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Glass transition in metallic glasses: A microscopic model of topological fluctuations in the bonding network

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Understanding of the structure and dynamics of liquids and glasses at an atomistic level lags well behind that of crystalline materials, even though they are important in many fields. Metallic liquids and glasses provide an opportunity to make significant advances because of its relative simplicity. We propose a microscopic model based on the concept of topological fluctuations in the bonding network. The predicted glass transition temperature, $T_g$, shows excellent agreement with experimental observations in metallic glasses. To our knowledge this is the first model to predict the glass transition temperature quantitatively from measurable macroscopic variables.

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I. INTRODUCTION

The atomic structure and dynamics of liquids and glasses are much less understood today compared to those of crystalline solids, for which the first-principle theories can answer many questions, and even make some predictions. In particular the nature of the glass and glass transition is considered to be one of the most challenging problems.1,2 In dealing with liquids and glasses we do not enjoy the benefit of the translational symmetry and resultant Bloch theorem, and have to face directly the complex many-body interactions. This frustrates most of conventional approaches which rely on the Bloch theorem. For this reason much of the recent progresses have been made in the theories of mesoscopic scale, in particular the mode-coupling theory (MCT).3,4 In the MCT the dynamics of liquid is described by nonlinear hydrodynamic equations with variables chosen to fit certain experimental results. While the physics of nonlinear coupling and feedback effect is very well described by the MCT, the atomistic underpinning of such interactions is usually disregarded as less important details. This may well be unavoidable for very complex fluids such as macromolecular systems, but for simpler liquids it should be possible to improve and deepen our understanding of the atomic level mechanisms. For this purpose we propose a unique approach based on the description of topological fluctuations in the atomic bonding network to describe the thermal evolution of the structure of metallic liquids and glasses. We show that the theory leads to expressions of the glass transition temperature that agree well with experimental observations in metallic glasses. A key idea to this success is to focus not on the dynamics of each atom but on the collective dynamics of the nearest neighbor shells, and take into account the dependence on Poisson’s ratio that characterizes the interaction between local density and shear fluctuations.

In order to describe and understand the atomic structure of liquids and glasses the choice of the key concepts or parameters that connect the bare atomic coordinates with the properties is critical. The most widely used one is the atomic pair-distribution function (PDF), which can be directly determined through diffraction experiments.5 However, the physical properties usually depend on more collective environment of atoms. For this purpose the Voronoi polyhedral analysis is frequently used.6,7 The underlying assumption of this approach is that the relative positions of the nearest neighbor atoms are most important in determining the behavior of an atom, and thus we should focus on the topology of atomic connectivity. This approach is natural for covalent glasses, but it is also applicable to metallic and ionic liquids and glasses as well, since the first nearest neighbors are fairly well defined by the first peak in the pair-density function.8 However, the local properties depends not only on the local topology but also on the geometrical distortion of the local polyhedra of nearest neighbor atoms. In order to include this effect we introduced the method of describing the local atomic packing in terms of atomic-level stresses.9

In the liquid state the topology of atomic connectivity is not static, but is fluctuating. Earlier we introduced basic concepts of describing thermal effects by using the fluctuations in atomic-level stresses and strains as local variables.10,11 In this way it became possible to connect the local topology to the local energy landscape. In the present work we make connections between the local fluctuations and the criterion of the local topological instability12,13 to discuss the glass transition. The results are compared to the experimental data, demonstrating that the theory is capable of quantitatively describing the glass transition temperature from measurable quantities (atomic volume and elastic moduli) with minimum assumptions.

