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On the Abnormal Response of Ti₃SiC₂ To High Strain-Rate Loading

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9	Keywords: Dynamic response, High strain rate, MAX Phase, Ripplocations
10 11	Abstract
12	Herein, we report for the first time on the response of the MAX phase, Ti ₃ SiC ₂ , to shock wave
13	compression at strain rates above 10^4 s ⁻¹ . The shock response was determined by measuring the rear,
14	free surface, velocity of samples - subjected to impact by high-velocity projectiles launched by a gas-

gun - using interferometry. The effects of temperature and sample thickness on the dynamic yield and 15 dynamic tensile (spall) strengths were explored. The most important result of this work is the unique 16 dual-nature, at high strain rates, of the response of Ti₃SiC₂, in that it is reminiscent of both metals and 17 ceramics. For low energy impacts, the elastic response is reminiscent of ductile metals. However, for 18 high energy impacts, it performed like a hard-ceramic with quite high work hardening rates. In other 19 words, Ti₃SiC₂ behaves like nothing before it and thus must reflect its nanolayered structure. This work 20 not only provides results on the dynamic mechanical properties of Ti₃SiC₂ but is a critical first step 21 towards understanding the response of ripplocations in layered solids to high strain rates. 22

24 1. Introduction

The $M_{n+1}AX_n$ (MAX) phases are ternary, thermodynamically stable nanolayered hexagonal 25 carbides and nitrides, where M is an early transition metal, A is a group A element (mostly from groups 26 27 13 and 14), and X is C and/or N.[1] To date, there are over 155 known MAX phases, which can be further categorized by their *n* value as "211s" for M_2AX (n = 1), "312s" for M_3AX_2 (n = 2) and "413s" 28 for M_4AX_3 (n = 3) etc.[2] The MAX phases combine some of the more attractive properties of ceramics 29 and metals. They are all metallic-like conductors - in some cases, their conductivities are higher than 30 those of their pure *M* element. [3,4] They are relatively soft (Vickers hardness range from 1.4 to 8 GPa), 31 plastic at high temperatures and damage tolerant.[5,6] 32

In layered crystalline solids, such as graphite, mica, MAX phases and many others, basal 33 dislocations have long been considered to be the operational micro-mechanism for their 34 35 deformation.[7–10] However, in many other fields, where the layers can be either crystalline and/or amorphous (i.e., geology, [11,12] wood, [13] playing cards, [14] laminated composites, etc.), basal 36 dislocations (BDs) have never been invoked. Instead, it has been assumed that a buckling mechanism, 37 38 that typically results in kink band formation is operative. Moreover, the fact that many crystalline 39 layered solids fail in compression – but not tension- by forming kink bands, like those seen in other 40 fields, have all been clues that maybe BDs were not the main operational micro-mechanism in the 41 deformation of layered solids.

42 Kushima et al. [15] coined the word ripplocation (Fig. 1a), for a near surface defect in Van der 43 Waals solids. Using density functional theory, DFT, they demonstrated that ripplocation were 44 fundamentally different than dislocations by showing that two ripplocations of the same polarity 45 attracted, in contradistinction to two dislocations of the same sign that repel (Fig. 1a). Shortly 46 thereafter, we used molecular dynamics (MDs) on graphite and transmission electron microscopy 47 (TEM) observations on Ti_3SiC_2 after spherical nanoindentation,[16] to extend the idea to most layered solids and introduced the idea of a bulk ripplocation (Fig. 1b).[14,16–21] Our MD calculations showed that ripplocations, unlike dislocations, have no polarities and that they were surprisingly mobile in graphite even at 10 K. In the nanoindentation experiments, we also showed extremely high hardening rates. Along the same lines, a hallmark of deformation by ripplocations is the strong effect of confinement on yield points,[8,22–25] a feature that was not easily reconciled with the basal dislocations.[18]

Single ripplocations are probably as rare as isolated dislocations. Instead we showed that 54 lavered solids require an inhomogeneous state of stress (e.g. Fig. 1c) and that the response depends on 55 the relative size of the perturbation or applied strain field, A, relative to the thicknesses of the 56 individual sheets, τ . If $\Lambda \approx \tau$, the layers respond by nucleating evenly spaced ripplocation boundaries, 57 RBs, that propagate, wave-like, from the source of the stress as shown in Fig. 1c, for plastic cards, thin 58 59 steel sheets and graphite. The movies/simulations shown in Ref.[14], are quite instructive in that they 60 show that the process is one of confined buckling. RBs are defined as the locus of points joining the highest curvature in each layer.[19] In this case, the wavelengths of the RBs, λ_{RB} , is of the order of A 61 and the RBs.[26] However, if $\Lambda \gg \tau$, the response is quite complex wherein a large number of RBs 62 interact together in quite complex ways as shown in MD simulations that can be found in Refs.[16,17]. 63 In all cases, however, upon removal of the stress, the system reverts back to its pristine state, i.e. the 64 process is fully reversible. This reversibility, however, has a limit; once the amplitudes of the RBs are 65 high enough, they will cause quasi-brittle failure of polycrystalline solids or kink bands in single 66 67 crystals.[27-29]



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Figure 1. (a) Comparison between edge dislocations and ripplocations in a generic lattice. Both 69 deformation micromechanisms allow the relative movement of one plane of atoms over another. 70 However, in the case of dislocation, there is only in-plane strain, and climb is required to deform out of 71 plane. Additionally, two dislocations of the same sign repel, but two ripplocations of the same sign 72 attract.[15] (b) For ripplocations, c-axis strain is embedded, and can be observed in MD simulated 73 ripplocation (c) Snapshots of ripplocation boundaries formed when a cylindrical indenter is loaded 74 edge-on into, from left-to-right, plastic cards, thin steel sheets, and graphite (MD simulation). Images 75 adapted with permission from Refs. [14,15,17] 76

Somewhat surprisingly, and even though the MAX phases have attracted much attention for over two decades, little of that attention was dedicated to understanding their response to dynamic loading environments. As far as we are aware, apart from the work of Bhattacharya et al., Naik et al. and Shannahan et al. – who tested some MAX phases at maximum strain rates of $\approx 10^3$ s⁻¹, and by Jordan et al. on the Hugoniot measurements of Ti₃SiC₂ - there is little else on the subject in the literature. [30–34] The purpose of this work was to study the response of one of the most studied of the MAX phases, viz. Ti₃SiC₂, to shock experiments at strain rates as high as 10^4 s⁻¹. To that effect, its dynamic compressive and tensile (spall) strengths were tested using a Velocity Interferometer System
for Any Reflector (VISAR) [35] planar impact (shock) experiments as a function of sample thickness
and temperature, that varied between 300 to 1200 K.[35]

87 2. Materials and Methods

Polycrystalline Ti₃SiC₂ samples were prepared by ball milling a 1.95:1:1 stoichiometric mixture 88 89 of TiC (325 mesh, 99%, Alfa Aesar), Si (325 mesh, 99.5%, Alfa Aesar), and Ti (325 mesh, 99%, Alfa 90 Aesar) powders for 24 h in a polyethylene jar using zirconia milling balls. The mixed powders were 91 poured into a boron nitride-coated graphite die. The die was placed in a graphite element heated hot 92 press, HP, and pre-compacted at 5 MPa. The die was then heated at a rate of 400 °C/h, to a maximum 93 temperature of 1550 °C under a uniaxial load corresponding to a stress of 25 MPa. This temperature 94 and pressure were maintained for 4 h before cooling down at a rate of 400 °C/h. The HPed samples' 95 surfaces were ground to remove any boron nitride, graphite and/or carbide residue on the surface before further preparation for dynamic testing. 96

The bulk polycrystalline Ti_3SiC_2 sample was cut with a diamond saw to ~3 mm thick, 11×11 mm² squares. The squares were polished down to the required nominal thicknesses (0.5, 1, 2 and 3 mm) and 1 arc min parallelism. In order to enhance the reflection of the laser light, on one surface of the square samples a micron layer of gold was vacuum deposited.

