Microrheology of DNA hydrogels

Zhongyang Xing1, Alessio Caciagli1,2, Tianyang Cao1,2, Iliya Stoew, Mykolas Zupkauskas, Thomas O'Neil, Tobias Wenzel, Robin Lamboll, Dongsheng Liu, and Erika Eiser1

1Cavendish Laboratory, University of Cambridge, Cambridge CB3 0HE, United Kingdom; and 2Department of Chemistry, University of Tsinghua, Beijing 100084, China

Edited by Michael L. Klein, Temple University, Philadelphia, PA, and approved June 26, 2018 (received for review December 20, 2017)

A key objective in DNA-based material science is understanding and precisely controlling the mechanical properties of DNA hydrogels. We perform microrheology measurements using diffusing wave spectroscopy (DWS) to investigate the viscoelastic behavior of a hydrogel made of Y-shaped DNA (Y-DNA) nanostars over a wide range of frequencies and temperatures. We observe a clear liquid-to-gel transition across the melting temperature region for which the Y-DNA bind to each other. Our measurements reveal a cross-over between the elastic G′(ω) and loss modulus G″(ω) around the melting temperature Tm of the DNA building blocks, which coincides with the system’s percolation transition. This transition can be easily shifted in temperature by changing the DNA bond length between the Y shapes. Using bulk rheology as well, we further show that, by reducing the flexibility between the Y-DNA bonds, we can go from a semiflexible transient network to a more energy-driven hydrogel with higher elasticity while keeping the microstructure the same. This level of control in mechanical properties will facilitate the design of more sensitive molecular sensing tools and controlled release systems.

DNA nanotechnology | self-assembly | micro rheology | hydrogels | semiflexible polymers

DNA hydrogels made of well-defined small building blocks are a type of tenuous, semiflexible polymeric network that consists of precisely designed synthetic nucleotide strands as chemical or physical cross-linkers (1-5). These manmade bulk DNA hydrogels have been widely studied as functional materials that can be potentially used for controlled drug delivery, tissue engineering, biosensing, and other applications in the fields of nanotechnology and bioengineering mainly because of their biocompatibility and the ability to mix them with other (bio-)polymers (6-10). In particular, the vast combinations of the Watson–Crick pairing provide a unique way to achieve programmable self-assembly of thermally reversible gels with precise functionality (11, 12). While current studies mostly focus on the fabrication and utilization of DNA hydrogels, the fundamental physics relating the microstructure of these gels to their macroscopically observed viscoelastic properties still lacks good understanding (5, 8, 13). In recent years, a series of computational and experimental studies was carried out on the phase diagram of DNA hydrogels made of two-, three- and n-valent nanostars and their mixtures, providing a good reference for creating volume-spanning, percolating gels (11, 14-18).

The study of transient networks has been at the heart of many theoretical (19-21) and experimental studies (22, 23) for their display of complex phase diagrams and dynamics (24). The reversible cross-links in transient networks can be mediated by the short sticky ends of telechelic polymers (25), telechelic dendrimers (26), triblock copolymers (27), or charged end groups (28). These cross-links have a finite lifetime, and transient networks therefore behave like yield-stress fluids. In contrast, chemically cross-linked networks, such as rubbers, do not flow, but break like a solid when deformed extensively. Other transient but active networks are formed by semiflexible actin filaments cross-linked via proteins that give shape to cells and provide their locomotion (29). Vitrimeric, in which the bonds or cross-links can be exchanged through a catalytic process, are another class of transient networks with self-healing properties relevant in biological tissue engineering (30-32). However, with the ability to program DNA with an almost unlimited amount of highly specific interactions, different DNA nanostars can be designed to form arbitrarily complex networks, which cannot be achieved with other natural thermally reversible hydrogels made, for instance, of agarose or gelatin (33) or for that matter, vitrimeres. In particular, local inhomogeneities with well-defined nanopores and mechanical properties may be built into the network, as different DNA blocks can be made to hybridize at different temperatures (34). Indeed, recent shear rheological studies on macroscopic samples (3, 35, 36), here referred to as bulk rheological studies, showed that the specific structure and connectivity of DNA nanostars have a strong influence on their macroscopic mechanical response. However, these low-frequency measurements only describe the systems macroscopic, long-time viscoelastic response and cannot describe how more local relaxation times change as the system goes through the melting transition.

