Materials' Physics in Extremes: Akrology

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An understanding of the behavior of materials in mechanical extremes has become a pressing need in order to exploit new environments. Any impulse consists of a cascade of deformation mechanisms starting with ultrafast and concluding with slower ones, yet these have not been suitably defined over the past years. This requirement has prompted the design of new experimental platforms and diagnostics and an increase in modern computer power. However, this effort has removed necessary focus on the operating suite of deformation mechanisms activated in loaded materials. This article reviews the material response and attempts to order physical pathways according to the length and time scales they operate within. A dimensionless constant is introduced to scale the contributions of component pathways by quantifying their completion with respect to the loading impulse applied. This concept is extended to suggest a new framework to describe the response to arbitrary insult and to show the relevance of particular techniques to component parts of the problem. The application of a step impulse *via* shock loading is shown to be the primary derivation experiment to address these needs and map components of the response.

DOI: 10.1007/s11661-011-0720-1

I. INTRODUCTION

MODERN challenges push research toward performance in novel environments that require new materials to respond to mechanical, thermal, or electromagnetic extremes.[1] The need to understand the physics of matter under these extreme conditions defines the remit of the expanding field of *akrology* within the study of materials' physics. In this article, the impulses with mechanical or thermal origin are divorced from electromagnetic interactions, which will form later considerations. In this loading environment, materials' response to transient loading is the result of mechanical thresholds and the kinetics operating to establish them.[2] Stronger materials can be developed to meet such insults and progress has been made over the past 50 years in this area.[3] However, science has not explicitly addressed the kinetic pathways by which materials arrive at their continuum state save in isolated cases where a few mechanisms were identified and understood.[4–7] On the other hand, interaction of materials and their jointing into structures has achieved some success with the advent of hydrocodes, which account for the mechanical interaction of component phases using conservation relations for mass, momentum, and energy and stress equilibration across component boundaries.[8] To do so completely would allow the design of a material or structure for theaters where the loading may be extreme but transient to inform the engineering of the future.[1]

The operating environment lies within an envelope within thermodynamic phase space that encompasses the pressure, volume, and temperature excursions experienced by a material under loading of differing intensities and application rates. The bounds of this phase space define the requirements for experimental investigation, which are driven by the need both to derive the necessary parameters for the suite of existing material models and better define operating mechanisms within the material under load. These will inform physical theoretical descriptions of deformation for future applications in design codes. Such a suite of experimental requirements for properties of inert and energetic materials has driven the need for novel platforms and diagnostics to deliver the required data for the modeling sets. This capability goes in tandem with the design of further experiments to test how the derived description performs, which itself constitutes a further closely aligned requirement to validate the model for use.

The nature of real, condensed-phase materials is that they are inhomogeneous at the smallest length scales, so that study at atomic length scales can never capture more than a small set of properties that are mirrored in those at the continuum.[9] At extremes of temperature, vaporization to the plasma state allows for representative thermodynamic description to apply across the population of material states sampled. The two are joined through an inhomogeneous transition regime dubbed warm dense matter (WDM), where new composite states can exist (terms used include lumpy matter to emphasize this distinction), but this region exists beyond bounds of pressure and temperature considered further in this article.[10] The observations discussed here concern the loading of matter from a condensed-phase primary state with a defined initial temperature to a subsequent changed condition at some later time. The response is the result of a defined mechanical impulse of total duration determined by the application considered. When the impulse completes, material recovers with

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Manuscript submitted October 24, 2010.
Article published online June 2, 2011

further release mechanisms operating as the material returns to ambient conditions. The primary ambient state controls the initial material strength that decays over time as mechanisms with successively longer kinetics operate within the material. As time extends to its limit at infinity, response accrues the results of these, through plasticity, damage, and creep, until strength tends to zero and matter shows a purely hydrodynamic response governed only by gravitation forces and determined by its density and the external pressures it experiences. Thus, "mountains flow before the lord," as the prophet Deborah foretold in the old testament; a concept used in rheology to define a dimensionless number to characterize the observed fluidity of a material as mountain morphology develops over time.[11] Whereas this concept is a useful guide, quantitative application was due to Weertman, who applied concepts of creep to predict the shrinkage of morphology on this and other planets and to the flow of glaciers.[11,12] That the form of dislocation creep and that of some high-rate strength models is similar is no accident given the thermally activated processes involved. Thus, the properties of condensed matter evolve with their environment, and the final state is the result of the suite of mechanisms that operate throughout the processes. Material constants, which relate to these mechanisms, evolve with time too until the processes that they describe have been completed.

While time controls evolution, the presence of inhomogeneities triggers mechanisms to operate at defined length scales at which impulse and matter can interact. This balance of cause and effect is particularly evident in the case where the response is within the regime dubbed dynamic, which is defined here to be that in which the loading pulse or observational response exceeds the time required for stresses to equilibrate by waves propagating within the target to transfer momentum to the target allowing it to act in a Newtonian manner from its center of mass. The dynamic response of materials constitutes the regime of behavior where wave interaction determines response, constituting one of the most active branches of akrological research.

Understanding material response under dynamic loading concerns predicting behavior and deformation to arbitrary insult. In this arena, typical engineering insults considered are found in two groups: impact and those generated from energetic reaction (explosives). Both induce compression in a material as momentum is transferred across its bulk to apply mechanical conditions at the continuum where the target may be treated as a point mass. In the initial state, a material has a defined strength and the deformation of condensed phase materials in the first moments is always elastic. As observational times tend to infinity and length scales come to represent continuum structures, the response tends to that observed hydrodynamically; in this view, strength is a transient quantity that decreases with time.

