Assembly of 3D micro/nanostructures in advanced functional materials has important implications across broad areas of technology. Existing approaches are compatible, however, only with narrow classes of materials and/or 3D geometries. This paper introduces ideas for a form of Kirigami that allows precise, mechanically driven assembly of 3D mesostructures of diverse materials from 2D micro/nanomembranes with strategically designed geometries and patterns of cuts. Theoretical and experimental studies demonstrate applicability of the methods across length scales from macro to nano, in materials ranging from monocrystalline silicon to plastic, with levels of topographical complexity that significantly exceed those that can be achieved using other approaches. A broad set of examples includes 3D silicon mesostructures and hybrid nanomembrane–nanoribbon systems, including heterogeneous combinations with polymers and metals, with critical dimensions that range from 100 nm to 30 mm. A 3D mechanically tunable optical transmission window provides an application example of this Kirigami process, enabled by theoretically guided design.

Three-dimensional micro/nanostructures are of growing interest (1–10), motivated by their increasingly widespread applications in biomedical devices (11–13), energy storage systems (14–19), photonics and optoelectronics (20–24), microelectromechanical systems (MEMS) (25–27), metamaterials (21, 28–30), and electronic (33–35). Of the many methods for fabricating such structures, few are compatible with the highest-performance classes of electronic materials, such as monocrystalline inorganic semiconductors, and only a subset of these can operate at high speeds, across length scales, from centimeters to nanometers. For example, although approaches (36–39) that rely on self-actuating materials for programmable shape changes provide access to a wide range of 3D geometries, they apply only to certain types of materials [e.g., gels (36, 37), liquid crystal elastomers (39), and shape memory alloys (38)], generally not directly relevant to high-quality electronics, optoelectronics, or photonics. Techniques that exploit bending/folding of thin plates via the action of residual stresses or capillary effects are, by contrast, naturally compatible with these modern planar technologies, but they are currently most well developed only for certain classes of hollow polyhedral or cylindrical geometries (1, 10, 40–44). Other approaches (45, 46) rely on compressive buckling in narrow ribbons (i.e., structures with lateral aspect ratios of >5:1) or filaments to yield complex 3D structures, but of primary utility in open-network mesh type layouts. Attempts to apply this type of scheme to sheets/membranes (i.e., structures with lateral aspect ratios of <5:1) lead to “kink-induced” stress concentrations that cause mechanical fracture. The concepts of Kirigami, an ancient aesthetic pursuit, involve strategically configured arrays of cuts to guide buckling/folding processes in a manner that reduces such stresses, to enable broad and interesting classes of 3D structures, primarily in paper at centimeter and millimeter dimensions. Traditional means for defining these cuts and for performing the folds do not extend into the micro/nanoscale regime, nor do they work effectively with advanced materials, particularly brittle semiconductors. This paper introduces ideas for a form of Kirigami that can be used in these contexts. Here, precisely controlled compressive forces transform 2D micro/nanomembranes with lithographically defined geometries and patterns of cuts into 3D structures across length scales from macro to micro and nano, with levels of complexity and control that significantly exceed those that can be achieved with alternative methods. This Kirigami approach is...
different from conventional macroscopic analogs [e.g., including lattice Kirigami methods (47, 48) that solve the inverse problem of folding a flat plate into a complex targeted 3D configuration], where negligible deformations occur in the uncut regions of the folded structures and from recently reported microscale Kirigami methods that use 2D forms for stretchable conductors (49). The current approach is also fully compatible with previously reported schemes based on residual stresses and on buckling of filamentary ribbons. Demonstrations include a diverse set of structures formed using silicon nanomembranes, plates, and ribbons and heterogeneous combinations of them with micro/nanopatterned metal films and dielectrics. A mechanically tunable optical transmission window illustrates the extent to which theoretical modeling can be used as a design tool to create targeted geometries that offer adaptable shapes and desired modes of operation.

