Soft Electronics



Materials and Structures toward Soft Electronics

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Soft electronics are intensively studied as the integration of electronics with dynamic nonplanar surfaces has become necessary. Here, a discussion of the strategies in materials innovation and structural design to build soft electronic devices and systems is provided. For each strategy, the presentation focuses on the fundamental materials science and mechanics, and example device applications are highlighted where possible. Finally, perspectives on the key challenges and future directions of this field are presented.

1. Introduction

Nonplanar surfaces, either static (e.g., complex shaped objects) or dynamic (e.g., biology), are prevalent in nature. Soft electronics that allow interfacing with nonplanar surfaces significantly expand the capabilities of conventional rigid electronics in sensing, monitoring, diagnosing, and potentially intervening functions.^[1-6] Nonplanar surfaces can be divided into two categories: developable and nondevelopable.^[7,8] Developable surfaces can be flattened onto a plane without stretching or compression.^[7] They have zero Gaussian curvature and are characterized by only bending in one direction at a time. Examples of developable surfaces include cylindrical and conical surfaces. Surfaces that do not satisfy this criterion are nondevelopable surfaces.^[8] Examples include spherical surfaces and any curvilinear surfaces (such as the human body). By definition, flexible devices can only conform to developable surfaces. To integrate seamlessly with a general nonplanar surface, stretchability is required.^[9] A series of review articles have already covered flexible electronics



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very well,^[10–14] and therefore this article will mainly focus on the stretchable aspect of soft electronic devices.

The implications of soft electronics integrating with nonplanar objects are multifold. First, the intimate contact between the device and the nonplanar object will allow high-quality data to be collected.^[15] With rigid electronics, air gaps at the interface between the device and the object reduce the contact area, and can

potentially introduce noise and artifacts, which compromise signal quality.^[5] Second, foldable, low-profile devices can enable mobile and distributed sensing, which hold great promise for Internet-of-Things technology.^[16] Finally, in the area of medical devices, which is probably the major driving force of this field at present, soft electronics have similar mechanical properties and thus cause minimal irritation to the human skin, which can be a key enabling technology for continuous healthcare.^[17,18]

Two strategies can be applied to achieve stretchability in electronics: 1) materials innovation, by developing novel materials that are stretchable in single or aggregated forms; 2) structural design, by making nonstretchable materials into specific structures that can absorb the applied strain without fracturing, e.g., by inducing mixed modes of mechanical deformations.^[9] In this review, we summarize state-of-the-art advances in both strategies, covering a broad range of topics (**Figure 1**), including hydrogels, liquid metals, conductive polymers, and nanomaterials for the material approach, and waves/wrinkles, "island-bridges," textiles, origami, kirigami, cracks, and interlocks for the structural approach. We provide outlooks on the challenges in the field and possible future research directions at the end of this review.

2. Materials for Soft Electronics

The most intuitive approach for soft electronics is to exploit intrinsically soft materials as building blocks. Common building blocks that are necessary for conventional rigid electronic devices include conductors, dielectrics, and most importantly, functional materials, e.g., semiconductors. The intrinsically soft counterparts of these building blocks have been identified and largely developed recently. Additionally, attractive features that are absent in conventional rigid electronics are emerging in their soft counterparts, such as toughness, self-healing, and stimuli responsibility, as discussed in the following.

2.1. Hydrogels

Hydrogels are hydrophilic polymeric networks with 3D microstructures.^[17,19] Since their first report in the 1960s,^[20]



hydrogels have been widely used as biomaterials in tissue engineering,^[21] synthetic extracellular matrices,^[22] cell culture,^[23] and drug delivery.^[24] Due to their excellent physicochemical properties such as super stretchability, self-healing, biocompatibility, stimuli-responsibility, and outstanding conductivity given by innovations in polymer chemistry, composite physics, and micro- and nanofabrication, new applications of hydrogels in soft electronics have emerged (**Figure 2**).^[25–32]

Traditional hydrogel synthesis methods include chemical crosslinking by covalent bonds and ionic interactions, and physical crosslinking by entanglement. The chemically crosslinked hydrogels normally exhibit excellent mechanical properties including toughness, stiffness, and strength.^[33,34] On the other hand, the strong and irreversible covalent bonds usually lead to slow stimuli response,^[35] limited stretchability,^[36] and poor selfhealing properties.^[37] The physical crosslinking method such as by hydrogen bonds are much easier to break and rebuild than chemical bonds,^[38] and thus provides an effective approach for the self-healing.^[39,40] Physically crosslinked hydrogels typically have low mechanical properties, which present a challenge for their applications in soft electronics.^[33]

Novel strategies including nanocomposites,^[41-44] double networks,^[45–48] and sliding crosslinking^[49,50] have been developed to tailor the mechanical properties of the hydrogels. By adding silicate nanoparticles to the covalently crosslinkable poly(ethylene) network, highly stretchable (1500%) and mechanically robust hydrogels have been demonstrated.^[51] In this case, the covalent crosslinking restricts the polymer chain movements, resulting in great elastic properties. At the same time, the physical crosslinking between the silicate nanoparticles and the polymer chains forms a viscoelastic network that enhances the hydrogel elongation. In another example, a double network hydrogel that combines ionic and chemical crosslinking can be stretched to over 2000% with a fracture energies of 9000 J $\,m^{-2}.^{[45]}$ When this hydrogel is stretched, the ionically crosslinked alginate network is ruptured to substantially dissipate the strain energy, and the covalently crosslinked polyacrylamide network allows the hydrogel to return to its original configuration. This double network method makes the hydrogel highly resistive to defects. The hydrogel maintains a stretchability of 1700% with a notch (Figure 2a). By introducing slide-ring polyrotaxane crosslinkers and ionic groups into the polymer network,^[49] an extremely stretchable and tough hydrogel was demonstrated. In this case, the crosslinked α -cyclodextrin molecules can move along the polyethylene glycol chains and the ionic groups help the extension of the polyrotaxane crosslinkers in the polymer network, leading to excellent mechanical properties of the hydrogel.

Self-healing allows the devices to recover their functions/ morphologies after damage, which are highly imperative for their practical applications with improved safety and extended lifetime.^[52] Self-healing capabilities in hydrogels (Figure 2b) are commonly realized by dynamic crosslinkages^[39,53] or noncovalent interactions.^[54,55] The dynamic crosslinkages mostly require additional interventions/stimuli (e.g., heat, pH, or light) or healing agents, to trigger the reversible selfhealing processes.^[56] The noncovalent interactions enable autonomously self-healing, where the network rebuilds by a variety of driving forces, such as host–guest recognition, hydrogen bonding, π – π stacking, metal–ligand interaction, or





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electrostatic interactions.^[34,57–59] For example, the self-healing poly(ι -glutamic acid) hydrogel synthesized via the host–guest interaction can recover their initial strength upon injury in 1 min without any additional stimuli.^[56]

Functional devices are typically composed of multilayers of materials. Large bonding strength between the hydrogel and the other materials, such as elastomers, fabrics, and soft tissues, becomes essential.^[60–64] Figure 2c shows bonding







Figure 1. Summary of the review. Two strategies can be applied to realize soft electronics: materials innovation, by developing novel materials that are stretchable in single or aggregated forms; structural design, by making nonstretchable materials into specific structures that absorbs externally applied strain without fracturing. Advanced materials include hydrogels, liquid metals, conductive polymers, and nanomaterials. Structural designs commonly used in soft electronics include waves/wrinkles, islands–bridges, textiles, origami, kirigami, cracks, and interlocks. These material and structural strategies provide soft electronic devices with an elegant balance of electronic functionality and mechanical deformability. Hydrogel image: Reproduced with permission.^[63] Copyright 2016, Nature Publishing Group. Liquid metal image: Reproduced with permission.^[144] Copyright 2013, Wiley-VCH. Polymer image: Reproduced with permission.^[183] Copyright 2012, Nature Publishing Group. Nanomaterial image: Reproduced with permission.^[164] Copyright 2014, American Chemical Society. Kirigami image: Reproduced with permission.^[541] Copyright 2017, Elsevier. Origami image: Reproduced with permission.^[541] Copyright 2017, Elsevier. Textile image: Reproduced with permission.^[608] Copyright 2016, Nature Publishing Group. Island-bridge image: Reproduced with permission.^[641] Copyright 2017, Elsevier. Textile image: Reproduced with permission.^[608] Copyright 2016, Nature Publishing Group. Island-bridge image: Reproduced with permission.^[641] Copyright 2017, Elsevier. Textile image: Reproduced with permission.^[608] Copyright 2016, Nature Publishing Group. Island-bridge image: Reproduced with permission.^[641] Copyright 2017, Elsevier. Textile image: Reproduced with permission.^[608] Copyright 2016, Nature Publishing Group. Island-bridge image: Reproduced with permission.^[641] Copyright 2017, Elsevier. Textile image: Reproduced with permission.^[608] Copyright 2016, Nature Publishing Group. Island-bridge ima

of hydrogels to a smooth glass substrate with an interfacial toughness value of over 1000 J m⁻², which is even higher than that between tendon and bone in human body (800 J m⁻²). The strong bonding results from the covalent anchorage between the hydrogel and the glass.^[63] Similar strong bonding between hydrogels and elastomers was achieved by covalently grafting the benzophenone-modified elastomers to a pre-shaped hydrogel.^[64] The use of benzophenone alleviates oxygen inhibition effects and activates elastomer surfaces. The use

case includes an extremely stretchable hydrogel–elastomer functional circuit (Figure 2d). This strategy is feasible for various kinds of common elastomers and hydrogels. The strong adhesion of hydrogels on the human skin can be achieved by the use of nanocomposite networks due to the induced viscoelastic properties by nanoparticles.^[51,65]

Because of their water-rich nature and structural similarity to the natural extracellular matrices, hydrogels are highly biocompatible and best for biomimetic applications.^[38,66,67] They can







Figure 2. Hydrogels for soft electronics. a) Highly stretchable and notch-insensitive hydrogels from polymers with ionically and covalently crosslinked networks. Reproduced with permission.^[45] Copyright 2012, Nature Publishing Group. b) Self-healing of the color alternating hydrogels with the colorant diffusing for 1 h (left) and 36 h (right). Reproduced with permission.^[56] Copyright 2015, American Chemical Society. c) Tough bonding of hydrogels with a smooth surface. Reproduced with permission.^[63] Copyright 2016, Nature Publishing Group. d) Hydrogel circuit under severe deformation, which benefits from the tough bonding at the hydrogel–elastomer interface. Reproduced with permission.^[64] Copyright 2016, Nature Publishing Group. e) An artificial electronic organ based on stacked polyacrylamide hydrogel compositions by mimicking the anatomy of the electric eel. Reproduced with permission.^[71] Copyright 2017, Nature Publishing Group. f) Soft and invisible hydrogel conductors as electrodes. Reproduced with permission.^[82] Copyright 2016, American Association for the Advancement of Science. h) 4D biomimetic printing hydrogels with time-lapse sequences of the flowers during the swelling process. Reproduced with permission.^[95] Copyright 2016, Nature Publishing Group.

be degraded in controlled modes and rates through designed enzymatic hydrolysis, ester hydrolysis, photolytic cleavage, or a combination of them.^[68] Hydrogels can resist the formation of foreign-body reaction-induced collagenous capsule after subcutaneous implantation.^[69] Stretchable living sensors and interactive genetic circuits by integrating genetically engineered cells with hydrogel–elastomer hybrids have been demonstrated.^[70] Recently, artificial electrocytes based on the stacked polyacrylamide hydrogels have been demonstrated to mimic the anatomy of electric eels (Figure 2e).^[71] In this case, gels with high and low salinity, and cation and anion selectivity mimicked the four major components of an electrocyte: intracellular and extracellular salt compartments, and anterior and posterior ion selective membranes, respectively. Each of the gel cells can generate an open-circuit voltage of 130–185 mV and a power density of 27 mW m⁻² upon contact activation. A total open-circuit voltage of 110 V was demonstrated by connecting 2449 gel cells in series. The artificial electrocyte is soft and biocompatible, suggesting the possibility of integrating power sources inside biological systems.

Transparency not only allows potential interaction of the soft electronics with visible light, but also enables the stealthy or visually customizable on-body electronics. Figure 2f shows a hydraulic actuator made of soft and invisible hydrogels. The hydrogel actuator can easily capture a goldfish due to the transparent nature of the hydrogel.^[72] In addition, the hydrogel

(more than 90 wt% is water) is ultrasonically undetectable in the water owing to the very low acoustic impedance mismatch between the water and hydrogel, which indicates their potential applications in the military.^[72]

Conductive hydrogels can be obtained by mixing hydrogels with different conductors such as carbon nanotubes, graphene, metal nanowires, polyaniline (PAni), p-poly(3,4ethylenedioxythiophene)-polystyrene sulfonate (PEDOT:PSS), and ionic solutions.^[52,73,74] Semitransparent PAni-containing conductive hydrogels can be used as the electrode for stretchable (150%) supercapacitors with superior charge-discharge stability of up to 35 000 cycles at a high current density of 8 A g^{-1.[75]} By dissolving acrylamide (AAm) monomer and NaCl in deionized water followed by crosslinking AAm using N,N-methylenebisacrylamide (MBAA), the hydrogel-based ionic conductor is highly conductive, transparent, and extraordinary stretchable,[76] which enables various interesting functions for soft electronics, such as stretchable ionic touch panels,[77] transparent loudspeakers,[76] artificial skins,[78] and electroluminescence devices.^[79–81] Figure 2g shows a representative electroluminescent display with giant stretchability using the ionically conductive hydrogel as the electrodes.^[82]

One drawback of the conductive hydrogels is their environmental instability since their primary constituent is water, which can easily evaporate over time, causing the hydrogels to dry and stiffen. To overcome this challenge, organogels were reported by crosslinking PEDOT:PSS with polyacrylamide in ethylene glycol solution.^[83] Because of the much lower vapor pressure (0.06 mmHg at 20 °C) and much higher boiling point (197 °C) of ethylene glycol compared with those (17.3 mmHg at 20 °C and a boiling point of 100 °C) of water, the organogels are more stable. Other properties of hydrogels, such as stimuli response,^[84,85] antibiosis,^[86] shape memory,^[30] photochromic,^[87] and freeze tolerance,^[88] empower them a myriad of potential applications for soft electronics.

A series of techniques have been employed for programmable and controllable deposition of hydrogels, such as self-assembly, microfabrication, inkjet printing, and 3D printing.^[17,33,89–94] Specifically, hydrogels with sophisticated structures in a predictable manner by 4D printing have been demonstrated (Figure 2h).^[95] The preprinted hydrogel layouts were transformed into dynamic, complex 3D shapes when swelled in water owning to the encoded anisotropic stress in the hydrogel filaments containing aligned cellulose fibrils.^[95] DNA sequence-directed shape change of photopatternable hydrogels via high-degree swelling has also been reported.^[96] The DNA-crosslinked hydrogels can have a 100-fold volumetric expansion induced by the binding of specific DNA molecules and can therefore undergo different shape changes in response to different DNA sequences. The spatiotemporal control of hydrogel geometry provides a powerful approach for the design and fabrications of soft robotics.

2.2. Liquid Metals

Liquid metals provide the best combination of conductivity and deformability owning to their intrinsic nature.^[97,98] There are five known metallic elements that are in liquid phase at or near room temperature: francium (Fr), cesium (Cs), rubidium (Rb),

mercury (Hg), and gallium (Ga). Among them, Fr is radioactive, and Rb and Cs are both violently reactive with air, rendering them unsuitable for common device applications.^[99,100] With a high electrical conductivity of 1.04×10^6 S m⁻¹, Hg has been extensively utilized for electronic devices.^[101,102] However, Hg is toxic and has a very high surface tension (487 mN m^{-1}), which makes it difficult to process and be compatible with other materials.^[103-105] Compared with Hg, Ga and its alloys such as EGaIn (75% gallium and 25% indium) and Galinstan (68.5% gallium, 21.5% indium, and 10% tin) are much less toxic^[106,107] and have high electrical conductivities of $(3.4-6.7) \times 10^6$ S m^{-1.[108]} Additionally, Ga and its alloys have negligible vapor pressure,^[109] rendering them much safer than Hg. More importantly, Gabased alloys can form an atomically thin oxide layer (Ga₂O₃) on their surfaces upon contacting with air.[110-112] The oxide layer can provide mechanical stability and ease of patterning by simple methods.^[108,113] Thus, Ga and its alloys are the focus in this review and are not distinguished in the discussions because of their very similar properties.

The liquid metals are typically embedded in fluidic channels of elastomers such as Ecoflex, poly(dimethylsiloxane) (PDMS), polyacrylates, and block copolymer elastomers to form an intrinsically stretchable conductor.[114-117] The liquid metals do not add any mechanical loading to the carrying elastomers due to their fluidity (Figure 3a). Thus, mechanical properties of the resulting conductor mainly depend on the carrier elastomeric matrix.[118-120] Ultrastretchable (600%) conductors by combining the liquid metal and highly stretchable triblock copolymer gels have been demonstrated (Figure 3b).^[121] The liquid metal can be injected into hollow fibers to enable mass production of conductive fibers with robust stretchability (700%), negligible stress-strain cyclic hysteresis, and metallic conductivity.^[114] Figure 3c compares the conductivity at the maximum strain of various stretchable conductors.^[97] Liquid metals are the only material in the upper-right quadrant of this plot, indicating its combined high conductivity and stretchability.

Self-healing of the conductors can be achieved by combining the liquid metal with the self-healing carrying elastomer. The oxide thin layer of the liquid metal can hold the metal flush at the damaged interface and allow merging again when contacting with each other (Figure 3d).^[122–124] Another approach to selfhealing conductors relies on the triggered release and transport of microencapsulated liquid metals to the sites of damage.^[125] Although there are very few studies on the biocompatibility of the liquid metals, gallium salts have been approved by the United States Food and Drug Administration as magnetic resonance imaging (MRI) contrast agents.^[97] Drug delivery composed of light-driven shape-transforming liquid metals have been reported, indicating the biocompatibility of liquid metals.^[126,127]

Many electronic devices generate excessive heat during operation. Accumulation of heat can shorten device lifetime and lead to catastrophic failure. Especially for the soft electronics, elastomers are known to be poor thermal conductors, so heat dissipation of soft electronics becomes critical.^[128] Current thermal management techniques in rigid electronics include heat sinks, thermoelectric coolers, forced air systems and fans, and heat pipes, which are not suitable for the soft electronics due to their rigidity, bulky sizes, and structural complexities.