II. MECHANISMS

Consider first the deformation of crystalline solids under mechanical loading. The first mechanisms excited are purely elastic as the first planes of atoms within a microstructure compact. If a unit cell can minimize its potential energy by rearranging atoms within it at this stage, a denser phase may quickly form within the structure. In regions where the material has no point defects, the lattice responds with its theoretical strength controlled only by that of the interatomic bonding, which is influenced by the atomic environment.[12] These processes occur regardless of loading amplitude and at all times in the loading, since some motion is imparted to the first layers of the solid at the start of the process.

At inhomogeneities in the lattice, however, local deformation can begin and nucleation of dislocations occurs at discrete sites. Slip along interatomic planes begins and plastic deformation is established around the nucleation site. Thus, at later times, defect interactions will nucleate dislocations in crystalline domains and slip can then proceed under the shear stresses present in the material. This defines the transit from the initially purely elastic state within the material to an inelastic one at lower pressure as work is done compressing the lattice. As plasticity is then defined, processes from this point onward may be described by concepts developed for the mathematical descriptions of those states defining the solid mechanics developed from the nineteenth century to today.[13]

In metals, at later times and on greater length scales, dislocation motion ceases as they become blocked by precipitates or when they tangle with one another, pinning their motion and giving rise to transient structures within the solid.[14] This has the effect of making deformation more difficult, thus increasing the strength of the material. Such structures relax when unloaded and the remaining relaxation morphologies may be observed in samples recovered from targets loaded in idealized square pulse experiments (Figure 1).[2,15] Differences are observed between classes of metal where the interatomic stacking results in differing interatomic strength and local order. The integrated forces across planes required to overcome the interatomic bonding determine the Peierls stress for slip on close-packed planes, in particular, unit cell geometries. Clearly, the atomic packing density is highest in fcc and hcp metals, although bcc has the highest number of possible slip systems. However, there are no truly close-packed planes in the bcc crystal structure and thus a slip system requires energy input to activate since the Peierls stress is higher. This results in a high density of dislocations and their rapid propagation in fcc structures, with low dislocation density and slower propagation speeds in bcc metals.

Considering loading in single-phase metals up to 10 GPa (Figure 1), different material classes exhibit characteristic behaviors, while all show pronounced shock strengthening with pressure over this range. Behind the front, pure fcc metals such as Cu and Ni harden, while pure bcc metals (Ta discussed here, but also Nb and W) soften over the microsecond taken for the microstructure to develop (stress histories to the right). Differences in the Peierls stress are reflected in contrasting substructure evolution in the metals. Materials with higher values have greater resistance to defect

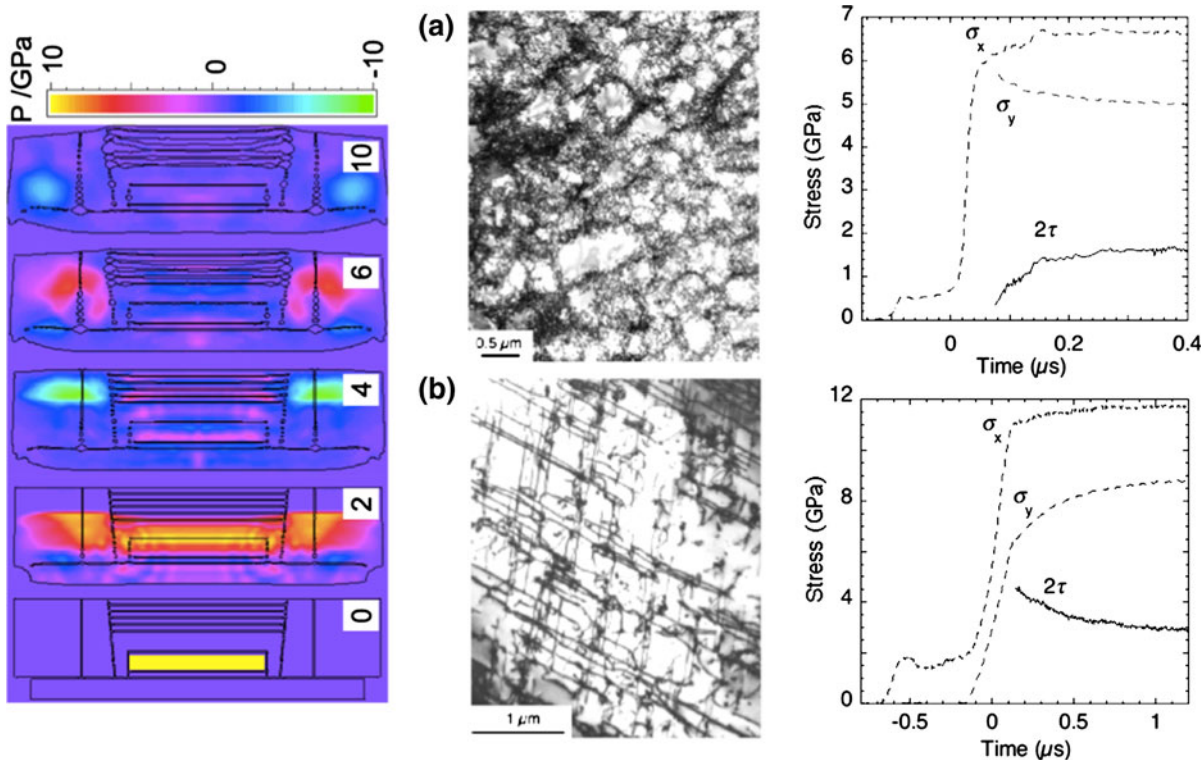

Fig. 1—Left: recovery of targets in 1-D strain; plate hits sectioned target from the rear, loading a sample (yellow rectangle first frame) recovered for metallographic analysis. Colors show pressure: yellow—high compression, and green—tensile component. Center: TEM of recovered microstructures from shock-loaded targets. Right: longitudinal and lateral (dashed) and shear stress components (solid) in (*a*) Ni and (*b*) Ta loaded to 10 GPa for 1 μs. Details in Refs. 2 and 15.

motion in the lattice and show a greatly reduced ability to cross-slip and, therefore, store additional line defects. Lower symmetry materials (such as Ta) possess higher Peierls stress, which restricts dislocation motion when there is insufficient thermal activation (such as is found with short impulses or at low temperatures).

