PERSPECTIVE

On the failure and fracture of hydrogels for cartilage replacement

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Abstract

The failure and fracture properties of hydrogels and hydrogel composites are considered in the contexts of applicable fracture mechanics and biomaterials engineering. Distinction is made between material failure properties, characterized by a work of failure independent of mechanism, and fracture properties, characterized by fracture resistance that requires clear identification of crack propagation. Although advanced hydrogels can exhibit very large works of failure relative to conventional single-network hydrogels, they do so only at large strains and are not well characterized by fracture properties alone. The large failure strains are not obviously relevant to many biomaterial applications such as cartilage replacement, for which the engineering requirements involve limited physiological strains. An example is given of fiber-reinforced hydrogel composites that demonstrate increased work of failure at small strains.

1. Introduction

In the 60 years since their introduction, hydrogels have generated substantial interest as materials with potential for use in biomedical applications, including implant coatings, tissue engineering scaffolds, and drug delivery vehicles [1]. The potential applications now also include the nonmedical, such as in agriculture, separation science, sensors, and soft electronics [2]. Hydrogel functionality arises from an underlying structure of a water-swollen hydrophilic polymer network forming a relatively sparse gel, in which both bound and unbound water is incorporated [3]. Due to the unbound water, there is an intrinsic solid-fluid dual nature to these materials and the dynamic flow properties of the fluid phase are often as important as the solid phase physical properties in determining mechanical behavior [3]. Hydrogel-based materials can be formed from a vast array of natural and synthetic gel chemistries. The resulting microstructures have features at a broad range of length scales, from molecular to macroscopic, such as single polymer network structures, multiple network structures, functionally graded structures, and composite material structures. The latter microstructural categories have emerged largely in response to the perceived poor physical properties of single network hydrogels that limit the potential usefulness of such materials for a variety of engineering applications.

In addition to chemistry and structure, there is also a growing interest in the failure and fracture behavior of hydrogel-based materials [4–6], along with a similar interest in natural soft tissues such as cartilage [7]. The relatively large water content in many hydrogels (up to 99% by volume) makes them attractive materials in a biocompatibility sense, but the consequent small polymer content (1%) renders their physical properties less robust than many structural materials. Single network hydrogels are frequently compliant and brittle—failing catastrophically at small strains, \( \epsilon_f \approx 0.3 \), and strengths, \( \sigma_f \approx 20 \) kPa [8] without appreciable plastic deformation—and their constitutive mechanical responses can be complicated by substantial time-dependent deformation that is viscoelastic, poroelastic, or poroviscoelastic [3]. Further, although a fracture resistance or toughness is commonly calculated and reported for hydrogel systems [4, 6, 8–11], hydrogel behavior may not always be consistent with a simple fracture mechanics framework or interpretation: Frequently, events occurring at a crack tip associated with the creation of new surfaces consume far less energy during loading of a hydrogel to failure than the irreversible deformation distributed throughout a test specimen. This limitation is particularly true in testing of advanced, ‘tough,’ hydrogels for
which the extensions to failure are extremely large, i.e. at large stretch ratios $\lambda$, where for simple tension $\lambda = 1 + \varepsilon$, of approximately 10–20 or greater [4, 8–11].

Here, the failure and fracture behavior of hydrogel systems is considered from a perspective based on solid mechanics, with a view to optimizing mechanical properties of hydrogels for biomedical applications such as cartilage replacement. First the basic framework of fracture mechanics is examined for hydrogel material systems, making clear that terms such as toughness and fracture resistance are well defined, but may not always be useful in comparing hydrogel failure properties. Next, experimental measurements of the failure and fracture behavior of a range of hydrogels are considered in terms of an objective measure of failure resistance, the work of failure. This measure is combined with the failure strains of hydrogels to generate a map that provides a pictorial overview of hydrogel fracture and failure. The map allows hydrogel microstructural development schemes to be compared and hydrogel suitability as a replacement material for natural cartilage in tissue engineering applications to be assessed.

