500*eVforallcases.Allgeometrieswereoptimizedusingtheconjugated*−*gradientmethoduntilaHellman*−*Feynmanforceconvergencethresholdof*10^{-3}$ eV/Å was reached, with energy differences converged within 10^{-6} eV for each self-consistency iteration. To avoid interactions between adjacent layers, the vacuum region exceeded 25 Å.

Experimental Methods

Growth of Monolayer MoS_2 Monolayer MoS_2 was grown in a BEQ double temperature zone tubular furnace by chemical vapor deposition (Fig. S3). The tube furnace was evacuated to 1 Pa and purged with 200 sccm Ar gas for 15 minutes to minimize O_2 content. SiO_2/Si substrates were soaked in piranha solution for 2 hours, then cleaned sequentially with deionized water, acetone, absolute ethanol, and deionized water for 10 minutes each, and dried with N_2. The prepared Na_2MoO_4 solution (4 mg/ml) was spin-coated on SiO_2/Si substrates, which were then placed in the double temperature zone tubular furnace with 50 mg sulfur powder in the upstream heating zone. The sulfur and Na_2MoO_4 areas were heated to 150°C and 800°C, respectively. Growth proceeded for 15 minutes under Ar/H_2 gas flow (60 sccm), after which the flow rate was increased to 400 sccm for rapid cooling.

Fabrication of M@MO NPs and MoS_2 FETs Ag films were deposited on Si/SiO_2 substrates (300 nm SiO_2) by DC magnetron sputtering, then annealed in a tubular furnace at 200°C for 30 minutes in air to form M@MO NPs. As-grown monolayer MoS_2 was transferred to Si/SiO_2 substrates with Ag NPs using a polymethylmethacrylate (PMMA)-assisted wet transfer method. FET metal electrodes were fabricated by UV exposure and thermal evaporation of 100 nm Au electrodes. The Au electrodes on Si/SiO_2 substrates were transferred to M@MO NPs/MoS_2 by PMMA-assisted dry transfer to complete FET fabrication. The process involved: (1) spin-coating PMMA on the electrodes, (2) immersing PMMA/Au electrodes in 2 mol/L KOH solution for 30 minutes to separate them from the Si/SiO_2 substrate, (3) diluting with deionized water multiple times and transferring to MoS_2 using a homemade transfer platform, (4) heating transferred samples at 60°C for 30 minutes to improve electrode-MoS_2 contact, and (5) removing residual PMMA with acetone and isopropanol.

Laser Shock Treatment A Nd:YAG laser (pulse width: 9 ns, wavelength: 1064 nm) was used for laser shock treatment of M@MO NPs/MoS_2. A 4 m

Al foil sacrificial layer coated with graphite was placed on the MoS_2 FET surface to absorb laser energy. Under pulsed laser irradiation, the graphite layer evaporates and ionizes instantaneously, and the expanded plasma is confined by a transparent limiting layer, generating an instantaneous pressure shock wave. The Al foil served as a momentum transfer layer to transmit the pressure shock wave to the metal electrode and induce deformation of both the metal electrode and MoS_2. The beam diameter (typically ~5 mm) was controlled by a focusing lens and calibrated with photosensitive paper.

Characterization The morphology of M@MO NPs and MoS_2 was examined by scanning electron microscopy (SEM, Tescan Mira3). Surface profiles before and after laser shock were measured in touch mode using a MicroNano D-5A atomic force microscope. Raman scattering and photoluminescence spectra were obtained on a Zolix RTS2 micro-confocal Raman system with a 532 nm laser source. Raman laser energy was kept low to avoid heating effects. MoS_2 and Ag NPs/MoS_2 FETs were tested in a vacuum probe station at ~1$×10^{-2}$ Pa using a Keithley 2636B System SourceMeter for electrical measurements.

Supplementary Figures

Fig. S1 (a) Atomic snapshot from MD simulation of laser shock response of multilayer 2D materials on a rigid template. (b) Detailed view of fracture location in 2D materials after laser shock. SEM images of multilayer 2D materials on rigid template before (c) and after (d-e) laser shock. (e) SEM image after strain release showing broken parts at corner edges. The black stripe in post-impact SEM images results from 2D materials being pressed into the mold (Fig. S1(d)).

Fig. S2 (a) Atomic snapshots of Ag NPs/MoS_2 (cross-section) during LSICD at different time steps (0 ps, 40 ps, 60 ps, and 80 ps). Impactor speed: 0.25 km/s. (b-c) Time evolution of strain tensor XX and strain tensor YY in MoS_2 (top view).

Fig. S3 Schematic illustration of monolayer MoS_2 film growth process.