These considerations mean that bcc metals have fewer slip systems activated on short time scales than fcc crystals, where a low critical shear stress relative to the loading direction favors slip. Further, since there are no close-packed planes in bcc crystals, a greater stress is required to initiate the process. The structure and the shock-loading direction thus determine the population of glissile dislocations and cross-slip is made more difficult, resulting in a more planar result. Finally, in the case of fcc metals, more homogeneous nucleation or widespread multiplication from existing sources will lead to reduced travel distances prior to tangling with other defects. Thus, in terms of dislocation density and motion, the response of pure tantalum over the first 100 ns is analogous to the stage of deformation occurring in the first nanosecond (and at lower length scales) in nickel. Similarly, the response of nickel over hundreds of nanoseconds may be viewed as the longer-term effect (at a lower length scale) of deformation that would occur in bcc metals over times tens or hundreds of times longer if the shock pulse duration could be sustained and if an amplitude were reached such that suitable dislocation density were activated (refer to the reviews in Refs. 2 and 15, and the references therein).

In microstructures where defect densities are lower, dislocations do not become close and thus do not lock to form barriers to further motion. Crystal plasticity is composed of group atomic interactions that are best treated theoretically with molecular dynamics (MD), until effects from second phases (which are on a larger dimension than can interact within the shock front) become important. At this time, the stress states become ill-defined within each phase until equilibration by wave propagation has occurred within a representative microstructural volume element. One may consider the time at which information can travel from areas of differing impedance to interact with processes within the crystal to define a temporal and a physical boundary between scales within the material.

Having discussed a suite of mechanisms, principally crystal deformation, it is worth adding times and corresponding length scales that one must resolve to observe these processes in operation in representative single-phase metals. Table I shows a suite of typical mechanisms with corresponding order of magnitude estimates of time scales for such processes and length scales over which they operate. The times themselves clearly vary markedly between metals of differing structure and density and with the applied compression in individual impulses, so that these values are for ranking purposes only. This is caused since the wave speeds that drive the faster mechanisms are defined by the atomic mass and the interatomic bonding of the atoms that are displaced. A second group relies on the

Table I. Characteristic Length and Time Scales for Observing Some Typical Deformation Mechanisms in a Single-Phase Metal After Applying Step Compression Loading at $t = 0$

Mechanism	Representative Time Scale	Representative Length Scale
Impact; material to elastic state	0	0
Phase transformation	10 ps	10 nm
Dislocation nucleation	50 ps	50 nm
Twin formation	1 ns	1 nm
Interaction of dislocations	100 ns	100 nm
Spallation; tensile damage	100 ns	1 μm
Adiabatic shear	10 μs	100 μm

Table II. Representative Length and Time Scales for Observing Some Typical Deformation Mechanisms Under Compression in an Amorphous Brittle Material After Applying Step Loading at $t = 0$

Mechanism	Representative Time Scale	Representative Length Scale
Impact; material to elastic state	0	0
Crack nucleation	10 ns	100 μm
Crack coalescence	100 ns	1 mm
Comminution	1 μs	1 mm
Fragment flow	1 μs	1 mm
Interfragment friction and rotation	10 μs	10 mm

slower operating mechanisms accompanying the process driven by surface creation or deformation. Thus, order of magnitude estimates for typical single-phase metals are indicated, reflecting the speed at which mechanisms operate, and are connected to spatial scales through wave speeds in the materials (which are of order thousands of m s^{-1} in these metals).

Thus, impulses of greater temporal or spatial length than each of these values will be required to observe the corresponding deformation in these cases, and experimental platforms used to investigate each mechanism need to have suitable capability to deliver the correct impulses (amplitude and pulse length) to observe the required response if continuum material models are to be constructed from their observations.

In the case of amorphous solids (such as glasses, for example) with no crystalline structure, the microstructure compresses to a denser state until fracture propagates from some defect capable of nucleating a crack.[16] Processes operate to fail the bulk at less than the theoretical strength, and these are determined by defects generated by inelastic processes, in this case, fracture. Table II shows a number of mechanisms that result from impulsive loading of amorphous targets showing brittle response.[17] A glass has no ability to slip so that inelasticity results from the nucleation and propagation of fracture within the bulk of the material. Cracks propagate from flaws swept by the compression front, which concentrates strain around sites that nucleate fracture under local tension within the solid.[18] Since the highest concentration of such sites is on the impact face, fracture is observed to propagate initially from these points, but these macroscopic sites take some time to concentrate tensile stress under impulsive loading since a front must sweep them first. The propagating fractures eventually coincide to fail the bulk, but it is only after fragments can move that inelastic flow is established and such mechanisms as penetration can operate.[19] The later time processes must involve the translation and rotation of failed material. These occur at speeds typical of the particle velocities behind the flow, not at the wave speeds of the materials of the fragments themselves. They are thus much slower and are dominated by dissipative mechanisms such as friction, since the grains themselves are so strong.

