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Facile Fabrication of Tough Hydrogels Physically Cross-Linked by Strong Cooperative Hydrogen Bonding

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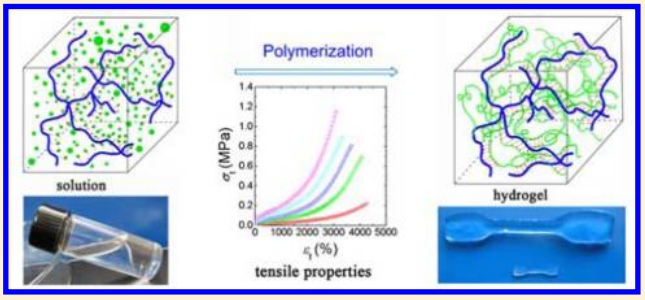
Supporting Information

ABSTRACT: Novel hydrogels with excellent mechanical properties have prompted applications in biomedical and other fields. The reported tough hydrogels are usually fabricated by complicated chemical and/or physical methods. To develop more facile fabrication methods is very important for the practical applications of tough hydrogels. We report a very simple yet novel method for fabricating tough hydrogels that are totally physically cross-linked by cooperative hydrogen bonding between a pre-existing polymer and an *in situ* polymerized polymer. In this work, tough hydrogels are prepared by heating aqueous acrylamide (AAm) solution in the presence of poly(*N*-vinylpyrrolidone) (PVP) but without any chemical initiators or covalent bonding cross-linking agents. Mechanical tests of the as-prepared and swollen PVP-*in situ*-PAAm hydrogels show that they exhibit very high tensile strengths, high tensile extensibility, high compressive strengths, and low moduli. Comparative synthesis experiments, DSC characterization, and molecular modeling indicate that the formation of strong cooperative hydrogen bonding between the pre-existing PVP and the *in situ* formed PAAm chains contributes to the gel formation and the toughening of the hydrogels. The unique microstructure of the gels with evenly distributed flexible cross-linking sites and long polymer chains attached to them endow the hydrogels with an excellent mechanism of distributing the applied load.



1. INTRODUCTION

Hydrogels have many applications in pharmaceutical, biomedical, and industrial fields;[1] they are used as superabsorbents,[2] contact lenses, drug delivery systems,[3] sensors and actuators,[4] scaffolds for tissue engineering,[5] etc. However, conventional synthetic hydrogels are generally mechanically very weak due to the structural inhomogeneity in the gel network. To broaden the applications of hydrogels, especially in load bearing applications, many efforts have been devoted to develop novel mechanically strong hydrogels.[6] Typical tough hydrogels include slide-ring (SR) gel,[7] nanocomposite (NC) gel,[8] double-network (DN) gel,[9] tetra-arm poly(ethylene glycol) (PEG) gel,[10] and the hydrogels synthesized by using polyfunctional initiating and cross-linking centers (PFICC).[11]

The main strategy used for fabricating these tough hydrogels is to synthesize or introduce special and uniformly distributed cross-linking points or centers, which are effective in distributing the applied load, into the gels. Tetra-arm gels have a homogeneous network structure, which is created by using two kinds of tetrahedron-like macromonomers with monodispersed chain extenders.[10b] SR gels are made by using sliding cross-linkers, which are covalently joined α-cyclodextrin (α-CD) molecules with PEG chains threaded through.[7] NC gels employ nanoscopic clay slabs as polyfunctional cross-linking centers (PFC), on which chemical initiators are adsorbed to initiate the polymerization of monomer molecules and the formed polymer chains are physically adsorbed.[8] For the PFICC-based gels,[11] part of the polymer chains is covalently joined to the cross-linking centers. DN gels are synthesized via a two-step sequential free-radical polymerization process by combining a rigid and brittle first network with a soft and ductile second network.[9] The toughing mechanism of DN gels is different to the above-mentioned tough hydrogels, in which the first brittle network serves as sacrificial bonds, which breaks into small clusters to efficiently disperse the stress.[12]

There is a current trend to fabricate tough synthetic hydrogels based on physical cross-linking. The physically reversible interactions capable of constructing and maintaining the gel network include hydrogen bonding,[13] ionic inter-

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action,[14] hydrophobic interactions,[15] crystallization,[16] inclusion complexation,[17] and interpolymer complexation.[18] However, in most cases, chemical cross-linking is combined with physical cross-linking to improve the mechanical properties of the gels. For example, Suo and co-workers[19] reported the synthesis of highly stretchable and tough hydrogels from polymers forming ionically and covalently cross-linked networks. Only a few tough hydrogels are totally physically cross-linked. One example is the ionically cross-linked triblock copolymer hydrogels;[14b] another example is poly(vinyl alcohol) (PVA) hydrogels made by the repeated freezing−thawing method, in which part of PVA chains form crystallites held by very strong hydrogen bonds (H-bonds).[16,20]

The synthesis methods for the chemically and/or physically cross-linked tough gels are usually multistepped and time-consuming, and for some methods monodispersed polymers[10] or triblock copolymers with ionic domains[14b] are required. To develop more facile fabrication methods is very important for the practical applications of tough hydrogels.

It is well-known that there is a cooperative effect in H-bonds of small molecules and polymers which exist in chains, rings, or other shapes where the molecule can participate concertedly as a donor and as an acceptor.[21] The cooperative effect significantly enhances the H-bond strength. A nonlinear or higher-order cooperativity is even found in the hydrogen bonding between two polymers.[22] Hydrogen-bonding cooperativity plays an important role in stabilizing secondary and tertiary structures of proteins, DNA duplexes, and ligand−receptor complexes.[23] For example, the strong cooperative hydrogen bonding between the base pairs of the double helix of DNA is the main reason for the formation of chiral assemblies that are stable in water, where solvent molecules can compete effectively for hydrogen bonds.[21d]

The H-bond cooperativity between two polymers leads to the formation of interpolymer complexes with properties entirely different to the properties of their component polymers.[24] There are some reports on the fabrication of hydrogels utilizing the interpolymer complexation as one of the cross-linking mechanisms, usually combined with chemical cross-linking.[18b,25] However, these hydrogels usually do not typically exhibit high mechanical strengths. One exception is the interpenetrating polymer network (IPN) systems reported by Frank and co-workers, in which interpolymer complexes are formed between a cross-linked poly(acrylic acid) (PAA) network and a chemically cross-linked poly(ethylene glycol) (PEG) network.[18c−e] There is no report on the fabrication of tough hydrogels physically cross-linked by only interpolymer complexation (cooperative hydrogen bonding) yet.

Herein, we report a very simple yet novel method for fabricating tough hydrogels cross-linked by cooperative hydrogen bonding. We found with some surprise that when a polymer—poly(*N*-vinylpyrrolidone) (PVP)—and a monomer—acrylamide (AAm)—were dissolved into water to form a solution, a hydrogel could be formed when the deaerated solution was kept at a moderately elevated temperature without using any specific chemical initiators or covalent bonding cross-linking agents. Tensile and compressive tests show that the mechanical properties of the hydrogels are among the highest ever reported. Comparative synthesis experiments, molecular modeling, and DSC characterization were performed to understand the formation and toughening mechanisms of the hydrogels.

2. EXPERIMENTAL SECTION

Materials. Poly(*N*-vinylpyrrolidone) (PVP, $M_w = 4.0 \times 10^4$, high pure grade) and acrylamide (AAm, ultra pure grade) were purchased from Amresco Inc. (OH). Polyacrylamide (PAAm, $M_w = 8.3 \times 10^5$, 10% in water) was purchased from TCI (Shanghai, China). *N*-Methylpyrrolidone (AR grade) was from Sinopharm Chemical Reagent Co., Ltd. (Shanghai, China). *N*-Vinyl-2-pyrrolidone (AR grade) was from Acros Organics (Morris Plains, NJ); it was redistilled before use.

Hydrogel Synthesis. PVP was dissolved in deionized water, and then AAm was added to the PVP solution. The concentrations of PVP and AAm are specified in the corresponding main text. The homogeneous transparent PVP−AAm solution was transferred into a mold made by placing a silicone spacer with a height of 2 or 4 mm between two glass plates. Dissolved oxygen in the solution was then removed by three cycles of vacuum evacuation and exchange with high-purity nitrogen. Finally, the mold was kept at 56 °C for 36 h to transform the solution into a hydrogel. The hydrogels synthesized by this approach are referred to here as PVP-*in situ*-PAAm gels.