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Figure 3. Liquid metals for soft electronics. a) Comparison of Young's modulus among different materials. Reproduced with permission.^[97] Copyright 2017, Wiley-VCH. b) Ultrastretchable conductors composed of liquid metals injected into microchannels composed of triblock copolymer gels. Reproduced with permission.^[121] Copyright 2013, Royal Society of Chemistry. c) Comparison of conductivity at maximum strain of various stretchable conductors. Reproduced with permission.^[97] Copyright 2017, Wiley-VCH. d) Self-healing of liquid-metal-based electrical interconnects. Reproduced with

Thus, the development of elastomers with low profile yet efficient thermal dissipation capabilities is highly valuable.^[129] Generally, its thermal conductivity is lower when the elastomer becomes softer due to the reduced phonon transport dynamics.^[130] Therefore, there is a tradeoff between the thermal conductivity and elastic modulus. This dilemma can be overcome by using the liquid metals for thermal transport thanks to their high thermal conductivity (26.4 W m⁻¹ K⁻¹ at 30 °C) and liquidity.^[131–134] Elastomers embedded with liquid metals that function as thermoconductive highways have been reported.^[135] The resulting composites exhibit an extraordinary combination of low stiffness (<100 kPa), high strain limit (>600%), and metal-like thermal conductivity (up to 9.8 W m⁻¹ K⁻¹) that far exceeds any other soft materials (Figure 3e,f).^[135]

As mentioned above, liquid metals will rapidly and inevitably form an atomically thin layer of oxides when exposed to air even with a ppm level of oxygen.^[136] The oxide skin, with a typical thickness of 0.5-3 nm,^[137,138] has a negligible impact on the charge transport of the liquid metals but can support a maximum surface stress of 0.5–0.6 N m^{-1.[104,139]} This attribute enables various methods of shaping and patterning the liquid metals. These patterning methods can be divided into four categories, as have been well reviewed in other references:[97,98,113] i) lithography-based processes, either directly or indirectly;^[115,140] ii) molding by the use of pneumatic pressure or other forces to inject the metal into predefined templates:^[108,141,142] iii) additive deposition by printing or coating the liquid metal only in desired regions; [115,143,144] and iv) subtractive etching by selective removal of the liquid metals from a substrate, e.g., by laser ablation.^[145,146] Recent advances of these methods can achieve complex 3D structures^[144] and high resolutions (minimal line width of 2 µm)^[140,147,148] of the liquid metals, as shown in Figure 3g,h, respectively. As an example, complex patterns of liquid metals with high precision and resolution can be readily formed (Figure 3i).^[149]

In addition to stretchable conductors, the liquid metals can be used as soft sensors,^[150-152] electronic components,^[148,153] and reconfigurable devices.^[124,154] Associated changes in resistance or capacitance when liquid metals are deformed can be exploited for sensing touch, pressure, and strain. Figure 3j shows a liquid-metal-based sensing glove for human hand tactile feedback.^[155] Memristors composed of two liquid metal electrodes separated by two thin layers of hydrogels have been used for integrated circuits, showing the potential applications in bioelectronic systems.^[156] Reconfigurable electronics allow reversibly changing the device functions with the liquid metal shapes. For example, the resonant frequency of a stretchable antenna can be mechanically tuned over a wide range by shaping the liquid metal in the elastomer (Figure 3k).[154,157] Additionally, the oxide skin that stabilizes the shape of the liquid metals can be controllably removed and deposited by physical or chemical methods, which allow steering and manipulating the liquid metal flow in the microfluidic channels, opening up exciting opportunities for adaptive and intelligent electronic devices.^[158–161]

2.3. Conductive Polymers

With combined electronic properties of conductors and mechanical and processing attributes of polymers, conductive polymers (CPs) have been investigated extensively for diverse electronic and optoelectronic applications^[162-168] since their discovery in 1976,^[169,170] especially after the Nobel Prize was granted to Alan J. Heeger, Alan G. Macdiarmid, and Hideki Shirakawa in 2000. CPs are composed of highly π -conjugated polymeric chains,^[171] along which charge carriers from doping move, giving rise to electrical conductivity.^[172] CPs can be synthesized either chemically or electrochemically with tunable conductivities, e.g., conductive or semiconductive, depending on the dopants used and the level of doping.^[171,173] There are dozens of CPs available now,^[174] including PEDOT:PSS, PAni, polypyrrole (PPy), poly(3-hexylthiophene) (P3HT), polyacetylene (PA), polythiophene (PTh), etc. As one of the most widely investigated CPs, PEDOT:PSS has emerged as an ideal electrode material because of its state-of-the-art high conductivity,^[175] transparency,^[176] chemical stability,^[177] biocompatibility,^[178] ease of deposition,^[179] cost-efficiency, and commercial availability. However, the accompanying cracks when the PEDOT:PSS thin film is stretched limit its applications in stretchable electronics. To deal with this issue, plasticizers such as Zonvl and Trion are blended into PEDOT:PSS to enable their stretchability by decreasing the PEDOT:PSS modulus and increasing its crackonset strain.^[180,181] For example, dough-like CPs by incorporating an excess amount of Triton X-100 (C14H22O(C2H4O)n n = 9-10) in the PEDOT:PSS has been demonstrated, which exhibited desirable conductivity, stretchability (50%), moldability, and self-healing properties.^[181] To further improve the stretchability, ionic additives and conductivity enhancers were added to the PEDOT:PSS aqueous dispersion, resulting in highly stretchable CPs with conductivity of over 3100 S cm⁻¹ at 0% strain and over 4100 S $\rm cm^{-1}$ at 100% strain. $^{[182]}$ From 0% to 100%, the strain enhances the PEDOT chain alignment along the direction of stretching, which leads to the increased conductivity of the PEDOT:PSS. Further increasing the strain beyond 100% will rupture the polymer chains and thus result in decreased conductivity. The conductivity is highly durable and remains above 100 S cm⁻¹ under 600% strain. Figure 4ai shows the excellent conductivity of the stretchable PEDOT:PSS conductor under poking or twisting while stretching. Using PEDOT:PSS as the electrode, stretchable ultrathin solar cells

permission.^[122] Copyright 2016, Royal Society of Chemistry. e) Stretchable thermoconductive composites by impregnating liquid metal droplets in the elastomer. The lower-left inset shows the schematic illustration of the liquid metal elongation under stretching. f) Summary of the thermal conductivity as a function of the ratio of strain limit to Young's modulus for different materials. Reproduced with permission.^[135] Copyright 2017, National Academy of Sciences. g) A freestanding 3D structure of liquid metals. Reproduced with permission.^[144] Copyright 2013, Wiley-VCH. h) Liquid metal lines with a high pattern resolution. Reproduced with permission.^[147] Copyright 2015, Elsevier. i) Printed electronics based on liquid metals with sophisticated layouts. Reproduced with permission.^[149] Copyright 2014, Nature Publishing Group. j) A soft tactile sensing glove based on the microfluidic liquid metal. Reproduced with permission.^[155] Copyright 2017, Wiley-VCH. k) The antenna frequency as a function of the tensile strain. Reproduced with permission.^[154] Copyright 2010, Wiley-VCH.







Figure 4. Conductive polymers for soft electronics. a) Conducting polymers are ideal electrode materials and their applications in: i) stretchable conductor (Reproduced with permission.^[182] Copyright 2017, American Association for the Advancement of Science.), ii) solar cells (Reproduced with permission.^[184] Copyright 2012, Nature Publishing Group.), and iii) LEDs (Reproduced with permission.^[184] Copyright 2013, Nature Publishing Group.), and iii) LEDs (Reproduced with permission.^[184] Copyright 2013, Nature Publishing Group.). b) Semiconducting polymers are indispensable building blocks for electronic devices. i) Intrinsically stretchable semiconductor whose mobility is minimally affected by the strain under 100%. Reproduced with permission.^[189] Copyright 2017, American Association for the Advancement of Science. ii) Transient FETs based on biocompatible and totally disintegrable semiconductive polymers. Reproduced with permission.^[190] Copyright 2017, National Academy of Sciences. c) Conductive polymers that allow ionic and electronic transport. i) Metal microelectrode arrays coated with conducting polymers for electrophysiological recordings in vivo. Reproduced with permission.^[200] Copyright 2017, Wiley-VCH. ii) Conducting polymers as the channel to fabricate organic electrochemical transistors for brain activity recording. Reproduced with permission.^[217] Copyright 2013, Nature Publishing Group. d) Thermoelectric properties of conducting polymers. i) Thermoelectric energy harvesting. Reproduced with permission.^[215] Copyright 2013, Royal Society of Chemistry. ii) A temperature–pressure dual-parameter sensor. Reproduced with permission.^[217] Copyright 2015, Nature Publishing Group.

and light-emitting diodes have been demonstrated, as shown in Figure 4aii,iii, respectively.^[183,184]

In addition to the stretchable conductors, stretchable semiconductors are indispensable for building functional components and devices such as diodes and field-effect transistors (FETs).^[185] Polymers can be semiconductive by tuning the dopant type and the level of doping. Additionally, semiconductive polymers can be molecularly tailored for different levels of stretchability, making them promising for stretchable functional components. For example, P3HT has a crack onset strain of over 150%, which can be tuned by tailoring its molecular structure.^[186,187] Another example is intrinsically stretchable semiconductor based on 6-di(thiophen-2-yl)-2,5dihydropyrrolo[3,4-c]pyrrole-1,4-dione (DPP) and 2,6-pyridine dicarboxamide (PDCA), in which PDCA introduces hydrogen bonds to promote dynamic noncovalent crosslinking of DPP.^[188] The stretchable semiconductor is able to recover its high fieldeffect mobility of more than 1 cm² V⁻¹ s⁻¹ after 100 cycles at 100% strain. However, the stretchable semiconductor has anisotropic charge transport behavior and its mobility decreases from 1.32 cm² V⁻¹ s⁻¹ perpendicular to the applied strain direction to 0.11 cm² V⁻¹ s⁻¹ along the strain direction. Recently, highly stretchable polymer semiconductors with strain-insensitive mobility were demonstrated using a nanoconfinement effect.^[189] The resulting FET exhibits high biaxial stretchability and constant charge mobility under stretching (1.08 cm² V⁻¹ s⁻¹ up to 100% strain), as shown in Figure 4bi. Semiconducting polymers that consist of reversible imine bonds can easily be decomposed under mild acidic conditions and have been used as building blocks for disintegrable FETs,^[190] which may find applications in environmental friendly electronics and transient and biocompatible implants (Figure 4bii).

In biological systems, the electrical signals are controlled and transported through ionic transport across cell membranes and neurotransmitter diffusion between synapses in the nervous systems.^[191] Materials that can accommodate both electronic and ionic transport are, therefore, of particular value for interfacing biology with electronics.^[192] The mixed ionic-electronic conductivity of CPs makes them unique in bioelectronic applications, enabling low electrical impedance at the biotic/abiotic interface and therefore improved quality of signal recording.^[192-195] Combining other attributes of CPs such as biocompatibility, high flexibility, thin thickness, and oxide-free interfaces with aqueous electrolytes, abundant examples of CPs in bioelectronic applications have been demonstrated.^[195-197] One of them is to use CPs to coat the metal microelectrode arrays for electrophysiological recordings in vitro and in vivo,^[198-200] exhibiting higher sensitivity, spatial resolution, and signal-to-noise ratio than bare metal electrodes (Figure 4ci). Another example is to use CPs as the channel to fabricate organic electrochemical transistors (OECTs). In the OECTs, aqueous electrolyte functioning as the dielectric material allows ions of interest to diffuse in and out of the channel, which modulates the CPs' conductivity.^[201,202] The OECTs offer intimate interfacing between solid-state electronics and biological milieu,^[203] thereby making them ideal for biological sensing,^[204] such as electrocardiogram monitoring,^[205,206] brain activity recording,^[207] and neurotransmitters detecting.^[208] Figure 4cii shows an OECT conformably laminated on the somatosensory cortex of a rat.^[207] Compared with the metal microelectrodes, OECTs show a higher transconductance and signal-to-noise ratio due to this bulk coupling between ionic and electronic charges.^[209] Moreover, OECTs and FETs have been integrated for detecting electromyogram with a cutoff frequency of 3 kHz,^[210] paving the way to large-scale integrated electronics and high-frequency signal measurements of OECTs.

CPs have intrinsically low thermal conductivity.^[211] In stark contrast to their inorganic counterparts, the thermoelectric properties of CPs can be decoupled from electrical conductivity, which holds great implications for high-performance thermoelectric generators (TEGs) and temperature sensors.^[212-214] TEGs are devices that can produce electrical power from temperature gradients, which offer a desirable solution for waste heat harvesting. Flexible TEGs based on the PEDOT film have been demonstrated, which allow conformal contact to nonplanar surfaces for maximum waste heat scavenging (Figure 4di).^[215] A voltage output of 10 μ V was generated by the touch of fingertips.^[215] For temperature sensors, an array of PAni nanofiber-based temperature sensors interfaced by a stretchable active matrix can provide 2D temperature mapping.^[216] When supported on a microstructural frame, the PEDOT:PSS film shows dualparameter sensing properties, i.e., thermoelectricity for temperature and piezoresistance for pressure. Additionally, by harvesting thermal energy, a self-powered temperature–pressure dual sensor was demonstrated, with a temperature resolution of <0.1 K and a pressure sensitivity of 28.9 kPa⁻¹ (Figure 4dii).^[217] Other than those, CPs have found applications in a variety of soft electronic devices, such as electrochromic devices,^[218–220] pressure sensors,^[215,220–222] pH sensors,^[223] and supercapacitors,^[224] documenting their versatile roles in soft electronics.

2.4. Nanomaterials

With advances in the large-scale bottom-up synthesis of functional nanomaterials, conductive nanomaterials become integral building blocks for stretchable electronics.^[225] Nanomaterial-based stretchable conductors are different from the intrinsically stretchable conductors.^[226] Instead, they are created in the form of composites, i.e., by mixing conductive fillers, usually inorganic materials,^[227] in a stretchable polymeric matrix. Those conductive nanofillers can form a percolated network in the elastomer, granting electrical current pathways to the composite.^[231-234] Such nanofillers with distinctive morphologies can be divided into three major categories: 0D nanoparticles (NPs); 1D nanowires (NWs) and nanotubes (NTs) with high aspect ratios; and 2D nanosheets with large junction areas. Elastomers such as PDMS, Ecoflex, and polyvinylalcohol (PVA) are generally used as the matrix to provide elastic recoverability for the percolation network.^[228] The self-assembly approach to construct the percolation network can be large area, low cost, and high throughput.^[229] Due to their intertwined physical overlap and entanglement, those selfassembled nanostructured conductive pathways are robust for common modes of mechanical deformation.^[230] Moreover, due to the involved nanoscale dimensions, the resulting devices can have an ultrathin profile, leading to outstanding mechanical compliance,^[230] which entails significantly reduced strain. In all, these attributes are crucial for those conductive nanocomposites to integrate on the human body that requires conformal intimate contact with curvilinear and time-dynamic surfaces. Design parameters and fabrication strategies are discussed in the following sections.

2.4.1. Stretchable Nanocomposite Conductivity

The source of resistance in the stretchable nanocomposites is twofold: the resistance of filler itself, also known as the intraresistance, and the resistance between the fillers, called the interresistance or contact resistance. The latter one is more dominant. There are two kinds of contact mode: physical contact that has nearly no resistance and incomplete contact or noncontact mode whose conductivity depends on its tunneling resistance. The tunneling current between two noncontact fillers is an exponential decay function of the interfiller distance.^[236,237]

According to the percolation theory, the nanocomposite conductivity is mainly determined by the nanofiller volume fraction $^{\left[235\right] }$

 $\sigma = \sigma_0 (V_{\rm f} - V_{\rm c})^s \tag{1}$



where σ is the conductivity of the composite film, σ_0 is a scaling factor proportional to the intrinsic conductivity of the filler, $V_{\rm f}$ is the volume fraction of the filler, $V_{\rm c}$ is the percolation threshold, and s is the critical exponent of the conductivity. From Equation (1), it becomes clear that higher loading of the conductive filler will bring higher conductivity. However, the nanofillers normally have higher Young's modulus than the elastomer matrix and will add mechanical loading to the elastomer matrix, which will thus increase the overall rigidity of the composite. Therefore, a delicate equilibrium should be achieved between the electrical conductivity and mechanical compliance. It is worth noting that Equation (1) is under strain-free situation. Under tensile strain, the nanocomposite conductivity highly depends on the design of the conductive network.^[238] This can be controlled by several major aspects, such as the morphology of the conductive fillers (0D, 1D, or 2D),^[239] volume fraction of the conductive fillers,^[240] and structure of the percolated network (2D or 3D),^[231] as discussed in the following sections.

2.4.2. Stretchable Nanocomposite Fabrication

To incorporate the conductive nanomaterials into the stretchable matrix, there are several representative strategies: blending, in situ formation, and implantation. The most straightforward way is direct blending^[242] or infiltrating^[233,241] the elastomer into an already percolated conductive-filler networks (**Figure 5**a). The performance of the composite can be tuned by several parameters, such as the filler volume fraction and geometry, and subsequent annealing.^[233,241,243] A representative stretchable composite of single-wall nanotubes (SWNTs) in polytertbutyl-acrylate is shown in Figure 5b. The composite has high conductivity due to the high volume fraction of the SWNTs.^[243]

The nanofillers tend to aggregate at low volume fractions before reaching the percolation threshold, leading to poor dispersion of the nanofillers and thus reduced conductivity. To solve this challenge, besides using mechanical homogenization with the help of surfactant,^[242] an alternative method is to directly synthesize these nanofillers inside the polymer matrix (Figure 5c), by taking advantage of the good miscibility between the nanofiller precursors and the elastomer. This miscibility helps prevent the filler from growing into large sizes because the precursor diffusion is retarded by the elastomer polymer chains. The in situ formation mechanisms of the nanofillers typically include chemical redox reactions,[244-247] UV treatment, [245,248] catalyst moieties assisted reaction, [249,250] and laser carbonization.^[251] The crosslinking of elastomer is induced by thermal annealing or UV exposure. Those composites can carry a large volume fraction of the filler, leading to high electrical conductivities. Representative results are shown in Figure 5d,e. First, polymeric fibers absorbed with silver precursors in alcohol are fabricated (Figure 5d). After chemical treatment, the precursors are then reduced to form percolated silver NPs (Figure 5e).^[252]

Another effective technique is ion implantation, which provides a powerful way to embed conductive nanofillers directly into the surface region of the polymer matrix (Figure 5f). Metal cations are generated by plasma colliding the target source. These cations are subsequently accelerated typically to 0.2–10 eV, which allow easily penetrating the porous polymer chains.^[253] By modulating the acceleration voltage, the total implantation depth can be controlled from 10 nm to 1 μ m.^[254] This method has great feasibility and flexibility that apply to nearly any metals and any polymer substrates.^[253] Another key feature with ion implantation is that it forms the conduction path only close to the elastomer surface, without wasting a lot of metal NPs in the bulk.^[254,255] The SEM image in Figure 5g shows a good example of implanting Au NPs into PDMS.^[255] In this case, the implanted NPs form densely distributed conductive networks and concentrate at a particular depth in the substrate.