Of course all of the preceding has assumed a step in deformation and stress on the impact face of the material with attendant temperature increases due to the adiabatic state in, and immediately behind the shock.[20] However, this is the only impulse that will excite all available operating mechanisms and must be that of choice for investigating operating kinetics. When the rise time of the wave is longer, a series of slower mechanisms becomes important of which the primary consideration is that appreciable heat conduction away from the deformation zone becomes possible and the mechanical response approaches isothermal deformation in the loaded solid.

The preceding discussion was reserved for possible activated mechanisms when a step pulse is introduced onto the face of a target. Further, the examples given are applied only to a subset of possible responses that the material may adopt, since the preceding discusses a series of mechanisms observed in a range up to approximately ten times the yield stress with the present suite of techniques. It does not address those that may operate at higher levels and assumes that the initial material temperature is far from the melting point of the solid. However, there are further comments on the nature of the impulse that should be emphasized. First, applying the pulse more slowly will excite mechanisms triggered at different stress amplitudes and different times. Thus, the temperature state achieved will be lower than that induced by a shock and the flow will be quasi-isentropic. However, while the equation of state (EoS) may be derived in this way, details of mechanistic kinetics will be difficult to extract since the response of the mechanisms will be serial rather than parallel. Thus, the choice for investigation of deformation kinetics must be application of a shock impulse, while that for EoS studies of a final state best employs a suite of ramp compression experiments to control temperature with shock as a limit as the rise time reduces.

Thus, a criterion is required to reflect the provenance of the data collected for a loading mode that reflects the completed mechanisms operating within the material subjected to a defined impulse. The input parameters for such considerations concern the length and the amplitude of that pulse but also the time scale of the loading applied. The ranges of stress that activate each

mechanism will determine which commences and the times of the loading will determine what states have been achieved within the solid as it deforms. Whereas the Deborah number addresses observation times for geological processes and is therefore qualitative, a new criteria includes the impulse properties and the kinetics of the operating mechanisms to determine the result of the loading.[11] A nondimensional number, F, is defined to represent the extent to which a mechanism is driven to completion by a stress or temperature excursion during the time for which the impulse is active. It is defined in the following manner:

$$F = \frac{t_{\text{relax}}}{t_{\text{impulse}}} \qquad [1]$$

where t_{relax} refers to the characteristic relaxation time for the step in the rate limiting process and t_{impulse} is the length of the impulse applied to the structure. Clearly, the thresholds in stress, pressure, or temperature need to be exceeded in order for the mechanism to be activated, and this is assumed in the use of the quantity in the examples following. When the critical process (since there, generally, will be a suite of them) is complete long before the impulse releases back to an ambient state, the stable final form will be an equilibrium state of the material under loading. If this is the case, F will be small and the state will be defined. Conversely, if the impulse is short relative to the completion time of a mechanism assumed to act (or the shortest available at that stress level), then F will be large and the state observed will be transient and similar to the initial state of the material at the start of the process.

Such a criterion can be used to rank experiments to derive the quantities necessary to construct an equation of state for a material, since it measures the degree to which loading has reached an equilibrium state within it. Thus, it is a litmus test of the ability of an experiment to excite a response to investigate a process on the one hand, and provides a means of tailoring an impulse to optimize a material's properties for resisting dynamic deformation, on the other.

III. APPLICATIONS

Examples are presented here to illustrate this view of deformation applied to dynamic failure in tension and conversely that in compression, where the interplay of kinetics and a nested suite of operating mechanisms are critical to understanding response.

The first example illustrates plastic deformation on recovered targets of copper after shock loading to two stress levels (approximately 30 and 60 GPa). In this case, different platforms were compared in their response at the same stress levels with impulse durations of 1 μs for plate impact and 3 ns for laser loading. The recovery and examination of shock loaded targets is a complex matter, since both thermal excursions during the loading process and lateral wave interactions from surfaces must be controlled in order to deduce operating mechanisms in a purely one-dimensional loading phase.[21] However, in one piece of work, this has been successfully accomplished.[22] The recovered targets identified a series of features in the microstructures of loaded single-crystal copper targets and these included stacking faults, twins, recrystallized regions, and microshear bands. The results showed that while the copper showed dislocation structures and twinning within the crystals regardless of the mode of loading, the recrystallization and microshear were not present in the target subject to laser loading. In the previous discussion of plasticity in metals (Table I), representative times for dislocation nucleation and twinning were indicated, which in both the laser and plate impact case give $F < 1$. Clearly, the mechanisms were complete in both cases, as seen in the recovered microstructures. However, shear banding was not fully developed in either case with $F >> 1$ in the laser case and greater than one for the plate impact case. The time quoted in Table II represents the formation of a discrete band of tens of micron dimension in which localized deformation occurs. The value of F in the case of these experiments is around 5 for the plate impact test quoted, and this corresponds to a mechanism that is still in the process of completion. Thus, it is likely that the authors have captured an initial stage of localization, as they discuss in their article. The recrystallization regions observed in the recovered targets reflect the greater energy deposited by the plate impact impulse over that in the laser loading. The pulse length is approximately 1000 times longer in the former case, making the impulse delivered the same factor greater. The consequent extra heating results in greater temperature rise and, thus, has a much greater microstructural effect on cooling back to the ambient state.