2. Fracture mechanics and hydrogels

A clear distinction has not always been made between failure measurements and fracture testing in hydrogel-based materials. A failure measurement is unambiguous and model-independent, as failure is defined by the inability of a specimen to sustain further applied force or deformation. The force and deformation observed during failure of a hydrogel, or any material, are often simply converted to nominal or engineering stress and strain for between-materials comparisons. A fracture test, on the other hand, connotes a very specific crack extension experiment that is analyzed within a specified fracture framework. The measured forces, deformations, and crack lengths are combined so as to generate fracture parameters for between materials comparisons.

The simplest material fracture property is the surface energy density, $2\gamma$, the energy per area required to fracture the material; the factor of 2 arises as two surfaces are created by crack propagation, such that $\gamma$ is the excess energy of a surface over that of the bound bulk state (usually the word ‘density’ is implied and omitted). Based on scaling arguments from bond energy and the bulk density of bonds, a calculated surface energy for single network hydrogels is approximately $2\gamma = 10 \text{ J m}^{-2}$ [12]. This value is consistent with observed experimental values for alginate [8, 13]. For comparison, diamond is approximately $2\gamma = 18 \text{ J m}^{-2}$ [14], Si approximately $2\gamma = 5 \text{ J m}^{-2}$ [14], and SiO$_2$ glass approximately $2\gamma = 8 \text{ J m}^{-2}$ [15].

If a cracked material in a fracture system is linear elastic, fracture equilibrium is given by $G = 2\gamma$, where $G$ is the mechanical energy release rate, the derivative of the mechanical energy in the system with respect to crack area. Mechanical energy is the work performed by the applied loading less elastic energy stored in the specimen [15]. If the material is linear elastic, the stress field adjacent to the crack is well known and, in particular, the amplitude of the dominant divergent component of the field is described by a stress-intensity factor $K = (GE)^{1/2}$, where $E$ is the elastic Young's modulus of the material. By analogy, the toughness of the material, $T$, is defined by $T = (2\gamma E)^{1/2}$, such that fracture equilibrium can be written in stress terms as $K = T$. Single network hydrogels are well described by this framework and are thus regarded as brittle. However, if the deformation behavior of a cracked material in a fracture system is non-linear, the fracture equilibrium expression is generalized to $J = R$, where $J$ is a generalized non-linear mechanical energy release rate, $R$ is the fracture resistance, and detailed knowledge of the crack related stress field is usually not known. Both terms are recognized as dependent on crack length, specimen geometry, and system history [16]. In particular, $R$ may reflect time-dependent energy dissipation distributed throughout a specimen. If the energy dissipation is restricted to a zone of plastic deformation or environmental reaction that is invariant and localized to the crack tip, a hybrid 'small scale' equilibrium statement is $G = R$, with the recognition that linear elasticity only pertains exterior to the zone. For both distributed and small-scale dissipation, the baseline value of $R$ is $2\gamma$.

The implicit goal in development of many advanced hydrogels is thus to significantly increase $R$, by adding additional networks or reinforcement phases, to create multi-network hydrogels or hydrogel composites, respectively [8–13, 17]. At a minimum, these additions could lead to greater bond density and fracture resistance increases in a simple brittle fracture manner by multiplicative increase of $2\gamma$. Additionally, in the case of composites, crack opening displacements at fracture could lead to energy dissipation as a reinforcing phase is dragged with friction from the gel matrix and lead to increases of fracture resistance in a quasi-brittle manner by additive increase to $2\gamma$ [15]. In both cases, however, these crack localized effects are likely not dominant as a substantial synergistic effect occurs and the additional network or phase leads to greater deformation throughout a specimen. The distributed deformation leads to greater distributed fluid flow and network re-arrangement with associated poroelastic and viscoelastic losses [18] and very large potential increases in fracture resistance. The mechanics is similar to the distributed dissipation in ductile metal specimens [16] although the dissipation mechanism is completely different.
Figure 1. Experimental configurations for measurement of hydrogel strength and fracture resistance: (a) tension, (b) disc compression, (c) sphere compression, (d) tensile fracture (e) trouser tearing, (f) fracture in constrained tension.