Results and Discussion
Assembly Concepts and Design Principles. Fig. 1 A–E and SI Appendix, Figs. S1 and S2, present examples of this type of Kirigami process for assembly of 3D mesostructures from corresponding 2D bilayers of nanomembranes of monocrystalline silicon [Si nanomembranes (NMs); 300 nm in thickness] and films of a photodefinable epoxy (SU8; 300 nm in thickness). Here, photolithography and etching define patterns of cuts in these structures to yield enhanced flexibility in certain orientations, at specific locations. Compressive forces imparted in the plane at selected points (anchors; red, in SI Appendix, Fig. S2) deform the systems into engineered 3D configurations via lateral buckling (50), using a concept similar to the one exploited in 3D filamentary networks (46). The left frame of Fig. 1A illustrates a simple case that includes five square regions connected
by narrow joints. Here, the outer squares attach to small rectangular anchors that adhere strongly to a biaxially prestrained elastomer substrate through covalent surface chemical bonding. All other regions release from the substrate via elimination of a sacrificial interface layer, as described in Methods. Relaxing the prestrain generates compressive stresses that induce these regions to buckle out of the plane, as shown in an intermediate state of assembly [results from finite-element analysis (FEA); Fig. 1A, Center Left]. The final configuration corresponds to that of a square cuboid, as given by the FEA result in Fig. 1A, Center Right and the scanning electron micrograph (SEM) (colorized; Fig. 1A, Right). The color indicates the maximum principal strains. Peak values remain well below the fracture thresholds (~2%) for the Si NMs, owing to the stress-reducing effects of the Kirigami cuts that form the narrow hinges between the sides. (The silicon is assumed to exhibit linear responses up to strains that approach the fracture point.) A failure criterion based on the maximum principal strain is adopted here for simplicity, in which the threshold is assumed to be independent of thickness, for membrane thicknesses down to ~100 nm. The dimensions of the anchor regions must be carefully selected to avoid delamination and surface buckling. In choosing patterns of cuts, the locations should (i) eliminate localized deformations that might occur otherwise and (ii) avoid any possible self-locking of different subcomponents during the compressive buckling. The cuts also play critical roles in defining the final 3D geometries. Fig. 1B shows an example in which a cross-cut pattern divides a large square into four smaller ones. Here, assembly forms a curved pyramidal mesostructure. Repeated implementation of such crosses (Fig. 1B) in the subsquares (i.e., the smaller squares) yields a fractal-inspired pattern of cuts (51) that divides the original square into interconnected small pieces with similar shapes and sizes. The resulting 3D structure adopts a highly complex configuration, where bending and twisting deformations localize at the joints defined by the cuts, with strains that remain below the fracture threshold. The image in Fig. 1C, Right and those in SI Appendix, Fig. S3 offer multiple viewing angles. Many other geometries are possible, including those with circular symmetry, as shown in Fig. 1D and E. The latter case has a well-defined chirality, set by the configuration of Kirigami cuts. In all cases examined in this paper, 2D precursors without carefully placed cuts tend to undergo sharp, localized deformations with associated stress concentrations (SI Appendix, Fig. S4) that lead to fracture. For unpatterned, circular 2D precursors, the maximum strains reach values that are nearly 4 times larger than those with Kirigami designs under otherwise similar conditions. Consequently, even at the largest level of prestrain (~35%) that can be accommodated without cuts, the corresponding maximum 3D extension is small (SI Appendix, Fig. S5A), as defined by the aspect ratio, \( \alpha = \frac{d_{\text{out-of-plane}}}{d_{\text{in-plane}}} \), where \( d_{\text{out-of-plane}} \) and \( d_{\text{in-plane}} \) denote the maximum out-of-plane dimension and in-plane extent, respectively. Specifically, the value of \( \alpha \) without Kirigami cuts (0.3) is nearly 2.5 times smaller than that achievable with cut geometries demonstrated in Fig. 1D and E and SI Appendix, Fig. S5B.

These concepts can be implemented across a broad range of lengths scales, with nearly any type of material, and in systems that include filamentary 3D networks and/or hierarchical layouts.