A second example examines the onset of dynamic tensile failure in materials. This phenomenon is frequently dubbed spallation by analogy with the processes that occur when material is ejected from the rear surface of the targets as a compression pulse reflects from it during the loading sequence.[20] It is regarded as "the process of internal failure or rupture of condensed media through a mechanism of cavitation due to stresses in excess of the tensile strength of the material" (Grady) for the case where dynamic failure occurs due to transient states of tensile stress within the body brought about by the interaction of stress waves.[23,24] A wealth of literature exists on the subject, and the inception of new voids growing from defect sites is embodied in the nucleation and growth (NAG) model of Curran *et al.*[24] Such a mechanism has a series of operating timescales within the applied impulse, and as the duration of that pulse becomes small, the observed voids are smaller as the growth phase is reduced. In a series of experiments, two different pulse durations, an order of magnitude different, were introduced to cavitate SS316 at three different stress levels.[25] The results are shown in Figure 2. The recovered targets showed two differing degrees of void formation, with those from the longer pulses (upper micrograph) larger and some interconnected and those in the central section much smaller in diameter. The pullback signal recorded in the two cases was very different. In the long pulse case, it was well

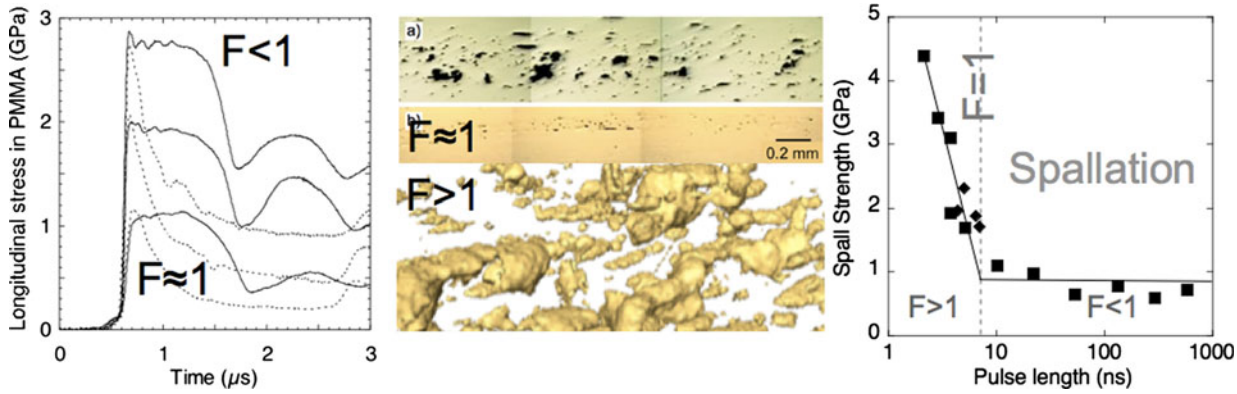

Fig. 2—Spallation as a function of pulse length in SS316 (Ref. 25). Left side shows three pulse levels; two groups of F. Center shows void populations for two pulse durations. Top and bottom: $F > 1$. Center: $F = 1$. Right shows spall strength in aluminum as a function of pulse length[27–30].

developed, but it was much smaller in the case of the shorter one.[26]

The relaxation time for the mechanism for failure is of the order of 10 ns to nucleate a new free surface in this case, which means that the state observed with the longer pulse is in a stable regime. However, although the tensile pulse is of the same magnitude for the second impulse, the void volume is very small since F is of order 1. This then represents a critical state where the void NAG is on the cusp of not reaching completion.

Other work has used a range of shock impulses to induce tensile failure into pure metals with different pulse lengths.[27–30] Results for such loading onto aluminum are also presented in Figure 2, summarizing a series of experiments conducted on a range of platforms that deliver pullback signals, which have been used to recover dynamic tensile strength. The trends show that the NAG of voids is consistent with delivering a signal, which implies an almost constant value of tensile strength. This is derived from experiments across all the different launchers used to assemble the data that, in turn, deliver differing square pulses into targets before wave interaction subsequently fails them. The pulse lengths from the laser loading, however, increase markedly over those from other platforms, and the recovered targets do not show the same voiding on the failure surfaces, although postmortem samples have seen a large temperature excursion.[31] Plotting the data in this fashion, as a function of the logarithm of the impulse, allows regions of the response to be assigned an F number for the NAG mechanisms assumed to describe the spallation process in ductile metals. The divide comes at the 10 ns pulse length discussed previously. When F becomes large, the observed spall strength rises since the mechanisms failing the metal change. This relates to the defect population from which failure ensues at the mesoscale, on the one part, but also since the trade between energy of the surface created and work done growing the void do not favor cavitation at this length scale (of order 1 μm). If the pulse is sufficiently short that it does not start nucleation at a site, or open the voids significantly, then the energy required to open a new surface increases bounded by the theoretical shear strength of the solid. This value is taken here to the shear modulus, $G/2\pi$, which, in the case of aluminum, is approximately 4.5 GPa (the asymptote to the values recorded at the smallest pulse lengths).[12] Thus, these short laser-induced impulses offer a means of accessing the limits of strength in a material in a manner not possible by other means.[30]

The precise mechanisms for failure in short pulse (including laser driven) spallation are not as well developed as those in the longer pulse (plate impact) regime, where the duration is of microsecond duration, since the former field is developing techniques to describe the processes occurring at the small length and timescales. It is clear that these impulses are accessing different mechanisms to those at the meso-scale, and further work is required to understand these processes and complete the suite of mechanisms at the smallest length scales. Response over much longer times moves from a wave to a bulk dominated process in which the surfaces of the metal become significant, with localization occurring first within the material and then in the bulk, resulting in necking of the target before subsequent ductile fracture ensues. All these processes result in reduced tensile strength in the material as time scales increase.