3. Failure of hydrogels

Quantifying the fracture properties of hydrogels in the above framework is difficult as measurement of a well-defined crack in a specimen is required. Most hydrogels are tested as uniform bars in simple tension, figure 1(a). Alternatives include prismatic discs in simple compression, figure 1(b), in order to avoid gripping issues and spheres in diametral compression, figure 1(c), in order to avoid challenges in fabrication of parallel faces on water-swollen gels. The applied force and displacement in such tests are easily measured, but the deformation of the gel is distributed and no crack is measured. Tensile fracture tests, figure 1(d) are possible for hydrogels, but introducing and measuring a crack (as opposed to a notch) is difficult. Shear fracture tests, figure 1(e), and constrained tension tests, figure 1(f) are possible, but although introducing a long crack is somewhat easier than the short cracks required for simple tension, specimen gripping issues are still a problem.

As a consequence, the majority of hydrogel mechanical measurements have been performed in simple tension from which little can be concluded about material fracture properties as there is no crack. Nominal or engineering stress and strain are commonly calculated from initial specimen dimensions and elastic modulus often determined \[3\]. The limiting properties of materials, such as engineering strength and engineering failure strain are also thus easily determined, as is the work performed by the testing machine through the integrated force-displacement response. The work of failure density, \(w^{*}\), expressed in terms of the initial specimen volume, is also easily obtained from the integrated nominal stress–strain response \[4\]. Without a deformation localizing crack, however, it is clear that such strength, strain to failure, and work of failure (again, the term ‘density’ is usually omitted) are usually not good indicators of material fracture properties but indicators of distributed deformation and damage response \[19\]. From the above, it is also clear that \(w^{*}\) is not the toughness \((T\) or \(R))\) of a hydrogel or even necessarily a good indicator of ‘toughening’ (increases in \(T\) or \(R\)) in a hydrogel, despite some common usage \[6, 8–13, 17\]. This point is particularly important in notched uniaxial tests, a variant of figure 1(d), in which there is significant inelastic deformation and energy dissipation throughout the specimen.

The failure properties of hydrogels can be characterized by the work of failure, although referencing \(w^{*}\) to the initial dimensions can greatly underestimate the active (Cauchy) stress and overestimate the strain in a specimen, thereby leading to potential misinterpretation of data in terms of deformation mechanisms.
Figure 2. Work of failure $w^*$ (MJ m$^{-3}$) versus engineering failure strain $\varepsilon_f$ for hydrogels. Small open symbols represent data from the compilation and review of [4]. Small filled symbols represent data from [10, 20–24]. Shaded areas represent the ranges of behavior of single network hydrogels (left) and advanced hydrogels (right). Hatched band represents the physiological strain range of normal cartilage [25–27]. Connected large open circles represent data characterizing fibrous hydrogel composites from [8, 28].

Nevertheless, $w^*$ can be used to interpret material failure behavior and, as an approximation here,

$$w^* \approx \frac{\sigma f \varepsilon_f}{2}$$

which permits simple comparison of published measurements. The definition of $w^*$ makes no assumptions about the mechanical work utilization within the sample: elastic versus permanent deformation, viscous drag, or creation of new surfaces via fracture of the specimen. The quantity $w^*$ represents a reasonable first-order estimate for an engineering stress–strain curve of any shape in the absence of detailed information about that shape. Unlike ductile metals, the stress–strain responses of advanced hydrogels are often strain-stiffening with no discernible yield stress, and failure is often catastrophic at the peak force without any distinction between maximum stress and final stress at failure [4]. The stress–strain responses of simple, single network, hydrogels are similar to those of brittle ceramics and glasses, in which cases equation (1) is exact.