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**Fig. 2.** Computational results that highlight fundamental aspects of Kirigami designs in examples with purely radial and purely circumferential cuts. (A) Maximum material strain as a function of the square root of the compressive strain for membranes with \( t/L = 0.0011, w_{\text{cut}}/L = 0.044, \) and \( l_{\text{cut}}/L = 1.68 \) (for circumferential cuts) or 0.76 (for radial cuts) and the associated strain distributions. (B) Maximum material strain as a function of the dimensionless thickness for membranes with \( w_{\text{cut}}/L = 0.044 \) and \( l_{\text{cut}}/L = 1.68 \) (for circumferential cuts) or 0.76 (for radial cuts), under a compressive strain of 50%, and the associated strain distributions. (C) Maximum material strain as a function of the dimensionless widths of Kirigami cuts in membranes with \( t/L = 0.0011 \) and \( l_{\text{cut}}/L = 1.68 \) (for circumferential cuts) or 0.76 (for radial cuts), under a compressive strain of 50%, and the associated strain distributions. In all cases, the color in the FEA results corresponds to the magnitude of the maximum principal strain.
An example of a polymer structure appears in Fig. 1F, where a Kirigami-based circular pattern connects to serpentine ribbons organized in a circular, closed form to yield an elaborate 3D mesostructure that resembles a jellyfish (Fig. 1F). The buckling begins with the ribbons at the periphery, followed by the eight straight ribbons in the central circular membrane, leading to a 3D configuration with multiple levels. Fig. 1G shows structures that have characteristic dimensions ranging from 100 nm (thickness of the bare Si NM in the example in Fig. 1G, Left) to ~30 mm (lateral dimensions of the 3D plastic sheet in Fig. 1G, Right), each overlaid with results from FEA simulations. Throughout all examined geometries, materials, and length scales, experimental results exhibit excellent quantitative agreement with FEA predictions, thereby establishing computation as a means for rapid design iterations, as demonstrated subsequently in the engineering of a tunable optical device. The 2D precursors of all examples in Fig. 1 are in SI Appendix, Fig. S6.

FEA can also reveal the dependence of the maximum principal strains on the prestrain in the elastomer substrate, as a function of geometric parameters related to the membrane structure and Kirigami cuts, as shown in Fig. 2. During lateral buckling, the 2D precursors undergo complex out-of-plane bending deformations, with associated spatially dependent variations in the curvature. The maximum strains occur at locations with highest change in curvature; these locations typically remain constant throughout the buckling process. Quantitative analyses of representative Kirigami patterns (Fig. 2A and B) with purely radial and circumferential cuts show that the maximum strains ($e_{\text{max-material}}$) are proportional to the normalized thickness for a single-layer Si membrane, i.e., $t/L$, where $L$ measures the overall dimension of the 2D precursor (e.g., the radii of the circular geometries in Fig. 2), and the square root of the compressive strain ($e_{\text{comp}}$) applied to the 2D precursor, where $e_{\text{comp}} = e_{\text{pre}}/(1 + e_{\text{pre}}).$ This scaling, i.e., $e_{\text{max-material}} \propto \sqrt{e_{\text{comp}}/L}$, also applies to the other 3D structures examined here (SI Appendix, Figs. S7 and S8), including those with various Kirigami cuts (e.g., antisymmetric cuts in a serpentine configuration or with combinations of radial and circumferential cuts) as well as uniaxial and biaxial prestrains in the elastomer substrate. Although the effect of the widths of the cuts ($w_{\text{cut}}$) cannot be captured with a simple scaling law, the qualitative dependence consistently involves a decrease in the maximum strain with an increase in $w_{\text{cut}}$ (e.g., Fig. 2C and SI Appendix, Fig. S9). This trend further highlights the critical, enabling role of Kirigami concepts in this approach to 3D assembly. The effect of cut length is even more complicated, partly because this parameter significantly affects the nature of deformation modes in a qualitative sense, as shown in the results of Fig. 2D and SI Appendix, Fig. S10. These calculations indicate, in fact, that the lengths must be sufficiently large to avoid stress concentrations (e.g., in Fig. 2D and SI Appendix, Fig. S10 B and D). These qualitative and quantitative rules, together with the high accuracy in the FEA, provide a strong foundation for systematic, engineering design.

### Three-Dimensional Mesostructures in Membranes and in Membranes/Ribbons, with Diverse Geometries

Fig. 3 presents a collection of experimental results and FEA predictions for dozens of 3D structures formed with Si NM/polymer precursors (both layers, 300 nm in thickness). The nature of the Kirigami cuts in the 2D precursors provides the basis for a classification scheme: (i) membranes without any cuts, (ii) membranes with symmetric cuts, (iii) membranes with antisymmetric cuts, and (iv) membranes with asymmetric cuts. Without cuts (Fig. 3A), the bonding locations, the overall shapes, and/or the addition of holes must be selected carefully to avoid the type of stress concentrations mentioned previously. These considerations impose tight restrictions on the 3D geometries that are possible. Kirigami cuts avoid these constraints, such that even for a given overall membrane shape and set of bonding locations, as shown in Fig. 3 B–D (except for the last two designs in Fig. 3B), a rich range of 3D topologies can be realized. For circular shapes, cuts along the radial or circumferential directions serve as the basis for