A third example may be found in the penetration of brittle solids by high density projectiles in ballistics. The impact onto a brittle solid instigates inelastic flow in the material just as is the case in metals. The strength of ceramic facing materials was related to ballistic performance of a laminate target.[19] Continuum measurements of strength histories near the impact face of metals and ceramics showed that the strength decays from an elastic to an inelastic state with kinetics dependent upon operating mechanisms. Shock and recovery of AD995 alumina has shown evidence of twinning in the grains above the lower elastic limit of the composite ceramic and trans- and intergranular fracture within the microstructure in this range.[17] Micromechanics controls the conditioning of the impact zone ahead of an incoming penetrator and the density of nucleation sites and nature of fracture in the projectile's path. This phase is governed by the slowest operating

mechanism, that of intergranular fracture in the ceramic that is governed by the speed of crack propagation in the material. This is bounded by the Rayleigh wave speed, close to the shear wave velocity in the material. The process fails a target in a typical plate impact experiment in and around the same time as the impulse that applies the shock. Thus, F is around 1 for this loading, and results for this experiment have frequently been shown not to show the failure phenomenon due the pulse applied being of insufficient length to fail the target. For short pulses, material is observed to show much higher strength around the theoretical strength for the material with a period of microseconds necessary before failed strength is observed.

On the other hand, rod penetration depth into the ceramic can be shown to scale with the failed strength of the materials independent of whether the targets are metals, brittle glasses, or polycrystalline ceramics.[19] The processes described previously are operating in the initial stages of the impact process and the kinetics of damage and flow are set up in these initial states, and the times taken for failure are of a different magnitude than those that occur in the steady-state penetration that occurs later. This can be observed in the flash X-ray sequence for penetration into glass presented in Figure 3.

Response over much longer times again moves from a wave to a bulk dominated process in which the surfaces of the target become significant. The sequence clearly shows that the first part of the process is dominated by the fragmentation of the brittle target by compression and release form the interfaces. Until this phase is complete, the steady penetration phase cannot begin, which occurs at the particle speed driven by the impactor. In such a process, the failure and flow of the fragments occurs in a regime where $F < 1$. Thus, the impact and penetration process is characterized by an elastic period ($F > 1$), after which the material fails by fracture, followed by a hydrodynamic phase in which $F < 1$.

To access the operating suite of mechanisms in condensed matter requires the full suite of loading platforms to understand the response of a material to a load, since the correct impulse must be applied to activate the mechanisms necessary to describe complex response.

IV. DISCUSSION

As the previous sections indicated, key aspects of any material are inhomogeneous at one length scale or another. Thus, studies at atomic dimension probing structures consisting of assemblages of a hundred or so atoms can never capture more than isolated features of the response at the continuum. Further, all classes of materials are to some extent defective when viewed through the correct magnifier (nanoscale through to bulk composites). An impulse provides a stimulus that exceeds the strength of some localities and allows inelastic states to develop. The speed of deformation in these cases is controlled by the kinetics of the available mechanisms that have been excited and transfers compression to deformation during that part of the impulse where the stresses remains unrelieved. In what follows, the discussion will assume ordered structures, but the reader can provide analogous comments for amorphous ones or materials where chemical reaction might also occur.

To discuss the provenance of defective microstructure requires a framework of terminology to describe relevant length and timescales within the material. Between the various regimes are boundaries across which behavior is described in different manners. The smallest scale considered here requires individual atoms to be resolved. Across this boundary lies atomic physics ruled by quantum mechanics. The multidimensional nature of these states collapses at this scale, since interaction from other atoms becomes important. This process is known as quantum decoherence.[32] Interactions within the

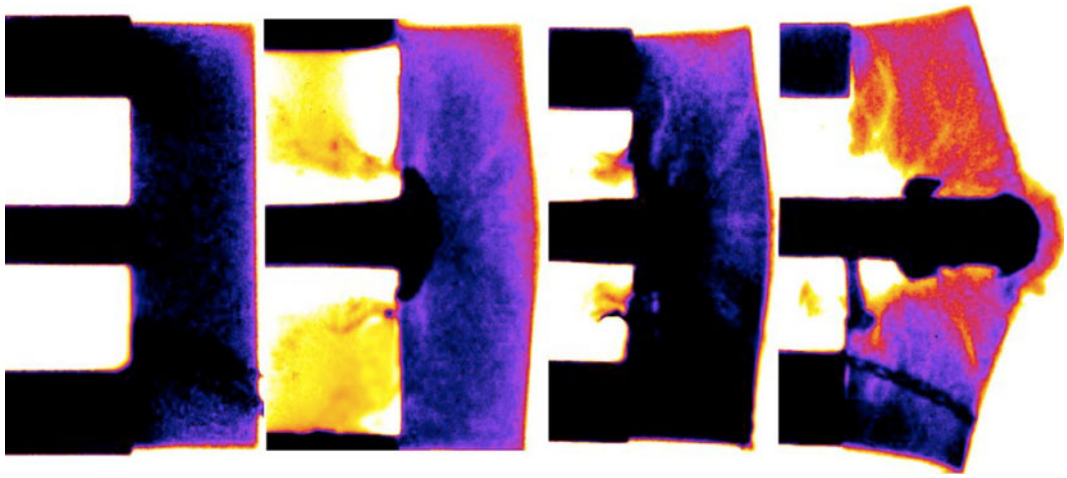

Fig. 3—X-ray image showing a steel rod 10 mm in diameter hitting a block of glass at a velocity of about 500 m s^{-1}. The first image was taken 3 μs after impact, and the second 30 μs later. By the third frame, the rod has eroded a pit in the glass and the block begins to fail. In the final frame, 80 μs after impact, a disc of metal sheared off by the glass can be seen left behind in the surface crater.