For microstructure analysis the sample was mounted, ground, polished and etched for 2-3 s by a 1:1:1 part solution of hydrofluoric acid (48%, Sigma Aldrich, St. Louis, MO), nitric acid (68%, Alfa Aesar, Ward Hill, MA) and water. The microstructures were imaged in a scanning electron microscope, SEM, (XL30, FEI, Hillsboro, OR). The SEM images were processed using ImageJ software to obtain the average grain size.[36] Up to 10 images were analyzed and several hundreds of individual grains were examined. The average and standard deviations of the grain size, determined from the SEM images, was found to be $\approx 50\pm13 \mu m$ (Fig. 2a). Phase composition was measured by X-ray diffraction,

- 108 XRD, using a Rigaku SmartLab diffractometer (Tokyo, Japan) over the range of 5 to 90 20 degree,
- with a step size of 0.02 degree and a dwell time of 0.5 s per step. The only impurity was found to beTiC (3.6%).



Figure 2. (a) Typical SEM microstructure of a polished and etched Ti₃SiC₂ sample studied herein. (b)
 XRD pattern of the Ti₃SiC₂ sample.

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Prior to the impact tests, the longitudinal, $c_l = 9.22 \pm 0.01$ km/s, shear $c_s = 5.69 \pm 0.02$ km/s and bulk ($c_b = \sqrt{c_l^2 - (4/3)c_s^2} = 6.47 \pm 0.02$ km/s) speeds of sound were determined using an ultrasonic pulse-echo technique. The initial sample density, ρ_o , measured by the Archimedes method was 4.54±0.01 Mg/m³. All these values are on par with previously reported ones by Radovic et al.[37]

Since no information on the behavior of Ti_3SiC_2 under planar impact loading was available, the first set of experiments were carried out with polymethylmethacrylate (PMMA) window. Here 3 mm thick samples were backed with a 6 mm thick PMMA windows preventing premature damage of the rear (un-impacted) sample surface upon the shock wave's arrival. In this first series of experiments, the samples - initially at ambient temperatures, ~ 300 K - were loaded by 0.5 mm copper, Cu, or 1 mm tungsten, W, impactors accelerated using a 25 mm smooth bore gun to velocities that ranged from 200 to 1040 m/s. The velocity of the sample/window interface was continuously monitored by VISAR having interferometric constants of 96.4 and 439.2 m/s per fringe. One of the outcomes of this first series of experiments was understanding that until the impact velocity exceeded the 1 km/s level, they could be performed with samples whose rear surfaces were free (windowless). This essentially simplified the next series of tests. The velocity monitored by VISAR in series II to IV (see below) is thus the velocity of the free sample surface. In these tests, the VISAR interferometric constant was 96.4 m/s/fringe. In all experiments, the impact velocity (with an uncertainty of about 0.5%) was controlled by two pairs of electrical charged pins and by two in-barrel fiber optic stations.

The second series of tests was also performed with 3 mm thick samples whose temperature was varied between 300 and 1200 K. These samples underwent relatively modest loading by the 1 mm Cu impactors with a velocity of 210 ± 10 m/s. In the third and fourth series of experiments, 0.5, 1, and 2 mm thick samples were exposed to the same loading conditions in the 300 to 900 K temperature range. The details of how the samples were pre-heated with a resistive heater up to temperatures of about 1400 K with ~ 10 K uncertainty can be found elsewhere.[38] For the ambient temperature tests, the same assembly, this time without heater, thermal screen, or thermocouple was used.

141 **3. Results and Analysis**

142 **3a.** *Principal Hugoniot of Ti₃SiC*₂

Because materials have different compressibility in the elastic and inelastic (plastic) states, the 143 144 front of the rectangular stress pulse generated at the sample surface by the planar impact is unstable and splits into, (i) an elastic precursor wave with an amplitude corresponding to the material's Hugoniot 145 elastic limit, HEL, propagating through the sample with a velocity close to the longitudinal speed of 146 sound, c_l , and, (ii) a plastic wave whose propagation velocity is close to the bulk speed of sound, c_h (147 $c_b < c_l$). This two-wave structure is apparent in the velocity histories u(t), recorded at the interfaces 148 between differently loaded samples and the PMMA windows shown in Fig. 3a. Using PMMA shock 149 Hugoniot data,[39] and a shock correction for the index of refraction of the PMMA window,[39,40] the 150 stress at HEL, σ_{HEL} , can be estimated from Eq. 1:[41] 151

152
$$\sigma_{HEL} = \frac{1}{2} \left(\rho_0^{\text{Ti}_3 \text{SiC}_2} c_l^{\text{Ti}_3 \text{SiC}_2} + \rho_0^{\text{PMMA}} c_l^{\text{PMMA}} \right) u_{HEL} = 0.183 \text{ GPa}$$
(1)

This value corresponds to a shear stress at HEL, $\tau_{HEL} = (c_s^2/c_l^2)\sigma_{HEL} = 70$ MPa and to a yield strength $Y = 2\tau_{HEL} = 140$ MPa ,[42] which is 4-6 times lower than values measured in quasi-static compression tests.[43–46]



Figure 3. (a) Velocity histories of Ti₃SiC₂/PMMA interfaces recorded after impacts of different 157 strengths. Sample and impactor thicknesses and impactor velocities are listed next to the waveforms. 158 Inset shows determination of u_{HEL} used for estimating σ_{HEL} . (b) Shock velocities U_s (open squares, 159 right hand y-axis) and stress σ_H (open circles, left-hand y-axis) at Hugoniot determined from the 160 velocity histories shown in (a) as a function of particle velocity u_p . Filled squares correspond to the 161 ultrasonic speed of sound in Ti₃SiC₂. Dashed line is a linear approximation. The dotted line is the 162 Hugoniot hydrostat $p_H = \rho_0 U_s u_p$. Inset shows departure of σ_H values from hydrostat. Error bars 163 correspond to the uncertainty in determinations of U_s and u_p . 164

Taking into account the physical properties of PMMA [39,40] and applying, in sequence, the mass and momentum conservation laws to the elastic and plastic waves,[47] one obtains the shock wave propagation velocity (U_s), compressive stress (σ_H), specific volume (V_H), and compressive strain (positive) $\eta_H = 1 - V_H / V_0$ at the final, Hugoniot state. Figure 3b plots the dependence of U_s and σ_H on particle velocity, $u_p = \sqrt{\sigma_H \eta_H / \rho_0}$. As apparent from Fig. 3b, the U_s values obtained after the two high energy impact tests, together with c_b , may be approximated - with reasonable accuracy with a Pearson correlation coefficient > 0.999 - by the linear expression: $U_s = C_0 + su_p = 6.47 + 1.61u_p$, where s is the slope and C_0 is the intercept. This means that C_0 of the linear Hugoniot virtually coincides with c_b and that for $u_p > 650-700$ m/s, the sample behind the shock front is close to being in a state of pure *hydrostatic compression*. Accordingly, the expression for the Hugoniot hydrostat, is: $p_H = \rho_0 (C_0 + su_p) u_p = 29.37 u_p + 7.32 u_p^2$.