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**Fig. 3.** Experimental and computational studies of various 3D silicon/polymer mesostructures and their classification according to geometric characteristics of the Kirigami cuts. (A–D) Two-dimensional precursors, SEM images, and FEA predictions for 27 3D mesostructures formed with precursor patterns without any cuts (A), with symmetric cuts (B), with antisymmetric cuts (C), and with asymmetric cuts (D). (Scale bars, 200 μm.)
symmetric Kirigami patterns. Cuts with serpentine configurations provide antisymmetric examples. Fig. 3 B–D demonstrates how the orientations of the cuts dictate the assembly process. Including additional bonding locations at the inner regions of the precursors further enhances the spatial variations in the modes of deformation. The last two cases in Fig. 3B provide examples where the positions of holes help to avoid self-contact of the membrane during the 3D assembly.

As demonstrated in Fig. 1, these Kirigami concepts can naturally include ribbon-shaped precursors (46), to yield complex 3D structures, including those with multiple levels of buckling. Fig. 4A and SI Appendix, Fig. S11 present an additional 12 examples. For the first 5 (i.e., 4 in Fig. 4A, Left and 1 in Fig. 4A, Top Right), buckling occurs first in the membranes; these motions then induce compression in the supported ribbons, leading to subsequent buckling processes. The first structure represents an exception, where the untethered ends of the ribbons allow freedom of motion, with little intrinsic deformation throughout the assembly process. Here, the ribbons simply follow the supporting membranes, to final orientations that are almost perpendicular to the plane of substrate. In such designs, the ribbons have negligible effects on the 3D configurations of the membranes. The three examples in Fig. 4A, Upper Middle, Lower Middle, and Bottom Right represent cases where buckled ribbons play an essential role in the assembly, via their selective bonding to the substrate, to form a first level of construction. Membranes raised upward by these ribbons form a second level. The comparatively high stiffnesses of the membranes affect deformations of the supporting ribbons, as evidenced by their rotation with respect to the corresponding length directions.

Using the membrane and/or hybrid membrane–ribbon configurations as building blocks, arrays or nested architectures can be formed, as shown in Fig. 4B. Fig. 4B, Top Left involves an evenly spaced, triangular collection of double-level membrane–ribbon mesostructures (in Fig. 4A, 1), with five unit cells along each edge. Fig. 4B, Top Center shows a double-level architecture that resembles a “crown,” achieved with a 2D precursor illustrated in SI Appendix, Fig. S12A. Images at multiple view angles (SI Appendix, Fig. S12B) highlight the geometrical complexity. The third example (Fig. 4B, Top Right) represents a triangular array of membrane–ribbon mesostructures with raised circular disks that adopt nearly planar shapes, owing to their relatively large stiffnesses. Fig. 4B, Bottom corresponds to a mixed array composed of six membrane mesostructures without any cuts (in two different configurations), another four membrane mesostructures with antisymmetric cuts (with opposite chirality), and six hybrid membrane–ribbon mesostructures (in two different configurations). All of these results agree well with FEA predictions.