Among all of those techniques, one common and critical issue is the interfacial bonding between nanomaterial and its matrix, which can help the percolation network maintain structural integrity under large deformations. However, most of the elastomers such as PDMS and Ecoflex are not desirable for forming robust bonding with nanomaterial percolation network.^[231] Therefore, various treatment techniques, e.g., UV ozone,^[256,257] oxygen plasma,^[258,259] and adding surfactants,^[260,261] can be utilized to enhance either the interfacial chemical reaction or physical adsorption between those nanomaterials and the polymeric matrix.

2.4.3. NP-Based Stretchable Electronics

The fundamental physical picture of NP-based stretchable electronics is illustrated in **Figure 6**a. In this conducting network, it is assumed that the particles are homogeneously distributed and have a perfect bonding with the polymer matrix. Conceivably, the composite will be divided into cubic elements, each of which has an individual particle in the center. There are two critical parameters that decide the percolation threshold: the particle diameter (*D*) and the interparticle distance (*D*_{IP}), labeled in Figure 6a. The percolation threshold of the network *V*_c can be calculated by

$$V_{\rm c} = \frac{\pi D^3}{6(D + D_{\rm IP})^3},$$
 (2)

which indicates that the V_c monotonically increases as the particle size increased (Figure 6b). According to numerical studies,^[239] the interparticle distance has a strong impact on the tunneling current magnitude between the NPs due to their small aspect ratio. Larger interparticle distance leads to much smaller chance of effective tunneling and therefore lower electrical conductivity. When D_{IP} is equal to or less than 10 nm, electron hopping takes place, which results in a rapid increase in electrical conductivity of the composite.^[262,263] In this case, the resulting composite conductivity is independent of the polymer resistivity. Therefore, $D_{IP} = 10$ nm is widely adopted as the criterion for conduction in composite materials. Various types of NP-based high-performance stretchable conductors have been demonstrated. As illustrated in Figure 6c, Ag NPs were formed in situ from micrometer-sized 2D Ag flakes and were dispersed within the polymer matrix after adding surfactants and heating. Those NPs formed a dense percolation network with a spacing below 10 nm (Figure 6d), which was SCIENCE NEWS _____ www.advancedsciencenews.com

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Figure 5. Methods for incorporating nanomaterials in polymers. a) Schematic illustration of the blending or infiltrating process. b) The SWNT network visualized by atomic force microscopy (AFM). Reproduced with permission.^[243] Copyright 2009, Wiley-VCH. c) Schematics for in situ formation of nanofillers. d) SEM images of polymer fibers and NP precursors before chemical reduction. e) SEM images after chemical reduction of the NP precursors. Reproduced with permission.^[252] Copyright 2012, Nature Publishing Group. f) Schematics for ion implantation. g) A transmission electron microscopy (TEM) image of the cross-section of Au implanted PDMS, showing the formation of nanometer-sized Au clusters. Reproduced with permission.^[255] Copyright 2009, Wiley-VCH.

small enough to allow high-efficiency electron tunneling. These in situ formed NPs significantly improve the conductivity of the composite from 5.5×10^{-5} to 4919 S cm⁻¹·^[264]

More interestingly, when under strain, NPs have higher dynamic properties and thus greater freedom for reversible nanoscale reconstruction than its higher dimensional counterparts







Figure 6. Conduction mechanism and properties for NP-based stretchable conductors. a) Schematics showing individual NPs located in a unit cell, based on the assumption of a 3D random uniform distribution model. Key parameters deciding the conductivity and percolation threshold are labeled, including particle diameter *D* and interparticle distance D_{IP} . a) Redrawn based on ref. [239] with permission. b) Estimated percolation threshold as a function of the NP diameter. Small NPs lead to lower percolation threshold. b) Reproduced with permission.^[239] Copyright 2014, Elsevier. c) Ag NPs in situ synthesized from micrometer-sized Ag flakes. Reproduced with permission.^[264] Copyright 2009, Nature Publishing Group. d) TEM images of a stretchable conductor, with densely distributed Ag NPs. The inset shows the relatively close interparticle distance. Reproduced with permission.^[264] Copyright 2009, Nature Publishing Group. e) TEM images of self-aligned NP-based stretchable conductors under different levels of applied uniaxial strain (from 0% to 50%). Scale bars are all 50 nm. Reproduced with permission.^[265] Copyright 2013, Nature Publishing Group.

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(1D NWs and 2D nanosheets).^[265] As seen in Figure 6e, when there is no external strain, the NPs are capable of self-assembling into the chain-like structure at the phase boundary by van der Waals and steric forces, which can be modulated by the concentration of additives.^[266] Such chains provide conduction paths at the freestanding state. During stretching, although the large strain will increase the interparticle spacing, the NPs tend to reorganize in a different configuration. The NPs are so small that they share the surrounding polymer chains, which drive the NPs to gradually self-assemble into bands along the stretching direction, retaining structural and electrical integrity.^[264] Additionally, this strain induced self-assembly is dynamically reconfigurable. The following model describes this reorganization behavior under strain

$$V_{\rm c}(\varepsilon) = \frac{V_{\rm c}^0}{1 + \alpha \sqrt{\varepsilon}} \tag{3}$$

where α indicates reorganizing capability of the NPs under stress, ε is strain, V_c^0 is the percolation threshold under freestanding conditions, and $V_c(\varepsilon)$ is the strain-dependent percolation threshold. Those unique properties of NP-based stretchable conductors can be used to fabricate conductive materials with mechanical tunability to meet the demands of special use cases that require tunable electrical properties. Further improvement of stretchability can be realized by structural designs, e.g., by introducing prestrains to the substrate that is discussed in later sections.^[246,267]

2.4.4. NW- and NT-Based Stretchable Electronics

NWs/NTs have been widely used to construct soft electronic devices, such as biosensors,^[268] composite biomaterials,^[269] transistors,^[270,271] floating gate memory units and logic gates,^[272] generators,^[273] electronic skins,^[274] etc. The anisotropic high aspect ratio (10–100 times larger) NWs/NTs have a much lower percolation threshold V_c than the isotropic NPs,^[275,276] which can be explained by the theory of percolation

$$N_{\rm c}L^2 = 5.71$$
 (4)

where $N_{\rm c}$ is the percolation threshold density and *L* is the length of the NWs or NTs. This equation indicates that the percolation threshold is inversely proportional to the square of the filler length.^[235] The relationship between the network conductivity and conductive filler density can be defined by

$$\sigma \propto (N - N_c)^s \tag{5}$$

where σ is the network conductivity, *N* is the density of the conductive fillers, and *s* is the critical exponent of the conductivity as in Equation (1). For the same level of conductivity, 1D nanomaterial can significantly decrease the filler density compared with the 0D configurations. Furthermore, the 1D configuration provides a high persistence length, which ensures a direct charge transport path and decreases the need for interfiller contact areas. This is critical for reducing the piezoresistance of the composites because the contact areas are prone to forming defects and cracks when strain is applied.

Due to the low percolation threshold of NWs, it is suitable for fabricating stretchable conductor that is transparent. **Figure 7a** shows a stretchable elastomeric polymer light-emitting diode with Ag NWs as transparent and stretchable electrodes. The top left panel of Figure 7a illustrates the high transparency due to the low density of the NW network (top right panel). The large aspect ratio (\approx 500) of NWs allows the percolation network to maintain good light-emitting efficacy under 120% uniaxial strain (bottom panels).^[235]

Since NWs have high intrinsic conductivity, the NW-NW junction conductivity usually plays a dominating role for the stretchable composite. At the junction site of two NWs (left panel in Figure 7b), the insulating ligands used for the synthesis and solution dispersion of the NWs will usually create a nanometer-scale gap between the NWs and thus decrease the wire junction conductivity (inset in Figure 7b, left). To solve this problem, the NW junction can be welded by light-induced annealing (right panel in Figure 7b), which forms epitaxial recrystallization at the overlapping region. This significantly reduces the junction resistance by three orders of magnitude (from 14285.7 to 33.3 Ω).^[277] Besides light-induced annealing, other methods including plasmonic nanowelding,^[278] thermal annealing,^[279-281] mechanical pressing,^[282,283] and soldering by other types of materials^[284,285] have also been proven to be effective in decreasing the NW contact resistance.

Due to the ultrathin thickness of the NW-based percolation network, it is highly suitable for fabricating biological interfaces that have low mechanical footprint.^[286] A Au NW percolation network has been used to build a nanomesh electrical interface that supports conduction and circuit functions directly on the skin (Figure 7c). The percolation network, when under 20% tensile strain (labeled by the dashed circle in Figure 7d in the left panel), shows a large opening space. This is because the NWs will slip from the intertwined networks when stretched. When the tensile strain is released (Figure 7d, right panel), the fibrous network recovers and looks nearly identical to the initial state. Even under cycling, this phenomenon is reproducible, demonstrating the robustness of NW-based conductive composites.^[286,287]

NWs with strategic structural layouts can further enhance the mechanical and electrical performance.^[288,289] One of the effective techniques is to use prestrain. Figure 7e shows buckled carbon NTs (CNTs) fabricated by transferring the percolated CNT network onto a prestrained (tensile strain) PDMS substrate and then releasing the prestrain in the substrate. The buckled CNT conductors could accommodate around 100% tensile strain with only 4% resistance increase. The NW network can also form a reticulated structure, which possesses unique hierarchical structures with very long and good interbundle connections (Figure 7f). Those CNT bundles, synthesized by floating catalyst chemical vapor deposition, entangled randomly and created 3D interconnection.^[290,291] This kind of interconnection is effective for electronic transport and mechanical robustness, similar to the leaf veins (inset in Figure 7f). With this design, the electrical conductivity and mechanical strength can reach 2000 S cm⁻¹ and 350 MPa, respectively.^[292]

The composite resistance will usually increase significantly with stretching due to the increase in the nanofiller spacing.^[293] A strategy that solves this challenge is to exploit self-aligning properties under the tensile strain of the nanofillers, which was demonstrated by a stretchable sensing sheet



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Figure 7. Conduction mechanism and properties of 1D nanomaterial-based stretchable conductors. a) Elastomeric light-emitting diodes formed by using Ag NW percolation network as transparent electrodes, which can withstand 120% strain. Reproduced with permission.^[235] Copyright 2013, Nature Publishing Group. b) TEM images of Ag NW junctions before (left) and after (right) welding. The inset schematics illustrate the overlapping of adjacent NWs. Reproduced with permission.^[277] Copyright 2012, Nature Publishing Group. c) Images of patterned conductive traces of Au NWs directly on the skin. A magnified image of similar traces on the fingertip, illustrating the conformal integration of Au NWs to the fingerprint ridge patterns. The bottom right SEM image shows the mesh structure of Au NWs. Reproduced with permission.^[287] Copyright 2017, Nature Publishing Group. d) SEM images of the Au nanomesh conductor with cracks formed after 20% stretch (labeled in the dashed circle, left) and at same crack location when the strain is removed (right), showing its high adaptability to strain. Reproduced with permission.^[287] Copyright 2017, Nature Publishing Group. e) An SEM image of the CNT bundles after compressive buckling, forming a periodic wavy structure. Reproduced with permission.^[397] Copyright 2012, Wiley-VCH. f) SEM images showing the leaf veins (right), both of which can effectively absorb the applied strain. Reproduced with permission.^[274] Copyright 2012, Wiley-VCH. g) AFM images showing the evolution of NW morphologies due to the self-alignment under stretching. Reproduced with permission.^[277] Copyright 2012, Wiley-VCH. S) AFM images showing the evolution of NW morphologies due to the self-alignment under stretching. Reproduced with permission.^[277] Copyright 2017, Nature Publishing Group.

fabricated by directly spray-coating CNTs onto a PDMS substrate (Figure 7g).^[274] Before stretching, the fibrous CNT bundles behave in an isotropic crosslinking state (Figure 7g, left). After stretching, the isotropic pattern becomes anisotropic to accommodate the deformation (Figure 7g, right), as highlighted by the fibers in the dashed boxes.^[274] Therefore, the strain on the individual fiber and associated resistance change can be greatly reduced. This mechanism also paves the way for bending insensitive devices that can achieve accurate pressure measurements on complex and dynamic surfaces without the influence of external bending, twisting, or wrinkling.^[294]

Intriguingly, some NWs have intimate interaction with polymers, which allow fabricating textile-based stretchable conductors.^[295] CNTs have proven to have large van der Waals interactions with many types of polymers.^[296] Particularly SWNTs and cellulose fibers possess relatively large van de Waals interaction between each other.^[297] Furthermore, after acid treatment, SWNTs gain carboxyl functional groups on the surfaces and the ends, which can form strong hydrogen bonds with the hydroxyl groups on the cellulose fibers. Upon contact, large van der Waals forces and hydrogen bonds bind the SWNTs very tightly to the cellulose, improving the robustness of conductive fibers.^[297]

2.4.5. Nanosheet-Based Stretchable Electronics

1D nanomaterials sometimes suffer from large junction resistance when stretched due to the relatively small junction area.^[298] Even though sintering and welding approaches can alleviate this problem, the requirements of high temperature or advanced light sources present limitations on substrate compatibility and scaling up. 2D materials such as graphene and its derivatives^[299] with large surface area,^[300] high crystallinity,^[301] uniformity,^[302] extremely high mobility,^[303] and high elasticity^[304,305] have gained increasing popularity for stretchable electronic materials.^[306] Specifically, the large surface area provides a broader junction area and conductive path, and thus lower resistance. High crystallinity and uniformity bring stable electrical performance. High mobility is suitable for highefficiency electronic transport. Intrinsically high elasticity is valuable for accommodating mechanical deformation.

Graphene can be synthesized with various methods, such as mechanical exfoliation,[307] epitaxial growth,[308] chemical reduction from graphene oxide,^[309] and laser-induced reduction from organic polymers.^[310] However, the large area graphene normally shows relatively poor conductivity in stretchable electronics due to the presence of abundant structural defects and weak interlayer junction contact. To solve this challenge, chemical vapor deposition (CVD) was utilized to grow large-scale, low-defect, and uniform crystalline monolayer/multilayer graphene.^[311] The monolithic graphene sheet from CVD growth can be transferred to arbitrary substrates for stretchable electronic devices.[311-313] Electromechanical testing of the monolithic graphene is shown in Figure 8a (with an isotropic prestrain level of 12%). If there is no prestrain, the graphene sheet will fail after stretching to ≈6% (inset in Figure 8a). During stretching, the monolithic graphene shows a robust mechanical behavior due to its honeycomb structure. Both longitudinal and transverse resistances (R_v and R_x) appear stable up to 11% uniaxial strain. R_{ν} (along the stretching direction) shows only one order of magnitude increase at 25% uniaxial strain. Another interesting property of the monolithic graphene is that further stretching (>25%) will alternate its electronic band structure, introducing huge sheet resistance increase.^[311]

Monolithic, large area graphene sheet can only be stretched to a limited extent of 6% before structural failure starts to

appear, which presents limitations for fabricating highly stretchable conductors,^[314] especially in devices for human skin integration, which requires stretchability to be 20–30%.^[15] One method that solves this problem is to change its original 2D planar layout to 3D foam-like conductive network, which can be directly synthesized by template-directed chemical vapor deposition followed by template removal.^[315] This 3D interconnected stretchable network can greatly enhance its electromechanical properties. Another method to enhance the graphene stretchability is to transform the material morphology by prestrainenabled buckling (Figure 8b, top panel). The key point is that the large area graphene sheet is transferred to a prestrained rather than a freestanding substrate. Releasing the prestrain in the substrate will create reversibly crumpled graphene sheets. The stretchability of this electrode can therefore be greatly improved to reach even 200-400% (depending on the prestrain level) without structural failure or conductivity degradation.^[304] Besides graphene and its derivatives, this strategy also applies to other metallic 2D nanomaterials.[316-318]

Considering the cost and difficulty of fabricating monolithic sheets of 2D materials, the most popular approach is to homogeneously blend fragmented 2D nanosheets in the elastomers.^[301] A representative example is a G-putty material, which is fabricated by combining graphene nanosheets with lightly crosslinked polysilicone (Figure 8c, left). In this way, the nanosheets form a conductive percolation network (as illustrated in Figure 8c in the middle panel), which is similar to the percolation theory of 0D and 1D nanomaterial-based stretchable composites. However, there are some unique properties brought by this 2D morphology. For example, those 2D nanosheets can provide much larger junction areas than the NPs and NWs, which can allow more efficient electron tunneling between the nanosheets.[319] Additionally, some 2D materials, e.g., graphene, possess higher mobility than traditional semiconductors, and are thus very suitable for high-performance transistors and detectors.

The conductive mechanism of this 2D nanomaterial-based stretchable conductor is discussed as follows. In the percolation network (Figure 8c, middle), the conductive path is labeled. The total resistance of this conductive path can be calculated as the summation of intersheet resistance $R_{i,i+1}$ and intrasheet resistance $R_{i,i}$

$$R \approx \sum_{i=0}^{n} R_{i,i} + R_{i,i+1}$$
 (6)

While this represents the series resistance for one conductive path, there are of course many conductive paths in parallel. Therefore, the total network resistance is the sum of the contribution of all these parallel paths

$$\frac{1}{R_{\rm T}} = p_2 \frac{1}{p_1(R_{i,i} + R_{i,i+1})} \tag{7}$$

In Equation (7), given that p_1 is a measure of the number of nanosheets in a conducting path and p_2 is a measure of the number of parallel conducting paths.

Pure graphene is a high-performance conductor when compared with its low dimensional material counterpart (0D







Figure 8. Conduction mechanism and properties for 2D nanomaterial-based stretchable conductors. a) Resistance of a graphene film transferred to a PDMS substrate isotropically stretched by 12%. The left inset shows the case when the graphene film is transferred to an unstrained PDMS substrate and has only 6% stretchability. Reproduced with permission.^[311] Copyright 2017, Nature Publishing Group. b) Schematic illustration of macroscopic deformation of a graphene sheet on a biaxially prestrained substrate (top). SEM images of crumpled graphene when the bi-axial strain is released (bottom). A zoomed-in image of the crumpled graphene appears in the inset. Reproduced with permission.^[304] Copyright 2017, Nature Publishing Group. c) Putty and G-putty based on a stretchable graphene-based nanocomposite (left panel) formed by mixing graphene nanosheets with a lightly crosslinked polysilicone. Schematic illustration of the conduction model of fragmented graphene nanosheet-based stretchable conductors. The conduction path and junction areas formed by overlapping graphene nanosheets (No. 1 to No. 4) are labeled (middle panel). A TEM image of the graphene nanosheet (right panel), showing its relatively large junction area. Reproduced with permission.^[322] Copyright 2016, Nature Publishing Group.