Here, we present microrheology studies using diffusing wave spectroscopy (DWS) (37, 38) to study the equilibrium elastic and viscous moduli of our DNA gels over a much larger frequency range than that available in bulk rheology (39-42). In particular, using sealed sample chambers allowed us to perform these measurements over a large temperature range without fear of changes in the samples due to evaporation, which is problematic in bulk rheology (35, 43). We show that DWS enables us to link

Significance

While widely known as the molecule of life, DNA is also an amazing building block at the nanoscale, since it allows us to design and program the structure and dynamics of functional nanomaterials. We exploit the programmability of DNA to achieve control over the rheology of self-assembled hydrogels, which have elastic or viscous behavior (similar to that of slime) that is finely regulated by temperature. Using micro rheology to investigate the mechanical properties of DNA hydrogels at the microlength scale, we map the viscoelastic response over a broad range of frequencies and temperatures. The deep understanding in the fundamental physics provides a way to design DNA-based materials with precise control over the structure stability and rigidity at molecular level.

Author contributions: E.E. designed research; Z.X. performed research; T.C., M.Z., T.O., T.W., R.L., and D.L. contributed new reagents/analytic tools; A.C. and I.S. analyzed data; and Z.X. and E.E. wrote the paper.

The authors declare no conflict of interest.

This article is a PNAS Direct Submission.

This open access article is distributed under Creative Commons Attribution-NonCommercial-NoDerivatives License 4.0 (CC BY-NC-ND).

1To whom correspondence may be addressed. Email: z.x230@cam.ac.uk or ee247@cam.ac.uk.

2A.C. and T.C. contributed equally to this work.

This article contains supporting information online at www.pnas.org/lookup/suppl/doi:10.1073/pnas.1722206115/-/DCSupplemental.

Published online July 25, 2018.

www.pnas.org/cgi/doi/10.1073/pnas.1722206115
the system’s characteristic binding–unbinding processes with the local and global viscoelastic properties of the gel over a temperature range that covers the full melting region between the DNA nanostars. The measured $G'(\omega)$ and $G''(\omega)$ curves reveal a clear, temperature-reversible percolation transition coinciding with the melting temperature of the systems sticky ends. Moreover, we exploit the fact that ssDNA is much more flexible than dsDNA to show that, by keeping the total number density of Y-shaped DNA (Y-DNA) the same, we can render the network more rigid when removing the flexible linkers between them.

**Results**

**DNA Design.** The DNA hydrogels used in this study are composed of Y-shaped DNA building blocks made of three partially complementary oligonucleotides, here denoted $S_i$ (Fig. 1 and Table 1). These ss $S$-oligos are 43 bases long and consist of four functional parts: a sticky end (9 bases), a free joint of four Thymines providing flexibility, and two segments (I and II) that form the core of the Y shape. Segments I and II of, say, ssDNA $S_i$ are designed to be partially complementary to the respective segments of $S_2$ and $S_3$, thus making up the three dsDNA arms at the center of the Y shapes. Note that the center is created to be fully binding, leaving no nonbinding bases at the center and thus, reducing its flexibility. The respective binding sequences of the three dsDNA arms are given in Table 1, while the full sequences of the $S$-oligos are given in [SI Appendix]. The sticky ends can specifically bind to their complementary sequence due to Watson–Crick pairing. In all of the following gels, we use the same Y-shaped cores carrying either sticky ends named T or the complementary T′ DNA, such that Y shapes with T ends (T DNA) can only bind to those with T′ ends (T′ DNA). We keep the T to T′ DNA ratio 1:1, thus maintaining the system’s connectivity.

