Damage Evolution in Ductile Metals

Developing a quantitative and predictive understanding

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If one pulls hard enough on a bar of soft, ductile metal such as aluminum, it will stretch, and if one continues to pull, the bar will eventually break. What happens internally to cause it to break? To begin with, the pulling puts the material into a state of tension, which, if high enough, will cause tiny voids to form. This process, known as nucleation, typically begins near sites of defects such as impurities that are introduced during the original processing of the material. Following nucleation, the voids begin to expand, and, if close enough together, they coalesce to form microscopic cracks. In regions having a high density of voids and microcracks, this progression culminates with the development of a complete surface failure; that is, the bar breaks. That scenario is the currently accepted model of damage evolution in most ductile metals.

Metals that are subjected to shock waves can also fail via this pathway. In shocked materials, a state of high tension can be produced as pressure waves reflect off free surfaces and interact with each other. Shock-induced damage, or spall, as it is known among material physicists, occurs in metals shocked by lasers and in tank armor hit by conventional munitions. Because even the plutonium in a nuclear weapon can spall, this process is an important area of research for science-based stockpile stewardship. At Los Alamos collaboration between experimentalists and modelers is beginning to paint a detailed picture of the events leading up to spall. In this article we discuss recent results from gas-gun shock spall experiments specially designed to investigate the dynamics of ductile damage and failure.

Rather sophisticated models of damage evolution that incorporate many of the steps involved in metals spallation are being developed at Los Alamos. One of the authors is developing a new micromechanical model that includes void growth, void coalescence, and crack formation (Tonks et al. 2002). When validated, the model will replace simpler damage models currently employed in advanced simulation and computing codes, tensile plasticity codes, and others. To aid the validation process and provide direction for further improvements in the model, we performed a number of well-controlled gas-gun experiments on the evolution of spall in tantalum and copper targets. Ideally, one would like to have enough control to arrest the damage evolution at different stages of development. In our gas-gun experiments, we made the shock pressures large enough to initiate the damage evolution sequence, but not so large as to result in fracturing the samples. The resulting damage is called incipient spall. We also varied the shock loading to investigate the effects on damage from changes in peak pressure and shock duration. The targets were recovered and microscopically examined to determine the degree and type of damage produced under each loading condition.

Figure 1 shows optical micrographs of cross sections through the damaged region of two tantalum samples. Both samples were shocked to the same peak stress, but the duration of the shocks differed by a factor of two. Both samples show damage in the form of spherical voids, but the sample subjected to the longer period of shock loading developed discontinuities in the microstructure of the metal—see Figure 1(b). These "linkages" are attributed to strain localizations that presumably had time to develop during the longer loading period. Such areas

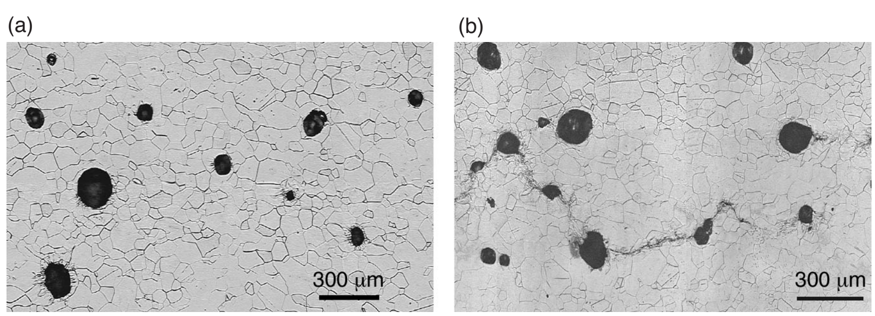


Figure 1. Optical Micrographs of Incipient Spall in Tantalum
The optical micrographs show the microstructure of damaged regions of tantalum samples after spall tests at a shock pressure of 5.6 GPa (56 kb). The samples were subjected to different shock durations: (a) 1.1 μs, resulting in a final porosity of 4.1% and (b) 2.2 μs, resulting in a final porosity of 11.6%. The sample in (b), which was subjected to a longer period of shock loading, shows a line of damage connecting two voids, a form of damage not included in most models of spall.