NPs and 1D NWs) and can maintain high conductivity even under large deformations (Figure 8c, right).^[320] Therefore, its intrasheet resistance is relatively small. The overall resistance is dominated by the intersheet resistance, i.e., the contact resistance. It has been proven that the intersheet resistance under the applied strain can be calculated by^[321]

$$R = ad_0(\varepsilon + 1) \times e^{bd(\varepsilon + 1)}$$
(8)

where d_0 is the intersheet separation during the strain-free state and is largely determined by the filler concentration. The contact resistance is largely determined by the applied strain ε , which can be calculated by $\varepsilon = (d - d_0)/d_0$, where *d* is the intersheet separation under strain. *a* and *b* are given by^[322]

$$a = \frac{h^2}{Ae^2 \sqrt{2mE_B}} \quad \text{and} \quad b = \frac{4\pi}{h} \sqrt{2mE_B} \tag{9}$$

where *h* is Planck's constant, *A* is the junction area, *m* is a constant related to the fractal structure of the network, and $E_{\rm B}$

is the height of the potential barrier (work function) between nanosheets.

Therefore, various parameters can be tuned to design the 2D material blended stretchable conductors, such as the filler volume fraction, nanosheet morphology, and nanosheet overlapping for junction area construction. Considering the high dependence of intersheet resistance on the intersheet separation, which is determined by the applied strain, strain sensing materials with extremely high gauge factors can be designed.^[322] Examples can be seen in the assembled 3D macroporous nanopapers^[314] and crumpled graphene nanosheets.^[312]

Another effective approach to design 2D nanomaterial-based stretchable conductors is to dope them with other materials. Taking graphene for example again, after blending with elastomers, graphene sheets always have line defects and disruptions such as wrinkles, ripples, and foldings that function as scattering centers for charge carriers and adversely affect the electronic transport properties.^[323] To solve this problem, it has been demonstrated that combining the graphene with other conductive fillers such as NWs and NPs can significantly



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minimize the defects in each individual material. While situated at the defect location, the NW and NP fillers can locally improve the conductivity of those boundaries.^[324] Based on this strategy, a wearable sensor and stimulator system is made by doped graphene and Ag NWs, whose electromechanical properties are much better than using graphene alone (with 2-7 times improvement on conductivity, depending on the NW material).^[325] Another wearable patch for sweat monitoring based on graphene doped with Au and Au mesh showed an improved conductivity in comparison with pure graphene (the resistance decreased from 10^5 to $5 \times 10^2 \Omega$).^[326] Finally, some 2D materials possess unique semiconducting properties and thus can be leveraged to fabricate stretchable FETs that meet both the mechanical and electrical needs.^[327-331] Applications ranging from stretchable LEDs^[332,333] to energy devices^[334,335] and temperature sensors^[336] demonstrate the rich potential of those 2D material in building stretchable electronics.

3. Structural Designs for Soft Electronics

The structural design is a powerful approach for soft electronics. It allows not only enhancing the stretchability of materials that are intrinsically soft, but also bestowing stretchability to materials that are intrinsically rigid. Recent advances in structural designs have enabled hybridization of all classes of materials, organic or inorganic, soft or rigid, lab-made or commercial-off-the-shelf, on the same stretchable platform, which combines the electrical properties of conventional rigid waferbased devices and the abilities of elastomers to be physically deformed. In this part, strategies and considerations are discussed in various structural designs, including wave/wrinkle, "island-bridge," origami, kirigami, textile, crack, and interlock. Exemplary use cases of each structure are highlighted when appropriate.

3.1. Wave/Wrinkle

The wave/wrinkle, similar to fold, ridge, or crease commonly seen in the skin or on fabric, inspires the first type of structural design for inorganic-material-based stretchable electronics.^[225] In this design, inorganic thin films are bonded on the surface of an elastomeric substrate with tensile prestrain, which drives the formation of wave/wrinkle patterns in the inorganic thin films when it is released.[337] Basically, the inorganic thin films convert the compressive strain from the substrate into its bending strain. The prestrain can usually be generated either by thermally induced expansion,^[337] solvent swelling,^[338] or mechanical prestretch.^[339-342] The prestrain levels are typically on the order of a few percent by the thermal expansion or solvent swelling,^[267,343] and can be as high as hundreds of percent by the mechanical prestretch.^[344,345] The inorganic thin films are normally deposited by vapor phase deposition^[337] or transfer printing.^[346] Kinetic growth of the wave/wrinkle includes three stages: initial growth, coursing, and equilibrium, which can be quantitatively understood by the combination of liner perturbation analysis and energy minimization.^[347-349] Both single- and multiwave patterns can occur (**Figure 9**a), corresponding to the global and local buckling modes of the hybrid system, respectively.^[350–352] Critical factors that separate these two modes are the modulus and thickness ratio between the substrate and the inorganic thin film.^[353] For a particular substrate and an inorganic thin film, e.g., PDMS and Si, a thickness ratio of more than 1000 favors the local buckling mode, and a ratio lower than 1000 favors the global buckling mode. The former is more common and is the focus in this review.

The wave/wrinkle configuration can accommodate the tensile or compressive strain by changing its profile (amplitude and wavelength; Figure 9b) to avoid fracturing the relatively stiff inorganic thin films.^[354] The wavelength and amplitude are determined by the mechanical properties^[355] and geometric features^[356] of both the inorganic thin films and the elastomeric substrates, and the level of prestrain.[357] Many mechanical models have been developed to predict the wave/wrinkle configurations as well as their stretchabilities.^[350,358-360] Generally, in the regime of small prestrain ($\varepsilon_{\rm pre} < 5\%$), the wave/wrinkle configuration can be determined by the small deformation theory.^[361,362] In this approach, outof-plane displacement with a sinusoidal profile is assumed for the buckled inorganic thin films whose wavelength (λ_0) and amplitude (A_0) can be determined through the minimization of total strain energy

$$\lambda_{0} = \frac{\pi h_{\rm f}}{\sqrt{\varepsilon_{\rm c}}} = 2\pi h_{\rm f} \left(\frac{\overline{E}_{\rm f}}{3\overline{E}_{\rm s}}\right)^{1/3} \tag{10}$$

$$A_0 = h_{\rm f} \sqrt{\frac{\varepsilon_{\rm pre}}{\varepsilon_{\rm c}} - 1} \tag{11}$$

where the subscripts "f" and "s" denote the thin film and the substrate, respectively; $\overline{E} = E/(1-v^2)$ is the plane-strain modulus, with *E* and *v* denoting the elastic modulus and Poisson ratio of the material, respectively; $h_{\rm f}$ is the thin film thickness; and $\varepsilon_{\rm c} = (3\overline{E}_{\rm s}/\overline{E}_{\rm f})^{2/3}/4$ is the critical strain needed to induce buckling. When $\varepsilon_{\rm pre} < \varepsilon_{\rm c}$, no bucking occurs and the thin film keeps flat. When $\varepsilon_{\rm pre} > \varepsilon_{\rm c}$, the thin film buckles with a compress strain of $\varepsilon_{\rm c}$.^[363] The $\varepsilon_{\rm c}$ is $\approx 0.034\%$ for Si ribbons on a PDMS substrate.^[340]

It is worth noting that the wavelength does not depend on the prestrain, as indicated in Equation (10), which deviates evidently from the experimental measurements at large prestrain levels ($\varepsilon_{\rm pre} > 5\%$).^[363] This is because of the finite deformations and geometrical nonlinearities of the elastomeric substrates that are ignored in the small deformation theory. By considering those effects, the modified equations for the regime of large prestrain can be described as follows^[357,363,364]

$$\lambda = \frac{\lambda_0}{\left(1 + \varepsilon_{\rm pre}\right)\left(1 + \xi\right)^{1/3}} \tag{12}$$

$$A = \frac{A_0}{\left(\sqrt{1 + \varepsilon_{\rm pre}}\right) \left(1 + \xi\right)^{1/3}} \tag{13}$$

where $\xi = 5\varepsilon_{\text{pre}}(1 + \varepsilon_{\text{pre}})/32$; λ_0 and A_0 represent the wavelength and amplitude in the small prestrain regime, respectively. The



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Figure 9. Wave/wrinkle designs for soft electronics. a) Schematic illustration of two different buckling modes of Si ribbons on the PDMS substrate after releasing the prestrain. Reproduced with permission.^[353] Copyright 2008, American Institute of Physics. b) Schematic illustration of the wave profile: amplitude and wavelength. Reproduced with permission.^[376] Copyright 2010, Wiley-VCH. c) Experiments and simulations of the wave profile of buckled Si ribbons on the PDMS as a function of the prestrain. d) SEM images of bulked Si ribbons on a PDMS substrate. Reproduced with permission.^[363] Copyright 2007, National Academy of Sciences. e) A 45° titled SEM image of Si ribbons with controlled delamination on the PDMS with a prestrain level of 50%. Reproduced with permission.^[369] Copyright 2006, Nature Publishing Group. f) SEM images of 2D wavy Si ribbons on the PDMS. Reproduced with permission.^[369] Copyright 2009, American Vacuum Society. g) A confocal laser scanning microscope image of the patterned wrinkle with the customized design. Reproduced with permission.^[381] Copyright 2017, American Association for the Advancement of Science. h) A SEM images of the stretchable integrated circuits consisting of a set of complementary ring oscillators and isolated transistors. The inset shows the wavy Si-CMOS inverters on the PDMS formed with a prestrain of 2.7%. Reproduced with permission.^[401] Copyright 2008, American Association for the Advancement of Science.

new equations agree well with the experiments (Figure 9c), providing accurate prediction of wavelength and amplitude at large prestrain levels.^[363]

The wave/wrinkle configuration can dynamically respond to the applied strain ($\varepsilon_{applied}$).^[359,365–367] Its wavelength and amplitude as a function of the applied strain can be expressed as

$$\lambda' = \frac{\lambda_0 \left(1 + \varepsilon_{\text{applied}}\right)}{\left(1 + \varepsilon_{\text{pre}}\right) \left(1 + \varepsilon_{\text{applied}} + \xi'\right)^{1/3}}$$
(14)

$$A' = h_{\rm f} \frac{\sqrt{\left(\varepsilon_{\rm pre} - \varepsilon_{\rm applied}\right)/\varepsilon_{\rm c} - 1}}{\sqrt{1 + \varepsilon_{\rm pre}} \left(1 + \varepsilon_{\rm applied} + \xi'\right)^{1/3}}$$
(15)

where $\xi' = 5(\varepsilon_{\text{pre}} - \varepsilon_{\text{applied}})(1 + \varepsilon_{\text{pre}})/32$, and $\varepsilon_{\text{applied}}$ is positive for stretching and negative for compressing. The theory agrees well with the experimental data and finite element simulations, in which the wavelength increases but the amplitude decreases as the tensile strain goes up.^[363] Once the tensile strain reaches the prestrain, the amplitude of the thin film becomes zero, beyond which fracture will occur with additional stretch of ε_c and $\varepsilon_{\rm fracture}$.^[359] Thus, the total stretchability of this system is determined by $\varepsilon_{\rm pre} + \varepsilon_c + \varepsilon_{\rm fracture}$.^[368] For the case of Si ribbon on PDMS, as shown in Figure 9d, the maximum stretchability is \approx 31% given by the maximum prestrain of \approx 29% for Si with a fracture strain of \approx 1.8% and a critical strain of \approx 0.034%.^[225,340,359]

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Controlled delamination allows releasing part of the compressive strain energy into large degrees of bending of the thin film, and thus improves the stretchability (Figure 9e).^[369] The thin film delaminates off the substrate at defined sites during the substrate compression, leading to buckled structures.[368] The predefined delamination is realized by patterning adhesion sites between the thin film and the substrate.^[370] For the case of Si and PDMS, such patterned adhesion sites can be achieved by selectively activating the surface sites of Si and PDMS via UV/ ozone to create -OH groups that condense to form covalent -O-Si-O- linkages.^[369,371] The stretchability of the resulting structure can be over 100%, which is much dependent on the prestrain level.^[369] However, excessive prestrain would lead to a bending curvature that is too large to achieve, and thus thin film fractures. The wavelength is set to be the pitch of the selective surface activation sites. The amplitude depends on the adhesion patterns and the prestrain, described as^[366]

$$A = \begin{cases} \frac{2}{\pi (1 + \varepsilon_{\text{pre}})} \sqrt{W_{\text{in}} [W_{\text{in}} + W_{\text{act}} (1 + \varepsilon_{\text{pre}})] (\varepsilon_{\text{pre}} - \varepsilon_{\text{c}})}, & \varepsilon_{\text{pre}} > \varepsilon_{\text{c}} \\ 0, & \varepsilon_{\text{pre}} < \varepsilon_{\text{c}} \end{cases}$$
(16)

where $W_{\rm act}$ and $W_{\rm in}$ denote the widths of activated and inactivated sites, respectively; $\varepsilon_{\rm c} = \frac{h_{\rm f}^2 \pi^2}{3W_{\rm in}^2}$ is the critical strain to trigger the buckling. The controlled delamination also provides more precise geometry engineering of the structures compared with the fully bonding ones.^[364]

Multidirectional stretchability is necessary when stochastic and dynamic strain is present. The multidirectional stretchability can be realized by 2D wave/wrinkle structures, following a similar mechanism to the 1D wave/wrinkle structures.^[354] The 2D wave/wrinkle geometries can be fabricated by applying biaxial prestrain to the substrates (Figure 9f).^[350] The wave/ wrinkle morphologies can change among checkerboard,^[372] zigzag,^[373] labyrinth,^[374] and herringbone,^[375] depending on the anisotropy and sequence of the applied forces.^[372] Simulations show that the checkerboard morphology appears when the strain level is slightly above the critical strain and then transforms into labyrinths or herringbone subsequently as the strain increases.^[372,376] The herringbone is frequently observed in experiments because of its lowest system energy among these configurations.^[377]

Spatial distribution of wave/wrinkle structures can be controlled by inducing non-uniform strain in the substrate. The non-uniform strain can be formed by substrate relief structures via micromolding or patterned photopolymerization.^[378–381] For example, a customized wrinkle design was realized by precisely controlling ridge-guiding structures of the substrate.^[381] In this case, the ridge-guiding structure composed of a groove array was prepatterned on the microparticle surface through the selective photopolymerization of the photocurable polymer. Then, each ridge decision point was randomly transformed to either an ending, a bend, a straight line, or a bifurcation type during the wrinkling. According to the layout of the ridge-guiding sites, wrinkles were spatially aligned in predesigned directions. Figure 9g shows the confocal laser scanning microscope image of a representative pattern.^[381]

Except for the Si ribbons as mentioned above, this strategy can presumably be generalized to any materials. Demonstrated examples include metals,^[337] GaAs,^[382] PZT,^[383] ZnO,^[384] diamond,^[385] graphene,^[386] Si NWs,^[387] CNTs,^[388] and colloidal NPs,^[389] ranging from nanoscale to macroscale.^[390] These features of the wave/wrinkle structures collectively enable numerous applications in stretchable electronics such as stretchable conductors,^[391–397] field-effect transistors,^[382,398] photodetectors,^[369] p-n diodes,^[340] electrical switches,^[399] supercapacitors,^[400] and integrated circuits.^[401,402] For example, stretchable metal-semiconductor field-effect transistors (MES-FETs) based on GaAs ribbon were demonstrated (Figure 9h).^[382] The wavy GaAs ribbon can accommodate a large level of tensile strain (20%). The fabricated MESFETs do not change their electrical characterizations at the demonstrated strain level of 4.7% that is well beyond the fracture strain of conventional GaAs devices. In another example, enabled by the 2D wavy structural configurations, stretchable integrated circuits consisting of a set of complementary ring oscillators and isolated transistors can function well at different levels of strains applied in any directions in the circuit plane (Figure 9i).^[401] The oscillator frequencies change less than 7% under severe buckling deformation and tensile strain of 5%.

3.2. Island-Bridge

The island–bridge design is probably the most widely studied structure for stretchable electronics. It can achieve high stretchability by accommodating presumably any functional components at the islands that are connected by the conductive bridges.^[15] Under applied strain, the conductive bridges effectively deform both in plane and out of plane, while the islands remain still, thereby providing a mechanical isolation for the functional components on the islands.^[403] The bridges can be either coplanar or noncoplanar with the islands. The system level stretchability can be up to 1600%, depending on their initial designs.^[404] We will highlight representative bridge designs, including coplanar geometries (e.g., serpentine, self-similar, and spiral) and noncoplanar ones (e.g., arc-shape, non-coplanar serpentine, and helix), and discuss their underlying principles to understand and optimize the stretchability.

3.2.1. Serpentine Designs

Serpentine interconnections have the meander-shaped layout consisting of a number of periodically distributed unit cells that comprise two half-circles connected by the straight lines (Figure 10a).^[405] When responding to the applied strain, the serpentine interconnections can rotate in plane and also buckle out of plane, resulting in greatly reduced strains in the serpentine material per se, as well as low effective stiffness on the







Figure 10. Serpentine designs for soft electronics. a) Schematic illustrations of a typical serpentine structure and a unit cell defined by four geometric parameters: the ribbon width *w*, the arc radius *R*, the arc angle α , and the arm length *l*. b) Schematics of various serpentines derived by changing the four geometric parameters. b) Reproduced with permission.^[414] Copyright 2014, Elsevier. c) Schematic deformation modes (top panel) of freestanding serpentines under stretching determined by the *t/w* ratio. Critical strain (bottom panel) of separating the buckling modes of a serpentine interconnect under stretching at different unit cell numbers (*m*) and length/spacing ratios of the serpentine, respectively. Reproduced with permission.^[420] Copyright 2013, Royal Society of Chemistry. d) Stretchability versus *l/R* and α of fully bonded ITO serpentines. Reproduced with permission.^[413] Copyright 2015, Elsevier. e) Schematic crosssection (top image) and corresponding stretchability as a function of the core thickness (bottom image) of the encapsulated serpentine. Reproduced with permission.^[420] Copyright 2016, National Academy of Sciences. f) The stress distribution of single- and multi-trace serpentines under stretching according to the FEA. Reproduced with permission.^[430] Copyright 2007, IEEE.

system level.^[225] The serpentine patterns are typically fabricated by lithography-based microfabrication,^[406] cut-and-paste manufacturing,^[407] and laser ablation.^[408] Metals are the most commonly used for the serpentine interconnections due to their excellent conductivity and low cost.^[409,410] Other materials such as graphene,^[411] silicon,^[15] ZnO,^[384] and ITO^[412,413] can also be patterned into serpentine layouts for special purposes such as transparency^[411] and energy harvesting.^[384]

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The serpentine unit cell can be defined by four geometric parameters: the ribbon width *w*, the arc radius *R*, the arc angle α , and the arm length *l*, as shown in the bottom image of Figure 10a.^[414] Based on these parameters, a series of serpentine patterns can be derived to meet different requirements (Figure 10b).^[414] The serpentines can be freestanding,^[128] fully bonded to the substrate,^[415,416] or fully encapsulated inside the elastomer,^[417,418] and their mechanical behaviors could be qualitatively different, as discussed in the following sections.