Three-Dimensional Mesostructures in Different Materials and Geometries and with Supported Micro/Nanopatterns. The physical nature of the Kirigami assembly process allows immediate application across a broad range of material types. Fig. 5 A and B presents examples, including those formed using both polymers and metals (Au), with membrane or hybrid membrane–ribbon configurations. Corresponding 2D precursors appear in SI Appendix, Fig. S13. Additional examples are in SI Appendix, Figs. S14 and S15. All of the structures in Fig. 3 reproduced in millimeter-scale plastic models are shown in SI Appendix, Fig. S16. Heterogeneous combinations of different materials are also possible, with two examples constructed with polymers and silicon in Fig. 5C. Furthermore, buckled membranes can be exploited as 3D platforms (with curved surfaces), for micro/nanopatterns of other materials, as demonstrated in polymer–silicon (Fig. 5D) and polymer–metal (Fig. 5E). In particular, Fig. 5D shows a square array of silicon nanodisks (~200 nm in thickness, ~200 nm in diameter) formed by soft lithography (SI Appendix, Fig. S17) on a 2D polymer precursor that transforms into a 3D structure with three untethered ribbons. Fig. 5E, Left corresponds to a square array of Au microdisks (~50 nm in thickness, ~5 μm in diameter) distributed across the area of a 2D precursor. The array follows the curved surfaces of the 3D architecture that forms by Kirigami assembly. Fig. 5E, Center Left involves a spiral pattern of Au microdisks (~50 nm in thickness, ~10 μm in diameter), consisting of eight unevenly spaced branches (each with ~20 microdisks) that adapt to the antisymmetric cuts of the supporting polymer membrane. The assembly process projects these patterns onto four petal-shaped structures, thereby placing them in a 3D configuration. Fig. 5E, Center Right is an example with the configuration of a square space-filling tree (with fifth order) as a complex Au network (with 5 μm width for each wire) that is then transformed into a 3D spatial form. Fig. 5E, Right corresponds to third-order fractal Cayley tree (52) microstructures (Au, ~5 μm width for each wire) on a 3D membrane with four identical parts. Similar hybrid architectures with first- and second-order Cayley tree configurations are in SI Appendix, Fig. S18 A and B. Two additional examples with an array of Au microdisks assembled in a 3D polymer layout appear in SI Appendix, Fig. S18 C and D.
Three-dimensional mesostructures of membrane and hybrid membrane–ribbon configurations with various material compositions. (A and B) Experimental images and overlaid FEA predictions of 3D mesostructures made of polymer and metal (Au). (Scale bars, 200 μm.) (C) Three-dimensional membrane mesostructures with heterogeneous combinations of silicon and polymer. (Scale bars, 200 μm.) (D) Three-dimensional membrane mesostructures consisting of a polymer membrane with a patterned array of silicon nanodisks on the surface, with a magnified view in Inset. (Scale bar, 200 μm; in Inset, 1 μm.) (E) Related 3D mesostructures with patterned arrays of Au microstructures. (Scale bars, 200 μm.)

**A Mechanically Tunable Optical Transmission Window.** The ability to dynamically and reversibly change the 3D shapes represents an important functional option associated with all of the mesostructures described previously. Fig. 6 provides a device demonstration in the form of a mechanically tunable optical transmission window (with a 3 × 3 array of shutter-like structures). A representative element (Fig. 6A, Top) consists of a reflective membrane (~50-nm-thick Au on an ~8-μm-thick layer of SU8) in a square shape, with two nonpenetrating Kirigami cuts. The two ends bond to a transparent, uniaxially prestrained elastomer substrate at rectangular anchors. Compressive buckling of the central ribbon forces rotational motion of the membrane, as shown by both experiment and FEA results in an intermediate state of assembly (Fig. 6A, Middle). As the membranes rotate upward, they block a decreasing fraction of normally incident light. FEA simulations identify a level of prestrain (~90%) that maximizes the range over which the transmittance can be tuned in this fashion. Here, the membranes, in their fully rotated state, are nearly perpendicular to the plane of the elastomer substrate (Fig. 6A, Bottom). The optical micrographs and FEA images show excellent agreement for both the intermediate and final states of assembly. Fig. 6B presents a variation of this design, in which short segments (~17%) in the centers of the ribbons have increased thicknesses (corresponding to a double layer of ~50-nm-thick Au and ~23-μm-thick SU8). This structure offers greatly enhanced rotations for a given level of strain, as a consequence of the reduced curvature in the thickened regions of the ribbons. Here, the thin segments accommodate an increased level of deformation. Experiment and FEA results (Fig. 6B, Middle and Bottom) illustrate this characteristic. Consequently, a comparatively low level of prestrain (~66%) actuates the full, 90° rotation of the membrane.

Fig. 6C shows measurements and modeling results for the dependence of optical transmittance on the uniaxial tensile strain ($\varepsilon_{\text{appl}}$) applied to the elastomeric substrate for these two different designs. The illumination spot (diameter ~1.0 mm) covers the entire active area throughout the experiments. In both cases, the optical transmittance decreases monotonically from ~97% in the zero-strain state to ~22% at the critical state ($\varepsilon_{\text{appl}} = 90%$ and 66% for the two designs). Linear fits of the data yield metrics (i.e., the slopes of the fitted lines) for the sensitivity of the transmittance to strain, indicating ~40% (relative) increase in sensitivity (~1.26 vs. ~0.90) enabled by the thickness-modulated design (Fig. 6B). This result indicates the potential of engineering variations in thickness to achieve desired mechanical behaviors. The measured optical properties agree reasonably well with modeling that involves calculation of the optical transmittance associated with 3D geometries predicted by FEA. Three representative states of the nonuniform design appear in Fig. 6C, Right. Effects of fatigue do not appear in the optical devices after they are stretched to ~65% strain repetitively at a frequency of ~0.04 Hz for ~150 cycles (SI Appendix, Fig. S19).