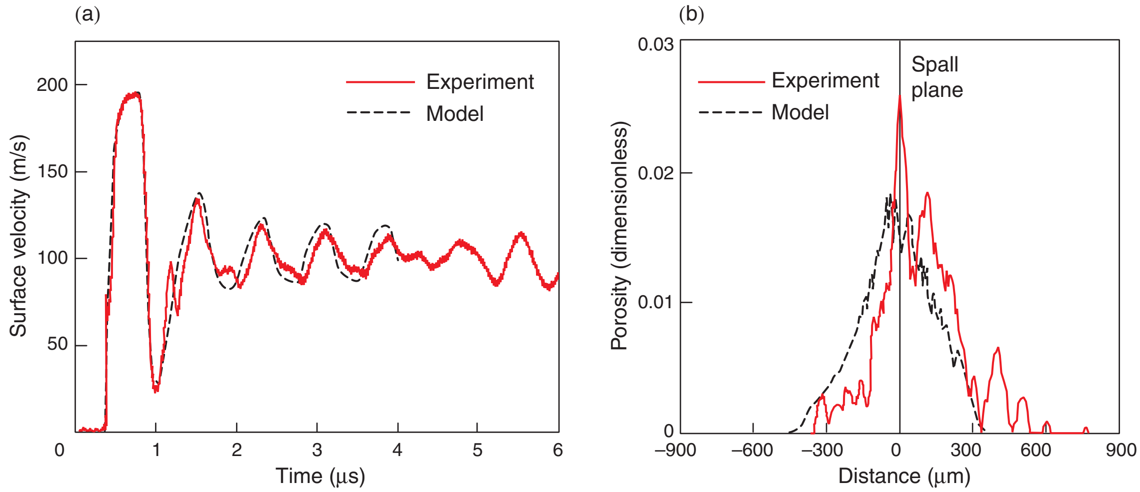


Figure 2. Comparison of Model Predictions and Experimental Results for Tantalum Spall
In the gas-gun experiment, a quartz flier plate, 1.5 mm thick, hit a tantalum target of the same thickness at 448 m/s. The impact produced in the target a shock pressure of 5.6 GPa for a duration of 0.4 μs. (a) The plots show close agreement between model predictions and experimental results for the velocity history of the target's back surface (VISAR trace). (b) The plots show the resulting porosity distribution in the region of the spall plane, with maximum porosity of 0.025. The standard error between the measured and modeled porosity was calculated as 0.05.

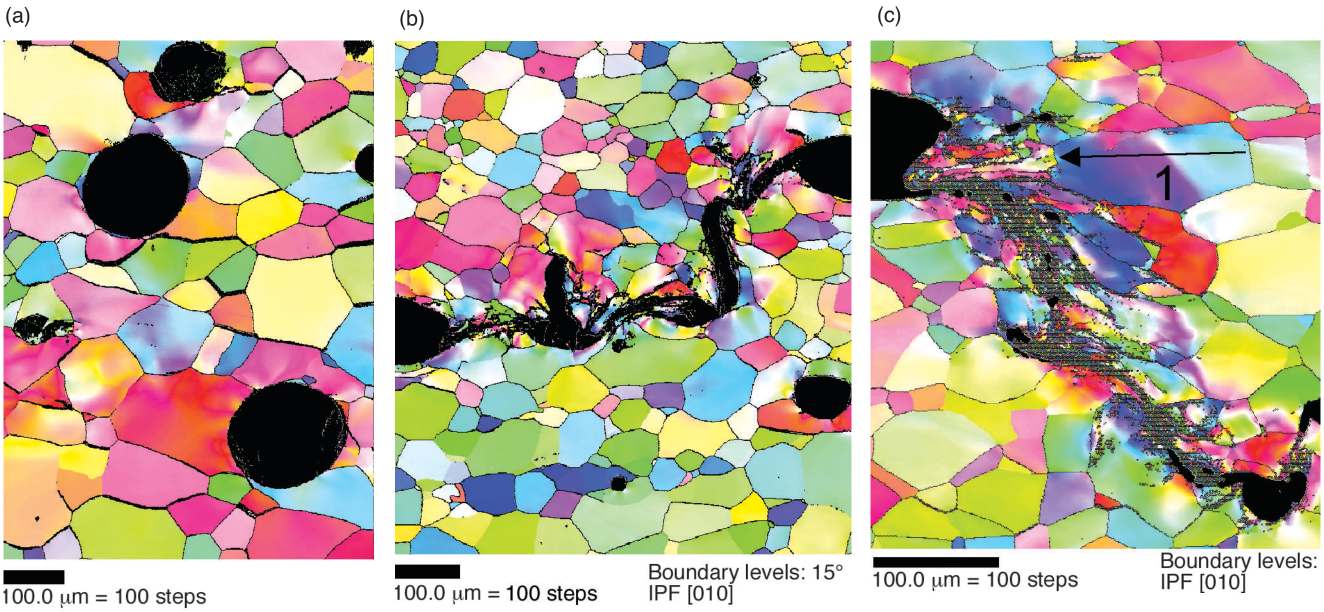
of high deformation between voids, although not accounted for in most models, might be the precursor to the coalescence of voids into microcracks. However, this hypothesis requires further investigation.