176 **3b.** Temperature dependence of dynamic yield and tensile (spall) strengths of Ti₃SiC₂

As noted above, only the first series of samples were backed with a PMMA window. The 177 velocities measured in all other series were those of free sample surfaces. Figure 4a displays the free 178 surface velocity histories for 3 mm thick samples preheated to temperatures (T_0), of 300, 600, 900 and 179 1200 K and shock-loaded with a 1 mm Cu impactor having velocities of \approx 210 m/s. As apparent from 180 Fig. 4a, varying T_0 does not change the waveforms substantially. The main changes take place during 181 the initial stages of shock compression. Inset in Fig. 4a, shows that the part of the waveforms 182 associated with the elastic-plastic transition varied systematically with T_0 . As in the case of the 183 waveforms (Fig. 4a) in the previous subsection, the elastic deformation is associated with a velocity 184 ramp is possibly caused by either some dispersion of the wave due to the presence of multiple grain 185 boundaries and/or ripplocations, or by a relatively low stresses needed to split low energy core 186 187 dislocations.[48] Note that the distinct signature of an elastic-to-plastic transition is absent in the waveforms of Fig. 4a. This results in substantial uncertainty in the determination of u_{HEL} by the method 188 shown in the inset of Fig. 4a. This uncertainly is propagated to estimates of σ_{HEL} and yield strength, Y, 189 given by: [42] 190

191
$$\sigma_{HEL} = \frac{1}{2} \rho_0 c_l u_{HEL}$$
(2a)

192
$$Y = 2\frac{c_s^2}{c_l^2}\sigma_{HEL}$$
(2b)

To make use of Eqs. 2a and 2b, c_l and c_s (or of moduli $E' = \rho_0 c_l^2$) and of the densities ρ_0 at T_0 need to be known.[37,49] It will be shown later that the two-wave, i.e. an elastic precursor wave, followed by a plastic wave, configuration obtained with the 3 mm thick samples may be considered as steady. In such a case, the simple wave approximation [50,51] may be applied to the waveforms of Fig. 4a for obtaining, based on the experimentally measured velocity histories, the compressive stress-strain function $\sigma(\varepsilon)$ of Ti₃SiC₂ and is described in parametric form in Appendix A.

At the impactor/sample interface, the planar impact generates a compressive stress pulse 200 travelling towards the free surfaces of both impactor and sample free surfaces. After reflection from the 201 surfaces, both compressive waves start to travel back towards the impactor/sample interface, now as 202 release (or rarefaction) waves. Arrival of the rarefaction wave from the impactor rear surface to the rear 203 surface of the sample starts to decrease its velocity from the maximum u_H value. The deceleration is, 204 however, limited by the material strength which is virtually proportional to the velocity pull-back Δu_{pb} 205 (Fig. 4a). The in-sample collision of two rarefaction waves results in the generation of a tensile stress 206 pulse at the site of collision. If the so-generated tensile stress exceeds the tensile strength of the sample, 207 208 dynamic tensile (spall) fracture takes place. The fracture is accompanied by the creation of two new surfaces at the collision site and by the generation of a compressive signal that changes deceleration of 209 the sample's rear surface into acceleration followed by velocity minima shown in Fig. 4a. 210

The dynamic tensile or spall strength σ_{sp} can thus be estimated assuming:[52]

212
$$\sigma_{sp} = \frac{1}{2}\rho_0 c_b \left(\Delta u_{pb} + \delta u\right), \qquad (3)$$

where $\delta u = d_{sp} (1/c_b - 1/c_l) |\dot{u}_1| \dot{u}_2 / (|\dot{u}_1| + \dot{u}_2)$ is a correction for the distortion of the waveform as a result of the different propagation velocities of the spall signal front and of the rarefaction wave ahead of it in the elastic-plastic medium,[52] and d_{sp} is the thickness of the spall plate. that is virtually equal to the impactor thickness. The mechanical properties of Ti₃SiC₂, namely σ_{HEL} , *Y*, *Y*_{0.1}, and σ_{sp} , determined based on the Eqs. (2)-(4) are plotted in Fig. 4b, together with the quasi-static (strain rate 10⁻⁴ s⁻¹) compressive strength results taken from Ref. [53] as a function of T_0 . What is quite intriguing and worth mentioning is the maximum in σ_{sp} observed at 600 K, that will be later discussed at Section 4b.

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Figure 4. (a) Free sample surface velocity histories of 3 mm thick Ti₃SiC₂ samples, preheated to different temperatures, T_o (labelled by the waveforms in inset). Inset zooms on short times. Arrows and dotted lines show the waveforms' parameters used for determining the yield and spall strengths. (b) Stress at σ_{HEL} (open circles), yield Y (filled circles), offset yield $Y_{0.1}$ (filled squares), and spall strength σ_{sp} (filled red triangles) as a function of T_o . Dashed and dotted lines are the linear fits of Y, $Y_{0.1}$ and σ_{HEL} , respectively. Solid line reproduces results of Ref. [53] obtained under compression using a strain rate of $\approx 10^{-4}$ s⁻¹. Slim error bars correspond to the typical uncertainty of the property determination.

Figure 5 displays the work hardening moduli (θ), normalized by the Young's modulus (*E*), viz. $\theta = (1/E)(\Delta Y/\Delta \gamma)$, employing two yield strength increments $Y_{0.1} - Y$ and $Y_{0.2} - Y_{0.1}$. Considering the uncertainties in the determination of the values of θ for the former and the latter intervals, these values can be considered as shown in Fig. 5, to be temperature independent and equal to 0.8 and 0.5,

233 respectively.



234

Figure 5. Normalized strain hardening moduli θ determined based on $Y_{0.1} - Y$ (filled circles) and $Y_{0.2} - Y_{0.1}$ (open circles) increments as a function of T_0 . Error bars correspond to the relatively high, \approx 10 %, uncertainty in the determination of θ .

238 **3c.** Relaxation of shear stress in shock-loaded Ti_3SiC_2

To better understand the processes governing plastic deformation we carried out two additional series of impact tests on samples of different thicknesses shock-loaded at 300 and 900 K. As noted above, the shock front in the elastic-plastic material is unstable and splits into elastic and plastic waves travelling with different velocities. Initially, at the impacted surface, the shear stresses are relatively high. In particular, in the case of the tests whose results are shown in Fig. 4a, it is about $\tau_{in} \approx (c_s^2/c_l^2)(\rho_0 c_l u_H/2)$ or, normalized by the shear modulus $G = \rho_0 c_s^2$, $\tau_{in}/G \approx u_H/2c_l \approx 6 \times 10^{-3}$. After the 3 mm traverse (for, e.g., 300 K test, Fig. 4a), $\tau_{HEL}/G \approx 9 \times 10^{-4}$, is significantly reduced, i.e. the elastic precursor wave decays with propagation distance and the shear stress behind the precursorfront relaxes. Evidently, the relaxation occurs at the expense of plastic deformation.