Figure 2 is a logarithmic plot of $w^*$ vs $\varepsilon_f$ generated by digitizing published data [4, 10, 20–24], including those from a 2011 review of hydrogel properties (open symbols) [4] and more recent works (solid symbols) [10, 20–24]. The data fall into two groups on the plot based on the calculated values. For small failure strains and small works of failure, the data are associated with ‘conventional’ hydrogels, and for large works of failure and also large failure strains, the data are associated with ‘advanced’ hydrogels such as double network gels and nanoclay composite gels. These separate groupings are indicated by shaded background regions. The failure strains are large compared with those of glasses and ceramics (about $10^{-3}$) and metals (about 1), especially for the advanced hydrogels. The works of failure are also large. For a typical tensile specimen, 5 cm long $\times$ 2 cm tall $\times$ 1 cm thick, of single network hydrogel with $2\gamma = 10$ J m$^{-2}$, the contribution to the work of failure of a plane of broken bonds in the specimen is $2 \times 10^{-4}$ MJ m$^{-3}$. The works of failure of some single network hydrogels are of this magnitude at the lower edge of the left shaded region, suggesting that these materials are indeed classically brittle and the work of failure is dominated by simple bond scission.
Clearly, however, the works of failure of most hydrogels are much greater than these values, especially in the right shaded region, suggesting that significant energy dissipation mechanisms are active in the failure of these gels beyond bond rupture, and that these dissipation mechanisms are associated with large strains. What cannot be inferred from figure 2 is that the advanced hydrogels are tougher or more fracture resistant than simple hydrogels as the behavior of specimens containing cracks is not known. Nor can it be known if, or to what extent, the dissipation mechanisms implied in figure 2 for some gels would be active in cracked gels, even in highly deformed crack tip regions.

4. Development of hydrogels for cartilage replacement

It is quite common when considering advanced hydrogels such as those in figure 2 to invoke the potential application of scaffolds for cartilage tissue engineering [7, 11, 17]. Natural articular cartilage is approximately 70% to 85% water, with the balance proteins (fibrillar type II collagen) and a hydrogel-like protein-sugar matrix in roughly a 2:1 protein to sugar ratio [29]. Natural joint cartilage has a fracture resistance of around 1 kJ m$^{-2}$ [29], a compressive elastic modulus of 0.3–1.0 MPa [30], a tensile elastic modulus of <1 MPa to 30 MPa [29], and a tensile failure strain $\varepsilon_f$ ranging from 0.3 to 2.5, but mostly <1.5 in the surface (articulating) layers of the tissue [31]. In physiological service, the daily compressive strain levels in knee joint cartilage are the order of 0.05–0.1 due to body weight in gravity and walking, and varying between 0.15 and 0.6 during vigorous activities such as jogging, and running [25–27]. This strain range is indicated in figure 2 as the vertical cross-hatched bar. Tensile forces arise in the cartilage due to the complex three-dimensional geometries and corresponding stress states associated with exercise and other daily activities, but the cartilage is constrained by collagen fibers anchored firmly into the underlying subchondral bone with an elastic modulus of 20–30 GPa. Hence, an upper bound to the natural tensile strains is indicated by the vertical bar.

The shaded regions indicating the ranges of behavior of the single (left) and advanced (right) hydrogels relative to the hatched bar in figure 2 highlight two different approaches to hydrogel materials development for cartilage applications. In both cases, a design criterion is optimizing the mechanical behavior while retaining the fluid transport properties. Advanced hydrogels seek to maximize the work of failure through microstructural manipulation and the development efforts have been very successful in that $w^*$ has been increased to many orders of magnitude greater than that of the simple hydrogels and required by bond breaking. The difficulty with implementing such materials and the microstructures they embody as cartilage replacements is that the very large works of failure are achieved at very large strains, much, much greater than those imposed physiologically. Conversely, it is clear that the simple, single network hydrogels usually exhibit failure strains comparable to those imposed physiologically, greatly increasing the risk of failure in use, although some simple hydrogels do exhibit considerable works of failure at small strains.

Hence, figure 2 makes clear that (a) increasing energy dissipation in conventional hydrogels at small strains or (b) decreasing the strains required for large energy dissipation in advanced hydrogels are both viable materials development paths. An example of the former path is shown by the connected open symbols in figure 2, illustrating the potential of hydrogel fibrous composites [8, 28]. The lowest point from that work in terms of $w^*$ is consistent with the other values for conventional single network gels, here for plain alginate gels, which were used as the matrix in the fibrous composites. The larger work density values from that research are, in increasing order, for random fiber composites, laminated composites with alternating layers of unidirectional fibers, and for unidirectional fibers aligned in the loading direction [8, 28]. Similar microstructural tailoring exhibits great promise for fabricating materials with greater work densities at smaller failure strains.