II. LOCAL TOPOLOGY OF ATOMIC LIQUIDS AND THEIR FLUCTUATIONS

An obvious choice of a parameter to characterize the local topology of atomic connectivity is the local coordination number, $N_C$, or the number of first neighbors. In a glass structure represented, for instance, by the dense random packed (DRP) structure $N_C$ varies locally. We note that the
local variation in $N_C$ has a very direct physical meaning: Consider a void surrounded by a certain number of atoms, $N_C$. It is obvious that when the void is large $N_C$ is also large. One would simply argue that $N_C$ should be proportional to the square of the radius of the void, $r_v$, since near neighbors would fill the inner surface of the void with a certain constant packing fraction. Indeed for a system with a short-range pairwise potential it was possible to derive a quantitative expression for the ideal coordination number of an impurity $A$ atom with the radius $r_A$ embedded in the metallic glass of $B$ atoms with the radius $r_B$ (Ref. 15)

$$N_C(x) = 4\pi \left(1 - \frac{\sqrt{3}}{2}\right)(1 + x)[1 + x + \sqrt{x(x + 2)}],$$

where $x=r_A/r_B$. This equation was derived by a heuristic argument and is not meant to be a rigorous mathematic statement. However, its approximate accuracy, as proven by a computer simulation, is sufficient for the purpose of the argument developed here.

The inverse of Eq. (1), $x(n)=N_C^{-1}(n)$, specifies the ideal size of an atom, $x(n)r_B$, to fit an atomic site, or a void, with $N_C$. If the radius of an atom is larger than $x(n)r_B$, the atom will be under compression, while if it is too small it will be under dilatational (negative) pressure. Thus the misfit between the local topology and the atomic size can be translated into a local stress, and then to a local elastic energy. For instance if an icosahedron cluster is formed with 13 atoms around a central atom, and the peripheral atoms will be under shear stresses, as expected from the fact that in the icosahedron the distance from the center to the apex is shorter than the distance between apexes. Indeed Eq. (1) gives $N_C^{-1}(12)=0.958$. The fact $N_C(1)>12$ reflects the frustration of the icosahedral environment as discussed, for instance, by Nelson. Now the converse of this situation is even more interesting. Note that $N_C(x)$ defined by Eq. (1) is a continuous function of $x$, while the actual local coordination has to be integral. This means only at special values of $x = x(n)$ at which $N_C(x(n))=n$, an integer, the Eq. (1) can be satisfied, while in-between it can be achieved only in the average. For instance if $N_C=n+m/k$ ($m<k$), it is possible to achieve the ideal coordination in the average, by forming a crystal with $k$-m number of atomic sites with $N_C=n$ and $m$ sites with $N_C=n +1$. But every site is under some pressure, negative for $N_C=n$ and positive for $N_C=n+1$, costing elastic energies associated with local distortion. For a monoatomic system ($x = 1$), $N_C(1)=4\pi=12.56\ldots$, an irrational number. Thus this condition can be satisfied only with a crystal with an infinitely large unit cell, possibly with a quasicrystal, and the DRP structure becomes a strong competitor as an alternative.

Thus in the DRP structure in which Eq. (1) describes the ideal coordination number, the local energy landscape of an atom, $E(x)$, is an oscillating function of $x$, with minima at each value of $x_n$, since only then the actual coordination number, which is an integer, is ideal. Then if in the Gedanken experiment the size of an atom $A$ is inflated and the value of $x$ is continuously increased, $N_C$ will increase step-wise since $N_C$ is always an integer, and the local topology is changed every time the coordination number is increased. This point of topological change defines the critical strain for the local topological instability, which is related to the Lindemann’s criterion for melting. It is the strain that corresponds to the change in the equilibrium $N_C$ by about 0.5. This concept of topological instability was successfully applied to predict the compositional limit for glass formation in binary glasses. In order to describe the misfit between the ideal local packing and the actual local atomic structure we introduced the local atomic pressure of an $i$th atom, $p(i)$, as the local increase in the energy due to volume strain as

$$p(i) = \frac{1}{V_i} \sum_j f_{ij} \cdot r_{ij},$$

where $V_i$ is the local atomic volume of the $i$th atom, $f_{ij}$ is the two-body force, and $r_{ij}$ is the separation, between the atom $i$ and $j$. The local pressure thus defined is indeed correlated with the local coordination number, $N_C$. The local topology of the atomic bonds can be described not only by the number of bonds around an atom, $N_C$, but also by the anisotropy of the bond connectivity. For instance a hoop of atoms in the $x$-$y$ plane around a central atom may be different from that in the $x$-$z$ plane; the central atom may be bound tightly in the $x$-$y$ plane, but loosely in the $x$-$z$ plane. This gives rise to a local shear stress, $\tau$. Similarly, local elastic moduli, $B_i$ the bulk modulus and $G_i$ the shear modulus, can be defined.