Freestanding serpentines can be fabricated by selectively bonding the islands on the substrate while leaving the serpentine bridges on the substrate by van der Waals forces.^[419]The mechanics of freestanding serpentines have been extensively investigated via both experimental and theoretical approaches.^[414,420] Depending on the ratio of thickness to width (t/w), two different deformation modes can occur in the freestanding serpentines upon stretching (top image of Figure 10c):^[420] the in-plane deformation for large t/w (typically comparable to or larger than 1);^[421] simultaneous in-plane and out-of-plane deformations for small t/w (typically smaller than 1/5).^[403] The critical strain of separating the two deformation modes is linearly proportional to the square of the thickness/width ratio (t^2/w^2), as shown in the bottom image of Figure 10c.^[420]

In the case of in-plane deformation, a model based on the curved beam theory is used, in which the stretchability and effective stiffness of freestanding serpentines are governed by three dimensionless variables: α , l/R, and w/R, where the serpentine thickness t is assumed to be unit in the analysis model.^[414] Theoretical results show that the stretchability increases monotonically as the l/R increases and w/R decreases, and has a nonmonotonic dependence on α . As the α increases, the stretchability decreases first due to increased bending and then increases due to enhanced rotational contributions. The effects of all three variables on the effective stiffness are monotonic: the larger l/R, the smaller w/R, and the bigger α would yield the lower effective stiffness. These theoretical results are consistent with experiments. Since dimensionless variables are used, regardless if the serpentine sizes are macro-, micro-, or nanoscale, their normalized stretchability and effective stiffness will not have any differences as long as their α , w/R, and l/R are the same, assuming no size-dependent material properties.^[414] In the case of simultaneous in-plane and out-of-plane deformations, a semianalytic solution is proposed based on FEA. The results show that the stretchability increases with decreasing *t* and increasing period spacing and yield strain of the serpentine material.^[420]

The serpentines can be fully bonded to the substrate by depositing an adhesion layer (e.g., SiO_2) that forms strong covalent bonds with the substrate.^[419] The mechanical behavior of fully bonded serpentines is more complicated than the free-standing ones because of the constraints from the substrate.^[225]

Experiments and simulations have provided insights into the involved mechanics.^[412,413,422–424] For the case of Tegadermbonded ITO serpentines, the stretchability increases monotonically with decreasing the w/R. The stretchability increases first and then decreases with the increasing the l/R and α , with threshold values of 1 and 20°, respectively (Figure 10d).^[413] The nonmonotonic behavior is because of the strong substrate constraints: due to Poisson's contraction in the substrate, the long arm buckles transversely, which induces rupture in ITO. Strainto-rupture is reached before strain-to-debond when α is too large. The longer the arm, the earlier the rupture.^[413]

In addition to these usual geometric parameters (w/R), l/R, and α), the thicknesses of the serpentine ribbon and the substrate can largely affect the stretchability of fully bonded serpentines as well.^[423,425] With a certain substrate thickness, increasing the thickness of fully bonded serpentine ribbons changes the deformation modes of different levels of stretchability from wrinkling (out-of-plane deformation, wavelength < l/2), to buckling (out-of-plane deformation, wavelength > l/2), and to scissoring (in-plane deformation).[425] In the wrinkling regime, the stretchability increases first and then decreases on increasing the ribbon thickness. In the buckling regime, the stretchability increases monotonically first and then saturates with increasing the ribbon thickness. In the scissoring regime, the stretchability is independent of the ribbon thickness.^[425] With a certain ribbon thickness, the stretchability increases with reducing the substrate thickness but exhibits a two-stage, nonlinear behavior.^[423] The nonlinearity depends on the deformation modes separated by a critical length (l_{cr}) related to the substrate thickness. When $l < l_{cr}$ local wrinkling deformation occurs. The stretchability increases slowly with decreasing the substrate thickness. When $l > l_{cr}$, global buckling deformation occurs. The stretchability increases drastically with decreasing the substrate thickness.^[423] Decreasing the modulus of the substrate and the serpentine would increase the stretchability due to the reduced bending stiffness.^[423]

Encapsulated serpentines that are bonded from both the bottom and top surfaces are valuable in that the encapsulation provides necessary mechanical protection and environmental barriers to the substrate.^[426] The device can stay at the neutral mechanical plane of a core-shell structure, which results in reduced bending strain in the device, as illustrated in the top images of Figure 10e.^[417] Usually, an ultralow modulus core material (such as Silbione with a Young's modulus of 5 kPa) exerts minimal constraints to the serpentine deformations, and a higher modulus thin shell layer (such as Ecoflex with a Young's modulus of 60 kPa) provides mechanical support.^[417] The mechanical behavior of encapsulated serpentines strongly depends on their layouts, as in the case of freestanding and fully bonded ones.^[425,427] Additionally, the core modulus and thickness play a critical role in the stretchability of the system.^[417,428] Intuitively, lower core modulus leads to higher stretchability due to reduced constraints to the serpentine deformations. The extreme would be a modulus of 0 Pa, which corresponds to the freestanding case. The stretchability increases with increasing the core thickness due to the diminishing constraint from the rigid shell, as shown in the bottom image of Figure 10e.^[428] With further increasing the core thickness, the stretchability then saturates at a critical value that corresponds to the case when the serpentines are encapsulated with only the soft core (without the rigid shell). This critical core thickness decreases as the serpentine thickness increases because the deformation of stiffer serpentine is more dominated by their intrinsic properties rather than the surrounding environment.^[428]

A general conclusion is that the narrower *w*, the thinner *t*, the larger *R* and α , and the longer *l* will normally yield larger serpentine stretchability. In practice, multitrace instead of the single-trace serpentines are often used, which exhibit an improved stretchability without sacrificing their electrical performances.^[429–431] The maximum internal stress upon stretching reduces drastically from the single-trace of 1942 MPa to the multitrace of 230 MPa, as indicated by the FEA results in Figure 10f.^[430] However, manufacturing challenges should be considered in practical designs, such as the width limits due to fabrication resolution and the arc angle limits due to possible overlap between adjacent unit cells if the arc angle is too large.

3.2.2. Self-Similar Structures

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In addition to the stretchability, another critical factor in the island–bridge design is the filling ratio of the islands, which determines the areal density of the functional components. Filling ratio is central for applications in optoelectronics and energy devices, which require high sensor integration level and energy density.^[432–434] Given by the relationship^[435]

System stretchability = $(1 - \sqrt{\text{filling ratio}})$ × (interconnection stretchability)

the filling ratio and system stretchability compete with each other. Bridges with self-similar layouts enable large stretchability and high functional density in the system at the same time.^[436] Self-similar bridges refer to a pattern where subdivision into small sections yields pieces with geometries that resemble the whole.^[298,436–439] This pattern can make full use of the limited space by increasing the self-similar order.^[367]

An example of self-similar bridges with an order n up to 4 is shown in **Figure 11a**.^[440] The simple serpentine is used as the first order. The *n*th-order structure can be created by reducing the scale of the (n - 1)th-order one, followed by 90° rotation, and then connecting their multiple copies in a fashion that reproduces the layout of the original geometry.^[441] The spacing and height at any order can be scaled from the spacing of the highest order by^[435]

$$h^{(i)} = \eta \left(\frac{\eta}{2m}\right)^{(n-i)} l^{(n)}, \quad l^{(i)} = \left(\frac{\eta}{2m}\right)^{(n-i)} l^{(n)} \quad (i = 1, \dots, n)$$
(17)

where $h^{(i)}$ and $l^{(i)}$ are the height and spacing of the *i*th-order self-similar structure, respectively; η is the ratio between height and spacing, which is the same for each order; *m* is the number of unit cells at each order. Upon stretching, the *n*th-order structure unravels first while the (n - 1)th-order essentially does not deform because higher orders have a larger torque and are easier to deform than lower orders.^[441] Only after the *n*th order is fully extended does the (n - 1)th order start to unravel and contribute to the stretchability in succession. The process

continues until the first-order structure unravels.^[441] A demonstration of this concept using a two-order self-similar bridge is shown in Figure 11b, in which the second-order and first-order structures unravel sequentially, corresponding to a stretchability of 150% and 300%, respectively.^[433]

The ordered unraveling mechanism of the freestanding selfsimilar bridges is illustrated by a simple but effective hierarchical computational model (HCM).^[435] In this model, collapsed selfsimilar structures are assumed as straight beams because of their much smaller width and thickness than length. The assumption substantially reduces the modal complexity and computational costs. The straight beams would remain undeformed during the unraveling of structures of higher orders. The elastic stretchability (defined by the onset of plastic deformations) increases ≈200 times, from ≈10.7% for the first order, to ≈2140% for the fourth order, which agrees very well with results by the conventional FEA for $n \le 3$, as shown in Figure 11c.^[435] The missing FEA data point at n = 4 is probably because of the extremely long computational time needed. It took 2340 min in the postbuckling analysis of the third-order self-similar structures by FEA as compared with ≈146 min by HCM. This result indicates that high-order selfsimilar structures can substantially improve the elastic stretchability of the bridges. The tensile stiffness of freestanding self-similar bridges is determined by analytical solutions, which are verified by the experiments. The results show that normalized tensile stiffness decreases with increasing the self-similar order and ratio of total length to apparent length of the self-similar structure.^[442]

The deformation behaviors of partially or fully bonded selfsimilar bridges have been studied by combining FEA and experiments,^[439,443] which are much more complicated than the freestanding case. Figure 11d shows the high consistency between the FEA predictions and the experimental outcomes of various fully bonded self-similar layouts, ranging from lines (Koch, Peano, and Hilbert) to loops (Moore and Vicsek) and branch-like meshes (Greek cross).^[439] The diversity of self-similar topologies enables versatile electronic applications by integrating and interdigitating multiple self-similar layouts.^[439]

The ordered unraveling mechanism plays a critical role in enhancing elastic stretchability of the self-similar bridges, far beyond a simple increase of the total bridge length.^[435] Figure 11e shows the comparison of strain levels between the self-similar and conventional serpentine bridges. ε_{max} is the maximum principal strain in the metal layer and $\varepsilon_{\rm appl}$ is the externally applied strain.^[433] With the same total length, amplitude, width, thickness, and island span, the self-similar bridge shows nearly two times elastic stretchability over the conventional serpentine bridge (528% vs 284%).^[433] The concurrent large stretchability and high island filling ratios of self-similar structures enable various applications.^[444,445] For example, stretchable lithium-ion batteries with filling ratio up to ≈50% and system stretchability up to ≈300% have been demonstrated,^[433] which are much higher than those made by the conventional serpentine bridges (\approx 30% system stretchability at a filling ratio of 50%).^[443,446,447]

3.2.3. 2D Spirals

2D spiral refers to a planar curve that winds around a fixed center point at a continuously increasing or decreasing distance







Figure 11. Self-similar designs for soft electronics. a) Schematic illustrations of a self-similar bridge with an order up to 4. Reproduced with permission.^[435] Copyright 2014, Elsevier. b) Experimental and computational studies to show the ordered unraveling mechanism of a self-similar bridge upon stretching. Reproduced with permission.^[433] Copyright 2013, Nature Publishing Group. c) Elastic stretchability versus the order of self-similar bridges by the FEA and the hierarchical computational model. Reproduced with permission.^[435] Copyright 2014, Elsevier. d) Representative self-similar topologies and their FEA and MicroXCT images under elastic tensile strain. Reproduced with permission.^[439] Copyright 2014, Nature Publishing Group. e) Comparison of principal strain levels in the serpentine and self-similar bridges with the same total length, amplitude, width, thickness, and island span. Reproduced with permission.^[433] Copyright 2013, Nature Publishing Group.

from the center point. Spiral-inspired bridges represent another captivating structure toward stretchable electronics due to their larger stretchability compared with conventional serpentines with the same cover area and contour length.^[448,449] By unwinding the spiral arms upon stretching, the bridge can absorb externally applied strain and keep the islands with functional elements strain-free (Figure 12a).^[450] Two types of spiral bridge designs have been introduced, as shown in the top^[451] and bottom^[452] panels of Figure 12b, respectively. These two designs show preferential deformation behavior, i.e., in-plane^[451] and out-of-plane^[453,454] deformations, during the unwinding process. Critical factors that determine stretchability of the spiral bridges include the width, radius, and the turn number of the spiral arms, which can easily be tuned.^[455] The ratio of width to radius must be considered in the design, which is limited by the maximum principal strain ε_{max} in the spiral arm: $\varepsilon_{\text{max}} = w/2R$, where *w* is the width of the spiral arm and *R* is the radius of the innermost circle.^[451] For example, with a maximum allowable principal strain of 1% for silicon, the spiral arm must be 50 times thinner than the radius of the innermost circle to avoid Si fracture when straightened.^[450,456]

By the spiral design, silicon networks with excellent lateral stretchability of more than 1000% and area expansion as high as 30 folds have been demonstrated. This set of attributes enables the network to completely conform to highly curved and nonlinear surfaces (Figure 12c).^[450] Structural optimizations can improve the robustness of the spiral bridges. Upon stretching, the strain is found to be localized at the starting and ending points of the spiral arms, as validated by the FEA.^[450] Localized high strain points would cause material fracture and reduce the overall stretchability. To solve this problem, fillet structures with gradient width are added at the two terminals of







Figure 12. Spiral designs for soft electronics. a) Optical images of three hexagons connected with spiral bridges before and after stretching to 170%. Reproduced with permission.^[450] Copyright 2014, American Institute of Physics. b) SEM images of two different spiral designs, corresponding to inplane (top) and out-of-plane (bottom) deformations upon stretching, respectively. Top image, reproduced with permission.^[451] Copyright 2007, IEEE. Bottom image, reproduced with permission.^[452] Copyright 2016, IOP Publishing. c) Optical images of hexagon islands connected with a spiral bridge array on a curvilinear surface. Reproduced with permission.^[450] Copyright 2014, American Institute of Physics. d) Stress and strain distribution comparison between the spiral with and without fillet structures along the arms. e) Stress and strain distribution comparison between the spiral and unequal serpentine arms. f) Stress and strain distribution comparison between the original spiral and the optimized compound spiral. Reproduced with permission.^[457] Copyright 2017, Elsevier. g) An optical image of the double spiral bridges when stretched. Reproduced with permission.^[456] Copyright 2015, SPIE. h) Experimental and numerical results of separation versus displacement of the islands when stretched. Insets show the sample images before and after stretching. i) Von Mises stress distribution of a square array with crosslinked spiral bridges under stretching. Reproduced with permission.^[458] Copyright 2017, American Institute of Physics.

the spiral arms (the top image in Figure 12d), minimizing the strain localization.^[457] Von Mises analysis of stress and strain distribution in the spiral arms with and without the triangular ends is shown in the middle and bottom panels of Figure 12d. By the fillet structure, the maximum stress is reduced from 8698 to 4284 MPa, and strain is reduced from 4.23% to 1.49%, respectively.^[457]

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Adding serpentine structures to the spiral arms helps enhance the stretchability. Two versions of serpentines (with equal or unequal halves) have been investigated, as shown in the top panel of Figure 12e.^[457] Both of the two serpentine arms show reduced stress and strain compared with their planar counterparts. The serpentine with equal halves is slightly better than that with unequal halves, as indicated by the Von Mises analysis of stress and strain in the middle and bottom panels of Figure 12e. Specifically, the stress and strain are reduced to 2265 MPa and 0.87% for the equal halves and 2309 MPa and 0.93% for the unequal halves, respectively.^[457] Replacing the half-circle-based serpentine arms with horseshoe-shaped serpentines at the terminals further reduces the stress and strain localization (the top panel of Figure 12f).[457] The stress and strain are reduced to 1779 MPa and 0.69% for this optimized compound spiral, respectively, which represent 58.47% and 53.37% less than the conventional spiral (middle and bottom panels of Figure 12f).^[457]

Double spiral bridges can enhance the device robustness by providing a backup conduction pathway in case one fails (Figure 12g).^[456] However, adding an extra bridge pays the price of reducing the functional component density. Besides structural design, functional surface treatments of the spiral arm sidewalls can also improve the structure robustness. Silicon spirals with smooth surfaces, obtained through thermal oxidation followed by HF etching, result in higher yield strain compared with those with rough surfaces.^[451] The stress and strain of spirals under stretching change linearly with the prescribed displacement as demonstrated by both numerical simulations and experiments (Figure 12h). And the spiral-island geometrical layouts largely influence their deformation behavior.^[458] Among different distribution layouts such as hexagon, diamond, square, and square with crosslinked spiral, the square layout with crosslinked spiral enables the best areal density and mechanical robustness, as indicated by the numerical simulation results in Figure 12i.^[458]

3.2.4. Arc-Shaped Bridges

3D bridges that freely deform out of plane in response to the applied strain can save real estate for functional components.^[434,459] One of the approaches is the arc-shaped bridge. By transferring the island–bridge structure on a prestrained elastomeric substrate with the island fully bonded while the bridge free standing by van der Waals forces, arcshaped bridges are formed after releasing the prestrain in the substrate (**Figure 13**a).^[460] Similar to the wrinkled structures, both global and local buckling modes can occur in the bridges, depending on the prestrain magnitude and the adhesion energy between the bridge and the substrate (Figure 13b).^[461] The involved mechanics have been well studied, which provide a simple criterion for predicting the buckling patterns of the bridges on an arbitrary surface.^[462] These arc-shaped bridges can reversibly bend up or down to absorb the corresponding compressive or tensile forces, and can twist and shear in response to more complex deformations (Figure 13c), leaving the islands almost intact.^[463]

The stretchability of arc-shaped bridges depends on the wavelength (L^0_{bridge} is the initial bridge length) and the amplitude (*A*), which is determined by the minimization of total energy.^[464] The total energy of the bridge consists of two parts: the bending energy and the membrane energy. Minimization of total energy in the bridge gives the amplitude

$$A = \frac{2L_{\text{bridge}}^0}{\pi} \sqrt{\left(\frac{L_{\text{bridge}}^0 - L_{\text{bridge}}}{L_{\text{bridge}}^0} - \varepsilon_c\right)}$$
(18)

where $\varepsilon_c = \pi^2 h_{\text{bridge}}^2 / [3(L_{\text{bridge}}^0)^2]$ is the critical strain for the buckling of the bridge and h_{bridge} is the bridge thickness.^[464] When $(L_{\text{bridge}}^0 - L_{\text{bridge}})/L_{\text{bridge}}^0 < \varepsilon_c$, the bridge does not buckle and the membrane strain is $\varepsilon_{\text{bridge}}^{\text{membrane}} = -(L_{\text{bridge}}^0 - L_{\text{bridge}})/L_{\text{bridge}}^0$. When $(L_{\text{bridge}}^0 - L_{\text{bridge}})/L_{\text{bridge}}^0 > \varepsilon_c$, the bridge buckles up and the membrane strain becomes a constant $-\varepsilon_c$, and the bending strain kicks in and increases with the degrees of deformation given by $\varepsilon_{\text{bridge}}^{\text{bending}} = 2\pi (h_{\text{bridge}}/L_{\text{bridge}}^0) \times \sqrt{(L_{\text{bridge}}^0 - L_{\text{bridge}})/L_{\text{bridge}}^0 - \varepsilon_c}$. The maximum strain in the bridge is the sum of the bending strain and membrane strain

$$\varepsilon_{\text{bridge}}^{\text{max}} = 2\pi \left(h_{\text{bridge}} / L_{\text{bridge}}^{0} \right) \times \sqrt{\left(L_{\text{bridge}}^{0} - L_{\text{bridge}} \right) / L_{\text{bridge}}^{0} - \varepsilon_{c}} + \varepsilon_{c}$$

$$\approx 2\pi \left(h_{\text{bridge}} / L_{\text{bridge}}^{0} \right) \times \sqrt{\varepsilon_{\text{pre}} / (1 + \varepsilon_{\text{pre}})}$$
(19)

where $\varepsilon_{\rm pre} = (L_{\rm bridge}^0 - L_{\rm bridge})/L_{\rm bridge}$ is the prestrain in the substrate, and the approximation is valid when $(L_{\rm bridge}^0 - L_{\rm bridge})/L_{\rm bridge}^0 \gg h_{\rm bridge}^2/(L_{\rm bridge}^0)^2$.^[464] Equation (19) indicates that the maximum strain in the bridge is proportional to its thickness to length ratio. Therefore, thin and long bridges give small strain and thus large stretchability.