For comparison, aqueous solutions containing *N*-methylpyrrolidone and AAm, *N*-vinyl-2-pyrrolidone and AAm, and that with only AAm were also prepared, and they were treated in the same way described above.

Polymer Yield and Molecular Weight Measurements. To measure the conversion of monomer to polymer and the molecular weights of the *in situ* formed PAAm in the presence of pre-existing PVP, the PAAm need to be isolated first. The PVP-*in situ*-PAAm hydrogels were dissolved in water in the presence of urea (5 mol L^{-1}) at 75 °C in 6 days. And then the dissolved solutions were dialyzed against pure water by using dialysis bag with the molecular weight cutoff of 5.0×10^5. The dialysis was performed for a week with refreshing water twice a day. Since the molecular weight of PVP used in our work is 4.0×10^4, the dialysis process is believed to be able to separate the pre-existing PVP from the *in situ* formed PAAm. The dialyzed solutions were collected for determining polymer yield (Y_p) and for GPC analysis. The molecular weights and polydispersity indices (PDIs) of the PAAm's were measured with a Waters 515 gel permeation chromatograph (GPC) equipped with a Wyatt DAWN HELEOS-II (laser 658.0 nm) laser light scattering detector and a Wyatt Optilab rEX refractive index detector using water as the eluent at a flow rate of 0.5 mL min^{-1}.

Mechanical Tests. An Instron 3366 electronic universal testing machine (Instron Corporation, Norwood, MA) was used for mechanical testing. The as-prepared hydrogels were cut into dumbbell-shaped specimens, in accordance to DIN-53504 S3 (inner width: 2 mm; gauge length: 10 mm, thickness: 2 mm) for tensile testing. Tensile tests were performed at a cross-head speed of 800% min^{-1}. The tensile stress σ_t was calculated as follows: σ_t = load/tw (t and w were the initial thickness and width of the dumbbell-shaped hydrogel sample, respectively). The tensile strain ε_t is defined as the change in the grip separation relative to the initial length, $\varepsilon_t = [\Delta l/l_0] \times 100\%$ (l is the grip displacement during testing and l_0 is the length of the sample before tests). Tensile fracture stress or tensile strength (σ_b) and the tensile fracture strain or elongation at break (ε_b) are the tensile stress and strain at which the specimen breaks, respectively. Stress−strain data between ε_t = 10%−30% were used to calculate initial tensile elastic modulus (E).

Tensile mechanical properties of the PVP-*in situ*-PAAm hydrogels swollen to 90% water content and equilibrium swollen samples were also measured. To prepare gel samples with 90% water content, the as-prepared gel samples were swollen to the required water content, and then the swollen samples were kept in sealed plastic bags for 7 days to ensure the uniform swelling throughout the samples. To prepare equilibrium swollen samples, dumbbell-shaped specimens (standardized as DIN-53504 S3) cut from the as-prepared hydrogels were immersed in deionized water for 8 days, and their dimensions (length, width, and height) were recorded.

Uniaxial compression tests were performed on cylindrically shaped specimens with a thickness of 4 mm and a diameter of 20 mm at a

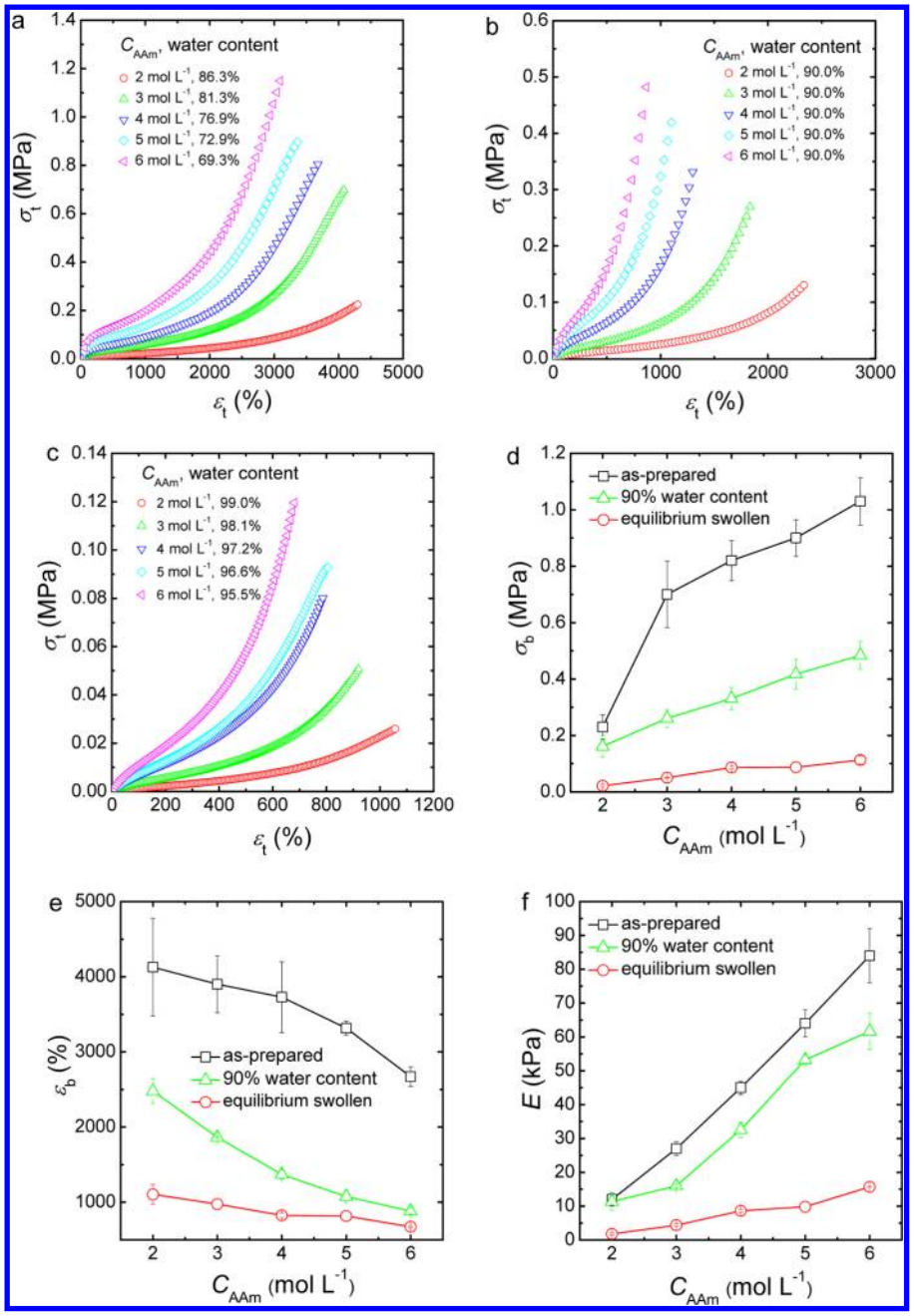

Figure 1. Tensile mechanical properties of the hydrogels synthesized with a fixed PVP concentration (0.017 g mL^{-1}) and different monomer concentrations (C_{AAm}). (a–c) Typical stress–strain (σ_t–ε_t) curves of the as-prepared (a), swollen to 90% water content (b), and equilibrium swollen hydrogels (c); (d–f) the tensile strength (σ_b), elongation at break (ε_b), and elastic modulus (E) of the hydrogels as a function of C_{AAm}, respectively.

crosshead speed of 4 mm min^{-1}. The compressive stress (σ_c) was calculated by σ_c = load/πr^2, where r is the original radius of the specimen. The strain (ε_c) under compression was defined as the change in the thickness (h) relative to the original thickness (h_0) of the freestanding specimen, $\varepsilon_c = (h_0 - h)/h_0$.

Prior to mechanical testing, the gel specimens were coated with a thin layer of silicon oil to prevent the evaporation of water. For each sample, at least four specimens were tested.

Computational Methods. The geometric parameters were optimized using B3LYP/6-311+G(d) methods and confirmed with vibrational analysis to ensure they are of correct curvatures in the corresponding potential energy surfaces. Single-point energy corrections have been performed with MP2/cc-PVQZ in order to consider the basis set effect and dispersion interaction. The optimized structures were also employed to perform the basis set superposition errors (BSSE) correction using counterpoise method[26] at the MP2/cc-PVQZ level, in order to calculate accurately the hydrogen-bonding interaction energies between donating polymers.