It was found that in the high-temperature liquid state the fluctuations in local pressure are related to temperature in a very simple manner

$$\frac{\langle p^2 \rangle}{2B} = \frac{kT}{4},$$

where $\langle \cdot \cdot \cdot \rangle$ is a thermal and temporal or ensemble average, $V=\langle V \rangle$, $B=\langle B \rangle$ and $k$ is the Boltzmann constant. This means that the total potential energy of the system, $3MKT/2$, and $N$ is the number of atoms, is well described as the sum of the local elastic energy due to the atomic level stresses, and it is equally divided among the six stress components, pressure and five shear stresses, that represent local topological fluctuations. However, Eq. (3) extrapolates to zero at $T=0$, which means that all the atomic bond lengths have to be equal to the ideal length at $T=0$, whereas it is impossible to achieve such a state in real metallic glasses because of topological frustration. This means that the system will not be able to achieve thermal equilibrium at low temperatures and becomes nonergodic, in other words kinetically freezes into a glassy structure, below a certain temperature which defines the glass transition. Thus in our earlier paper we used Eq. (3) to define the glass transition temperature, $T_g$, by

$$\frac{kT_g}{4} = \frac{\langle p^2 \rangle_{crit}}{2B} = B V \langle \epsilon_{crit}^2 \rangle.$$
long-range stress field to contain it.\textsuperscript{10,11} This long-range stress field can be calculated in the continuum approximation using Eshelby theory.\textsuperscript{18} For the local pressure the total elastic energy is given by

\begin{equation}
E_\varepsilon = \frac{V(p_0^2 - p)}{2B} K_\alpha = \frac{BV}{2K_\alpha} \left(\varepsilon_\varepsilon^T\right)^2,
\end{equation}

where $\varepsilon_\varepsilon^T$ is the volume strain before the environment relaxes (transformation strain).\textsuperscript{16} The energy to create such local density or pressure fluctuation, $E_\varepsilon$, must be related to the glass transition temperature. By combining Eqs. (4) and (5) we may express $T_\varepsilon$ by

\begin{equation}
T_\varepsilon = \frac{3(1 - \nu)}{2(1 - 2\nu)}.
\end{equation}
On the other hand it has been suggested that the glass transition temperature is proportional to the shear modulus $G$ (Ref. 19) and the Young's modulus $E$. Below these ideas are compared to the recent experimental values of the glass transition temperature and elastic moduli, and we show that Eq. (7) best describes the glass transition temperature.

$$\frac{kT_g}{4} = E_v^{\text{crit}} \frac{BV}{2K_a} \left(\frac{E_v^{\text{crit}}}{K_a}\right)^2.$$  \hspace{1cm} (7)

FIG. 1. Glass transition temperature, $T_g$, multiplied by $k/(2BV)$, as Eq. (1), plotted as a function of the Poisson’s ratio for various metallic glasses listed in Table I. The values of $\nu$ and $B$ used here are evaluated at $T_g$, corrected by the procedure described in the Appendix. The solid line indicates $(E_v^{\text{crit}})^2/K_a$, as in Eq. (7), with $E_v^{\text{crit}}=0.095$.

On the other hand it has been suggested that the glass transition temperature is proportional to the shear modulus $G$ (Ref. 19) and the Young’s modulus $E$. Below these ideas are compared to the recent experimental values of the glass transition temperature and elastic moduli, and we show that Eq. (7) best describes the glass transition temperature.

III. COMPARISON WITH EXPERIMENTAL DATA

For a long time metallic glasses available were thin ribbons produced by rapid quenching, and thus reliable values of elastic moduli to test these theories were not attainable. But recently the development of bulk metallic glasses and the resonant ultrasound spectroscopy (RUS) technique made it possible to determine $B$ and $G$ separately with high accuracy. Using recent data on metallic glasses, supplementing them with our own unpublished data obtained with the RUS measurements as tabulated in Table I we examined the relationship between the glass transition temperature and elastic moduli, and we show that Eq. (7) best describes the glass transition temperature.