At present, we can measure void sizes and distributions, volumetric void-number-density distributions, clustering, near-neighbor distances, strain localization distances, and final porosity in recovered targets. Those data are provided to the modelers to test the accuracy of their predictions. On each spall test, we used velocity interferometry (VISAR, or velocity interferometer system for any reflector) to measure the back free-surface velocity of the shocked target as a function of time, providing yet another piece of constraining data for the model predictions. (See Figure 3 in the Hixson article on page 117 for a discussion of this measurement technique.) Figure 2 shows an example in which the model very accurately predicted the free-surface velocity history as well as the incipient damage of tantalum. In that case, postshot metallurgical inspection revealed no evidence of strain localization in the sample. However, when shock loading produced more-extensive damage in the sample, including coalescence and strain localization, the model was significantly less accurate in predicting the results. In particular, the model overpredicted the amount of porosity developed in the sample. We surmise that the energy from the shock that went into making the extra voids in the model calculation was in reality partitioned into creating linkages between voids.

To gain additional quantitative information pertaining to strain localization and to guide further improvements to our damage models, we are developing the technique of automated electron backscatter diffraction (EBSD) (Adams et al. 1993). In regions where the strain in a material has localized, the dislocation density increases and causes degradation in the electron backscatter patterns. From these degraded patterns, one can extract strain localization information, such as information on their distributions, lengths, and widths.

Figure 3 shows examples of the details that can be analyzed by ESBD. False coloring is used to identify regions of specific crystallographic orientation within the individual grains of the sample tested. The highlighted grain boundaries have a misorientation (relative to the bulk) no larger than 15°. Regions where the misorientation is so high that the technique does not resolve the details appear as gray pixels. We equate those regions of a high deformation with regions of strain localization. Since the information from EBSD is in a digital form, we can clearly differentiate the portion of the energy consumed by the void formation from that consumed by the linkages, or the strain instabilities.

We are making significant progress in our quest to understand the phenomenon of ductile damage evolution and failure. More spall experiments are planned, as are experiments to investigate spall in shocked plutonium. To appreciate the complexity of this problem, consider that the first studies of material failure have been attributed to Leonardo da Vinci. Nearly five hundred years later we are still at the beginning; however, with better models, new diagnostic techniques, and well-controlled experiments, the future looks promising. ■



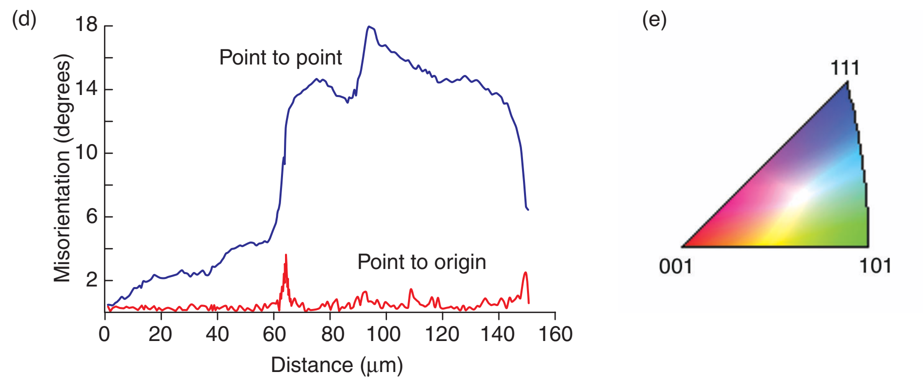


Figure 3. EBSD Images of Incipient Spall in Tantalum
(a) The EBSD micrograph of the sample in Figure 1(a) confirms the absence of any significant strain localization between individual voids. (b) A fragment of the damaged sample in Figure 1(b), which was exposed to the higher stresses, exhibits continuous and tortuous (black) features between two voids. We equate those features with strain localization. (c) In a magnified view of (b), one large grain near the region of strain localization (overlaid with an arrow) displays a particularly high level of misorientation as represented by the continuous change in color. (d) The graph plots the misorientation angle along the arrow measured from right to left in step sizes of 1 μm. The blue curve shows orientation changes from the origin of the arrow to points along the arrow, and the red curve shows point-to-point changes in orientation. (e) The color key correlates color with crystallographic orientation in the individual grains.

Further Reading

Adams, B. L., S. I. Wright, and K. Kunze. 1993. Orientation Imaging—The Emergence of a New Microscopy. *Metall. Trans. A* **24** (4): 819.

Tonks, D. L., A. K. Zurek, W. R. Thissell, J. E. Vorthman, and R. S. Hixson. 2002. “The Tonks Ductile Damage Model,” Los Alamos National Laboratory document LA-UR-03-0809.

Anna Zurek received her M.S. degree in 1976 from the University of Mining and Technology in Krakow, Poland, and her Ph.D. in 1981 from the University of Texas at Austin. In 1985, Anna joined the Laboratory’s Materials Science and Technology Division, where her current interests include the spall process and materials microstructure/property relationship in the environment of high strain rate and complex stress state. Anna has authored and coauthored more than 100 technical publications.

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