A method of visualizing the comparative behavior of materials from the two regions in figure 2 is shown in figure 3, illustrating two hypothetical elastic stress–strain curves. Both curves represent works of failure of $w^* = 1.5$ MJ m$^{-3}$, the center of the range in figure 2, but they arrive at this value in different ways. The left response represents a material with an elastic modulus $E = 3$ MPa and a failure strain $\varepsilon_f = 1$, similar to values reported for a hydrogel composite [8]. The right response represents a material with an elastic modulus $E = 30$ kPa and a failure strain $\varepsilon_f = 10$, consistent with values reported for some double network hydrogels [10]. Both hypothetical materials have the same work of failure. Overall, cartilage more closely resembles the stiffer material with the smaller failure strain in figure 3 than the more compliant and extensible material. There is no physiological mechanism associated with a potential tensile strain of one or greater under physiological loading conditions, and thus no obvious role for a ‘tough’ but highly extensible hydrogel or hydrogel composite in cartilage tissue engineering. This raises a fundamental question about the potential engineering applications for which a highly extensible material could be utilized, not to mention one in which one component of a multi-network system is considered ‘sacrificial’ and thus irreversible and inelastic. In addition, it is not clear that the greatly constrained cartilage in normal joint function could ever experience the very large strains required to implement the failure resistance of compliant advanced hydrogels.
Figure 3. Hypothetical stress–strain responses for a fiber-reinforced cartilage-like hydrogel with $E = 3.0$ MPa and $\sigma_f = 1$ and for a double network hydrogel with $E = 30$ kPa and $\sigma_f = 10$. The work of failure $w^*$ is $1.5 \text{ MJ m}^{-3}$ for both.

5. Outlook

The limitations in failure strength and fracture resistance in single-component hydrogels has led to a boom in research aimed at generating increasingly complex multi-network hydrogels and hydrogel composites. The research has developed a wide array of chemical synthesis routes and thus an equally wide range of bonding types within the hydrogel network, ionic, covalent, and physical entanglement [4, 8–12], and demonstrated impressive increases of many orders of magnitude in failure strain and work of failure (figure 2). No doubt, innovative chemical syntheses will lead to further increases in these metrics characterizing hydrogel failure, and this is an area of opportunity for further research.

However, a greater opportunity for hydrogel mechanical properties development probably lies in overcoming a limitation highlighted above: A clear distinction between failure and fracture is often not made in testing, reporting, or interpreting the behavior of candidate microstructures for advanced hydrogels. The failure characteristics of deformation specimens such as figure 1(a) are often reported as material toughness properties although the specimens do not contain a well-defined (or, often, any) crack as in the fracture specimen of figure 1(b). An opportunity thus exists to develop a fracture test for hydrogels that explicitly measures crack propagation. Such a test would enable quantitative comparison of fracture resistance of different gels and provide clear guidance on the microstructural features and mechanisms that enhance fracture properties. In developing such a test, the obvious experimental challenges of specimen gripping and crack initiation and measurement are clear. In addition, analytical challenges associated with energy partition during testing between surface fracture and bulk deformation also exist, as well as relating the energy changes to existing fracture mechanics. The large deformations exhibited by hydrogels on swelling and in tension and compression probably require numerical finite element analyses to determine the active stress and strain and internal fluid flow behavior during fracture and deformation tests and in hydrogel component operation [32–35]. Developing microstructures that dissipate energy at large imposed strains, through irreversible deformation or viscous or poroelastic flow, is clearly the primary method of increasing the work of failure. It is not clear that such microstructures would generate increased fracture resistance.
Finally, there is also a fundamental question that emerges about the potential applications of advanced hydrogels in service; if advanced hydrogels only exhibit large works of failure at extremely large strains or stretches, their use may be limited in applications in which the service strains are small, such as in tissue engineering of cartilage. A recent work [36] considering the mechanical behavior of hydrogels for cartilage replacement emphasizes this fact, as the imposed maximum strains for a comprehensive set of experiments were approximately 0.2, consistent with physiological activities. Strategies for improving hydrogel systems for such applications requires consideration of increasing the strength and work to failure without dramatically increasing the failure strain.

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