$T_g$ depends not only on $B$ but on both $B$ and $G$. The plot of $kT_g/2GV$ against $\nu$ is shown in Fig. 2. This plot shows a weaker dependence on $\nu$, but $kT_g/2GV$ is not constant. A similar plot for $kT_g/2E$ is shown in Fig. 3 (above). However, for a dimensional reason $kT_g$ should be related, if any, to $EV$ rather than to $E$. Indeed the plot of $kT_g/2EV$ in Fig. 3 (below)

FIG. 2. The same data as Fig. 1, but expressed as $kT_g/2GV$. The values of $\nu$ and $G$ used here are at $T_g$, corrected by the procedure described in the Appendix. The solid line indicates Eq. (7), with $E_v^{\text{crit}}=0.095$.

FIG. 3. The same data as Fig. 1, but expressed as $T_g/E$ (above, in the units of K/GPa) and $kT_g/2EV$ (below). The values of $\nu$ and $E$ used here are at $T_g$, corrected by the procedure described in the Appendix. The solid line in the lower figure indicates Eq. (7), with $E_v^{\text{crit}}=0.095$. 

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thatations of a continuous network. Moreover, these plots suggest
Young’s modulus describes a response of a free object to
uniaxial stress, which is not consistent with the fluctuations
of a continuous network. Therefore, these plots suggest
that $k_T$ is not a function of a single parameter, such as $BV$, $GV$, or $EV$, but depends on both $BV$ and $GV$, or equivalently
on $BV$ and $\nu$. Figs. 1–3 show also Eq. (7) as solid curves. It
is evident that Eq. (7) describes the experimental data excellently,
with the value of $\varepsilon^c_{\nu} \approx 0.095 \pm 0.004$. As far as we know this is the first time that the
transition temperature, quantitatively expressed in terms of measurable
variables, was shown to agree with a wide range of experimental
data.

IV. CRITICAL STRAIN

Equation (7) which explains the data so well has a single parameter
$\varepsilon^c_{\nu}$, in addition to the measurable parameters, $B$, $V$, and $\nu$. To explain the physical meaning of $\varepsilon^c_{\nu}$ and to
deduce its value let us go back to Eq. (1). As we discussed
earlier in DFR glasses for an atom to fit into the site with a
certain value of $N_C$ it has to be elastically strained by
$\varepsilon^c_{\nu} = 3\Delta \nu x = 3(x_{\nu} - 1)$, which is the transformation strain in the
Eshelby theory. By expanding Eq. (1) with $\Delta \nu$ around $x = 1$, it is given by

$$e^c_{\nu} = 3\Delta \nu x = 3\Delta N_C \left( \frac{\partial N_C(x)}{\partial x} \right)_{x=1} = \frac{3 \Delta N_C}{2\pi} (2\sqrt{3} - 3),$$

(8)

where $\Delta N_C = N_C(x) - N_C(1)$. The critical increment of $N_C$ for
the topological instability, $\Delta N_C = 0.5$, gives the local critical
volume strain, $\varepsilon^c_{\nu} \approx 0.11$. This means that if the local transforma-
tion volume strain is larger than 11% or smaller than
$-11\%$ the site is topologically unstable, and the local coordi-
nate number may change. For this reason these sites are
liquidlike, whereas the sites with the volume strain less than
11% in magnitude are solidlike. Since the atomic level volume strain $\varepsilon_{\nu}$ has a Gaussian distribution, the fraction of
the liquidlike atomic sites, $p(\text{liq})$, is given by the
complementary error function, $CE(y_{\nu})$, where
$y_{\nu} = e^{\nu} \sqrt{2} (\langle \varepsilon_{\nu}^c \rangle^2)^{1/2}$. For the standard deviation of the Gaussian
distribution $\langle \varepsilon_{\nu}^c \rangle^2$ equal to $e^c_{\nu} = 0.995$ as in Figs. 1–3,
$y_{\nu} = 0.825$. This means that the total fraction of the liquidlike
sites is given by $p(\text{liq}) = CE(y_{\nu}) = 0.243$. This value is in the
range of the percolation limit for the DFR structure, which is estimated
to be $p_c = 0.198$ for $N_C = 12$, and $p_c = 0.246$ for
$N_C = 8.73$. This result implies that the glass transition
occurs through the percolation transition of the liquidlike
states, as predicted by Cohen and Grest. The value of $p(\text{liq})$
is slightly higher than the estimate for $N_C = 12$, but it probably
originates from the kinetic nature of the glass transition that
it slightly depends upon the cooling rate. The exact per-
colation concentration may predict the ideal glass transition
temperature (Kauzmann temperature) rather than the real
glass transition temperature.