Duvall et al.[54] showed that in an acoustic approximation, the variation of σ_{HEL} (and τ_{HEL} , as well) with propagation distance (*h*), is related to the initial plastic strain rate $\dot{\gamma}$ as:

$$\frac{d\sigma_{HEL}}{dh} = -\frac{4}{3}\frac{G\dot{\gamma}}{c_l} \tag{4}$$

The two-test series with samples of various thicknesses were performed with the goal to obtain the dependences $\sigma_{HEL}(h)$ at different T_0 . The sample free surface velocity histories recorded in these series are presented in Fig. 6a and b for T₀ of 300 and 900 K, respectively. Note that the time along the *x*-axis is normalized as $t^* = tc_l/d$ where *t* is time and *d* is sample thickness. This is done to account for the additional time that takes for the shock wave to travel in sample of different thickness.[42]



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Figure 6. Free surface velocity histories for samples of different thicknesses (indicated to right of waveforms) shock loaded at T_0 of, (a) 300 and, (b) 900 K.

In order to quantify the amplitude decay of the elastic wave with *h* - Eq. (4), we used both σ_{HEL} (as defined in Eq. a and b), and the $\sigma_{HEL0.1}$ (equal to the compressive stress corresponding to a plastic strain $\gamma = 0.001$, see Eqs. (A1) and (A5)). The latter, however, requires a reservation. Equations A - A are applicable when the flow is steady,[50] i.e. when all the waveforms shown in Figs. 6a or 6b coincide. Using Eqs. A2 and A6 in the case of a relaxing solid, is limited to very low levels of plastic strain, when the uncertainties in the stress and plastic strain are not too large. It may be some 2-3% in the case of the 2 or 3 mm samples but may approach 10% for the 0.5 mm samples.

The values of σ_{HEL} and $\sigma_{HEL0.1}$ estimated at 300 and 900 K are shown in Fig. 7, as a function of *h*, for different *d*. As apparent from Fig. 7, with reasonable accuracy (a Pearson correlation coefficient better than 0.99) all the dependencies may be fit by a power function

$$\sigma(h) = \sigma_0 \left(h/h_0 \right)^{-\alpha} \tag{5}$$

where $h_0 = 1 \text{ mm}$ and σ_0 and α are fitting parameters whose values for both σ_{HEL} and $\sigma_{HEL0.1}$ are listed in Table 1 as a function of T_0 .



- Figure 7. Log-log plot of σ_{HEL} (filled symbols) and $\sigma_{HEL0.1}$ (open symbols) as a function of propagation
- distance at 300 K and 900 K. Solid and dashed lines are the power fits $\sigma(h) = \sigma_0 (h/h_0)^{-\alpha}$. Error bars correspond to typical uncertainties in stress measurements for the 0.5 and 3 mm thick samples.

276 Combining (4) and (5) it is possible to relate σ_{HEL} with the initial plastic strain rate:

277
$$\sigma_{HEL} = \sigma_0 \left(\frac{4}{3\alpha} \frac{h_0}{c_l} \frac{G}{\sigma_0} \dot{\gamma} \right)^{\frac{\alpha}{\alpha+1}} = A \dot{\gamma}^m \tag{6}$$

where *A* is the strain rate factor with units GPa·s^{*m*}, and *m* is the strain rate exponent. The values of *m*, *A*, and, for reference, the strain rate after a 1 mm traverse, are listed, respectively, in the three rightmost columns of Table 1.

Had the deformation been dislocation-based (not in our case), then the decay parameters shown in Table 1, can be used to shed more light on the problem. As mentioned above, all determined α values are typical for the control of dislocation motion and multiplication by the phonon viscous drag. In such case, the average dislocation velocity $\overline{\nu}$ is related to the applied shear stress τ via linear expression [55]

$$\overline{v} = \frac{b}{B}\tau,$$
(7)

where *b* is the dislocation Burgers vector, and *B* is the phonon drag coefficient, responsible for the increase in both σ_{HEL} and $\sigma_{HEL0.1}$ with increasing temperatures in metals. Substituting Eq. 7 into Orowan's equation one obtains:

290
$$\dot{\gamma} = \rho_m b \overline{\nu} = \rho_m \frac{b^2}{B} \tau$$
, (8)

where ρ_m is the density of mobile dislocations. Since information on *B* in Ti₃SiC₂ is unavailable, estimating ρ_m from Eq. 8 is impossible. It is possible, however, to estimate the ratio ρ_{mf}/ρ_{mi} between the density corresponding to the very beginning of plastic deformation, ρ_{mi} , and the density ρ_{mf} , after some finite plastic deformation, e.g. $\gamma = 0.001$. Combining Eqs. (4), (5) and (8) for the same *h* one obtains:

296
$$\frac{\rho_{mf}}{\rho_{mi}} = \frac{\dot{\gamma}_f \sigma_{HELi}}{\dot{\gamma}_i \sigma_{HELf}} = \frac{\alpha_f}{\alpha_i}$$
(9)

Substituting the values of α from Table 1 into Eq. (9) leads to conclusion that increasing the strain from virtually zero to 0.001, *decreases* ρ_m by 30 % at 300 K and by 10 % and 900 K. Given the very high hardening observed, such a conclusion is physically untenable (see below).

<i>T</i> ₀ , K	Stress	σ ₀ , GPa	α	т	A (GPa·s ^m)	$\dot{\gamma}(1 \text{ mm}), \text{ s}^{-1}$
300	$\sigma_{_{H\!E\!L}}$	$0.34(1)^{a}$	0.64(2)	0.39	9.28×10^{6}	1.01×10^{4}
300	$\sigma_{_{HEL0.1}}$	0.97(5)	0.46(3)	0.32	4.01×10^{7}	2.11×10^4
900	$\sigma_{_{H\!E\!L}}$	0.67(2)	0.38(2)	0.28	4.73×10^{7}	1.29×10^{4}
900	$\sigma_{_{HEL0.1}}$	1.8(1)	0.33(3)	0.25	1.37×10 ⁸	2.98×10 ⁴

Table 1. Parameters of elastic precursor decay in Ti_3SiC_2 at different temperatures.

^{*} Values in parentheses correspond to uncertainty of the last digit of the measured property.

302 4. Discussion

The most important result of this work is the uniqueness of the response of Ti₃SiC₂ vis-à-vis 303 what is known to date about how metals and ceramics respond to high strain rates. This is best seen in 304 305 Fig. 8, where the normalized strain hardening moduli are plotted as a function of σ_{HEL} for various metals and ceramics. From this plot alone, it is clear that Ti₃SiC₂, and presumably other MAX phases, 306 is an outlier. The leitmotiv of what follows is that in some respects Ti₃SiC₂ behaves like a ductile metal 307 but, crucially, in others, it behaves more as a hard, brittle ceramic material. In other words, Ti_3SiC_2 308 behaves like nothing before it and thus most probably reflects how ripplocations respond to very high 309 strain rates. We note in passing that in the case of the carbides and borides shown in Fig. 8, the post 310 HEL behavior is brittle and immediately after HEL the material is damaged. Moreover, the post-HEL 311 ramping of the signal (for example, particle velocity) is terminated in the compressive front whose 312 velocity is essentially lower than the bulk speed of sound of pristine material. 313



314

Figure 8. Normalized strain hardening moduli as a function of HEL for various materials.[56–62]

316 4a. Principal Hugoniot of Ti₃SiC₂

The first surprise is the fact that all the waveforms of the velocity histories (Fig. 3a) are *atypical for ceramic materials and are more reminiscent of metals*. The elastic precursor wave - with a very modest amplitude ($u_{HEL} \approx 8$ m/s vs. 100's of m/s typical for ceramics) - is followed by a plastic wave terminated at the Ti₃SiC₂ Hugoniot.