DSC Characterizations. Glass transition temperatures of vacuum-dried samples were determined with differential scanning calorimetry (DSC). DSC analyses were performed with DSC 1 (Mettler-Toledo, Switzerland) over the temperature range of 100–220 °C at a heating rate of 10 °C min^{-1} in a nitrogen atmosphere. Two consecutive scans were performed on each sample, and the middle point value of the transition in the second scan was taken as the glass transition temperature.

3. RESULTS AND DISCUSSION

Mechanical Properties of the Hydrogels. Hydrogels were made by heating the aqueous acrylamide (AAm) solution in the presence of poly(*N*-vinylpyrrolidone) (PVP) but without

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any chemical initiators or covalent bonding cross-linking agents. The hydrogels synthesized by this approach are referred to here as PVP-*in situ*-PAAm gels. We synthesized two series of PVP-*in situ*-PAAm hydrogels by fixing the concentration of one component but changing the other one. Both the as-prepared and swollen hydrogels displayed good mechanical properties, which can be felt by simple stretching or compression by hand. Tensile and compressive tests were performed on standardized gel specimens to obtain their mechanical properties.

We initially tested the PVP-*in situ*-PAAm hydrogels synthesized with a fixed PVP concentration (0.017 g mL^{-1}) but varying AAm concentrations (C_{AAm}). The typical tensile stress−strain (σ_t−ε_t) curves of the as-prepared, swollen to 90% water content and equilibrium swollen PVP-*in situ*-PAAm hydrogels are shown in Figure 1a−c, respectively, and the water contents of the hydrogels are also shown. Significant strain hardening is observed for both as-prepared and swollen hydrogels. The tensile strength (σ_b), fracture strain (ε_b), and elastic modulus (E) of the hydrogels are summarized in Figure 1d−f, respectively. The hydrogels exhibited very good mechanical properties. The as-prepared hydrogels showed very high σ_b's, which increased with increasing C_{AAm}, from 0.25 MPa at 2.0 mol L^{-1} to 1.20 MPa at 6.0 mol L^{-1}. Moreover, the hydrogels had extremely high extensibility (Figure S1 and Movie S1), as shown by the ultrahigh elongations, which were generally more than 3000% and the maximum was up to 4200%. The as-prepared hydrogels showed low E, in a range of 12−84 kPa, which increased gradually with increasing C_{AAm}.

The swollen hydrogels still had excellent mechanical properties. For the hydrogels swollen to 90% water content, their σ_b's increased with the increase of C_{AAm}, from 0.13 MPa at 2.0 mol L^{-1} to 0.53 MPa at 6.0 mol L^{-1}, the ε_b's were usually more than 880%, reaching a maximum of 2480%, and the E's were from 11.3 to 61.7 kPa. For the equilibrium swollen hydrogels with very high water content in the range of 95.5−99.0%, their σ_b's increased from 0.03 MPa at 2.0 mol L^{-1} to 0.13 MPa at 6.0 mol L^{-1}; the ε_b's were usually more than 700%, reaching a maximum of 1050%. The E's of the equilibrium swollen gels were between one-sixth and one-seventh of the corresponding as-prepared hydrogels.

Figure 2 shows the σ_b and E of the as-prepared hydrogels synthesized with a fixed C_{AAm} (5 mol L^{-1}) but varying PVP concentrations (C_{PVP}). The σ_b of the gels initially increased with C_{PVP} until 0.017 g mL^{-1} and then decreased as C_{PVP} was further increased; an abrupt decrease was found at 0.067 g mL^{-1}. E remained almost constant as the concentration was increased from 8.3×10^{-3} to 0.05 g mL^{-1}; then it also abruptly decreased when the concentration was increased beyond 0.05 g mL^{-1}. By calculation, the mass ratio of PAAm to PVP is about 10 at the C_{AAm} of 5 mol L^{-1} and the C_{PVP} of 0.05 g mL^{-1}. When the mass ratio of AAm to PVP was close to 1 or less, no tough hydrogels could be formed. In addition, when C_{PVP} was less than 8.3×10^{-3} g mL^{-1}, no tough hydrogels could be formed either. These results suggest a proper PVP concentration range is required for forming tough hydrogels.

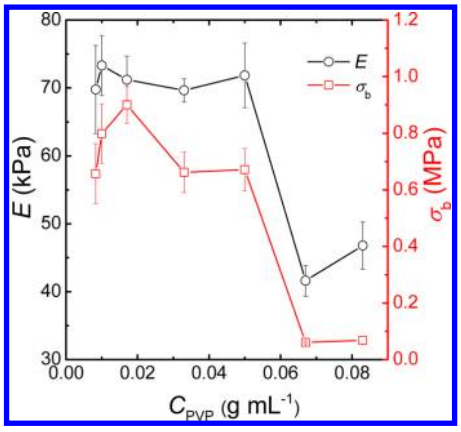

Figure 2. σ_b and E of the as-prepared hydrogels synthesized with a fixed AAm concentration (5 mol L^{-1}) but varying PVP concentrations (C_{PVP}). The water content of the hydrogels was 70−73%.

The compressive mechanical properties of the as-prepared PVP-*in situ*-PAAm hydrogels were also measured. Figure 3a shows the compressive stress−strain (σ_c−ε_c) curves of the PVP-*in situ*-PAAm hydrogels synthesized with different C_{AAm}, and Figure 3b gives the σ_c's of the gels at a ε_c of 0.95. The PVP-*in situ*-PAAm hydrogels did not fracture during the tests, and they could recover their original shapes immediately after release of the load. The σ_c's increased rapidly with increasing C_{AAm} from 2 to 5 mol L^{-1} and then slowly, and the highest one at 6 mol L^{-1} is about 17.0 MPa (Figure 3b).

The PVP-*in situ*-PAAm hydrogels also exhibited excellent shape recoverability. After being elongated to the fracture strain, the hydrogel specimens had almost the same length as the original specimens, and the gels also exhibited low hysteresis ratio and residual strains during the cyclic tensile tests (Figure S2). The PVP-*in situ*-PAAm hydrogels contain a high volume fraction of water, and they exhibit rubber-like elastic deformation over strains up to the breaking strain, implying that the cross-linked points are stable. In contrast, a concentrated polymer solution containing polymer chains of high molar mass would exhibit viscoelastic deformation under similar conditions as the entanglements translate as a function of time. Hence, for the hydrogels reported here, the interaction energy at the cross-linked points is much higher than the thermal energy.

Both tensile and compressive mechanical testing results show that the PVP-*in situ*-PAAm hydrogels have excellent mechanical properties. By comparing our testing results with reported values,[6c] it is observed that the mechanical properties of the PVP-*in situ*-PAAm hydrogels are much better than those of conventional synthetic hydrogels and are comparable to those of the toughest reported hydrogels. For example, the σ_b's of the as-prepared PVP-*in situ*-PAAm hydrogels are in the range 0.25−1.20 MPa, which are among the highest values for tough hydrogels, and are 2−3 orders of magnitude higher than those of the conventional synthetic hydrogels (~10 kPa). The as-prepared PVP-*in situ*-PAAm hydrogels show extremely high extensibility, with ε_b generally more than 3000%, similar to the most extensible NC gels and the graphene oxide nanocomposite hydrogel based on PFICC.[11c] The PVP-*in situ*-PAAm hydrogels show relatively low E (<90 kPa), similar to those of NC gels (1−50 kPa).[16] For the swollen hydrogels with water contents even more than 95%, their σ_b's are still high (0.03−0.13 MPa). There is an interesting comparison with a biological gel−jellyfish mesogloea, which has a σ_b of 0.17 MPa at a water content of 99%.[27]

We have also made a comparison of our gels with the PEG/PAA IPN gels described by Frank and co-workers.[18c,d] For the convenience of comparison, we calculated the true tensile strengths (σ_{true}) of our gels (Figure S3). The maximum σ_{true} reported by Frank and co-workers is about 13 MPa, whereas