Fig. S4 (a) Optical image of monolayer MoS_2 films grown by CVD. (b) Optical image of monolayer MoS_2 films transferred to Ag nanoparticles. (c) High-resolution TEM image of MoS_2 films. (d) SAED pattern of MoS_2 films. (e-f) AFM images of monolayer MoS_2 films transferred to Ag nanoparticles. The CVD-grown monolayer MoS_2 used in experiments had triangle side lengths of ~60 m. HR-TEM images show three interplanar distances of 0.27 nm. Optical and AFM images of M@MO NPs/MoS_2 reveal the rough substrate, with clear distinction between areas with and without MoS_2.

Fig. S5 (a) AFM images of Ag NPs/MoS_2 formed by transferring monolayer MoS_2 onto Ag nanoparticles created by magnetron sputtering for 30 s and annealing. (c-d) AFM images of Ag NPs/MoS_2 after LSICD.

Fig. S6 (a) Intensities of characteristic Raman and PL peaks of MoS_2 under LSICD. (b) Distributions of squared electric field ($|E|^2/|E_0|^2$) at NPs/MoS_2 before and after LSICD under 532 nm excitation (top view).

Fig. S7 SEM images of Ag NPs/MoS_2 formed by transferring monolayer MoS_2 onto Ag nanoparticles created by magnetron sputtering for 15 s (a-d) and 45 s (e-h) and annealing, before and after LSICD. Scale bar: 500 nm. The transferred monolayer MoS_2 films are supported by nanoparticles like tents. After laser shock, nanoparticles deform due to impact force, and MoS_2 films bond to the three-dimensional nanoparticle surfaces forming core-shell-like structures.

Fig. S8 AFM images of Ag NPs/MoS_2 formed by transferring monolayer MoS_2 onto Ag nanoparticles created by magnetron sputtering for 15 s (a-d) and 45 s (e-h) and annealing, before and after LSICD. Scale bar: 500 nm. With increasing laser shock energy, nanoparticles exhibit varying degrees of deformation: particle height decreases, transverse size increases, and tops become flatter.

Fig. S9-11 Statistical area distributions of Ag nanoparticles formed by magnetron sputtering for 30 s (S9), 15 s (S10), and 45 s (S11) and annealing, before and after LSICD. With increasing laser shock energy, flattening of nanoparticles increases the area occupied by single nanoparticles and decreases interparticle spacing, consistent with AFM measurements.

Fig. S12 Raman (a-b) and photoluminescence (c-d) spectra of monolayer MoS_2 films on different-sized nanoparticles with laser shock, without and with MoS_2 participation. As deposition time increases, nanoparticle size and height increase, and the red shift of MoS_2 Raman and PL characteristic peaks also increases.

Fig. S13 (a) MD simulation results of Ag NPs/MoS_2 structure during laser shock. (b-c) SEM images of Ag NPs/MoS_2 structure before and after LSICD. (d) MD simulation results of single Ag NPs during laser shock. (e-f) SEM images of Ag NPs/MoS_2 structure before and after laser shock. Scale bar: 200 nm. Deformation in NPs/MoS_2 structures is larger with smaller surface curvature than in single nanoparticles during laser shock, resulting in much lower strain levels if laser shock is performed without MoS_2 participation.

Fig. S14 (a) Band structure of monolayer MoS_2 under different biaxial tensile strains. (b) Band structure of monolayer MoS_2 with different bending heights.

Fig. S15 (a) MoS_2 crystal structures with different bending heights for DFT calculations. (b) Band structure of monolayer MoS_2 with different bending heights. (c) Band gap, electron and hole effective masses of monolayer MoS_2 under different biaxial strains. (d) Band gap, electron and hole effective masses of monolayer MoS_2 with different bending heights. Analysis shows that both band gap and electron effective mass of MoS_2 decrease with increasing strain, whether from biaxial strain or bending, corresponding to PL spectral and electrical performance measurements.

Supplementary Tables

Table S1 Field-effect mobility and carrier concentration of different MoS_2 FETs.

Sample	Mobility ($cm^2V^{-1}s^{-1}$)	Carrier concentration ($\times10^{12}$ cm^{-2})
MoS_2	1.9	3.22
NPs/MoS_2	36.1	3.69
LSICD1 NPs/MoS_2	44.1	3.93
LSICD3 NPs/MoS_2	(slightly decreased)	(slightly decreased)

Table S2 Statistics of _{FE} for monolayer MoS_2 FETs grown by CVD from literature.

Configuration	Dielectric	_{FE} ($cm^2V^{-1}s^{-1}$)	Reference
Monolayer by MOCVD	Back-gated	~11.2	Various
Monolayer by CVD	Back-gated	Various	Various
Monolayer by Au-assisted exfoliation	Top-gated	Various	Various
M@MO NPs/MoS_2	Back-gated	44.1 (this work)	This work
LSICD M@MO NPs/MoS_2	Back-gated	44.1 (maximum)	This work

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