From Fig. 3b it is clear that σ_H of the lower velocity tests (232 and 443 m/s) lie some 0.5 GPa 321 apart from the hydrostat. This, in turn, suggests that at the Hugoniot state, achieved by relatively low 322 energy impacts, corresponds to a shear stress of about $\tau = 3(\sigma_H - p_H)/4 \approx 380$ MPa. By comparing 323 this value, corresponding to the final stage of shock-induced plastic deformation, with that at its very 324 beginning (HEL) it is reasonable to conclude that plastic deformation of Ti₃SiC₂, under relatively low 325 impact energies, is accompanied by quite strong work hardening more typical of ceramics than 326 metals (Fig. 8). As discussed in the introduction section, we have recently shown that such high 327 hardening rates and moduli are a hallmark of deformation by ripplocations.[16] 328

329 Lastly in this section, it is important to note under more intense loading conditions, the complete comminution of the sample occurs accompanied by the vanishing of the deviatoric stresses and a loss 330 of shear strength. This statement is justified by estimating the spall strength of Ti₃SiC₂ based on the 331 velocity histories in Fig. 3a. While measuring the spall strength in experiments with low-impedance 332 windows, like herein, are accompanied by substantial uncertainties, [41] the spall strength of Ti₃SiC₂ for 333 the 232 and 443 m/s impacts is small but finite, viz. 0.29±0.05 and 0.32±0.06 GPa, respectively. The 334 spall strength determined for stronger impacts is thus definitely zero, which is *typical of ceramics* 335 shock-compressed above their HEL. 336

4b. Temperature dependence of dynamic yield and tensile (spall) strengths of Ti_3SiC_2

The temperature dependencies shown in Fig. 4b are striking in that the ambient values of Y and $Y_{0.1}$ are low and increase with increasing temperatures. Such a response is atypical for ceramics and *more typical of pure metals,* with relatively low Peierls stresses. In contradistinction, under static loading conditions (Fig. 4b) the strengths decreases with increasing temperatures, which suggests thermal activation.

To date, two mechanisms have been proposed to explain of the increases in *Y* and *Y*_{0.1} with temperature (Fig. 2b): i) an increased stacking fault energy with temperature, resulting in a decrease in dislocations mobility, and/or ii) viscous phonon drag of dislocations that increases with increasing temperatures.[55] This mechanism controls dislocation motion when the acting shear stress is higher than the Peierls stress and dislocations glide is the over-barrier type.[63] Since neither apply to Ti₃SiC₂, they cannot be invoked here.

As apparent from Fig. 4b, and although the temperature dependences of *Y* and *Y*_{0.1} are parallel, the absolute values of *Y*_{0.1} are significantly higher. It was noted in subsection 4a that the RT impact response is associated with substantial work hardening, quantified in Fig. 5. Usually the values of the strain hardening moduli in strongly work hardened FCC metals are about 10^{-3} - 10^{-2} , [61,64,65] i.e. the strain hardening moduli of Ti_3SiC_2 (0.3 to 0.6, Table 1) are extraordinarily high, even by metal standards.

The temperature dependence of the spall strength σ_{sp} shown in Fig. 4b points to one more 355 aspect of the dual nature of Ti₃SiC₂. Although the spall strength is more than modest, the values of 356 σ_{sp} over the 300 – 600 K temperature interval are twice as high as its σ_{HEL} . Such a ratio between the 357 dynamic tensile and compressive strengths is again typical of metals. Moreover, taking into account 358 that the spall process comprises nucleation of voids and their growth, followed by voids 359 coalescence,[52] and remarking that over this temperature interval (up to 600 K), σ_{sp} and Y correlate, 360 one can conclude that over this temperature interval the spall process is governed by the growth of 361 voids. However, it is not the case at temperatures higher than 600 K, where the behavior changes to 362 that more common of ceramics. In ceramics the σ_{sp} values are typically smaller than σ_{HEL} and no 363 σ_{sp} -Y correlations are observed. It follows that at higher temperatures, control of the spall process 364 seems to be passed to void or micro-cracks nucleation. It is worth mentioning that information about 365 366 temperature variation of spall strength of any material (not only MAX ceramics) is extremely scarce in the literature. 367

368 4c. Relaxation of shear stress in shock-loaded Ti_3SiC_2

From the first glance at Fig. 6 three things are apparent: i) in both cases, the amplitudes of the elastic precursor waves decrease with propagation distance, ii) the amplitudes of the elastic waves are greater at 900 K than those obtained at ambient temperatures, iii) the time intervals between the elastic and plastic waves are smaller at higher temperatures. The latter is possibly related to a faster decline with temperature of the longitudinal modulus $E' = \rho_0 c_l^2$ vs. the bulk modulus $B = \rho_0 c_b^2$ [66].

Regardless of whether σ_{HEL} or $\sigma_{HEL0.1}$ is considered, their decrease with *h* (Fig. 7), is *reminiscent* of

which is typical for over-barrier glide of dislocations and also shows some decline with temperature as occurs in pure Al [67] or Cu [68]. However, substituting these α values into Eq. (10) leads to the physically untenable result that increasing from the strain from virtually zero to 0.001, the density of mobile dislocations – if invoked – should *decrease* by some 30 % at 300 K and by 10 % and 900 K. Since such a conclusion is at odds with the large hardening rates observed herein, it is reasonable to conclude that the deformation is *not* mediated by basal dislocations.

Returning to Fig. 8, it is obvious that Ti_3SiC_2 is an outlier, with θ values more typical of ceramics, but σ_{HEL} values that are reminiscent of metals. Based on these crucial results, we reach the following tentative conclusion: The signature of layered solids that deform by ripplocations is one where σ_{HEL} is quite low and comparable to that of metals, while simultaneously having normalized hardening rates comparable to typical ceramics.

387 Currently, there is an existing body of work, that is rapidly expanding that suggests that deformation in layered solids is not mediated by BDs, but rather by ripplocations [14,16-21]. The 388 results of this work comprise one more set of results that cannot be explained by dislocations in 389 general, or more specifically, BDs (the only kind assumed present in layered solids). The very high 390 hardening rates observed herein have been observed previously in nanoindentation experiments of 391 Ti₃SiC₂ [16]. Since in this case, no external confinement was applied, like in a nanoindentation 392 experiment, the bulk of the sample itself acts as its own confinement. We have also previously argued 393 that the dependence of the failure stresses on confinement can be taken as a good indication that 394 ripplocations and not BDs, are the operative micromechanism [16,18]. Nothing in this work contradicts 395 any of these conclusions, if anything they confirm them. 396

398 5. Summary & Conclusions

Herein the dynamic properties, at strain rates of $> 10^4$ s⁻¹, such as Hugoniot hydrostat, Hugoniot elastic limit, yield strength, spallation strength and normalized work hardening moduli of the MAX phase Ti₃SiC₂ were determined for the first time. Additionally, the temperature dependencies of dynamic yield and tensile spallation strengths were established all for the first time.