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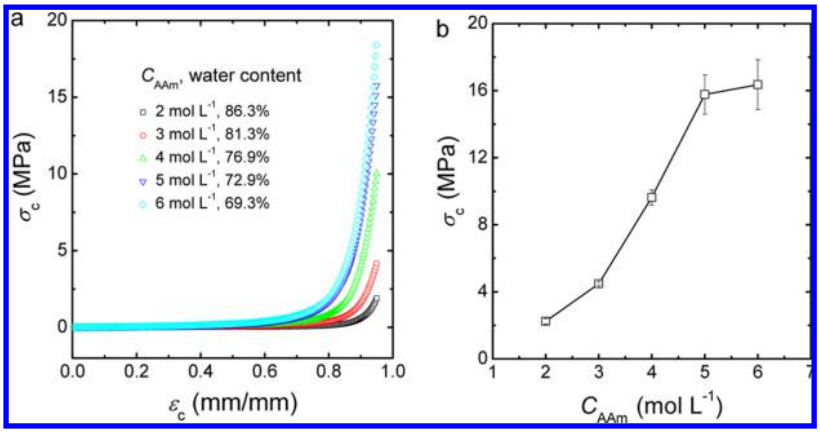

Figure 3. Typical compressive stress−strain curves of the as-prepared PVP-*in situ*-PAAm hydrogels (a) and the σ_c of the gels at a ε_c of 0.95 (b).

the minimum σ_{true} of our gels is about 10 MPa and the maximum value is about 38 MPa. It is also necessary to mention that the tensile mechanical properties of the gels in this work are even better than the gels synthesized by using peroxidized PVP chains as PFICC, in which both chemical and physical cross-links exist.[11d]

Hydrogel Formation and Toughening Mechanism. A new finding in this work is that very tough hydrogels can be formed by heating aqueous solutions of PVP and AAm in the absence of any chemical initiators and cross-linking agents. Very interesting and important questions arising here are why can the hydrogels be formed in the absence of any chemical initiators and covalent cross-linkers and why are the hydrogels so tough?

Comparative Synthesis Experiments. For the synthesis of a hydrogel starting from a monomer solution, the monomer molecules need to be polymerized and cross-linked. The polymerization of AAm is easy to understand, since AAm can be easily polymerized by self-initiated polymerization at an elevated temperature. When an AAm aqueous solution was heated, it became highly viscous at a low monomer concentration (2 mol L^{-1}) (Figure 4a,b) or was transformed into a gel-like material (at higher concentrations). Similarly, when AAm was polymerized in the presence of *N*-methylpyrrolidone (NMP), a small molecule containing a pyrrolidone ring, a gel-like material could be formed that could flow very slowly when the reactor was inverted or tilted (Figure 4c). The gel-like materials could be easily deformed and could be dissolved in water easily. In contrast, when sufficient PVP was present in the AAm solution during polymerization, the system formed a tough gel which could not flow even at the lowest concentration tested (Figure 4b,d).

These experiments proved that AAm can be polymerized by self-initiation under heating, but tough hydrogels can only be obtained in the presence of a proper amount of PVP, suggesting that PVP plays a vital role in the cross-linking and hence the formation of the tough hydrogels. The side group of PVP is a pyrrolidone ring, which is a lactam. There is not a mechanism for the formation of covalent bonds between PVP and PAAm under the mild reaction conditions, and hence the hydrogels are not chemically cross-linked. The obtained hydrogels did not dissolve when stored in a large excess amount of deionized water at ambient temperature for more than 8 months (Figure S4), instead reaching equilibrium swelling. However, in the presence of urea (5 mol L^{-1}), an efficient hydrogen-bond-breaking reagent, the PVP-*in situ*-

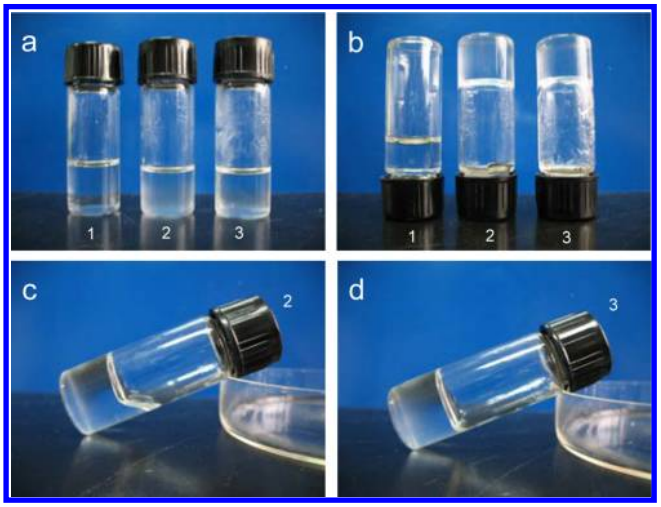

Figure 4. Comparative study of the formation of gels with the solutions containing only AAm (1), *N*-methylpyrrolidone and AAm (2), and PVP and AAm (3). (a) The solutions, (b) the reacted systems, (c, d) the reacted systems 2 (c) and 3 (d) after being tilted for about 2 h. Reaction conditions: C_{AAm} = 2 mol L^{-1}, C_{PVP} = C_{NMP} = 0.017 g mL^{-1}.

PAAm hydrogels dissolved in less than 25 days at room temperature, forming a viscous solution. Therefore, the hydrogels are physically cross-linked, and the most likely physical interaction is hydrogen bonding, since both PVP and PAAm have appropriate functional groups (i.e., pyrrolidone ring and amide) capable of forming hydrogen bonding. This experiment proves that the PVP-*in situ*-PAAm hydrogel is a physical gel cross-linked by H-bonds. It is necessary to mention that H-bonds can also be formed between PAAm chains, but it seems that they are not sufficiently stable to transform a PAAm solution into a hydrogel.

We isolated the PAAm's in the PVP-*in situ*-PAAm hydrogels synthesized with different C_{AAm} and then measured their yields (Y_p) and molecular weights, shown in Table S1. The yields of PAAm are usually more than 97%, indicating that the monomer AAm was almost completely transformed into PAAm. The GPC results show that the *in situ* formed PAAm's have high molecular weights, which decreases with increasing C_{AAm}, with the M_w ranging from 1.70 × 10^6 g mol^{-1} at 2 mol L^{-1} to 8.01 × 10^5 g mol^{-1} at 6 mol L^{-1}. In contrast, the polymerization of neat AAm in the absence of PVP led to lower polymer yields (Y_p) and much lower molecular weights of the PAAm's (Table

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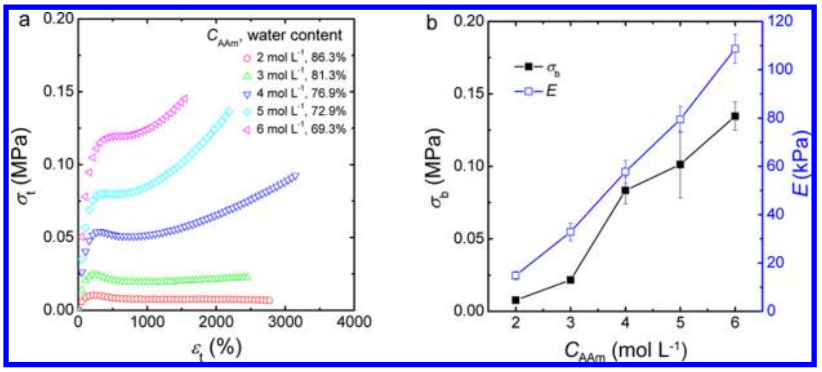


Figure 5. Tensile mechanical properties of the as-prepared P(AAm-*co*-NVP) hydrogels synthesized with a fixed NVP concentration (0.017 g mL^{-1}) but different C_{AAm}. (a) Typical stress−strain (σ_t−ε_t) curves; (b) the tensile strength (σ_b) and the elastic modulus (E) of the hydrogels as a function of C_{AAm}.