When subjected to applied strain, the bridge wavelength changes from L_{bridge} to \dot{L}_{bridge} (Figure 13a). The applied strain can be expressed as $\varepsilon_{\text{app}} = (\dot{L}_{\text{bridge}} - L_{\text{bridge}}) / (L_{\text{bridge}} + L_{\text{island}}^0)$, where L_{island}^0 is the island length and is assumed to be unchanged when the bridge is deformed.^[464] The maximal stretchability ($\varepsilon_{\text{stretchability}}^{\text{max}}$) of the system is the applied strain when the bridge returns to a flat state ($\dot{L}_{\text{bridge}} = L_{\text{bridge}}^0$ and $\varepsilon_{\text{app}} = \varepsilon_{\text{stretchability}}^{\text{max}}$)

$$\varepsilon_{\text{stretchability}}^{\text{max}} = \frac{\left(L_{\text{bridge}}^{0} - L_{\text{bridge}}\right)}{\left(L_{\text{bridge}} + L_{\text{island}}^{0}\right)} = \frac{\varepsilon_{\text{pre}}}{1 + \left(1 + \varepsilon_{\text{pre}}\right)\frac{L_{\text{island}}^{0}}{L_{\text{bridge}}^{0}}}$$
(20)

The message is that long bridge, short island, and large prestrain increase the maximal stretchability of the system. And the maximal stretchability is equal to the prestrain $\varepsilon_{\rm pre}$ when $L_{\rm bridge}^0 \gg L_{\rm island}^0$ and becomes $(L_{\rm bridge}^0/L_{\rm island}^0) [\varepsilon_{\rm pre}/(1+\varepsilon_{\rm pre})]$ when $L_{\rm bridge}^0 \ll L_{\rm island}^{0}$.^[464]

In practical applications, an encapsulation layer can provide protection to the device. The arc-shaped bridges embedded in a compliant elastomeric material such as PDMS have the similar mechanics as the freestanding ones, but with slightly decreased



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Figure 13. Arc-shaped designs for soft electronics. a) Schematic mechanical model for the island–arc-shaped bridge structure. b) Colorized SEM images of locally and globally buckled arc-shaped bridges. b) Reproduced with permission.^[461] Copyright 2009, Wiley-VCH. c) Different deformation modes of arc-shaped bridges under the applied strain: bending, diagonal stretching, and twisting. Reproduced with permission.^[463] Copyright 2008, National Academy of Sciences. d) SEM images of the arc-shaped bridges facing down into the trenches of the substrate before and after encapsulating. Reproduced with permission.^[459] Copyright 2011, Wiley-VCH. e) An SEM image of silicon islands connected with the double arc-shaped bridges. Reproduced with permission.^[466] Copyright 2010, Wiley-VCH. f) An SEM image of the irregular arc-shaped bridge for the amplifier applications. Reproduced with permission.^[463] Copyright 2008, National Academy of Sciences. g) A hemispherical electronic eye enabled by the island–arc-shaped bridge structure, showing a larger field of view, better illumination uniformity, and improved focusing behavior compared with the planar device. Reproduced with permission.^[470] Copyright 2008, Nature Publishing Group. h) Arc-shaped structures prepared by the multidirectional writing method. Left: An optical image of bonding Cu wires with a submicrometer diameter. Reproduced with permission.^[472] Copyright 2010, American Association for the Advancement of Science. i) A stretchable UV–vis–NIR photodetector array based on arc-shaped PbS-P3HT hybrid NWs prepared via the multidirectional writing method. Reproduced with permission.^[473] Copyright 2015, Wiley-VCH.

stretchability due to the restricted deformation from the encapsulation.^[463,465] Most notably, substrate trenches can be fabricated to accommodate the arc-shaped bridge (Figure 13d).^[459] In this case, the encapsulation layer has negligible influence on the mechanical behavior of the bridge. Double bridges can be used to enhance robustness of the system (Figure 13e), but the tradeoff between the system robustness and filling ratio need to be revisited.^[466] Irregular/asymmetrical designs have been demonstrated for the niche applications (Figure 13f).^[463] Based on the arc-shaped bridges, diverse system-level applications have been demonstrated such as high-resolution hemispherical electronic eyes with a larger field of view, better illumination uniformity, and improved focusing behavior compared to the planar device (Figure 13g);^[432] and stretchable solar cells with a filling ratio of ~67% and a system level stretchability of 60%.^[467]

Besides transfer printing,^[468] multidirectional writing methods have been used to "write" the arc-shaped bridges based on various

kinds of functional materials such as Ag NP inks, copper and platinum electrolytes, and PEDOT:PSS.^[469–471] This technique offers not only high stretchability but also accurately controlled material, size, shape, and position of each bridge (Figure 13h).^[470,472] A UV–vis–NIR photodetector array based on arc-shaped PbS-P3HT hybrid NWs has been demonstrated, which exhibits no performance degradation under 100% strain and up to 100 cycles of stretching (Figure 13i).^[473] Please note that in this case the electronic roles of islands and bridges swap with each other.

3.2.5. Noncoplanar Serpentines

Noncoplanar serpentines can be further stretched after they become planar under tensile strain,^[419,466] and therefore provide larger stretchability than the arc-shaped bridges. Noncoplanar serpentines can be fabricated following a similar process to the arc-shaped bridges (**Figure 14**a).^[463,474] The different layout design results in different mechanics.^[419,463] Figure 14b shows a system built with 35% prestrain, in which the serpentine returns to a geometry close to its original state at an applied strain of 40%, and can deform further to 70% strain through rotation and out-of-plane buckling of the serpentine.^[463] The strain distribution evaluated by the FEA reveals that, even under the tensile strain of 70%, the peak principal strain in the bridges is well below their fracture strain (Figure 14c).^[463]

The prestrain strategy can also be applied to the fully bonded serpentines, enabling a larger stretchability compared with the bonded and freestanding coplanar designs without the



prestrain.^[475] Figure 14d shows SEM images of the serpentines fully bonded on Ecoflex substrates before and after releasing the prestrain of 160%.^[475] The mechanics of the compressive serpentines under different prestrain levels have been studied by both FEA and experiments.^[475] The compressive noncoplanar serpentine with a prestrain level of 85% shows a total stretchability of 283%, which is more than two times higher than that (137%) of the coplanar serpentines with the same layout (Figure 14e), and the result agrees well with the FEA simulations. More importantly, both cyclic fatigue testing and FEA simulations show that the noncoplanar serpentine can sustain cyclic loading of 185% strain for about 25 000 cycles, while the coplanar serpentine can only sustain a much lower cyclic loading of 53%.^[475]

The analytic model and experiments show that the serpentine arms buckle into wavy shapes under stretching due to the Poisson effect, and the buckled wavelength is given by

$$\lambda = \frac{2\pi \left(\frac{4S_{\text{serpentine}}}{\overline{E}_{\text{sub}}}\right)^{\frac{1}{3}}}{\sqrt{1 + \varepsilon_{\text{app}}} \left[1 + \frac{5}{32} \left(1 + \varepsilon_{\text{app}} - \sqrt{1 + \varepsilon_{\text{app}}}\right)\right]^{\frac{1}{3}}}$$
(21)

where \overline{E}_{sub} is the plane-strain Young's modulus of the substrate, and $S_{serpentine}$ is the effective bending stiffness of the serpentine determined by its Young's modulus and the thickness.^[368,475] Equation (21) indicates that the buckled wavelength increases with increasing the serpentine thickness. For thin serpentines,



Figure 14. Noncoplanar serpentines for soft electronics. a) An SEM image of an array of stretchable CMOS inverters with serpentine noncoplanar bridges. b) Optical images of stretching tests for the noncoplanar serpentine bridge. c) FEA simulation results of a noncoplanar serpentine bridge before (35% prestrain) and after (70% applied strain) stretching. a–c) Reproduced with permission.^[463] Copyright 2008, National Academy of Sciences. d) SEM images of fully bonded serpentines before and after releasing the prestrain of 160%. e) Stretchability comparison of serpentines with and without a prestrain of 85%. f) Dependence of the buckling wavelength on the metal thickness based on experiments, numerical simulations, and analytic models. d–f) Reproduced with permission.^[475] Copyright 2014, Wiley-VCH.





 λ is smaller than the length of serpentine arm, so that the local buckling is observed. For thick serpentines, global buckling occurs because λ is larger than the arm length. For Cu serpentines sandwiched by two layers of 1.2 µm thick polyimide (PI), the calculated critical thickness that separates the two buckling modes based on the Equation (21) is 0.9 µm, which is slightly underestimated as compared with experiments and FEA results (Figure 14f).^[475]

3.2.6. Helices

Helices have very small physical coupling to the substrate and can effectively suppress strain concentration, which leads to highly elastic mechanical behavior and very compact design of the system.^[476] Under the applied strain, the helix behaves like a spring/telephone cord that meritoriously absorbs the strain (**Figure 15**a).^[365] A variety of approaches have been developed for fabricating helical structures, including 3D additive printing (Figure 15b),^[477–480] direct laser writing,^[481] manual winding,^[482] microcontact printing,^[483,484] mold processing (Figure 15c),^[485,486] chemical synthesizing,^[487,488] and compressive buckling.^[489] Among them, 3D additive printing and direct laser writing can only be applied to certain classes of materials such as metals and/or polymers, but not to other advanced materials such as single-crystalline semiconductors.^[490] Manual winding and microcontact printing are



Figure 15. Helices for soft electronics. a) Schematic illustration of helical structures for stretchability. Reproduced with permission.^[365] Copyright 2007, Royal Society of Chemistry. b) An SEM image of 3D helices made of CNT composites via 3D additive printing. Reproduced with permission.^[480] Copyright 2012, IOP Publishing. c) Photographs of a helical conductor prepared by molding Cu NW composites. Inset shows the screw mold. Reproduced with permission.^[485] Copyright 2014, Nature Publishing Group. d) Working principle and results of compressive buckling. Top: FEA results illustrate the formation of 3D helices from 2D serpentine Si ribbons bonded at selected points to a stretched slab of silicone elastomer. Bottom: An SEM image of as-fabricated helices. Reproduced with permission.^[489] Copyright 2015, American Association for the Advancement of Science. e) Schematic mechanics model and coordinate system for buckling-guided helices. Reproduced with permission.^[490] Copyright 2016, Wiley-VCH. f) Optical images of a soft electronic system at a tensile strain of 50%, consisting of ≈250 3D helical bridges and ≈50 discrete component chips. Reproduced with permission.^[476] Copyright 2017, Nature Publishing Group.

difficult to offer large-scale throughputs.^[491] Mold processing is challenging to achieve the sub-micrometer level of feature resolutions. Chemical synthesizing typically has a low yield and modest geometrical and compositional control.^[365] In comparison, mechanical buckling provides a strategy to forming helical structures over large areas out of a wide range of materials (from soft polymers to rigid inorganic semiconductors) and dimension scales (from nanometer to centimeter).^[490,492]

Figure 15d presents the schematic assembly procedures for controlled buckling of the helical structures.^[489] Patterned 2D serpentine precursors with predefined bonding sites are prepared by the conventional microfabrication process and then transferred onto a prestrained substrate. Releasing the prestrain in the substrate triggers simultaneous in-plane and out-of-plane translational and rotational motions of the nonbonded regions of the 2D serpentine precursor, which leads to the formation of 3D helical structures. Specifically, each unit cell of the 2D serpentine precursor transforms into a single turn of the corresponding 3D helical structure. The corresponding FEA simulation indicates the small physical coupling between the helix structure and the substrate.^[489] An SEM image of a pair of helices made of monocrystalline silicon is shown in the bottom panel of Figure 15d.^[489]

Based on the principle of minimizing the total strain energy, a theoretical model has been developed for predicting the geometry and analyzing the postbuckling behaviors of the helix.^[490] In this model, only a unit cell in the 2D serpentine precursor is analyzed due to the structural periodicity. As shown in Figure 15e, the unit cell consists of two identical arcs, each with a radius of R and top angle of θ_0 (typically $0 < \theta_0 < 5\pi/4$).^[490] A Cartesian coordinate system (X, Y, Z), with the origin at the joint of two arcs, determines the coordinates of every point on the serpentine by $\mathbf{r}_0(\theta) = X_0(\theta)\tilde{\mathbf{E}}_1 + Y_0(\theta)\tilde{\mathbf{E}}_2 + Z_0(\theta)\tilde{\mathbf{E}}_3$, where $\tilde{\mathbf{E}}_i$ (i = 1, 2, 3) are the unit vectors. After deformation, r_0 moves to $\mathbf{r} = \mathbf{r}_0 + \mathbf{U} = X_0(\theta)\tilde{\mathbf{E}}_1 + Y_0(\theta)\tilde{\mathbf{E}}_2 + Z_0(\theta)\tilde{\mathbf{E}}_3$, and the unit vectors become $\mathbf{e}_i = d\mathbf{r}/ds$, where U is the axial displacement and s is the arc length after deformation. The e_i is related to the curvature vectors $\mathbf{\kappa}_i$ by^[493] $\mathbf{e}'_i / \lambda = \mathbf{\kappa}_i \times \mathbf{e}_i$, where ()' = d()/dS; $\lambda = ds/$ dS is the stretch along the central axis; κ_1 and κ_2 denote the curvatures in the (e_2, e_3) and (e_1, e_3) surfaces, respectively; and the twisting curvature κ_3 is related to the twist angle Φ of the cross-section by $\kappa_3 = \Phi' / \lambda$.^[490] Moreover, the work conjugate of the bending moment and torque is $\hat{\kappa} = \lambda \kappa$,^[494] governed by the relations below

$$\hat{\kappa}_{1} = -U_{2}^{"} + \left(\Phi - \frac{U_{1}}{R}\right)U_{1}^{"} - \frac{U_{1}^{'}}{2R} + \frac{U_{3}^{'}}{R} - \frac{1}{2R}\left(\Phi - \frac{U_{1}}{R}\right)^{2} + \frac{1}{R}$$

$$\hat{\kappa}_{2} = U_{1}^{"} - \frac{R\Phi - U_{1}}{R^{2}}$$

$$\hat{\kappa}_{3} = \Phi'$$
(22)

This equation describes the distribution of curvature along the 3D helix structure. $\ensuremath{^{[490]}}$

The total strain energy in a serpentine unit cell mainly consists of in-plane bending energy, out-of-plane bending energy, and twisting energy, which can be written as



$$\Pi_{\text{total}} = EI_1 \int_0^{\theta_0 R} \left(\hat{\kappa}_1 - \frac{1}{R} \right)^2 dS + EI_2 \int_0^{\theta_0 R} \hat{\kappa}_2^2 dS + GI_p \int_0^{\theta_0 R} \hat{\kappa}_3^2 dS$$
$$= EI_2 R \left[\left(\frac{w}{t} \right)^2 \int_0^{\theta_0} \left(\hat{\kappa}_1 - \frac{1}{R} \right) d\theta + \int_0^{\theta_0} \hat{\kappa}_2^2 d\theta + \frac{2}{1+\nu} \int_0^{\theta_0} \hat{\kappa}_3^2 d\theta \right]$$
(23)

where $EI_1 = (Ew^3t)/12$ and $EI_2 = (Ewt^3)/12$ are the in-plane and out-of-plane bending stiffness, respectively; $GI_p \approx (Gwt^3)/3$ is the twisting stiffness for thin serpentine ribbons; and ν is the Poisson ratio.^[490] For any prescribed geometrical parameters $(w/t, w/R, \text{ and } \theta_0)$ and strain $\varepsilon_{app} = \varepsilon_{pre}/(1 + \varepsilon_{pre})$, the minimization of the total energy determines the displacement components (Φ , U_1 , U_2 , and U_3) and thus the coordinates of every point in the serpentine microstructure. Therefore, Equation (23) provides the full 3D helical structure during postbuckling.^[490] It also indicates that the precursor modulus has no impact on the 3D helical structure geometry.