The melting temperature profiles of the Y shapes carrying either T or T′ sticky ends (Fig. 1C) were determined by measuring the absorbance of 260 nm in 1 μM DNA in Tris-EDTA buffer solution at pH 8.0 containing 150 mM NaCl, as ssDNA adsorbs 260 nm stronger than dsDNA. Starting at 90 °C, we observe a plateau in the absorbance until roughly 65 °C, marking the point at which hybridization (binding) sets in. Until this temperature, the individual ssDNAs are all in the unbound state. On further cooling, the absorption decreases continuously until the low-temperature plateau is reached, at which all single strands have hybridized into Y shapes. The melt temperature (denoted $T_m$) is defined as the point at which one-half of all possible base pairs are dissociated. $T_m$ was obtained from the median between the linear curves fitting the low- and high-temperature plateaus and found to be $T_m \approx 58$ °C for both Y shapes with T or T′ overhangs for the concentrations used in the melt temperature measurements. As shown previously (44), nonbinding ssDNA tails slightly lowered the value of $T_m$, with respect to the value based on tabulated data by SantaLucia (45), which are, averaged over all three arms, $T_m \geq 60$ °C—details are given in [SI Appendix].

**DWS Microhrology.** DWS measurements rely on time correlations of diffusively scattered light caused by submicrometer-sized spherical tracer particles [here, 600-nm large, charge-stabilized polystyrene (PS) particles] embedded in the DNA hydrogel sample. The thermal motion of the particles is described by the mean-square displacement (MSD), denoted $\langle \Delta r(t) \rangle^2$. This is measured by the real-time fluctuations of the scattered light collected by a photodetector and presented as the intensity autocorrelation function (ICF), $g_2(\tau) = (I(\tau)/\langle I(0) \rangle)^2$, where $I(0)$ and $I(\tau)$ are the scattered intensities at time 0 and some delay time $\tau$ later. The decay of $g_2(\tau)$ is related to the time evolution of particle motion, allowing for the MSD of the particle to be measured (37). The complex viscoelastic modulus and MSD are related by the generalized Stokes–Einstein relation given in Eq. 1 (38, 39):

$$\tilde{G}(s) \approx \frac{k_BT}{\pi Rs} (\Delta \tilde{r}^2(s)),$$

where $G(s)$ and $\langle \Delta \tilde{r}^2(s) \rangle$ are the Laplace transforms of the complex shear modulus and the MSD, $R$ is the radius of the tracer particles, and $s$ is the Laplace frequency. Replacing $s$ with $i\omega$, we obtain the $G'(\omega)$ and $G''(\omega)$ from the complex shear modulus $G^*(\omega) = G'(\omega) + iG''(\omega)$ (38, 39) (Materials and Methods).

Our DNA hydrogels were studied using a DWS RheoLab (LS Instrument AG) in echo mode, which allowed for long correlation times (Fig. 24). The light source was a 685-nm-wavelength diode laser. The measured ICFs where fitted and then converted to $G'\omega$ and $G''\omega$}

---

**Fig. 1.** Design and characterization of the DNA hydrogel building blocks. (A) Schematic of the ssDNA $S_i$ used. Each oligo strand consists of four functional parts: the sticky end, the free joint, segment I, and segment II. Segments I and II are part of the ds core DNA; sticky ends are for cross-linking the Y shapes to a network. (B) Cartoon of T and T′ DNA connected via hybridization of complementary sticky ends. (C) Melting (cooling) and heating (hybridization) curves of T and T′ DNA in Tris-EDTA buffer containing 150 mM NaCl measured using ultraviolet-visible (UV-vis) spectrometer. (D) From left to right are photographs of DNA hydrogels without free joint, with free joint, and with only one component. All three samples are made of Y-shaped DNA $\approx 500$ μM. For clarity, the left and center gels were colored with the SYBR Safe DNA Gel Stain from Invitrogen. The sample maintains its original shape at a timescale of several minutes.
Table 1. The sequences of dsDNA arms

| Name | Segment I | Segment II |
|------|-----------|------------|
| $S_1$ | 5'-TGG ATC CGC ATG ATC | CAT TGC CCG TAA GTA-3' |
| $S_2$ | 5'-TAC TTA CGG CGA ATG | ACA CGG AAT CAG CCT-3' |
| $S_3$ | 5'-AGG CTG ATT CGG TGT | GAT CAT GCG GAT CCA-3' |

we also present data on the same Y shapes but with 12-base overhang.