S1), which might be the main reason for the formation of highly viscous or gel-like materials. van de Grampel et al.[28] reported that in the template polymerization of *N*-vinylimidazole along poly(methacrylic acid) the monomer conversions and the molecular weights of the polymers are much higher than those of polymers produced in the absence of a template, and the reason is thought to be the retardation of termination of radicals propagating alongside the template. The decreasing molecular weights with C_{AAm} of PAAm in the PVP-*in situ*-PAAm hydrogels might be due to the more possible occurrence of cross-termination by small radicals at a higher C_{AAm}.[28b]

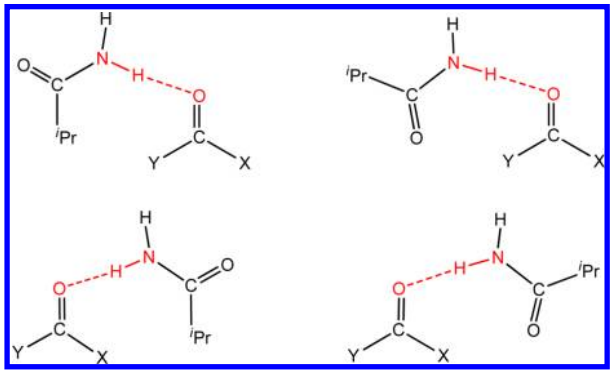
Materials were also prepared by copolymerizing *N*-vinyl-2-pyrrolidone (NVP) and AAm in the absence of any chemical initiators and cross-linking agents, referred to as P(AAm-*co*-NVP). Hydrogels were obtained when C_{AAm} exceeded C_{NVP}, and these materials dissolved easily in water at room temperature forming sticky solutions. When C_{NVP} was equal to or higher than C_{AAm}, only viscous solutions were formed (Figure S5).

The P(AAm-*co*-NVP) gels were strong enough for mechanical tests. The tensile properties of the gels synthesized with a fixed NVP concentration (C_{NVP}) (0.017 g mL^{-1} or 0.15 mol L^{-1}) but varying C_{AAm} are shown in Figure 5. The σ_t−ε_t curves of the P(AAm-*co*-NVP) gels (Figure 5a) are quite different to those of the PVP-*in situ*-PAAm gels (Figure 1a), exhibiting an obvious yielding phenomenon with high plasticity. These materials were not elastic beyond strains of 200%. The σ_b's of the P(AAm-*co*-NVP) gels (Figure 5b) were an order of magnitude lower than the PVP-*in situ*-PAAm gels (Figure 1c). Interestingly, the E's of the P(AAm-*co*-NVP) gels (Figure 5b) were slightly higher than the PVP-*in situ*-PAAm gels (Figure 1e). Hydrogels were also synthesized with a fixed C_{AAm} (5 mol L^{-1}) but varying C_{NVP}. The mechanical test results are provided in Figure S6. More obvious yielding is shown in the σ_t−ε_t curves of the P(AAm-*co*-NVP) gels synthesized with a higher C_{NVP}. Both σ_b and ε_b decreased with an increase of C_{NVP}, whereas E increased. These results suggest that there are H-bonds between the *in situ* formed PVP and PAAm chains, but they are not as strong as those formed between the pre-existing PVP chains and *in situ* formed PAAm chains.

Molecular Modeling. To obtain more quantitative information on the hydrogen bonding between PVP and PAAm, the following calculations were carried out to reveal the geometric features and interaction energies of H-bonds formed between the amide group of AAm (or PAAm) and the pyrrolidone ring of PVP to compare with those occurring in the PAAm autoassociation system. To simplify the calculations, 2-methylpropanamide (MP) and *N*-isopropylpyrrolidone (PP) are used as the model molecules of PAAm and PVP, respectively. As the first step, some probable conformers of MP and PP molecules were located and characterized by geometries and relative energies depicted in Figure S7. The energetically favorable structures MP-1 and PP-1 were selected to model PAAm and PVP chains for constructing all the H-bond complexes afterward.

MP molecules are able to serve as a proton donor of H-bond manifested by the two nonequivalent hydrogen atoms in amide (−NH_2) group; the carbonyl oxygen could interact with the proton donor from either of two directions, where one is on the side of substituent X group and another is on the side of substituent Y group when X is not equal to Y. Therefore, four types of H-bond complex bearing one molecule of donor and acceptor could be anticipated (Scheme 1).

Scheme 1. General Representations of Four Types of H-Bond Occurring in the Bimolecular Complexes of MP−MP or MP−PP



The optimized geometries for the dimer of MP are represented in Figure 6, along with the calculated interaction energies in kJ/mol. These data indicate that MP-MP-1 and MP-MP-2 are more favorable than MP-MP-3 by about 4 kJ/mol, which might indicate that the region syn to isopropyl (or anti to −NH_2) group is more inclined to attract the proton of N−H··· O than the other side around carbonyl oxygen. MP-MP-4 represents the structure of the eight-membered cyclic dimer of MP, and it is doubly hydrogen-bonded. Although energetically far more stable than other three isomers, MP-MP-4 may not

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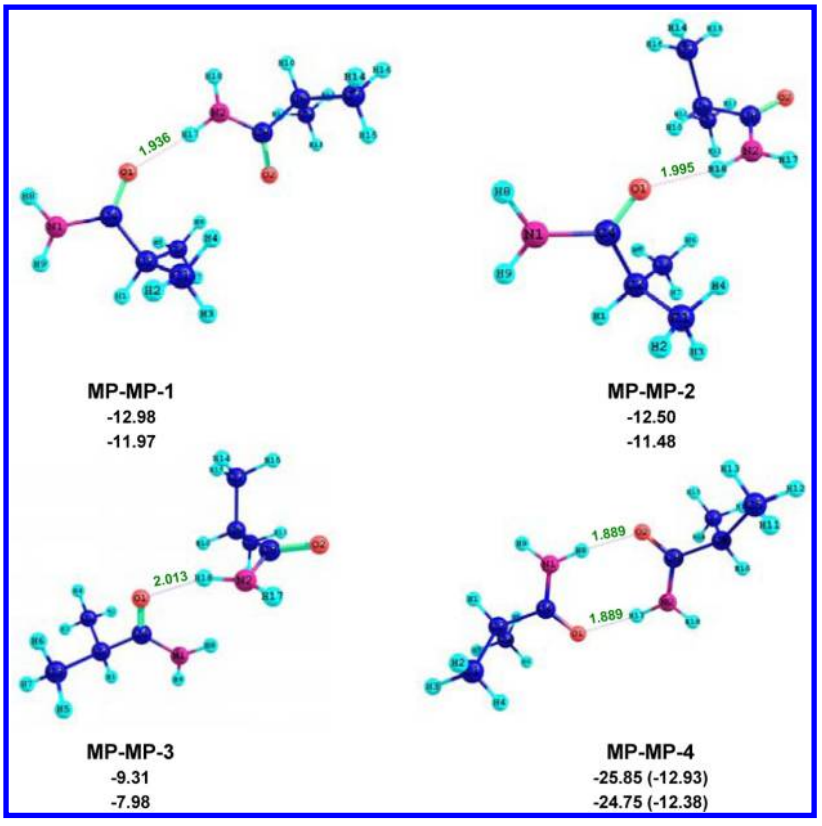

Figure 6. The 3D geometries of four stable structures for dimer of MP, along with the ZPE corrected interaction energies without BSSE (above) and with BSSE (below) in kJ/mol, in which the average interaction energies per H-bond for MP-MP-4 are given in parentheses.

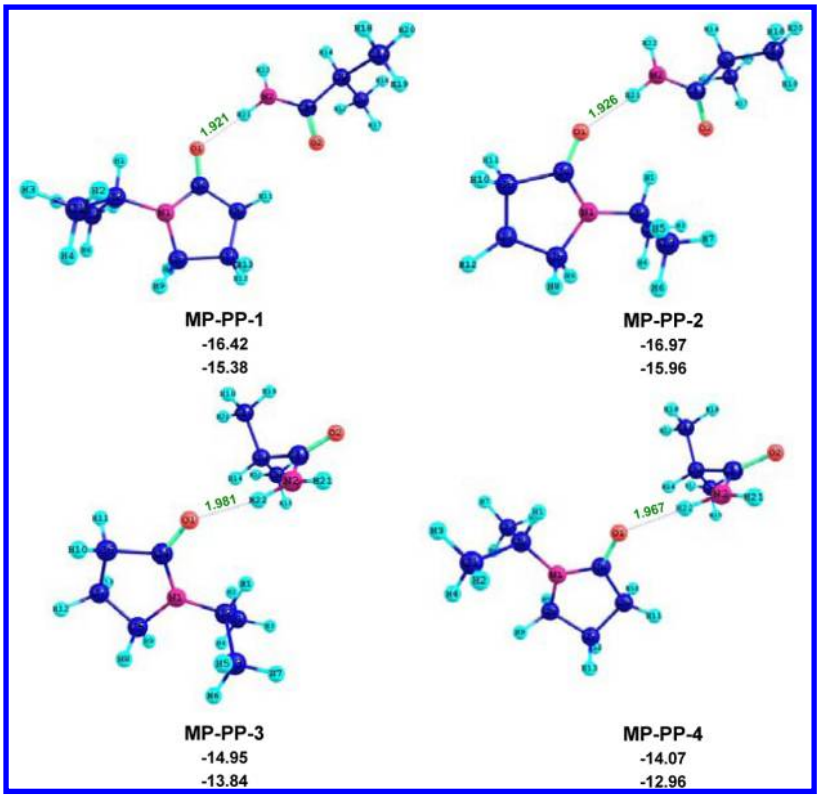

Figure 7. The 3D geometries of four stable structures for the MP−PP system, along with the ZPE corrected interaction energies without BSSE (above) and with BSSE (below) in kJ/mol.