It was found that the elastic precursor wave has a relatively low amplitude that was followed by a plastic wave, terminated at the Hugoniot. Such behavior is typical of ductile metals. However, during plastic deformation, the work hardening rates were quite high and more typical of ceramics subjected to dynamic loading.

The yield strengths were also found to increase linearly with increasing test temperatures, which is atypical for ceramics. The temperature dependency of the spallation strength, however, was found to be quite similar to the quasi-static yield strength dependency. However, the values of σ_{SP} were found to be double the σ_{HEL} values, which again is more typical of metals.

Lastly, we show that if standard dislocation-based theory is used to estimate the mobile dislocation density, ρ_m , we reach the untenable conclusion that ρ_m decreases by 30% as the plastic strain rate increases from 0 to 0.001 %. Our results, however, are at least qualitatively consistent with the presence of ripplocations and ripplocation boundaries.

This conclusion is of paramount importance for several reasons. First and foremost, a totally new fracture mechanics at high strain rates – and slow ones for that matter – of layered solids has to be developed. As shown herein dislocation theory is of little use. This work could also be of great importance to seismology and seismologists at both the geological and nanoscales. We have recently shown using transmission electron microscopy, that at the atomic level deformed biotite was riddled by ripplocation bridges that in some cases were a few atomic layers thick and in other much thicker. In

- 421 general, most researchers working with layered solids have assumed implicitly or explicitly that BD
- 422 were responsible for their deformation.

423 Appendix A:

424 *Compressive stress-strain function* $\sigma(\varepsilon)$:

425 Assuming that the compressive part of the velocity history (u(t)) is a result of propagating of a simple 426 wave, having Lagrangian velocity (a(u)) through the sample, the mass-momentum conservation 427 equations is as follows:

428
$$V = -\frac{1}{2} \int_{0}^{u_{H}} \frac{du}{\rho_{0}a(u)}$$
(A1)

429
$$\sigma = \frac{1}{2} \rho_0 \int_{0}^{u_H} a(u) du , \qquad (A2)$$

where σ is considered as positive in compression, the factor $\frac{1}{2}$ reflects the assumption that the particle velocity (u_p) is equal to one half of the free surface velocity (u), u_H is the particle velocity at the top (on Hugoniot) of the shock wave. The specific volume V is related to the engineering strain by $\varepsilon = (V_o - V)/V_o$, where, V_0 is the initial specific volume.

434 The a(u) of propagation of the signal - with time t(u) - corresponding to the particle velocity value 435 $u_p = u/2$ through a sample of thickness d is

436
$$a = c_l \frac{d/c_l - \Delta t(u)/2}{d/c_l + \Delta t(u)/2},$$
 (A3)

437 where
$$\Delta t(u) = t(u) - t(u = 0) = t(u) - d/c_1$$
.

438 Knowledge of the material Hugoniot hydrostat, $U_s = C_0 + su_p$ gives the pressure

439
$$p(\varepsilon) = \rho_0 C_0^2 \frac{\varepsilon}{(1 - s\varepsilon)^2}$$
(A4)

440 at the temperature of interest and allows us to find the dependencies of shear, τ

441
$$\tau = \frac{3}{4} \left[\sigma(\varepsilon) - p(\varepsilon) \right], \tag{A5}$$

442 and yield $Y = 2\tau$ stresses on the plastic strain

443
$$\gamma = \int_{0}^{u} \left(\frac{du}{a} - \frac{d\tau}{G}\right).$$
 (A6)

When the dependence $Y(\gamma)$ is obtained, the 0.1 % proof stress $Y_{0.1} = Y(\gamma = 0.001)$ can be found. The Ti₃SiC₂ Hugoniot hydrostat $U_s = C_0 + su_p = 6.47 + 1.61u_p$ is known only for 300 K. Here we assume that this hydrostat and c_b are not functions of temperature.

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451 **References**

- M.W. Barsoum, The MN+1AXN phases: A new class of solids: Thermodynamically stable
 nanolaminates, Prog. Solid State Chem. 28 (2000) 201–281. doi:10.1016/S0079-6786(00)000066.
- M. Sokol, V. Natu, S. Kota, M.W. Barsoum, On the Chemical Diversity of the MAX Phases,
 Trends Chem. xx (2019) 1–14. doi:10.1016/j.trechm.2019.02.016.
- 457 [3] X.H. Wang, Y.C. Zhou, Layered Machinable and Electrically Conductive Ti2AlC and Ti3AlC2
 458 Ceramics: a Review, J. Mater. Sci. Technol. 26 (2010) 385–416. doi:10.1016/S1005459 0302(10)60064-3.
- Z.M. Sun, Progress in research and development on MAX phases: a family of layered ternary
 compounds, Int. Mater. Rev. 56 (2011) 143–166. doi:10.1179/1743280410Y.0000000001.
- 462 [5] M. Barsoum, T. El-Raghy, The MAX Phases: Unique New Carbide and Nitride Materials, Am.
 463 Sci. 89 (2001) 334. doi:10.1511/2001.4.334.
- 464 [6] M.W. Barsoum, M. Radovic, Elastic and Mechanical Properties of the MAX Phases, (2011).
 465 doi:10.1146/annurev-matsci-062910-100448.
- 466 [7] B.T. (Brian T. Kelly, Physics of graphite, Applied Science, 1981.
- 467 https://books.google.com/books/about/Physics_of_Graphite.html?id=rGR3QgAACAAJ
 468 (accessed February 26, 2019).
- [8] R. Christoffersen, A.K. Kronenberg, Dislocation interactions in experimentally deformed biotite,
 J. Struct. Geol. 15 (1993) 1077–1095. doi:10.1016/0191-8141(93)90157-6.
- 471 [9] F.C. Frank, A.N. Stroh, On the Theory of Kinking, Proc. Phys. Soc. Sect. B. 65 (1952) 811–821.
 472 doi:10.1088/0370-1301/65/10/311.
- 473 [10] M.W. Barsoum, T. Zhen, A. Zhou, S. Basu, S.R. Kalidindi, Microscale modeling of kinking
 474 nonlinear elastic solids, (n.d.). doi:10.1103/PhysRevB.71.134101.
- 475 [11] M.A. BIOT, FURTHER DEVELOPMENT OF THE THEORY OF INTERNAL BUCKLING
 476 OF MULTILAYERS, GSA Bull. 76 (1965) 833–840. doi:10.1130/0016477 7606(1965)76[833:fdotto]2.0.co;2.
- 478 [12] P.J. Hudleston, S.H. Treagus, Information from folds: A review, J. Struct. Geol. 32 (2010)
 479 2042–2071. doi:10.1016/J.JSG.2010.08.011.
- 480 [13] S.R. Reid, C. Peng, Dynamic uniaxial crushing of wood, Int. J. Impact Eng. 19 (1997) 531–570.
 481 doi:10.1016/S0734-743X(97)00016-X.
- 482 [14] M.W. Barsoum, X. Zhao, S. Shanazarov, A. Romanchuk, S. Koumlis, S.J. Pagano, L.