The MSD results extracted from the ICF curves are shown in Fig. 3C. At temperatures well above $T_{m2}$ (Fig. 3C, orange lines), the MSD curves depend linearly on the lag time $\tau$ over the whole measured region, confirming the $\langle \Delta r(\tau)^2 \rangle \propto \tau$ relation for Newtonian fluids and thus proving that, in the temperature window between about 55 °C and 45 °C, our system behaves like a fluid of disconnected Y shapes dispersed in a buffer colloidal solution. Note that our sample displays a diffusivity that is two orders of magnitude lower than that extracted from MSDs for 600-nm PS particles dispersed in pure water having a diffusion constant $D = 0.15 \mu^2 \text{ms}^{-1}$ at 50 °C. A similar decrease is observed in control measurements using 230-nm-large PS particles and confirmed in bulk rheology measurements of a nongelling sample (SI Appendix). The related increased viscosity is purely due to the high DNA concentration. Indeed, assuming that the Y shapes take up an effective spherical volume due to rotational diffusion (assuming that each arm is ∼5 nm long), the approximate volume fraction occupied by the Y shapes is some 40%, although the actual DNA content is only 2 wt %.

At $T < 30$ °C and short lag times, the MSD curves are similar to the high-$T$ measurements, increasing with increasing $\omega$; however, they have a slightly lower exponent, indicating sub-diffusive motion of the local bridges between cross-links. At
intermediate ω corresponding to longer relaxation times (for instance, of the “cages” formed by the cross-links), the MSD curves reach a plateau. Holding the sample at this lower temperature over 20 min and measuring the ICF in 5-min intervals show that there is no further increase in the plateau value (SI Appendix). This is also expressed in the flattening of the corresponding elastic moduli G′ presented in Fig. 4. This means that the tracer particles remain locally diffusive on short timescales (the diffusion coefficient of the particles in pure water is 1.65 μm² ms⁻¹ at 20 °C) but are confined by the percolating DNA network on long timescales. The transition region marked by the changing colors in Figs. 3 and 4 represents the melt region to be centered around the melting temperature Tm2 = 35 °C of the sticky ends. The calculated MSD for the same 600-nm-large PS colloids in pure water at 30 °C is presented by the dashed line as guide to the eye.

Fig. 3. (A) Schematic phase diagram of the Y-shaped DNA. The arrow indicates the concentration and temperature range of the ICF measurements shown in Fig. 2. The red area represents the two-phase region, and Tc is the critical point. (B) Illustration of the hybridization range for the sticky ends of the two different Y shapes. The graded area signifies the range over which a fraction of base pairs is formed. (C) MSD extracted from the ICF curves in Fig. 2. The color of the lines gradually changes from orange to blue, standing for the transition region ranging from about 45 °C to 20 °C, which is centered around the melting temperature Tm2 = 35 °C of the sticky ends. The transition region marked by the changing colors in Figs. 3 and 4 represents the melt temperature region over which the fraction of hydrogen bonds formed between two Y shapes with complementary sticky ends is gradually increasing as T decreases. Using refined SantaLucia rules for hybridization (44), we estimate the width of this transition region to be ΔT ~ 25 °C (Fig. 3B). With a Tm2 ~ 35 °C, this means that we should reach a fully bonded state and thus, a maximum network stiffness at around T ~ 25 °C. A detailed discussion of the value of the stiffness is given in the following.