represent the most frequent H-bond formation occurring in bulk MP, since it has two N−H···O H-bonds, and the average interaction energy per H-bond instead of the total interaction energy is responsible for estimating their relative stabilities. In

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addition, if the cooperative effect in hydrogen bonding is considered, which will be discussed in the following part, the dimer MP-MP-4 with two H-bonds would not be the dominant complex. The computed energetic data indicate that MP-MP-1, MP-MP-2, and MP-MP-4 are of nearly the same interaction energy per H-bond if BSSE were included and the binding energy for single H-bond in MP-MP system is estimated to be about −12 kJ/mol.

A similar study has been conducted for the MP−PP complex in order to compare the strength of N−H···O in different systems. Figure 7 shows the optimized geometries for four types of complex and their interaction energies in kJ/mol. It is in good agreement with the chemical intuition that MP-PP-1 and MP-PP-2 are the dominant structures, in that two carbonyl groups are close to antiparallel with each other and thus induce weak intermolecular attraction. The calculated interaction energies predicted the binding energy for a single H-bond in MP-PP system to be about −15 kJ/mol, 3.0 kJ/mol larger than that in a dimer of MP.

The interaction energies without and with zero-point energy (ZPE) corrections (E_{int}^{e} and E_{int}^{e+v}), the ZPE and BSSE corrected interaction energies (E_{int}^{cor}), and the average interaction energy per H-bond (E_{int}^{av}) of all the complexes in our calculations are summarized in Table 1.

Table 1. Computed Interaction Energies without and with ZPE Corrections (E_{int}^{e} and E_{int}^{e+v}), the ZPE and BSSE Corrected Interaction Energies (E_{int}^{cor}), and the Average Interaction Energy per H-Bond (E_{int}^{av}) for Some Selected H-Bond Complexes (in kJ/mol)

H-bond complex	E_{int}^{e}	E_{int}^{e+v}	BSSE	E_{int}^{cor}	E_{int}^{av}
MP-MP-1	−17.81	−12.98	1.01	−11.97	−11.97
MP-MP-2	−17.42	−12.50	1.02	−11.48	−11.48
MP-MP-3	−15.55	−9.31	1.33	−7.98	−7.98
MP-MP-4	−34.06	−25.85	1.11	−24.75	−12.38
MP-PP-1	−21.49	−16.42	1.04	−15.38	−15.38
MP-PP-2	−21.31	−16.97	1.01	−15.96	−15.96
MP-PP-3	−19.37	−14.95	1.14	−13.81	−13.81
MP-PP-4	−19.11	−14.07	1.12	−12.95	−12.95
PP-H_2O	−24.14	−16.50	1.22	−15.38	−15.38
MP-H_2O	−22.39	−13.17	1.63	−11.54	−11.54
PP-MP-H_2O	−49.45	−35.82	2.66	−33.16	−16.58
PP-MP-MP	−48.19	−38.54	2.09	−36.45	−18.23

Judging from the interaction energy per H-bond (E_{int}^{av}), which can be employed to estimate the relative H-bond strength in different systems, the H-bonds formed between PP and MP are stronger than those formed by the autoassociation of MP, and hence the MP-PP complexes are more favorable than the MP-MP complexes. However, we must note that in an aqueous solution the large amount of water molecules, which can also form MP-H_2O and/or PP-H_2O H-bonds, should compete with the proposed formation of H-bonds between MP and PP. We calculated the interaction energies of the complexes of PP and MP with water (Figure 8 and Table 1). The H-bond interaction energies of (MP) C═O···H−O−H and (PP) C═O···H−O−H are estimated to be −11.54 and −15.38 kJ/mol (with ZPE and BSSE corrections), very close to those occurring in the most stable MP-MP-1 and MP-PP-2 complexes (−11.97 and −15.96 kJ/mol), respectively. These results suggest that PP and MP can competitively form complexes with water.

To understand the cooperative effect in hydrogen bonding in our system, we calculated the interaction energies of the termolecular H-bond complexes of PP-MP-MP and PP-MP-H_2O (Figure 8 and Table 1). Note that the middle MP molecule concertedly serves as a donor to the left PP and an acceptor to the right MP or H−O−H for both cases. A significant cooperative effect was observed in the H-bonds of the complexes with only one more molecule; the average interaction energies per H-bond (E_{int}^{av}) for PP-MP-MP and PP-MP-H_2O are −18.23 and −16.58 kJ/mol, respectively, larger than those of the bimolecular complexes (−13 to −16 kJ/mol). The higher H-bond interaction energy of the PP-MP-MP than that of PP-MP-H_2O indicates that the former is the more favorable structure in our system. Our results are consistent with that obtained in a very recent theoretical calculation using formamide and *N*-methylacetamide as the model molecules.[23] The cooperativity of the H-bonded complexes become weaker when the N−H···O═C H-bonds are replaced by H-bonds with water, indicating that the N−H and C═O bonds prefer to form N−H···O═C H-bond rather than to form H-bonds with water.[23]

The modeling proves that the H-bonds formed between the acrylamide groups and the pyrrolidone groups are stronger than those formed between acrylamide groups; in addition, the significant cooperative effect makes the former further enhanced.

DSC Characterizations. To verify whether specific interactions occurred between the polymers PVP and PAAm, DSC characterizations were carried out on pure PVP, pure PAAm, and the dried PVP-*in situ*-PAAm hydrogels synthesized with different PVP/AAm mass ratios.

For normal miscible polymer blends, the glass transition temperatures (T_g) is between individual two polymers, and it shows a linear relationship against the polymer composition.[29] It has been shown by experiments[30] and theoretical calculations[31] that for some miscible polymer blends interacting with strong specific intermolecular forces the compositional dependence of T_g shows a maximum, exhibiting a positive deviation from linearity with blend composition, and if the specific interactions (e.g., hydrogen-bonding) are strong enough, the T_g of a blend can be much higher than component polymers.[30] For comparison, we prepared the blends of PVP and PAAm with different mass ratios by mixing the solutions of PVP and PAAm and then measured the T_g's of the vacuum-dried samples. The blends of PVP and PAAm always have a single T_g, which is positively deviated from linearity with blend composition (Figure S8), indicating there are strong intermolecular interactions (hydrogen bonding) between PVP and PAAm.

Figure 9a shows the DSC thermograms of the PVP-*in situ*-PAAm as well as pure PVP and pure PAAm, and Figure 9b shows the T_g as a function of PVP mass ratio. Pure PVP shows a T_g of 168.3 °C, and the T_g of the pure PAAm prepared in this work is measured to be 186.3 °C, which is very close to the reported value (188 °C),[32] but lower than that of the purchased PAAm with a higher molecular weight (192.9 °C, Figure S8). The dried hydrogels synthesized with the PVP mass ratio less than 40% show a single T_g, indicating the miscibility of PVP with PAAm. The T_g's of the dried hydrogels are always higher than those of the component polymers, and the T_g increases with the increasing mass ratio of PVP until 30%. Very interestingly, when the PVP mass ratio is 50% or more, a new T_g appears at a constant temperature (171. 3 °C) very