The maximum strain in the helix can be expressed as^[490]

$$\varepsilon_{\max} = F_2(\theta_0) \frac{t}{R} \sqrt{\varepsilon_{\text{app}}}$$
(24)

where $F_2(\theta_0)$ is a function of θ_0 . Equation (24) indicates that the maximum strain is determined by the applied strain (ε_{app}), the arc angle (θ_0), and the normalized thickness (t/R), while it is independent of other parameters such as the Poisson ratio (ν) and normalized width (w/R), which agree well with the FEA results.^[490]

A representative application of the helices is shown in Figure 15f, consisting of ≈250 3D helices for electrical interconnects and ≈50 discrete component chips for multifunctional sensing, electrophysiological recording, and wireless energy and data transmission.^[476] This soft electronic system, with elastomeric encapsulation, has a radial stretchability of 70% and a uniaxial stretchability of \approx 110. It is worth noting that encapsulation in this case undergoes a two-step process: elastomeric precursor infiltration at a state of partial release of the prestrain (ε_{encap}), and precursor crosslinking followed by completely releasing the prestrain. This strategy significantly improves the elastic stretchability of the system compared with the conventional encapsulation method that takes place when the prestrain is fully released (\approx 2.2 and 2.8 times more stretchable for uniaxial and radial stretching, respectively, when $\varepsilon_{\rm encap} = 30\%$ and $\varepsilon_{\rm pre} = 150\%$).^[476]

3.3. Origami

Despite the extraordinary success of the wave/wrinkle and island–bridge designs, the involvement of elastomeric substrates in the fabrication process poses challenges to align with the mainstream manufacturing infrastructures.^[495] Origami does not involve elastomeric substrates and is compatible with the well-established manufacturing processes. It provides a potentially game-changing approach to soft electronics with high throughout, large deformability, and comparable performance to rigid electronics.^[496] Origami is an ancient art of paper folding with *ori-* meaning fold and *-gami* meaning paper,^[497] which can transform a planar sheet into complex 3D





structures according to the defined hinge crease patterns. The crease pattern, with a periodic array of geometrically and elastically coupled mountain and valley folds (**Figure 16**a), allows the 3D structure to fold and unfold simultaneously without inducing large strain to the facets (functional areas of soft electronics) between the creases (Figure 16b).^[498,499]

The crease pattern design dictates the folds necessary to achieve the desired 3D origami structure.^[500] A series of mathematical algorithms have been developed to design the crease pattern,^[501–503] the most popular being the tree method.^[504–506] This tree method converts the crease pattern into some desired uniaxial base by finding the folding pattern of the smallest possible square. The projection of the uniaxial base onto a common plane is a shadow tree, to express the desired shape.^[500,507] Based on this idea, a program named *TreeMaker* has been developed that can generate the crease pattern for customized origami structures.^[508] Moreover, mathematical studies of origami have led to a topic called "computational origami" in recent computer science, which focuses on efficient algorithms for solving problems such as origami foldability and origami design.^[509,510]

Traditional manufacturing methods for assembling origami structures mainly rely on manual or machine-assisted folding, which is challenging to implement on small scales and advanced materials.^[511,512] Self-folding ranging from nanoscale^[513,514]



Figure 16. Origami for soft electronics. a) Geometry of the Miura-ori structure with mountain and valley folds. Reproduced with permission.^[540] Copyright 2016, Nature Publishing Group. b) Schematic illustration of the folding and unfolding processes for the Miura-ori structure. Reproduced with permission.^[499] Copyright 2017, Nature Publishing Group. c) Schematic illustration of the self-folding process of Al₂O₃ facets with Au patterns driven by the capillary force. Reproduced with permission.^[513] Copyright 2011, Wiley-VCH. d) Photographs of the Miura-ori paper model under in-plane and out-of-plane deformations. Reproduced with permission.^[541] Copyright 2017, Elsevier. e) Geometry of the folded Miura-ori unit cell (top) and the contour plot of the dimensionless stretching rigidity of the unit cell along the *x* direction (bottom). Reproduced with permission.^[543] Copyright 2013, American Physical Society. f) The Miura-ori lithium-ion battery with a linear foldability of 200% and 1300% (top) and the corresponding cycling stability test at a series of deformation levels (bottom). Reproduced with permission.^[545] Copyright 2014, Nature Publishing Group.



to microscale^[515–518] and macroscale^[519,520] of various materials including metals,^[513,514,518] semiconductors,^[515,517,521] and polymers^[516,519,522] have been achieved by self-actuating materials (e.g., shape memory materials),^[520] or using stimuli-active hinges (e.g., by heat, light, or electricity),^[518,519,523,524] capillary forces,^[513,515] or residual stresses^[514,516,517,525,526] in the thin film. Figure 16c illustrates the process of self-folding Al₂O₃ facets with Au patterns. The capillary force from grain coalescence or reflow of the Sn hinges caused by the heat in the etching of the sacrificing layer provides the driving force of this process.^[513] Furthermore, approaches that enable reprogrammable and reversible operation of origami structures have also been demonstrated by reversely encoding and triggering prescribed geometric information into the materials.^[527–530]

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A wide variety of origami structures with distinct crease patterns have been demonstrated, including Miura-ori, Yoshimura, diagonal, zigzag, square-twist, diamond, waterbomb, and eggbox,^[531–536] among which the Miura-ori pattern is the most commonly used for soft electronics due to its rigid foldability, simple geometry, and advanced theory.^[537–539] As shown in Figure 16a, the crease pattern of Miura-ori can be represented as a quadrilateral mesh given by a set of vertices, lines, and facets, where the blue lines and red lines correspond to the valley and mountain folds, respectively, which are formed when the paper is folded toward or away from the viewer.^[540] A unit cell (gray shaded area in Figure 16a) can be used to represent the entire Miura-ori structure due to its geometrical periodicity.

Under external mechanical loadings, two deformation modes can occur in Miura-ori structure: out-of-plane and in-plane, as shown in Figure 16d.^[541] In the out-of-plane deformation mode, the facets must bend and twist, which will induce strain to the functional areas and is thus not ideal for soft electronics.^[542] In the in-plane deformation mode, deform only occurs at the creases by folding and unfolding while the facets per se remain strain free, also known as rigid origami.^[534,542] The in-plane kinematics of Muria-ori structure may be characterized by two quantities: the Poisson's ratio that describes the deformations in orthogonal directions and the stretching rigidity that characterizes its planar mechanical stiffness.^[543] Both quantities have a strong relationship to the geometric parameters of the folded unit cell.^[543] As shown in the top image of Figure 16e, the folded Miura-ori unit cell is deterministically defined by the oblique angle of α , the dihedral angle of θ , and the ridge lengths of l_1 and l_2 .^[543] The other parameters of the unit cell including the length *l*, width w, height h, and the other dihedral angle β can be characterized as follows: $\beta = 2\sin^{-1}[\zeta \sin(\theta/2)], l = 2l_1\zeta, w = 2l_2\xi$, and h = $l_1\zeta \tan \alpha \cos (\theta/2)$, where $\xi = \sin \alpha \sin (\theta/2)$ and $\zeta = \cos \alpha (1 - \xi^2)^{-1/2}$ are the dimensionless width and height of the unit cell, respectively.^[543] The in-plane Poisson's ratio is defined as

$$\nu_{wl} \equiv -\frac{dw / w}{dl / l} = 1 - \xi^{-2}$$
(25)

Equation (25) indicates the in-plane Poisson's ratio is scaleindependent and is always negative because $\xi \leq 1$, which means the Miura-ori structure expands simultaneously in both the stretching and the orthogonal directions.^[543]

To calculate the in-plane stiffness of the unit cell, the potential energy of a unit cell deformed by a uniaxial force f_x in the *x* direction is $H = U - \int_{\theta_0}^{\theta} f_x (dl/d\theta') d\theta'$, where $U = kl_e(\theta - \theta_0)^2 + kl_e(\beta - \beta_0)^2$, *k* is the hinge spring constant, and θ_0 and β_0 are the natural dihedral angles in the flat state.^[543] Then, the external force f_x at equilibrium is determined by $\delta H/\delta\theta = 0$. The stretching rigidity in the *x* direction is given by

$$K_{x}(\alpha,\theta_{0}) \equiv \frac{\mathrm{d}f_{x}}{\mathrm{d}\theta}\Big|_{\theta_{0}} = \frac{4k\Big[\left(1-\xi_{0}^{2}\right)^{2}+\cos^{2}\alpha\Big]}{\left(1-\xi_{0}^{2}\right)^{1/2}\cos\alpha\sin^{2}\alpha\sin\theta_{0}}$$
(26)

where $\xi_0 = \xi(\alpha, \theta_0)$.^[543] Equation (26) indicates that K_x is also scale-independent and has a nonmonotonic dependence on α and θ_0 . Setting $\partial_{\theta_0} K_x |_{\alpha} = 0$ and $\partial_{\alpha} K_x |_{\theta_0} = 0$ allows determining the optimal design curves, i.e., $\theta_{0m}(\alpha)$ (green dotted curve in the bottom image of Figure 16e) and $\alpha_m(\theta_0)$ (red dashed curve in the bottom image of Figure 16e), which yield the minimum value of the stiffness K_x as a function of α and θ_0 .^[543] Analogous analysis allows determining the orthogonal stretching rigidity K_p , which is related geometrically to K_x via the design angles α and θ_0 .^[543]

The origami has found various applications in soft electronics, ranging from lithium-ion batteries^[544,545] to solar cells,^[495] photodetectors,^[538,546] and thermoelectric generators.^[547] As a representative example, the Miura-ori lithium-ion battery with a linear foldability of 200% and 1300% is shown in the top images of Figure 16f, which exhibit a steady output voltage of 2.65 V under severe deformations. The maximum output power (17.5 mW) of the battery remains stable after 50 cycles of folding and unfolding at a series of deformation levels (up to 1340%, the bottom image of Figure 16f).

3.4. Kirigami

Very similar to the origami, kirigami is an art combining paper folding and cutting to create artistic models, which has recently emerged as an avenue toward stretchable electronics.[548-551] Figure 17a shows the most popular kirigami pattern for stretchability. To dissipate the tensile energy, the patterned slits open up to allow the planar sheets to buckle out of plane, perpendicular to the slits (right image in Figure 17a).^[541] Figure 17b shows the stress-strain curves of a tracing paper with and without the tessellated kirigami slits, which exhibits markedly different deformation behaviors.^[552] The tracing paper with the kirigami design deforms elastically, similar to the pristine paper at small strain magnitudes (purple section in Figure 17b), after which the slits open up to allow the paper to deflect out of plane (green section in Figure 17b). The change in deformation mechanism from purely elastic stretching to a combination of stretching and out-of-plane buckling leads to decreased slope of the stress-strain curve.[552] Then, the alignment of the struts under further stretching causes the overall structure to densify perpendicular to the pulling direction (white section in Figure 17b), after which fracture begins at the ends of the slits owing to the high strain concentrations at these regions.^[552] The kirigami design enables the overall sheets to reach an ultimate stretchability of almost two orders of magnitude higher than that of the pristine one (from 4% to 370%).[552]







Figure 17. Kirigami for soft electronics. a) Photographs of the kirigami paper model, showing the out-of-plane buckling deformation when it is stretched. Reproduced with permission.^[541] Copyright 2017, Elsevier. b) Stress–strain curves of the tracing paper with and without the kirigami cuts. The one with the kirigami cuts exhibits almost two orders of magnitude higher stretchability than the one without (from 4% to 370%). Reproduced with permission.^[552] Copyright 2015, Nature Publishing Group. c) Schematic illustration of a kirigami with key geometric parameters labeled. d) Influence of two dimensionless parameters α and β on the yield and fracture strains for zigzag (ZZ) and armchair (AC) graphene kirigami. Reproduced with permission.^[555] Copyright 2014, American Physical Society. e) Schematic illustration of the kirigami design that enables in-plane rotations. Reproduced with permission.^[499] Copyright 2017, Nature Publishing Group. f) Finite element calculations of the self-similar kirigami design, showing the enhanced stretchability and decreased allowable rotation angle when the self-similar level gets higher. g) The self-similar kirigami with an alternating cut motif between the levels showing improved maximum lateral strain. Reproduced with permission.^[559] Copyright 2014, National Academy of Sciences. h) The kirigami design with simultaneous stretchability and compressibility by embedding circular cutouts to the solid square units. Reproduced with permission.^[561] Copyright 2017, Elsevier. i) Electrical properties of the stretchable kirigami graphene transistor before stretching and when stretched to 240% strain. Insets are TEM images of the corresponding transistors. Reproduced with permission.^[563] Copyright 2015, Nature Publishing Group.

The overall mechanical properties of the kirigami sheet material have been investigated by two dimensionless parameters: $\alpha = (w - 0.5b)/L_0$ and $\beta = (0.5d - c)/L_0$, where L_0 and b are the sheet length and width, respectively; d is the cut pitch;

and *w* and *c* are the interior cut width and height, respectively. The edge cut width is defined to be half of the interior cut width (Figure 17c).^[553–555] Apparently, α represents the ratio of the overlapping cut width to the sheet length. The yield and

fracture strain values of the sheet increase with increasing α when $\alpha > 0$ and become insensitive to α when $\alpha < 0$ (the top image of Figure 17d).^[555] This is because when $\alpha > 0$, the edge and interior cuts overlap with each other, allowing the sheets to buckle out of plane, which will not occur when $\alpha < 0.$ ^[555] However, the yield stress exhibits an opposite trend to the yield strain because when $\alpha < 0$ the sheet behaves more like the cutfree sheet, leading to an enhanced rigidness compared with the sheet with $\alpha > 0$.^[555] Therefore, there is a tradeoff between the stretchability and strength of the kirigami sheet. Differing from α that describes the geometry perpendicular to the stretching direction, β describes the geometry parallel to the stretching direction and is directly related to the width and density of the cuts, which controls the likelihood of out-of-plane buckling.^[555] The impact of β on the yield strain and fracture strain when $\alpha = 0.07$ is shown in the bottom image of Figure 17d. As the β increases, it is more difficult for the out-of-buckling to occur. Therefore, the yield and fracture strains decrease.^[555]

The patterned slits buckle out of plane to absorb the tensile strain, which renders the kirigami structures difficult to conform and maintain an intimate contact to a surface. To solve this problem, kirigami designs that are based on in-plane rotations have been developed.^[541] As shown in Figure 17e, the imposed cut pattern divides the sheet into units (typically square or triangular) connected through small "hinges."^[499] Upon stretching, the units rotate (almost) freely around the hinges, resulting in a stretchable mechanism that is driven mostly by the rigid unit rotation instead of stretching or bending the units. Hence, this design is referred to as rigid kirigami.^[541] The rigid kirigami design enables a biaxial expansion of the sheet due to a negative Poisson's ratio.^[556]

The self-similar concept has also been implemented in the kirigami to increase its expandability (Figure 17f), in which the units are hierarchically subdivided into smaller units by repeating the cut pattern within the original one. In this design, the expansion at a higher level starts only after the prior lower level is largely exhausted.[557,558] The self-similar kirigami with the level of three exhibits a nearly two times larger stretchability than that with the level of one (79% vs 43%) (Figure 18f).^[559] While the larger stretchability with an increased level of hierarchy, the allowable rotation angle (θ_i in Figure 17f) of the smallest units become smaller due to the constant cut motif across the entire structure, implying a finite limit to the expandability of the self-similar kirigami.^[559] This limit, however, can be addressed by alternating the cut motif between the levels, allowing larger rotation angles for the smallest units and thus an improved maximum expandability.^[559] For example, the maximum lateral strain in the level-four structure consisting of the alternating cut motif is ε_4 ($\beta\alpha\beta\alpha$) = 130%, whereas that for the single motif is ε_4 ($\alpha\alpha\alpha\alpha$) = 108%, as shown in Figure 17g.^[559] In addition, dog-bone-shaped cuts and varying the hinge width with the hierarchical levels have been demonstrated to be effective in increasing the strength and maximum expandability of the self-similar kirigami, even for brittle materials such as the acrylics.[560]

In spite of the large stretchability enabled by these aforementioned kirigami designs, a major drawback with the kirigami is the lack of compressibility. Recently, a kirigami design introduces circular cutouts to the conventional solid square units, which enables simultaneous stretchability and compressibility of the structure.^[561] The porous units can accommodate a compression of as high as 50% by compacting and closing these pores (Figure 17h), while the stretchability is maintained by the line cuts that allow rotating the porous units.^[561]

A myriad of soft electronic devices with kirigami inspired structures have been demonstrated, including pressure and thermal sensors,^[562] transistors,^[563] solar cells,^[564] lithium-ion batteries,^[565] supercapacitors,^[566,567] triboelectric generators,^[568] and electrocardiogram bioprobes.^[569] For example, the kirigami has been used to build an atomic-thin graphene transistor (Figure 17i).^[563] Gated by 10 mM KCl solution at a source–drain bias of 100 mV, the transistor does not change its conductance when stretched to 240%.

3.5. Textiles

The idea of using textiles for soft electronics is intuitive as they are naturally stretchable and indispensable for human life.^[570,571] Textiles are hierarchical, consisting of structures on five scales: fibril (nm), fiber (nm– μ m), yarn (μ m–mm), fabric (cm), and the final product (m) (**Figure 18**a),^[364,572,573] which enables structural optimization and function integration at many levels. Fibrils are clusters of partially aligned polymer molecules organized in ordered (crystalline) and disordered (noncrystalline) regions, which assemble to form fibers that are natural or synthetic.^[573] Fibers are subsequently fabricated into yarns through processes such as twisting or wrapping, and yarns can be woven or knitted into planar fabrics that can then be processed into textile products.^[573]

Embedding desired mechanical and electrical characteristics on the fiber level is the most effective yet inconspicuous strategy toward soft electronic applications.[573-577] Stretchable conductive fibers can be fabricated through wrapping conductive materials (e.g., CNT, graphene) over elastic fibers (e.g., poly(styrene-block-butadiene-block-styrene), polyurethane). The conductive materials can offer electrical conductivity, tensile strength, and mechanical and thermal stability.[578,579] For example, aligned multiwall CNT sheets synthesized by chemical vapor deposition can be closely wound onto an elastic rubber fiber with an accurately designed helical angle, resulting in a fiber with a stretchability of 100% and an optimal resistivity of 0.086 k Ω cm⁻¹ (Figure 18b).^[579] Multiple fibers can be wound onto a substrate or knotted into various forms.^[579] A stretchable dye-sensitized solar cell utilizing this fiber as the counter electrode has been demonstrated, which achieves a maximum energy conversion efficiency of 7.13%, after 20 cycles of stretching at 30% tensile strain.^[579]

Another method for fabricating stretchable conductive fibers, which is perhaps more suitable for large-scale production, is wet spinning. In this process, precursors consisting of skeleton polymers and conductive components are extruded into a liquid bath to precipitate and then drawn to form fibers.^[580–583] For example, wet-spun fibers prepared by precursors consisting of 100–500 nm Ag NPs, multiwalled CNTs decorated with 3–5 nm Ag NPs, and poly(vinylidene fluoride-*co*-hexafluoropropylene) matrix exhibit a maximum initial conductivity of 17 460 S cm⁻¹ with rupture tensile strain of 50%, and the maximum strain









can be increased to 490% by decreasing the conductivity to 236 S cm^{-1,[584]} Helical fibers with hierarchical structures have also been demonstrated by a combination of drawing and twisting the fibers in its wet states, which further increases the stretchability.^[585] The helical fibers consisting of poly(vinyl alcohol) and graphene can sustain longitudinal strains as high as 414% due to the multiscale deformation combining both nanoscale platelets sliding and microscale spiral curl unraveling (Figure 18c).^[585]

The wave/wrinkle strategy has been borrowed from the 2D substrate to the 1D fibers to produce superelastic conductive fibers. Aligned multiwall CNTs are attached to a highly prestrained rubbery fiber (styrene–(ethylene–butylene)– styrene copolymer) followed by releasing the prestrain (Figure 18d).^[573,586] The fibers exhibit a periodic hierarchical buckling morphology (SEM images in Figure 18d), which can be stretched to 1320% with almost unchanged electrical conductivity, even after thousands of cycles of stretching.^[586] The rubber/CNT core/sheath fibers find applications as strain sensors that generate an 860% capacitance change after stretched, and electrically powered torsional muscles that operate reversibly by a coupled tension-to-torsion actuation mechanism.^[586]

In addition to stretchability and conductivity, fibers with light-emitting and energy-harvesting functions have been demonstrated.^[587–591] For example, the organic light-emitting diode fibers can withstand a tensile strain of 4.3% and exhibit luminance and current efficiency of over 10 000 cd m⁻² and 11 cd A⁻¹, respectively (Figure 18e), which are on par with ITO-glass-based counterparts.^[587] The cable-type lithium-ion battery can achieve extreme levels of mechanical flexibility and a stable electrical output. The capacity of 1 mAh cm⁻¹ and an output voltage of 3.5 V can power an LED screen under severe deformation (Figure 18f).^[591] All of these advances in fibers provide multitudinous building blocks for functional yarns or directly for textile electronics.