solids are modeled here, considering potentials between adjacent atoms and applying statistical mechanics to generate bulk states from molecular dynamics calculations.[33] Such simulations can give some indication of operating mechanisms despite the limitations resulting from computational errors and inaccurate potentials. The limits of the considerations of atomic representations come when other materials have an influence through their differing impedances upon the response of the material. Solid mechanics may be applied to describe these processes, and the constants therein are representations of integrated atomic behaviors at the lower scales. Here, the microstructure can be represented by wave propagation codes, which treat the response of the different phases at the scale of interest to the observer. Another collapse in dimensionality occurs as these scales are crossed. This mesoscale is defined by the length scales of the different phases viewed as a composite material, but its place is typically at the micron level in metals. At some time, the wave dominated mesoscale will give way to the hydrostatic state at the continuum where Newtonian mechanics operates. Here, wave propagation has equilibrated the momentum through the bulk so that it may be treated as a point mass to which mechanics may be applied. A further collapse in dimensionality occurs since details of the microstructure are equilibrated at this scale. In summary, the hydrostatic state at the continuum bounds the mesoscale where equilibration between interfaces dominate behavior. Below the point at which crystalline or amorphous structure intervenes, one enters the atomic regime where slip of atomic assemblages dominates. Below this lie the subatomic states and nuclear physics dominates the interactions. In the following discussion, the response of a material at each of these scales, and the component mechanisms that give integral deformation, will be explored.

Defects occupy discrete features within the microstructure (Figure 4). In crystalline solids, they constitute point defects at the atomic scale and line defects at the boundary of slip planes moving under applied shear. These serve as the lowest length scale of relevance to the suite of mechanisms that occupy response at this scale. Dislocation nucleation and slip constitute a second. At the mesoscale, volume defects become the primary consideration. These include grain boundaries between regions of different crystallographic properties and inclusions into the predominant phase in the microstructure. One special case will be voids where free surfaces exist at the micron scale and above. At the continuum, jointing is a primary defect in the response of structures and components.

The various length scales bound a series of mechanisms that operate in different regimes to define the response of materials to a step impulse. Each of these operates for a characteristic time and these are represented in Figure 5 along with the length scales pertinent to the behavior. The connection between these is the wave speed defining communication between nucleation sites or the speed of processes occurring within the material. The fastest mechanism is phase transformation, since it is dependent upon rearrangement of the unit cell, the smallest agglomeration of atoms in this picture. The length scales are thus of the order of nanometers, and the corresponding times are of picoseconds for rearrangement of this unit. Clearly, an assemblage of atoms will take a longer period, but nevertheless, the process is rapid once sufficient shear has been applied to the material. Dislocation nucleation and slip are also processes that involve atomic rearrangements and are fast considered at this scale. Once interaction of several surfaces starts to occur, mesoscale considerations become dominant. Dislocation interaction and consequent hardening in fcc metals or glide and release of shear stress in the case of bcc ones are features of this regime.

Thermal conduction at longer times becomes an issue. Then bulk heating due to plastic work and cooling by conduction balance to allow adiabatic shear bands to form within the structure down loci of principal shear. Where wave interactions cease to dominate the response, solid mechanics operates since stresses are equilibrated through the material. Between these states lie thresholds in length scale and time scale, where the consideration of properties necessarily changes. These are the quantum-atomic transition at which the subatomic considerations collapse, and the atomic-mesoscale transition at which the presence of second phases

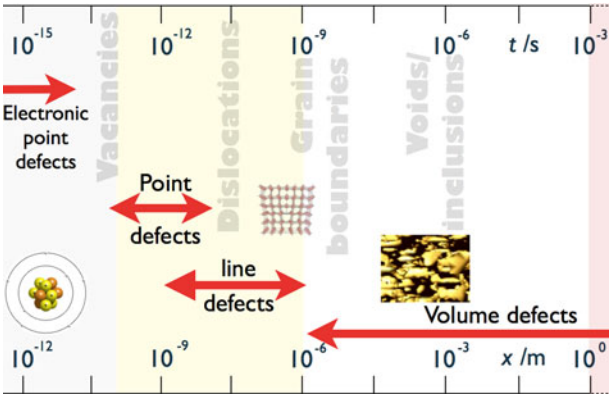

Fig. 4—Defect regimes. Distance scale x in meters and conversion to a corresponding time scale t *via* a wave speed of approximately 10^3 m s^{-1}.

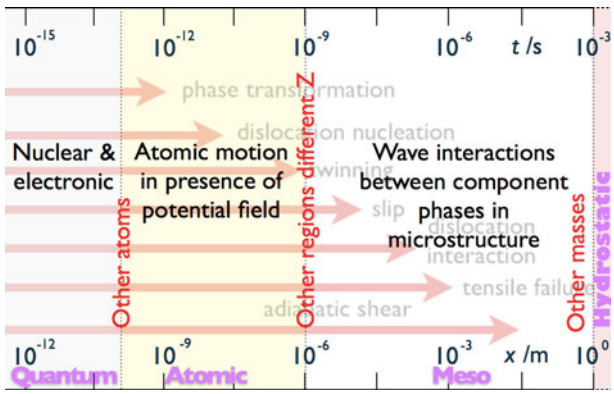

Fig. 5—Mechanisms in metal deformation operating over regions of space and time in consequence of a step loading at $t = 0$.

becomes an issue; finally, there is the mesoscale-continuum transition at which time deformation has ceased and the objects conserve momentum as point masses. Operating mechanisms sit within different regimes not crossing these boundaries, but that is merely a consequence of the integral nature of the mechanisms selected through a controlled set of experiments with defined conditions for observation.

From the continuum (engineering) perspective, the final geometry for a material, or of a structure of several materials must be predicted by understanding the constitutive behavior as a function of time that can then be applied to a representative volume element within it. With this information, a computer code can evolve the full stress state to a final form when the behavior's dimensionality collapses to classical Newtonian mechanics.