The elastic, G′(ω), and viscous moduli, G″(ω), measured in a cooling cycle are shown in Fig. 4. Again, similar results were obtained using the smaller tracer particles in a cooling and heating cycle (SI Appendix), suggesting that the equilibrium gels display only very small hysteresis effects. As expected, the elastic modulus, G′(ω), undergoes a significant change as the temperature changes over the melting temperature region, while G″(ω) retains the same linear trend until about 30 °C. At high temperatures (orange lines in Fig. 4), G′(ω) is nearly zero at long timescales, as the solution is in a fully fluid state of Y shapes; however, it is nonzero at high frequencies, reflecting the fact that the sample is a quasi-concentrated solution of elastic shapes behaving like soft colloids. Indeed, above Tm2, the loss modulus dominates (Fig. 4C). However, around the melting temperature (between 37 °C and 31 °C), G′(ω) and G″(ω) run parallel and on top of each other with a ω⁻½ dependence, which we identify as the point of full percolation. Such a behavior is also associated with the gel point of cross-linked polymers (47). This percolation can be understood when looking at Fig. 3B: in the melt transition region, increasingly more Y shapes bind to each other, forming many clusters that grow in size as the temperature decreases. At Tm2, one-half of all possible hydrogen bonds or base pairs are bound, which does not mean that one-half of all Y-shaped arms are bound at all times but that they continuously form and break partially; thus, on average, they form a single cluster. Below Tm2, the fraction of hybridized base pairs continues to increase until about 25 °C, and also, their lifetime becomes longer. At even lower temperatures (blue lines in Fig. 4), the G′ reaches a plateau value of ~200 Pa in the intermediate time range, which corresponds to a mesh size ξ ~ 21.5 nm, assuming the scaling behavior of the bulk modulus Gb(ω) ~ ω(1). This is in good agreement with the calculated mesh size from the design that suggests an average distance between bonded Y-shaped centers of ξ ~ 20 nm corresponding to a slightly higher elastic modulus.

Interestingly, below T ~ 25 °C, our fully formed network is very similar to that of classical transient networks of flexible polymers held together by cross-links through physical interactions that constantly form and break (Fig. 4C) (20). The frequency behavior of such transient networks shows a typical Maxwellian G′(ω) ~ ω(1), whereas at higher frequencies, reflecting the fact that the sample is a quasiconcentrated solution of elastic shapes behaving like soft colloids, the gel point of cross-linked polymers (47). This percolation can be understood when looking at Fig. 3B: in the melt transition region, increasingly more Y shapes bind to each other, forming many clusters that grow in size as the temperature decreases. At Tm2, one-half of all possible hydrogen bonds or base pairs are bound, which does not mean that one-half of all Y-shaped arms are bound at all times but that they continuously form and break partially; thus, on average, they form a single cluster. Below Tm2, the fraction of hybridized base pairs continues to increase until about 25 °C, and also, their lifetime becomes longer. At even lower temperatures (blue lines in Fig. 4), the G′ reaches a plateau value of ~200 Pa in the intermediate time range, which corresponds to a mesh size ξ ~ 21.5 nm, assuming the scaling behavior of the bulk modulus Gb(ω) ~ ω(1). This is in good agreement with the calculated mesh size from the design that suggests an average distance between bonded Y-shaped centers of ξ ~ 20 nm corresponding to a slightly higher elastic modulus.