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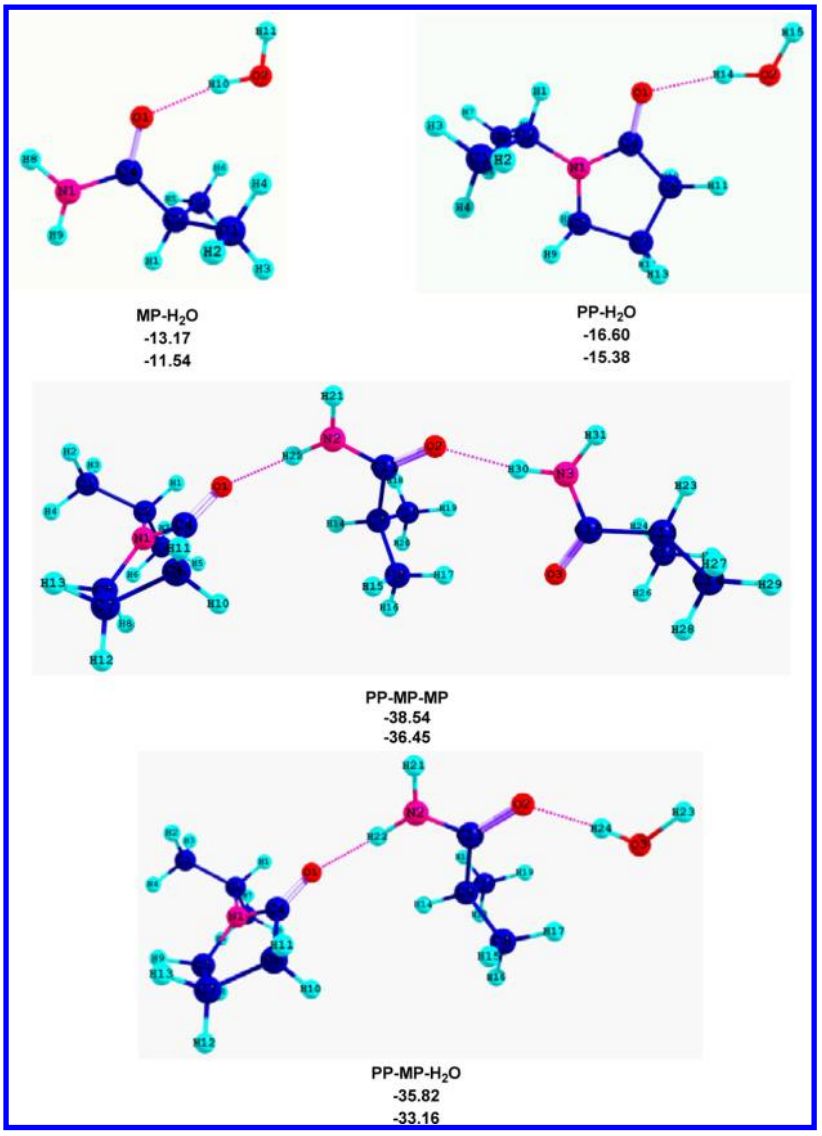

Figure 8. The 3D geometries of some bimolecular or termolecular complexes with water, along with the ZPE corrected interaction energies without BSSE (above) and with BSSE (below) in kJ/mol.

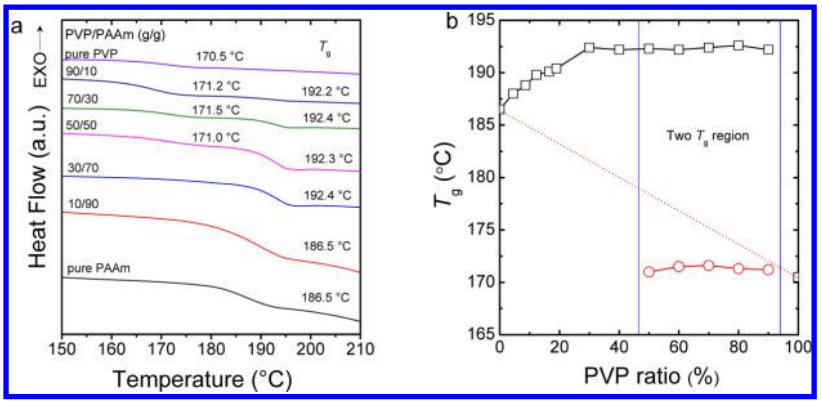

Figure 9. DSC characterizations. (a) Typical DSC thermograms of pure PVP, pure PAAm, and dried PVP-*in situ*-PAAm hydrogels synthesized with different PVP/AAm mass ratios. (b) The T_g's of the polymers. The dashed line shows the linearity with polymer composition.

close to that of pure PVP (Figure 9a, also see Figure S9 for more detailed information), while the other T_g keeps almost constant at about 192.4 °C as the PVP mass ratio increases, suggesting there are two distinctly different components in the PVP-*in situ*-PAAm. Since the new, lower T_g appears only after the mass ratio of PVP exceeds that of PAAm and it becomes more significant with the increase of PVP mass ratio (Figure S9), therefore it is very possibly attributed to the part of PVP

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which is not H-bonded with PAAm. The higher T_g is not due to PAAm with high molecular weights, since at a high PVP/AAm ratio the molecular weight of PAAm should decrease.[28b] Therefore, the higher T_g should be attributed to the hydrogen-bonded PVP and PAAm, as it appears at a temperature significantly higher than those of the component polymers, which is different than the case of blends of PVP and PAAm where the T_g is only positively deviated from linearity but less than that of PAAm. The appearance of two constant T_g's in the PVP-*in situ*-PAAm indicates that AAm molecules are consecutively rather than randomly H-bonded to PVP. The consecutive H-bonding enables the cooperative effect, leading to the stronger intermolecular interactions (hydrogen bonding) in the PVP-*in situ*-PAAm than those in the blends.

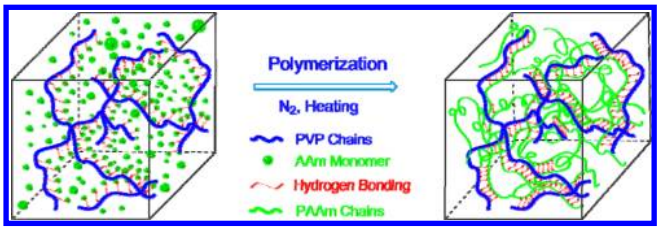
Discussion. The mechanical tests and the comparative synthesis experiments show that tough hydrogels can only be obtained when AAm is polymerized in the presence of an appropriate amount of PVP. In contrast, the polymerization of AAm itself or with NVP at similar concentrations does not form a tough gel.

The modeling proves that more significant cooperative effect exists in the stronger H-bonds formed between the acrylamide groups and the pyrrolidone groups than those formed between acrylamide groups. This can explain why the polymerization of AAm itself does not form a tough gel. However, the modeling does not explain why PVP-*in situ*-PAAm forms tough elastic gels, while P(AAm-*co*-NVP) forms relatively weak plastic gels. One major difference between the PVP in PVP-*in situ*-PAAm and P(AAm-*co*-NVP) is the number of consecutive NVP units in the former is large, while in the latter is small. Thus, some important differences in the hydrogen-bonding interactions and the topology of the gels induced by a large number of consecutive NVP units might be the main reason for the significant difference in the mechanical properties of the hydrogels.

The DSC studies on the T_g of the polymers have proven that intermolecular interactions (hydrogen bonding) in the PVP-*in situ*-PAAm is stronger than those in the blends of PVP and PAAm. As there are more than 300 consecutive NVP units in each PVP chain, during synthesis of PVP-*in situ*-PAAm many AAm molecules are attracted to each PVP chain by hydrogen bonding. The hydrogen-bonded AAm molecules are polymerized along the PVP chains with the PVP chains acting as templates for the polymerization of some of the AAm molecules.[33] Because of the large number of H-bonds, very strong hydrogen-bonding interactions are formed between PVP and PAAm chains. It is known that higher-order cooperativity exists in hydrogen bonding between a macromolecule and small molecules and between two macromolecules. In addition, the cooperativity becomes stronger with the increase of polymerization degree.[22] The main contribution to cooperativity is hypothesized to be due to a proximity effect, i.e., the lowering of the entropy demand of the next binding by the motional restriction imposed by the already existing bonds.[22] In addition, it is known that the hydrophobic interaction and the H-bond can mutually reinforce each other.[34] The cooperative effect of hydrophobic interaction of the hydrophobic parts of both PVP and PAAm chains concomitant to the formation of H-bonds might also contribute to the enhancement of the interactions between PVP and PAAm chains. Therefore, it is feasible that the very strong interactions between PVP and PAAm chains arise from cooperativity (or even higher-order cooperativity) between H-bonds and hydrophobic interactions lead to the formation of a stable interpolymer complex of PVP and PAAm.