If the range of mechanisms available for deformation in that material can be assembled, then a representation of the development of the stress tensor with time can be constructed to describe the response of the material to load. Some means of switching on the mechanisms at appropriate times in the response must be incorporated in order to filter the suite operating at any instant during the deformation. In this respect, F may be used as a coefficient to activate a term describing a material mechanism within the evolving stress state. The development of the total stress field for a material, σ, can be described by a sum over the i developing stress states for each operating mechanism with their ordering controlled by F for each mechanism over all positions x_j in the following manner:

$$\sigma(P, V, T, t) = \sum_{ij} Exp[-F_i(t)]\sigma_i\left(x_j, P, V, T, t\right) \quad [2]$$

In this way, a material's response is the sum of the mechanical effects due to a temporally stacked suite of mechanisms operating within the material. The natural division of the total stress into a hydrostatic and deviatoric component,

$$\sigma(t) = P(t) + \frac{4}{3}\tau(t) \quad [3]$$

separates the hydrodynamic and strength components. Typical states are described by adding a constitutive model to an equation of state, which is a step toward the description in Eq. [2]. In the formulations used at present, the hydrostatic component does not vary with time, and since such descriptions are generally applied for times short enough that these variations are not significant, the strength term does not either. However, as time extends to longer and longer values, experience argues that the strength term fades to zero in the limit.

Of course, nature has no boundaries in time and scale; the laws concerned map the potentials at the atomic level through to behavior at the planetary scale and beyond. To some extent, the boundaries between the regimes discussed here refer only to conceptual considerations in the construct with which we describe nature, which is based in the history of science over the last 500 years. However, there are scales at which there is a rapid change in the provenance of effects from neighboring materials, and it is here that these boundaries exist.

V. CONCLUSIONS

Natural science encompasses the response of composite structures to mechanical and radiative loads. This article reviews a perception of loading that has focused on the processes inherent in material deformation in extremes, from the yield surface up to the highest stresses and temperatures where matter changes phase from condensed state to liquid or gas. All these processes have thresholds in stress and temperature, with kinetics that depend upon the magnitude of the drive they experience. If the suite of operating mechanisms is understood, it may be possible to define the mechanical response of a material in terms of the impulse applied to it. This is a task only partially addressed by materials' physics and the shock process is the only drive that allows investigation and characterization of each pathway in the detail required to describe the behavior.

To start the process of doing this, it is necessary to reorder mechanics in terms of individual processes operating with different kinetics and over a range of length scales. Such a task is formidable, and at present, only a small number of mechanisms are accurately characterized using engineering approximations that make them only valid for particular length scales within a problem. To classify such observations, a dimensionless ratio of a relaxation time for the mechanism and the length of the impulse applied to the material, F, is used to indicate whether the resulting observation of the state achieved is appropriate to the mechanism considered. If F is not less than one, then the process is incomplete and the mechanism has no part in the observed deformation. The future for physically based models requires a range of platforms providing controlled impulses to map these mechanisms across material classes in order to construct an analytic representation of behavior across the required phase space, which can be applied at the laboratory scale and extended from there. In all of these studies, the shock is key to viewing these pathways, providing the step that switches on the processes.

A key observation is that disorder is present in all natural materials. This inhomogeneity gives rise to defects at distinct scales within most materials. These defects play a pivotal role in thermomechanical loading, since they act as starting points for the onset of deformation and chemical mechanisms within the microstructure. They are typical of the classes of material scale that they inhabit; the quantum, atomic, mesoscale, and hydrostatic regimes defined in terms of the scales of the inhomogeneities that interact to drive the deformation mechanisms' operation. As the applied impulse becomes longer, the material undergoes a larger number of operating mechanisms and the response observed at the end of process is the integrated suite of these. Thus, the postmortem state results from the range of pathways excited during the impulse delivered to the

body. The duration of this pulse also filters the suite available to operate in compression, since this state is relieved as expansion reaches sites within a target and components return to the elastic regime. It is the suite of mechanisms in compression that determines the state achieved, and it is the consequent shear of the structure that molds the new transient component of that. It is release that triggers tensile failures at later times, but consequently, these must occur on greater length scales.

The application of these concepts adds a framework to an already vibrant field. With new platforms available, the possibility of the study of hitherto unrealized ideals becomes possible. One caveat how is the need to ensure that there is good communication between the sectors of the field that use differing techniques with vastly differing time scales. However, there are great strengths too. Since the laser can probe for such short times, the grail of theoretical strength may be approached in material science; the short pulse allows one to filter the response of defects within the microstructure and to study a state only dreamed of or approximated in thin whiskers of material.[31] The canvas for the study is the vast array of ordered and disordered, inert, and reactive materials in all their states and with all their textures. The palettes include the range of platforms developed across the years to deliver pressures and temperatures over pulse lengths from nanoseconds upward, taking materials from yield to melt across pressures to terapascals and temperatures of tens of thousands of Kelvin. Also, the brushes include a range of experimental techniques to build the picture of the operating mechanisms transforming the microstructure and taking it to new horizons. This whole is contained within the discipline of Akrology; the materials' physics of extremes, which unites the science of this overarching endeavor.

ACKNOWLEDGMENTS

These ideas have developed over my time spent working in, and advising on, the future directions of this field. I thank all of the colleagues and students who have contributed, particularly my groups in Cambridge, Shrivenham, and across AWE. These ideas could not have progressed without my many friends at the United States national labs and academia who have discussed these concepts with me, and I thank them for their patience. This paper was presented at the first TMS-ABM Materials Congress held in 2010. I thank the organizers for their invitation and look forward to many further meetings over future years. The excellent and helpful comments of the reviewers have added considerably to the picture presented here; I thank them for their patience in this effort.

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