Interestingly, below T ~ 25 °C, our fully formed network is very similar to that of classical transient networks of flexible polymers held together by cross-links through physical interactions that constantly form and break (Fig. 4C) (20). The frequency behavior of such transient networks shows a typical Maxwellian G′(ω) ~ ω(1), whereas at higher frequencies, reflecting the fact that the sample is a quasiconcentrated solution of elastic shapes behaving like soft colloids, the gel point of cross-linked polymers (47). This percolation can be understood when looking at Fig. 3B: in the melt transition region, increasingly more Y shapes bind to each other, forming many clusters that grow in size as the temperature decreases. At Tm2, one-half of all possible hydrogen bonds or base pairs are bound, which does not mean that one-half of all Y-shaped arms are bound at all times but that they continuously form and break partially; thus, on average, they form a single cluster. Below Tm2, the fraction of hybridized base pairs continues to increase until about 25 °C, and also, their lifetime becomes longer. At even lower temperatures (blue lines in Fig. 4), the G′ reaches a plateau value of ~200 Pa in the intermediate time range, which corresponds to a mesh size ξ ~ 21.5 nm, assuming the scaling behavior of the bulk modulus Gb(ω) ~ ω(1). This is in good agreement with the calculated mesh size from the design that suggests an average distance between bonded Y-shaped centers of ξ ~ 20 nm corresponding to a slightly higher elastic modulus.
must come from the cluster phase, with clusters forming cages of the Y shapes. These completely disappear at even higher temperatures.

We made an exciting observation when performing low-frequency bulk rheology measurements on the very same system in which we removed the flexible linkers. The measured $G'(\omega)$ and $G''(\omega)$ curves showed an increase by a factor of up to 7, which was also observed when using samples with and without flexible linkers but 12 instead of 9 bases (data are shown in SI Appendix). This stiffening is particularly visible in the photograph in Fig. 1D. Moreover, further DWS data on the 12-bp sticky ends also confirm that the percolation transition coincides with the systems melting transition, only that now that transition occurs at $\sim 44 \, ^{\circ}C$, the $T_{m2}$ for the 12 bases.

Finally, at $T \gtrsim 35 \, ^{\circ}C$, we can plot the half time of the relaxation of the ICF as inverse function of temperature. The slope of the resulting Arrhenius plot (shown in SI Appendix) provides us with the strength of the bonds between two Y shapes, where the relaxation time $\tau = t_{1/2} = \tau_0 \exp(\Delta G / k_B T)$; here, $\Delta G$ is the Gibbs free energy. Following the arguments by Nava et al. (18), this will happen when at least two bonds per Y shape are broken, which corresponds to about 60 kcal/mol in our case and is very close to the tabulated value.

To summarize, our microrheological measurements show how a transient cross-linked hydrogel is formed as it is brought from the high-temperature range, where the Y-shaped building blocks form a viscous fluid, into an equilibrium gel phase. Remarkably, after all possible bonds are formed below the melt transition region, the DNA hydrogel shows a frequency behavior very similar to that of transient networks of flexible polymers when the Y-shaped centers are connected via a stiff dsDNA bond and two short, flexible ssDNA linkers. Interestingly, when the flexible ssDNA linkers are removed while keeping the same Y-shaped density, the elasticity of such a network increases sevenfold, indicating that the network goes from flexible and entropy driven to a more elastic system that is dominated by the energy of the semistiff connectors between the Y-shaped centers. Testing two differently long sticky bonds, we also show that their melt...
temperatures \( T_{m,2} \) could be identified as the percolation transition. It should be noted that the results presented here cannot be easily obtained by regular bulk rheology unless evaporation is controlled carefully, and also, single-bead microrheology would lead to ambiguous results due to local heating around the laser-trapped probe particle.

Our findings show that we can develop a class of hydrogels with more ordered local structure or if “coding in” a cascade of hybridization temperature hierarchical structures. These could be achieved by introducing locally more rigid DNA building blocks. Such networks could be envisaged as builders of thermoresponsive materials that could provide controlled drug release or act as micrometer-sized actuators with well-defined elastic modulus.