- 483 Lamberson, G.J. Tucker, Ripplocations: A universal deformation mechanism in layered solids,
 484 Phys. Rev. Mater. 3 (2019) 013602. doi:10.1103/PhysRevMaterials.3.013602.
- 485 [15] A. Kushima, X. Qian, P. Zhao, S. Zhang, J. Li, Ripplocations in van der Waals Layers, Nano
 486 Lett. 15 (2015) 1302–1308. doi:10.1021/nl5045082.
- 487 [16] J. Griggs, A.C. Lang, J. Gruber, G.J. Tucker, M.L. Taheri, M.W. Barsoum, Spherical
 488 nanoindentation, modeling and transmission electron microscopy evidence for ripplocations in
 489 Ti3SiC2, Acta Mater. 131 (2017) 141–155. doi:10.1016/J.ACTAMAT.2017.03.055.
- 490 [17] J. Gruber, A.C. Lang, J. Griggs, M.L. Taheri, G.J. Tucker, M.W. Barsoum, Evidence for Bulk
 491 Ripplocations in Layered Solids, Sci. Rep. 6 (2016) 33451. doi:10.1038/srep33451.
- 492 [18] M.W. Barsoum, G.J. Tucker, Deformation of layered solids: Ripplocations not basal
 493 dislocations, Scr. Mater. 139 (2017) 166–172. doi:10.1016/J.SCRIPTAMAT.2017.04.002.
- 494 [19] D. Freiberg, M.W. Barsoum, G.J. Tucker, Nucleation of ripplocations through atomistic
 495 modeling of surface nanoindentation in graphite, Phys. Rev. Mater. 2 (2018) 053602.
 496 doi:10.1103/PhysRevMaterials.2.053602.
- 497 [20] J. Aslin, E. Mariani, K. Dawson, M.W. Barsoum, Ripplocations provide a new mechanism for
 498 the deformation of phyllosilicates in the lithosphere, Nat. Commun. 10 (2019) 686.
 499 doi:10.1038/s41467-019-08587-2.
- [21] A.V. Alaferdov, R. Savu, M.A. Canesqui, Y.V. Kopelevich, R.R. da Silva, N.N. Rozhkova, D.A.
 Pavlov, Y.V. Usov, G.M. de Trindade, S.A. Moshkalev, Ripplocation in graphite nanoplatelets
 during sonication assisted liquid phase exfoliation, Carbon N. Y. 129 (2018) 826–829.
 doi:10.1016/J.CARBON.2017.12.100.
- 504 [22] A.K. Kronenberg, S.H. Kirby, J. Pinkston, Basal slip and mechanical anisotropy of biotite, J.
 505 Geophys. Res. 95 (1990) 19257. doi:10.1029/JB095iB12p19257.
- 506 [23] B. Budiansky, N.A. Fleck, J.C. Amazigo, On kink-band propagation in fiber composites, J.
 507 Mech. Phys. Solids. 46 (1998) 1637–1653. doi:10.1016/S0022-5096(97)00042-2.
- [24] Q. Yang, B. Cox, Cohesive models for damage evolution in laminated composites, Int. J. Fract.
 133 (2005) 107–137. doi:10.1007/s10704-005-4729-6.
- 510 [25] N.A. Fleck, Compressive Failure of Fiber Composites, Adv. Appl. Mech. 33 (1997) 43–117.
 511 doi:10.1016/S0065-2156(08)70385-5.
- 512 [26] G.W. Bentzel, M. Sokol, J. Griggs, A.C. Lang, M.W. Barsoum, On the interactions of Ti2AlC,
 513 Ti3AlC2, Ti3SiC2and Cr2AlC with palladium at 900 °C, J. Alloys Compd. (2019).
 514 doi:10.1016/j.jallcom.2018.08.127.
- 515 [27] M.W. Barsoum, T. Zhen, S.R. Kalidindi, M. Radovic, A. Murugaiah, Fully reversible,

- dislocation-based compressive deformation of Ti3SiC2 to 1 GPa, Nat. Mater. 2 (2003) 107–111.
 doi:10.1038/nmat814.
- 518 [28] A.G. Zhou, M.W. Barsoum, Kinking nonlinear elastic deformation of Ti3AlC2, Ti2AlC,
 519 Ti3Al(C0.5,N0.5)2 and Ti2Al(C0.5,N0.5), J. Alloys Compd. 498 (2010) 62–70.
- 520 doi:10.1016/J.JALLCOM.2010.03.099.
- [29] S. Basu, A. Zhou, M.W. Barsoum, On spherical nanoindentations, kinking nonlinear elasticity of
 mica single crystals and their geological implications, J. Struct. Geol. 31 (2009) 791–801.
 doi:10.1016/J.JSG.2009.05.008.
- [30] R. Bhattacharya, R. Benitez, M. Radovic, N.C. Goulbourne, High strain-rate response and
 deformation mechanisms in polycrystalline Ti2AlC, Mater. Sci. Eng. A. 598 (2014) 319–326.
 doi:10.1016/j.msea.2014.01.032.
- [31] P. Naik Parrikar, H. Gao, M. Radovic, A. Shukla, Static and Dynamic Thermo-Mechanical
 Behavior of Ti2AlC MAX Phase and Fiber Reinforced Ti2AlC Composites, in: Springer, Cham,
 2015: pp. 9–14. doi:10.1007/978-3-319-06995-1 3.
- [32] P. Naik Parrikar, R. Benitez, H. Gao, M. Radovic, A. Shukla, Mechanical response of fine
 grained Ti2AlC under extreme thermo-mechanical loading conditions, Mater. Sci. Eng. A. 658
 (2016) 176–184. doi:10.1016/J.MSEA.2016.01.106.
- [33] L. Shannahan, M.W. Barsoum, L. Lamberson, Dynamic fracture behavior of a MAX phase
 Ti3SiC2, Eng. Fract. Mech. 169 (2017) 54–66. doi:10.1016/J.ENGFRACMECH.2016.11.006.
- J.L. Jordan, Hugoniot Measurements of High Pressure Phase Stability of Titanium-Silicon
 Carbide (Ti3SiC2), 77 (2004) 77–80. doi:10.1063/1.1780188.
- L.M. Barker, R.E. Hollenbach, Laser interferometer for measuring high velocities of any
 reflecting surface, J. Appl. Phys. 43 (1972) 4669–4675. doi:10.1063/1.1660986.
- [36] W.S. Rasband, ImageJ, U.S. National Institutes of Health, Bethesda, ML, USA, 1997–2005,
 (n.d.). http://rsb.info.nih.gov/ij/.
- [37] M. Radovic, M. Barsoum, A. Ganguly, T. Zhen, On the elastic properties and mechanical
 damping of Ti3SiC2, Ti3GeC2, Ti3Si0. 5Al0. 5C2 and Ti2AlC in the 300-1573 K temperature
 range, Acta Mater. (2006).
- 544 [38] E.B. Zaretsky, High temperature impact response of 998 alumina, J. Appl. Phys. 114 (2013)
 545 183518. doi:10.1063/1.4830014.
- 546 [39] L.M. Barker, R.E. Hollenbach, Shock Wave Studies of PMMA, Fused Silica, and Sapphire, J.
 547 Appl. Phys. 41 (1970) 4208–4226. doi:10.1063/1.1658439.
- 548 [40] D.J. Chapman, D.E. Eakins, D.M. Williamson, W. Proud, Index of refraction measurements and