Based upon these results, we suggest that Eq. (7), deduced
by the topological fluctuation theory, predicts the glass transi-
tion temperature without a sample dependent adjustable parameter. Currently the most successful theory to describe
the behavior of liquids is the mode-coupling theory (e.g., Ref.
4). However, the mode-coupling theory requires input from the
experimental data for each composition. In comparison
the current theory needs only the knowledge of the elastic
moduli and atomic volume which can be determined from
the interatomic potentials or directly from the first
principles, thus bringing about the predictive capability. As
we demonstrate here this relatively simple model is capable of
quantitatively expressing the glass transition temperature
with high accuracy.

V. CONCLUSIONS

Understanding the mechanism of the glass transition at an
atomistic level is challenging because it is difficult enough to
describe the atomistic dynamics of liquid. We found earlier
that the atomic level stresses provide a good description of the
local structural fluctuations in liquids through their local
elastic energy. This description, however, fails below a
 Certain temperature because of the topological frustration of the
local structure, and we propose that this deviation from the
equilibrium defines the glass transition. Through this logic
we derived an expression of the glass transition temperature,
and found that it agrees excellently with the wide range of experimental data without an adjustable parameter. The success
reported here suggests that this approach may be extended to
more complex glasses, and could form the basis for a
long-sought general microscopic theory of liquids and
glasses.

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APPENDIX

Elastic moduli of a solid depend on temperature because of
the anharmonicity of interatomic interaction. For crystal-
line solids this dependence is well described by a phenom-
ological expression by Varshni
$$C(T) = C_0 - \frac{s}{e^{\theta_D T} - 1},$$

(A1)

where $C(T)$ is an elastic modulus at temperature $T$, $\theta_D$ is
the Debye temperature, and the value of $s$ is chosen so that
$C(T_m) = 0.55 C(0)$, with $T_m$ being the melting temperature.
We found that Eq. (A1) works for metallic glasses as well,
provided that we assume $B(T_m) = 0.78 B(0)$ for bulk modulus and $G(T_m) = 0.55 G(0)$ for shear modulus. $T_m$ was evaluated as a compositional average of the melting temperature of the components, rather than the actual melting temperature of the crystalline compound. The actual melting temperature depends on the delicate balance in the free energy between the solid and liquid, and is lowered by frustration near the eutectic composition. Elastic moduli, however, do not exhibit strong composition dependence, and follow more closely the composition dependence of the Debye temperature. Thus the compositional average of the melting temperature of the constituent elements provides a better energy scale for the elastic moduli. As shown in Table I the effect of this correction is small. If the elastic moduli measured at room temperature were in evaluating Eq. (7), we obtain $\varepsilon_{\text{Debye}} = 0.092 \pm 0.003$ rather than $0.095 \pm 0.004$, and $p(\text{liq}) = 0.227$ rather than 0.243. Thus this effect is within the margin of error. Volume expansion between room temperature and $T_g$ is even smaller, of the order of 0.1%, and was not considered in this work.

16 The ideal radius of the sphere to fit an icosahedron made of 12 spheres of radius unity is 0.91. However, since the space cannot be filled with icosahedra alone, when an icosahedron is embedded in the DRP structure it becomes distorted. Thus the ideal atomic size for $N_c = 12$ is more correctly given by Eq. (1), than the size of a sphere that fits a perfect icosahedron.
29 Z. Zhang, V. Keppens, and T. Egami (unpublished).