Acknowledgments. We thank D. Frenkel and C. Ness for useful discussions, P. Li for fabricating the macroscopic DNA network model, and A.C. and X. Z. K. receives financial support from the National University of Defense Technology Scholarship at Cambridge and the Associate Programme of the Doctoral Training Centre (NanoDTC) supported by the Engineering and Physical Science Research Council in Nanoscience and Nanotechnology. A.C. and E.E. acknowledge support from Marie Sklodowska-Curie European Training Network COLLIDENSE (H2020-MSCA-ITN-2014 Grant 642774). T.C. and D.L. thank National Basic Research Program of China 973 Program Grant 2013CB832802, National Natural Science Foundation of China Grant 21534001, and the Beijing Municipal Science & Technology Commission for financial support. I.S. and R.L. acknowledge support from Engineering and Physical Sciences Research Council (EPSRC) Grants RG90425 and 135307. M.Z. is funded by joint EPSRC and Unilever Cooperative Awards in Science and Technology Award RG748000. T.W. and E.E. thank the Winton Program for Sustainable Physics.

1. Um SH, et al. (2006) Enzyme-catalysed assembly of DNA hydrogel. Nat Mater 5:797–801.
2. Lee JH, et al. (2012) A mechanical metamaterial made from a DNA hydrogel. Nat Nanotechnol 7:816–820.
3. Li C, et al. (2011) Supramolecular hydrogel with identical cross-linking point density but distinctive rheological properties. Mater Chem Front 1:654–659.
4. Dong Y, Yang Z, Liu D (2014) DNA nanotechnology based on i-motif structures. Acc Chem Res 47:1853–1860.
5. Xing Y, et al. (2011) Self-assembled DNA hydrogels with designable thermal and enzymatic responsiveness. Adv Mater 23:1117–1121.
6. Okay O (2011) DNA hydrogels: New functional soft materials. J Polym Sci B Polym Phys 49:551–556.
7. Li C, et al. (2015) Rapid formation of a supramolecular polypeptide-DNA hydrogel for in situ three-dimensional multilayer bioprinting. Angew Chem Int Ed 54:3957–3961.
8. Li C, et al. (2015) Responsive double network hydrogels of interpenetrating DNA and polyelectrolyte for providing technical assistance in DWS measurements. ACS Nano 9:959–962.
9. Eiser E, Klein J, Witten TA, Fetters LJ (1999) Shear of telechelic brush. Phys Rev Lett 82:5076.
10. Vahabi M, et al. (2017) Normal stresses in semiflexible polymer hydrogels. arXiv:1712.02733.
11. Montanari D, Capelot M, Tournilhan F, Leibler L (2011) Silica-like maible materials from permanent organic networks. Science 334:965–968.
12. Capelot M, Unterlass MM, Tournilhan F, Leibler L (2012) Catalytic control of the vitrimer glass transition. ACS Macro Lett 1:789–792.
13. Rose S, et al. (2014) Nanoparticle solutions as adhesives for gels and biological tissues. Nature 505:382–385.
14. Klouda L, Mikos AG (2008) Thermoresponsive hydrogels in biomedical applications - A review. Eur J Pharm Biopharm 68:34–35.
15. Di Michele L, et al. (2013) Multiscale kinetic self-assembly of DNA-coated colloids. Nat Commun 4:2007.
16. Pan W, et al. (2016) Effects of chain flexibility on the properties of DNA hydrogels. Soft Matter 12:5537–5541.
17. Li J, Ngei T, Wu C (2010) The slow relaxation mode: From solutions to gel networks. Polym J 42:609–625.
18. Pine D, Weitz D, Chaikin P, Herbolzheimer E (1988) Diffusing wave spectroscopy. Phys Rev Lett 60:1134–1137.
19. Mason TG, Weitz D (1995) Optical measurements of frequency-dependent linear viscoelasticity of moduli of complex fluids. Phys Rev Lett 74:1250–1253.
20. Mason TG, Gang H, Weitz DA (1997) Diffusing wave-spectroscopy measurements of viscoelasticity of complex fluids. JOSA A 14:139–149.
21. Wei S, Tournilhan F, Leibler L (2004) Novel amphiphilic conetworks composed of telechelic polymers for self-assembly. Adv Mater 16:559–562.
22. Mason TG, Weitz D (1995) Optical measurements of frequency-dependent linear viscoelastic moduli of complex fluids. Phys Rev Lett 74:1250–1253.