- 549 window corrections for PMMA under shock compression, AIP Conf. Proc. 1426 (2012) 442–
- 445. https://www.tib.eu/en/search/id/BLCP:CN082480910/Index-of-refraction-measurementsand-window-corrections (accessed January 2, 2019).
- I. Girlitsky, E. Zaretsky, S. Kalabukhov, M.P. Dariel, N. Frage, Dynamic compressive and
 tensile strengths of spark plasma sintered alumina, J. Appl. Phys. 115 (2014) 243505.
 doi:10.1063/1.4885436.
- G.I. Kanel, V.E. Fortov, S. V. Razorenov, Shock-Wave Phenomena and the Properties of
 Condensed Matter, Springer New York, New York, NY, 2004. doi:10.1007/978-1-4757-4282-4.
- [43] M.W. Barsoum, T. El-Raghy, Synthesis and Characterization of a Remarkable Ceramic:
 Ti3SiC2, J. Am. Ceram. Soc. 79 (1996) 1953–1956. doi:10.1111/j.1151-2916.1996.tb08018.x.
- T. El-Raghy, M.W. Barsoum, A. Zavaliangos, S.R. Kalidindi, Processing and Mechanical
 Properties of Ti3SiC2: II, Effect of Grain Size and Deformation Temperature, J. Am. Ceram.
 Soc. 82 (2004) 2855–2860. doi:10.1111/j.1151-2916.1999.tb02167.x.
- 562 [45] Y. Bao, H. Zhang, Y. Zhou, Punch-shear tests and size effects for evaluating the shear strength
 563 of machinable ceramics, Zeitschrift Für Met. 95 (2004) 372–376. doi:10.3139/146.017960.
- [46] M.W. Barsoum, D. Brodkin, T. El-Raghy, Layered machinable ceramics for high temperature
 applications, Scr. Mater. 36 (1997) 535–541. doi:10.1016/S1359-6462(96)00418-6.
- 566 [47] G.E. Duvall, R.A. Graham, G.E. Duvail, R.A. Graham, G.E. Duvall, R.A. Graham, Phase
 567 transitions under shock-wave loading, American Physical Society, 1977.
 568 doi:10.1103/RevModPhys.49.523.
- 569 [48] B. Escaig, DISSOCIATION AND MECHANICAL PROPERTIES. DISLOCATION
- 570 SPLITTING AND THE PLASTIC GLIDE PROCESS IN CRYSTALS, J. Phys. Colloq. 35
 571 (1974). doi:10.1051/jphyscol:1974717.
- 572 [49] B. Manoun, S.K. Saxena, H.-P. Liermann, M.W. Barsoum, Thermal Expansion of
 573 Polycrystalline Ti3SiC2 in the 25o-1400oC Temperature Range, J. Am. Ceram. Soc. 88 (2005)
 574 3489–3491. doi:10.1111/j.1551-2916.2005.00642.x.
- 575 [50] R. Fowles, R.F. Williams, Plane Stress Wave Propagation in Solids, J. Appl. Phys. 41 (1970)
 576 360–363. doi:10.1063/1.1658348.
- 577 [51] E.B. Zaretsky, G.I. Kanel, S.V. Razorenov, K. Baumung, Impact strength properties of nickel578 based refractory superalloys at normal and elevated temperatures, Int. J. Impact Eng. 31 (2005)
 579 41–54. doi:10.1016/J.IJIMPENG.2003.11.004.
- [52] A.V. Antoun, T., Seaman, L., Curran, D.R., Kanel, G.I., Razorenov, S.V., Utkin, Spall fracture,
 Springer, new, 2002.

- [53] M.W. Barsoum, MAX Phases Properties of Machinable Ternary Carbides and Nitrides, 1st ed.,
 Wiley-VCH Verlag GmbH & Co., Weinheim, Germany, 2013.
- [54] G.E. Duvall, Propagation of Plane Shock Waves in a Stress-Relaxing Medium, in: Stress Waves
 Anelastic Solids, Springer Berlin Heidelberg, Berlin, Heidelberg, 1964: pp. 20–32.
 doi:10.1007/978-3-642-88288-3 2.
- 587 [55] A.D. Brailsford, Anharmonicity Contributions to Dislocation Drag, J. Appl. Phys. 43 (1972)
 588 1380–1393. doi:10.1063/1.1661329.
- [56] C.S. Alexander, L.C. Chhabildas, W.D. Reinhart, D.W. Templeton, Changes to the shock
 response of fused quartz due to glass modification, Int. J. Impact Eng. 35 (2008) 1376–1385.
 doi:10.1016/J.IJIMPENG.2008.07.019.
- 592 [57] D.E. Munson, R.J. Lawrence, Dynamic deformation of polycrystalline alumina, J. Appl. Phys.
 593 50 (1979) 6272. doi:10.1063/1.325766.
- 594 [58] D.R. Christman, N.H. Fnoulaf, Dynamic Properties of High-Purity Beryllium, 8 (n.d.).
 595 doi:10.2514/3.5693.
- [59] S. V. Razorenov, Hugoniot Elastic Limit and Spall Strength of Aluminum and Copper Single
 Crystals over a Wide Range of Strain Rates and Temperatures, 503 (2003) 503–506.
 doi:10.1063/1.1483587.
- [60] M.E. Kipp, D.E. Grady, Shock compression and release in high-strength ceramics, Albuquerque,
 NM, and Livermore, CA (United States), 1989. doi:10.2172/5777672.
- [61] C.H. Mok, J. Duffy, The dynamic stress-strain relation of metals as determined from impact tests
 with a hard ball, Int. J. Mech. Sci. 7 (1965) 355–371. doi:10.1016/0020-7403(65)90064-0.
- [62] M. Sokol, S. Kalabukhov, R. Shneck, E. Zaretsky, N. Frage, Effect of grain size on the static and
 dynamic mechanical properties of magnesium aluminate spinel (MgAl 2 O 4), J. Eur. Ceram.
 Soc. 37 (2017) 3417–3424. doi:10.1016/j.jeurceramsoc.2017.04.025.
- E.B. Zaretsky, G.I. Kanel, The high temperature impact response of tungsten and chromium, J.
 Appl. Phys. 122 (2017) 115901. doi:10.1063/1.4997674.
- [64] S. Asgari, E. El-Danaf, S.R. Kalidindi, R.D. Doherty, Strain hardening regimes and
 microstructural evolution during large strain compression of low stacking fault energy fcc alloys
- that form deformation twins, Metall. Mater. Trans. A. 28 (1997) 1781–1795.
- 611 doi:10.1007/s11661-997-0109-3.
- [65] M.H. Rice, R.G. McQueen, J.M. Walsh, Compression of Solids by Strong Shock Waves, Solid
 State Phys. 6 (1958) 1–63. doi:10.1016/S0081-1947(08)60724-9.
- 614 [66] G.I. Kanel, E.B. Zaretsky, S. V. Razorenov, A.S. Savinykh, G. V. Garkushin, Evolutions of

- elastic-plastic shock compression waves in different materials, in: AIP Conf. Proc., AIP
 Publishing LLC, 2017: p. 100030. doi:10.1063/1.4971655.
- E.B. Zaretsky, G.I. Kanel, Effect of temperature, strain, and strain rate on the flow stress of
 aluminum under shock-wave compression, J. Appl. Phys. 112 (2012) 073504.
- 619 doi:10.1063/1.4755792.
- E.B. Zaretsky, G.I. Kanel, Response of copper to shock-wave loading at temperatures up to the
 melting point, J. Appl. Phys. 114 (2013) 083511. doi:10.1063/1.4819328.