On the yarn level, multiple fibers can be twisted together to form a tight bundle.^[592–594] A supercapacitor is fabricated by twisting two GF@3D-G (a core of graphene covered with a sheath of 3D porous network-like graphene framework) electrodes solidified in H₂SO₄–poly(vinyl alcohol) gel polyelectrolyte (top image in Figure 18g), exhibiting a stable areaspecific capacitance of 1.2–1.7 mF cm⁻² after 500 cycles of bending.^[592] The supercapacitor can be shaped into a spring structure, which is both compressible (50%) and stretchable (200%) while maintaining its electrochemical features (bottom image in Figure 18g).^[592] The yarn twist (the orientation of the fiber relative to the yarn axis) and number of fibers within a yarn have a significant impact on the mechanical properties of the resulting yarn. In a study using the microscale model, due to the increase in yarn diameter, more fibers in the yarn give a larger tensile modulus of the yarn; because of the more compact and rigid yarn structure, a lower yarn twist angle leads to a larger tensile modulus.^[595]

On the fabric level, various methods can be used to produce textile fabrics from fibers and yarns, including weaving, knitting, felting, lacemaking, and braiding.^[571] Among them, woven fabric and knitted fabric are commonly used for textile electronics (image (iv) in Figure 18a).^[297,596,597] Woven fabrics are generally composed of two sets of yarns, interlaced at certain angles. Knitted fabrics are produced by interlocking one continuous yarn up and down in the weft direction.^[598] Woven fabrics are usually durable and provide a stable shape, which allows for accurate placement of individual yarns and dense integration of electronics, whereas knitted fabrics are characterized by high elasticity and elongation, good conformability in mechanically active environments, as well as decent air permeability, thermal retention, and humidity transport properties.^[571,599,600] Figure 18h shows the stress-strain responses of a single fiber and fabrics with woven or knitted constructions. It is clear that the knitted construction enables the largest elongation due to its more open and horseshoe-shaped structure.[601]

A myriad of wearable electronics has been demonstrated by weaving or knitting conductive and functional fibers/ yarns, including logic circuits,^[602] supercapacitors,^[603] solar cells,^[604,605] pressure sensors,^[606] and triboelectric energy harvesters.^[607,608] For example, a woven hybrid power textile composed of triboelectric energy harvesters and solar cells is shown in Figure 18i, which can simultaneously harvest mechanical and solar energies.^[608] The hybrid power textile with a size of 4 cm \times 5 cm can stably deliver an average output power of 0.5 mW under conditions of a human walking and a solar intensity of 80 mW cm⁻² (right panel in Figure 18i). The device can charge a 2 mF commercial capacitor up to 2 V in 1 min under ambient sunlight in the presence of mechanical excitations, such as motions from the human and wind.^[608]

Figure 18. Textiles for soft electronics. a) Hierarchical structures of textiles consisting of five levels of scales. Reproduced with permission.^[573] Copyright 2015, American Association for the Advancement of Science. b) Stretchable conductive fibers prepared by wrapping CNTs on an elastic fiber, which can be stretched to 100% strain and withstand various kinds of deformations. Reproduced with permission.^[579] Copyright 2014, Wiley-VCH. c) Helical stretchable fibers prepared via the wet-spinning method, which can sustain a longitudinal strain as high as 414% due to the multiscale deformation modes of both nanoscale platelet sliding and microscale spiral curl unraveling. Reproduced with permission.^[585] Copyright 2016, Nature Publishing Group. d) Superelastic conductive fibers enabled by applying aligned multiwall CNTs on a highly prestrained rubbery fiber, exhibiting a periodic hierarchical 2D buckled morphology and an ultrahigh stretchability of 1320% with almost unchanged electrical conductivity. d) Top: Reproduced with permission.^[573] Copyright 2015, American Association for the Advancement of Science. Bottom: Reproduced with permission.^[586] Copyright 2015, American Association for the Advancement of Science. e) A functional fiber with light-emitting characteristics. Reproduced with permission.[587] Copyright 2018, American Chemical Society. f) A flexible lithium-ion battery fiber. Reproduced with permission.[591] Copyright 2012, Wiley-VCH. g) A supercapacitor yarn fabricated by twisting two GF@3D-G fiber electrodes solidified in H2SO4-polyvinyl alcohol gel polyelectrolyte. Reproduced with permission.^[592] Copyright 2013, Wiley-VCH. h) Stress-strain responses of a single fiber and fabrics with woven and knitted structures. Reproduced with permission.^[601] Copyright 2017, American Association for the Advancement of Science. i) A hybrid power textile that can simultaneously harvest mechanical and solar energies (left) and its corresponding electrical power output curves (right). Reproduced with permission.^[608] Copyright 2016, Nature Publishing Group. j) A textile fabric with various patterns of nickel traces deposited by laser-scribing and electroless plating. Reproduced with permission.^[614] Copyright 2017, Wiley-VCH. k) A textile integrated with commercial rigid LEDs. Reproduced with permission.^[617] Copyright 2013, Trans Tech Publications Inc.

One main drawback with the weaving and knitting processes is their limited design freedom since the fiber/yarn paths are constrained to follow the warp or weft direction within the fabric.^[609] One way to solve this problem is to coat functional materials to textiles via inkjet or screen printing^[167,610–612] or electroless deposition,^[613,614] which give complete freedom for the design layouts since the functional materials are not tied to the fabric structure.^[609] Figure 18j shows a polyester fabric with different conductive patterns of nickel obtained by masked laser-scribing and subsequent electroless plating. This process enables customizable designs and high-performance pressure sensors in this case.^[614]

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Conventional rigid electronic devices can be integrated into the textile matrix to enable state-of-the-art functionalities.[615,616] As an example, a textile fabric integrated with commercial-offthe-shelf LEDs is shown in Figure 18k.^[617] The fundamental incompatibility of the rigid discrete electronic components and a soft textile matrix, however, creates strain at their interface, which would potentially cause delamination and device failure.^[618] Reducing the component size and foot print is effective to alleviate this problem.^[416,474] For example, an ultrathin and multifunctional physiological sensing system based on a serpentine open mesh structure has been demonstrated, which can conform to fabrics without influencing their original mechanical properties.^[416] The final system is breathable, washable, and reusable.^[416] Additionally, adding strain isolation mechanisms, such as a liquid layer, at the rigid/soft interface, represents promising directions as well.^[128,619]

3.6. Cracks

Cracks are generally considered as material defects due to the induced resistance increase or even device failure. In stretchable electronics, however, cracks with controlled distribution can provide a percolated conductive mesh in stretchable conductors.^[620-622] Figure 19a shows stretchable Au electrodes deposited on PDMS when uniaxially stretched to 60%. The connected lips between the random cracks guarantee continuous conductive paths of the Au thin films under a maximum strain of 120%.^[623] Moreover, the crack-based stretchable Au electrodes can be modified with dispersed Pt nanoislands via templateassisted electrodeposition (Figure 19b).^[623] The modified electrode maintains its stretchability and exhibits a much lower impedance compared to the bare Au because of the increased effective area and enhanced charge transfer capability of the Pt nanoislands. The nanostructured stretchable electrode allows conformal wrapping on a rat brain cortex and effective monitoring its intracranial electroencephalogram (Figure 19c).^[623]

Cracks can provide transformative capabilities in strain sensors.^[624] An analogous example in nature is the spider's slit organs that are located between the metatarsus and tarsus bones near the leg joints (Figure 19d). The slit organs consist of approximately parallel crack-shaped sensory lyriforms that enable ultrasensitive monitoring of vibrations in the spider web.^[625] Inspired by the spider's slit organ, an ultrasensitive crack-based strain sensor has been prepared by depositing a 20 nm thick Pt layer on top of the polyurethane acrylate substrate followed by bending the system to induce cracks, as illustrated in the inset of Figure 19e.^[625] The crack-based sensor exhibits a high gauge factor of over 2000 at the strain range of 0–2% (Figure 19e), where the gauge factor (GF = $(\Delta R/R_0)/\varepsilon$) is a key metric to evaluate the strain sensor performance. This gauge factor is orders of magnitude higher than strain sensors based on other mechanisms^[626,627] and thus holds great potential for diverse applications requiring ultrahigh displacement sensitivity such as physiological pulses and phonation recognitions.^[625]

The ultrahigh gauge factor is attributed to the disconnection-reconnection process under the applied strain, which can be explained using a theoretical model shown in Figure 19f.^[625] The nearly cut-through straight cracks are assumed to have inherent asperity, zip-like crack edges. Under applied axial strain (ε_v), separation of adjacent conductive segments causes a dynamic increase in resistance. At the same time, the transverse compressive strain $-v\varepsilon_{v}$ where v is the Poisson's ratio of the substrate, reconnects the conductive segments by the inherent asperity of the crack edges, which decreases the resistance.^[625] Overall, the total resistance will increase sensitively with the applied strain due to the reduced total overlapping area of adjacent conductive segments. Further increasing the applied strain will lead to fracture, when the displacement of the adjacent conductive segments overcomes the asperity height.^[625] Similar design strategies have been applied to other materials such as Au,^[624,628] graphene,^[629-632] ITO,^[633,634] and graphite,[635-637] which enable different strain sensing ranges and use cases.

The sensitivity of crack-based sensors can be further enhanced by controlling the crack geometry, such as the crack depth, density, and asperity.^[638,639] The sensor gauge factor generally goes up with increasing the relative crack depth (ratio of crack depth to conductive film thickness). The crack depth can be readily increased, without changing the other geometrical parameters, by applying additional tensile forces after the initial cracks are generated (Figure 19g).^[638] By this strategy, the gauge factor exhibits a high value of ~16 000 at 2% strain and a superior signal to noise ratio of ~35.^[638] Guided formation of straight cracks has been demonstrated by introducing patterned holes to the metal thin films (Figure 19h). The fast separation of adjacent conductive segments due to the straight crack geometry leads to an ultrahigh gauge factor of over 2 × 10⁶ at the strain range of 0–10%.^[639]

3.7. Interlocks

In human skin, various mechanoreceptors such as Merkel Disk (MD), Ruffini Ending (RE), Meissner Corpuscle (MC), and Pacinian Corpuscle (PC) distributed in the epidermal and dermal layers enable the spatiotemporal recognition of magnitude, location, and direction of tactile stimuli.^[640] The intermediate ridge of the interlocked microstructures present at the epidermal–dermal interface acts as a magnifying lever for the transmission of the tactile stimuli to the mechanoreceptors and enhances the tactile acuity of mechanoreceptors at the same time (**Figure 20**a).^[640,641] Inspired by the interlocked epidermal–dermal layers, stretchable tactile sensors with interlocked nanostructures have been demonstrated, enabling sensitive







Figure 19. Cracks for soft electronics. a) An SEM image of a stretchable Au electrode under tensile strain of 60%, exhibiting the percolated conductive paths due to the random crack structure. b) An SEM image of a stretchable Au electrode modified by dispersed Pt nanoislands, enabling simultaneous low impedance and high stretchability. c) Intracranial electroencephalogram recording of a rat utilizing the stretchable Au electrode with dispersed Pt nanoislands. Inset photograph shows the device wrapped on a rat brain cortex. a–c) Reproduced with permission.^[623] Copyright 2014, American Chemical Society. d) Schematics of a spider with the slit organ and parallel crack-shaped sensory lyriforms that enable ultrasensitive detection of vibrations in the spider web. e) The electrical response of a crack-inspired strain sensor to the applied strain. Inset shows the schematic crack-based strain sensor. f) Illustration of the proposed theoretical model for the disconnection–reconnection process of the crack-based strain sensor. d–f) Reproduced with permission.^[625] Copyright 2014, Nature Publishing Group. g) Schematic process of modulating the crack depth by secondary bending and stretching, to enhance the sensitivity of crack-based strain sensors. Reproduced with permission.^[638] Copyright 2016, Wiley-VCH. h) An SEM image of a metal thin film with patterned pores that can guide the formation of straight cracks, which can enhance the sensitivity of crack-based strain sensors. Reproduced with permission.^[639] Copyright 2017, Nature Publishing Group.

and directional detections of various tactile stimuli.^[640,642,643] For example, devices consisting of two interlocked arrays of high aspect-ratio Pt-coated polymeric nanofibers supported on thin PDMS substrates exhibit excellent sensing limit of pressure (5 Pa), shear force (0.001 N), and torsion (0.0002 N m), as shown in Figure 20b.^[641] The sensing mechanism is attributed

to piezoresistance change,^[644] the change in the degree of interconnection, and thus electrical conductivity between the nanofiber arrays under applied various stimuli.^[641] The nanofiber arrays undergo different kinds of deformation in response to the different nature of applied forces, which enables the differentiation of various mechanical stimuli such as







Figure 20. Interlocks for soft electronics. a) Schematic human skin structures showing the interlocked epidermal-dermal layers. Reproduced with permission.^[640] Copyright 2014, American Chemical Society. b) Curves of relative resistance change of a tactile sensor with interlocked structures, and associated dynamic pressure, shear, and torsion loads to the tactile sensor. Reproduced with permission.^[643] Copyright 2012, Nature Publishing Group. c) A tactile sensor with interlocked hierarchical structures of PDMS micropillars and ZnO NW for the simultaneous detection of static and dynamic stimuli. Reproduced with permission.^[646] Copyright 2015, Wiley-VCH. d) Illustration of a multimodal sensor for the simultaneous detection of static and dynamic pressure and temperature by combining the interlocked structural design and the piezoelectric and pyroelectric properties of PVDF composites and rGO. Reproduced with permission.^[650] Copyright 2015, American Association for the Advancement of Science.

shear, torsion, and pressure because of the different levels of deformation of nanofiber arrays responding to the direction of applied forces.^[641]

The aforementioned device based on the piezoresistance change has very limited capabilities in responding to dynamic high-frequency stimuli due to the time delay between the input and output responses.^[645] To solve this problem, tactile sensor with an interlocked hierarchical structure that integrates piezoresistance and piezoelectrics has been developed (Figure 20c).^[646] The device is based on the interlocked geometry of hierarchical PDMS micropillars decorated with ZnO NW arrays. The piezoelectric properties of the interlocked ZnO NWs contribute to the detection of fast dynamic stimuli (\approx 250 Hz) due to the extreme sensitivity to deformation and thus instantaneous generation of dipoles in ZnO NWs upon bending.^[647–649] The piezoresistive properties of both the interlocked PDMS micropillars and the ZnO NWs contribute to the detection of static pressure.^[646]

In addition to tactile sensing, devices with multimodal sensing capabilities such as pressure and temperature are highly desirable to mimic the human skin. By combining the interlocked structure and ferroelectric polymer composites, multimodal sensors have been demonstrated (Figure 20d).^[650]

In this device, composites of poly(vinylidene fluoride) (PVDF) and reduced graphene oxide (rGO) are used. The pyroelectric and piezoelectric effects of the PVDF/rGO composites enable temperature sensing and high-frequency dynamic pressure sensing, respectively. The piezoresistive effect enhanced by the interlocked structure allows the perception of static pressure. Moreover, parallel microridges of PDMS are fabricated on the device surface to mimic the fingerprint pattern, which amplifies the received signals of the device when scanned on objects with textured surface.^[650]

4. Conclusion and Outlook

This review, which is far from exhaustive, aims to provide a glance at the dynamic and vibrant advances that have been made in soft electronics to date. As we have shown, there are two strategies that can be applied to achieve stretchability in electronics. The materials innovation strategy utilizes either new materials that are specifically synthesized or existing materials with niche applications in soft electronics. The structural design strategy can provide a method to hybridize all classes of materials, either rigid or soft, on the same soft platform.





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There are still several grand challenges to be overcome. First, the current active component density of soft electronics is low in comparison with state-of-the-art rigid electronics driven by Moore's law. Enabling technologies for creating vertical interconnect access (VIA) for 3D integration and for assembling high-density electronic components such as bare dies on soft substrates will need to be developed.^[651,652] Second, even though there have been a lot of approaches for minimizing the strain level in soft electronics, they are still premature for daily use due to the delicate and fragile nature of the components. Few demonstrated devices can last through washing or long-term repeated use.^[653,654] Nanomaterials with peculiar mechanical properties may offer some opportunities in this regard. Third, as a common bottleneck for all devices that rely on electricity, the power source is still a big challenge for soft electronic systems. Current soft power solutions, either by harvesting or combined harvesting and local storage, offer insufficient power budget needed for long-term operating these devices.^[655,656] Efforts from two distinct directions are attempting to solve this problem: the device community tries to lower the power budget of the chips and systems, and the energy community aims to boost the output power and energy density of the harvesting and storage devices. These two directions are expected meet at some point in the middle of the energy landscape, which has yet to be reached today. Fourth, for human body integration, many practical factors need to be considered. For example, for implantable devices, biofouling and immune system response need to be minimized;[69] for wearable devices, breathability of the elastomers and reversible adhesion on sweaty skin needs to be improved;^[416] and for wireless devices, absorption of the electromagnetic waves by the human body needs to be thoroughly studied and minimized. Finally, to broaden the impact of this field, large-scale production of soft electronic devices at a reasonably low cost will have to be realized. Existing infrastructures are optimized for rigid substrates but may not be compatible with the emerging processes for soft substrates. New manufacturing machines and equipment will need to be explored.

We believe that the above-mentioned challenges represent great opportunities for researchers even though probably none of these challenges could be solved by a single researcher or group. Synergistic collaborations between cohorts of experts in cross-disciplinary fields are necessary and critical for pushing this burgeoning field forward.

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Conflict of Interest

The authors declare no conflict of interest.

Keywords

materials innovation, soft electronics, stretchable electronics, structural design, wearable electronics

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