The template polymerization of AAm molecules along PVP chains forms multiple H-bonded PVP-*in situ*-PAAm complexes. In this work AAm is in excess with respect to PVP; the large portion of AAm molecules, which are not H-bonded with PVP chains, polymerize to form unbound PAAm chains. The unbound sections of PAAm chains can be covalently joined to the PVP-*in situ*-PAAm complexes by mutual termination of growing polymer chains. The proposed formation mechanism and the microstructure of the PVP-*in situ*-PAAm hydrogels are shown in Scheme 2.

Scheme 2. Formation Mechanism and the Microstructure of the PVP-*in situ*-PAAm Hydrogels



The interaction energy of the PVP-*in situ*-PAAm interpolymer complex is from experimental evidence much higher than thermal energy. Hence, the PVP-*in situ*-PAAm interpolymer complexes function as cross-links forming a hydrogel. Different than the common cross-linking points formed by adding a chemical cross-linking agent and the polyfunctional cross-linking centers by using rigid inorganic clay sheets,[8] the flexible PVP-*in situ*-PAAm complexes in our gels can easily relocate or deform under an applied load. In addition, the cross-linking sites are evenly distributed due to the homogeneous distribution of PVP in the prereaction solution. The unbound sections of the PAAm chains are relatively long. The evenly distributed flexible cross-linking sites and the long polymer chains attached to them endow the hydrogels with an excellent mechanism of distributing the applied load evenly by the movement and deformation of all the polymer chains. The unique microstructure and the corresponding energy-dissipating mechanism account for the excellent mechanical property of the hydrogels.

The copolymerization of NVP and AAm generally produces a random copolymer due to the nature of radical copolymerization. For P(AAm-*co*-NVP) only weak plastic hydrogels are obtained. The most likely reason is that the average length of complexes formed between adjacent acrylamide and pyrolidone groups are smaller as it is known that the cooperativity in hydrogen bonding depends on polymerization degree[22] and steric factors.[21c] Therefore, relative to PVP-*in situ*-PAAm, there is a higher density of cross-linking sites arising from H-bond complexes, but these sites are less stable. In addition, mixing PAAm and PVP with different molecular weights and mass ratios also produced viscous liquid materials rather than tough hydrogels, since the steric hindrance of the pre-existing polymer chains inhibits the formation of consecutive hydrogen bonds between the chains.

The as-prepared tough hydrogels exhibit relatively low E (<90 kPa), indicating a low cross-linking density. We observed the increase of E with increasing C_{AAm} at a fixed C_{PVP} (Figure 1f), but the E remains constant or even decreases with increasing C_{PVP} at a fixed C_{AAm} (Figure 2). The possible reason

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is that the change of PVP/AAm mass ratio affects the numbers and molecular weights of PAAm formed, and hence the physical entanglement of the unbound sections of the PAAm chains. The strongly bound PVP-*in situ*-PAAm interpolymer complexes function as rigid cross-links and do not contribute an enthalpy term to the E of the hydrogels. During swelling in water, the weak physical cross-links between unbound sections of PAAm chains can be easily destroyed, leading to the dramatic decrease of the E of the gels. On the contrary, the strong PVP-*in situ*-PAAm interpolymer complexes are retained during swelling, leading to the equilibrium swelling rather than dissolution of the gels. Of course, when urea is present, the interpolymer complexes are broken.

It has been noted that the E's of the P(AAm-*co*-NVP) gels (Figure 5b) are slightly higher than those of the PVP-*in situ*-PAAm hydrogels (Figure 1e). A lower modulus indicates that there is a lower density of cross-links in the PVP-*in situ*-PAAm gels relative to the P(AAm-*co*-NVP) gels. The interaction energy of H-bonding between the NVP and AAm repeat units in the P(AAm-*co*-NVP) gels is probably comparable to thermal energy as the H-bonds are readily broken during swelling and under an applied force.

For the tensile strengths of the PVP-*in situ*-PAAm hydrogels, we observed the increase of σ_b with increasing C_{AAm} at a fixed C_{PVP} (Figure 1d), and the decrease of σ_b with increasing C_{PVP} at a fixed C_{AAm} (Figure 2). The DSC characterizations (Figure 9) suggest that the interactions (hydrogen bonding) between PVP and PAAm increase with increasing C_{PVP}, but the mechanical testing results show that excess PVP content leads to the decrease of the mechanical strength. These results can be understood from the proposed microstructure of the hydrogels. When the C_{PVP} is fixed, then the number of cross-linking sites is constant, the increase of C_{AAm} leads to the longer unbound sections of PAAm chains between the cross-linking sites. The longer unbound PAAm chains have a longer distance to move before reaching their maximum extension; therefore, they are more effectively in spreading the applied load and hence the higher mechanical strengths of the gels. On the other side, at a fixed C_{AAm}, the increase of PVP content leads to more bound PVP-*in situ*-PAAm strips, and hence the unbound sections of PAAm chains become fewer and shorter. The short chains are easier to break, and therefore the mechanical properties of the gels become worse. The tensile strength and modulus of the tough gels decrease dramatically when the C_{PVP} is greater than 0.05 g mL^{-1}. We can find that at this concentration the mass ratio of PAAm to PVP is about 10. There is a very interesting parallel with DN gels, which also show high mechanical strengths only when the molar ratio of the second loosely cross-linked network to the first highly cross-linked network is more than about 10, and the second network is in fact coupled to the first network by residual vinyl groups.[9] These results may suggest that the unbound sections of the PAAm chains in our PVP-*in situ*-PAAm gels and the polymer chains in the second network of DN gels can effectively distribute the applied force and stop cracks only when they are long enough; otherwise, they are easy to be broken due to their very limited deformation range. As an extreme example, no hydrogels will form when the mass ratio of AAm to PVP is close to 1 or less. The reason is that most AAm molecules are hydrogen-bonded to PVP chains, leading to the formation of only bound PVP-*in situ*-PAAm strips with no free unbound sections of PAAm chains linking them into a gel.

4. CONCLUSIONS

Our work demonstrates the fabrication of a novel type of tough hydrogels that are totally physically cross-linked by cooperative hydrogen bonding between a pre-existing polymer and an *in situ* polymerized polymer. This study provides an extremely simple method for fabricating tough hydrogels. Because of the wide choice of polymers and monomers that can form stable interpolymer complexes through hydrogen bonding,[18b] this method is versatile and can be applied to prepare many kinds of hydrogels. This method does not require any additional chemical initiators and cross-linkers; therefore, these hydrogels are more possible to find applications in biomedical fields.

The hydrogels exhibit excellent mechanical properties, as shown by the very high tensile and compressive strengths as well as extremely high extensibility, which are among the highest values for tough hydrogels. The mass ratio of the *in situ* formed polymer (PAAm) to the pre-existing polymer (PVP) is an important factor affecting the final mechanical properties of the gels, and tough hydrogels can only be obtained when the mass ratio is larger than 10.

The most important scientific problem arising from our work is that why are the hydrogels so tough. The microstructure and the toughening mechanism including the interactions between the polymer chains are quite different than other tough hydrogels. Our experimental and theoretical studies show that very strong interactions can be formed between the pre-existing PVP chains and the *in situ* formed PAAm chains due to the formation of cooperative H-bonds, which act as strong physical cross-linking sites in the gel formation. The evenly distributed flexible cross-linking sites and the long polymer chains attached to them endow the hydrogels with an excellent mechanism of distributing the applied load by the movement and deformation of all the polymer chains. The unique microstructure and the corresponding energy-dissipating mechanism account for the excellent mechanical property of the hydrogels. The detailed and precise mechanism is the subject of ongoing research.

■ ASSOCIATED CONTENT

Supporting Information

The Y_p, M_n, M_w, and PDI of PAAm's, photo showing the appearance of PVP-*in situ*-PAAm gel specimens, the resilience of the PVP-*in situ*-PAAm hydrogels, the true tensile strengths (σ_{true}) of the hydrogels, the photograph of the as-prepared and equilibrium swollen PVP-*in situ*-PAAm hydrogel specimens, the photographs of the reacted solutions containing an identical molar concentration of NVP and AAm, the tensile properties of the as-prepared P(AAm-*co*-NVP)hydrogels, the optimized geometries and relative energies (in kJ/mol) of two stable conformers for the molecule of MP (top) and PP (bottom), DSC thermograms, Movie S1 showing the elongation of a PVP-*in situ*-PAAm hydrogel with hand. This material is available free of charge via the Internet at http://pubs.acs.org.

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Notes

The authors declare no competing financial interest.

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