**Conclusions**

The Kirigami-inspired concepts, design principles, and micro/nanofabrication strategies reported here provide immediate access to diverse 3D membrane architectures with broad-ranging critical dimensions and material compositions, including high-performance semiconductor nanomaterials. The resulting engineering options in functional 3D mesostructures have sweeping implications for
The calculations used linear buckling analyses of 2D substrates with Young’s modulus $E = 4.02$ GPa and Poisson’s ratio $\nu = 0.27$ for SU8, $0.49$ for substrate; $0.44$ for gold; and $0.42$ for SU8. Transfer of 3D mesostructures in polymer membranes started with thermal oxidation to form a layer of $\text{SiO}_2$ (500 nm in thickness) on a silicon wafer. Next, spin casting and photolithography formed 2D precursors of SU8 (4 $\mu$m in thickness) on the $\text{SiO}_2$ wafer. Next, spin casting and photolithography formed 2D precursors with their top sides facing up, supported by the PVA. Exposing these precursors and a thin silicone elastomer (Dragon Skin, Smooth-On, 0.5 mm in thickness) to UV-induced ozone (UVO) yielded hydroxyl termination on their surfaces. A schematic illustration of steps is provided in SI Appendix, Fig. S20.

Preparation of 3D Si NM (100 nm in thickness) mesostructures involved defining 2D precursors on an SOI wafer (100-nm thicknesses of top silicon) and then following the procedures described above, except without the addition of SU8.

Preparation of 3D mesostructures in polymer membranes started with thermal oxidation to form a layer of $\text{SiO}_2$ (500 nm in thickness) on a silicon wafer. Next, spin casting and photolithography formed 2D precursors of SU8 (4 $\mu$m in thickness) on the $\text{SiO}_2$. Immersion in HF removed the $\text{SiO}_2$ film and then following the procedures described above, except without the addition of SU8.

Fabrication Methods for Silicon, Metals, Polymers, and Combinations of Them.

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Preparation of 3D mesostuctures in both silicon and polymer membranes began with spin coating a layer of photoreist (AZ 5214) on a silicon wafer (300-nm thickness of top silicon). Photolithography and RIE etching defined 2D patterns in the top silicon. Next, spin casting and photolithographic patterning formed a thin layer (4 µm in thickness) of SU8, in a distinct geometry spanning both the silicon and other regions. The remaining steps followed the procedures for 3D SU8 mesostructures described above. A schematic illustration is in SI Appendix, Fig. S22A.

Preparation of 3D mesostructures of SU with arrays of silicon nanodisks began with spin coating of a thin layer (200 nm in thickness) of SU8 on an SOI wafer (200-nm thickness of top silicon). Soft imprint lithography using a mold of PDMS with relief in the geometry of cylinders (period 300 nm, diameter 200 nm, height 200 nm) defined corresponding relief in the SU8. RIE etching of the residual layer of SU8 formed isolated disks of SU8 that served as masks for inductively coupled plasma reactive ion etching (Surface Technology Systems (STS)) to define arrays of silicon nanodisks in the top silicon layer. RIE eliminated the remaining SU8. Next, spin casting and photolithography defined patterns of SU8 (4 µm in thickness). The remaining steps followed the procedures for 3D SU8 structures described above. A schematic illustration of steps is in SI Appendix, Fig. S22B.

Preparation of 3D mesostructures in metal and polymer hybrid membranes began with thermal oxidation to form a layer of SiO₂ (500 nm in thickness) on a silicon wafer. Photolithography, electron beam evaporation, and liftoff defined patterns of Cr (5 nm in thickness) and Au (50 nm in thickness) on the SiO₂. Spin casting formed an adhesion-promoting layer (Omnicoat; Microchemicals, 30 nm in thickness) for spin casting and photolithographic patterning of a thin (4 µm in thickness) layer of SU8 in a geometry matched to the Cr/Au. RIE etching removed the exposed regions of the adhesion-promoting layer. The remaining steps followed the procedures for 3D SU8 structures described above. A schematic illustration of steps is in SI Appendix, Fig. S23.

Preparation of mechanically tunable optical transmission windows with uniform thickness followed steps similar to those for making 3D structures in hybrid membranes of metal and polymer, except that SU8 with Al₂O₃ thickness was used. Preparation of related structures with thickened regions involved photolithographic patterning of an additional layer of SU8 (15 µm in thickness).

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