COMMEMORATIVE LECTURE
BY THE TWENTY-FOURTH GOLD MEDALIST OF
THE JAPAN INSTITUTE OF METALS

Two Aspects of Impact Tests on Metals*

By Seita Sakui**

In a dynamic test (~ 1 m/s) fine grained iron and steel show only slight elongation while coarse grained specimens show large elongation. In the fine grained specimens, showing this apparent brittleness, the deformation is concentrated near the grips and is not propagated along the entire gage length. By using the wave propagation analysis, the quasi-critical impact velocity, adequate for materials with Lüders elongation and high yield stress, has been derived. It has been shown that the smaller the grain size, the lower the quasi-critical impact velocity, in accordance with the experimental observations. The existence of the apparent brittleness, pseudo-brittle fracture, has been found in aluminum. It agrees with the prediction based on the wave propagation analysis and the work hardening characteristics brought about by the combination of high strain rate deformation and subsequent slow deformation.

I. Introduction

A piece of wire elongates when it is pulled. This simple experience is described as follows; the wire starts to deform at the pulling end and the deformation propagates along the wire until it reaches the other end. In a static test, a very small amount of deformation propagates with a high velocity so that we practically observe the uniform elongation. However, in an impact test, a large amount of deformation travels at a relatively low velocity. This induces a heterogeneous elongation in the wire during a high speed test, as seen in Fig. 1⁽¹⁾. Wave propagation is essentially a mechanics problem although it is closely related to material properties as a constitutive equation. Conversely, one obtains information on material properties, intermingled with mechanics phenomena in a dynamic test. In other words, when we conduct an impact test, we encounter two aspects, both of the material properties and of the mechanics

Fig. 1 Distributions of elongation along 3 m long wire. Number on each curve shows the amount of elongation given by impact tensile load. Elongation along wire was measured at 50 cm intervals. ($7.7 \text{ m} \cdot \text{s}^{-1}$)

of wave propagation. These two aspects sometimes work together and sometimes overshadow each other. In this paper, I would like to show the coexistence of the two faces by demonstrating an interesting phenomenon observed in an impact test of metals.

* A part of the memorial lecture delivered at the Spring Meeting of Japan Institute of Metals, April 4, 1979, as the 1978 Gold Medalist of Japan Institute of Metals.

** Professor Emeritus, Tokyo Institute of Technology, Tokyo, Japan.

II. Strange Trait of Brittleness of Iron, Pseudo-brittleness

In Fig. 2, the elongation during static and dynamic tests of iron wire is plotted against annealing temperature⁽²⁾. In the static test, the lower the annealing temperature, the greater the elongation. This is what we usually expect. However, in the dynamic tests, the elongation is almost zero when the iron wire is annealed below 700°C and increases suddenly when the annealing temperature is over 750°C. The almost zero elongation at the lower annealing temperature is called pseudo-brittleness, the meaning of which will become clear later.

When I had an opportunity to discuss this phenomenon with research groups at some steel industries, no satisfactory explanation was offered. However, it was agreed that Fig. 2 was a strange result. The curiosity demanded me to have one of my boys run a similar test on mild steel. A plot of the elongation against the grain size in this test is shown in Fig. 3 to show the observation is closely related to the grain size⁽³⁾. This is also in striking contrast with the grain size effect on the ductile-brittle transition of steel. In a sense, the high speed test compares well with the low temperature test. It is taken for granted that steel becomes brittle at a low temperature. An example is given in Fig. 4⁽⁴⁾.

Fig. 2 Elongation-annealing temperature curves for pure iron under various strain rates. The falling heights of hammer were varied to change the strain rates on impact tensile tests.

Fig. 3 Overall elongation vs grain size curves on dynamic and static tensile tests for 0.03%C rimmed steel.

Fig. 4 Effect of grain size on the yield and fracture stresses of a low carbon steel in tension at -196°C . Grain size vs strain to fracture curve is also shown.

However, a close examination of the specimens which had exhibited pseudo-brittleness, produced a breakthrough; these specimens in fact deformed considerably at localized positions close to the impact or the fixed end. However, the rest of each specimen showed no evidence of plastic deformation. The specimens failed in a ductile manner. In short, the specimens were ductile but the plastic deformation did not propagate along the specimen. The mechanics of wave propagation is now on stage.

III. Propagation of Deformation

In a one-dimensional analysis, deformation

propagates with a velocity given by⁽⁵⁾⁽⁶⁾

$$C = \sqrt{\frac{1}{\rho} \frac{d\sigma}{d\varepsilon}}, \quad (1)$$

where ρ is the density, σ the nominal stress and ε the nominal strain. The slope of the stress-strain curve, $d\sigma/d\varepsilon$, decreases with increasing strain in most metals. However, some metals, such as steel, have an upwardly concave portion of their stress-strain curves. In this situation, a large strain catches up a small strain, forming a shock front. The Lüders front is a typical example of this⁽⁷⁾. This is pictured in Fig. 5. Thus, in the situation where there is an upwardly concave σ - ε curve the propagation mechanics must be formulated differently from the usual approach.

1. Critical impact velocity, regular type

When a bar of semi-infinite length is stretched at the impact end with the velocity V , the strain at the impact end is given by finding ε_v satisfying the following equation:

$$V = \int_0^{\varepsilon_v} \sqrt{\frac{1}{\rho} \frac{d\sigma}{d\varepsilon}} d\varepsilon. \quad (2)$$

Suppose V exceeds V_{cr} given by

$$V_{cr} = \int_0^{\varepsilon_m} \sqrt{\frac{1}{\rho} \frac{d\sigma}{d\varepsilon}} d\varepsilon, \quad (3)$$

where ε_m is the strain at $d\sigma/d\varepsilon=0$. Then the strain at the impact end becomes larger than ε_m . (ε_m corresponds to the uniform elongation in a static test). Since such a strain cannot

propagate, the bar will fail instantaneously at the impact end. That is, when the impact velocity is larger than V_{cr} , the (ductile) fracture occurs at the impact end, the remaining portion being undeformed, at least theoretically. There is not enough time for the deformation to propagate deep into the specimen. This is the important conclusion reached by Karman and Duwez⁽⁵⁾ and by Taylor⁽⁶⁾ and this V_{cr} is called the critical impact velocity.

2. Quasi-critical impact velocity⁽³⁾

The pseudo-brittle fracture mentioned in section II is similar in nature to the fracture occurring when the impact velocity exceeds V_{cr} . However, further consideration is necessary before developing a complete formulation. Annealed iron and steel have a dynamic yield stress larger than ~ 490.3 MPa. Thus, the impact velocity which gives the yield stress at the impact end is calculated to be > 12 m/s. The pseudo-brittle fracture occurs when the impact velocity is below 1.4 m/s, see Fig. 2. It is clear, therefore, that the fracture does not occur at the instant when the test has started. It must have occurred after the elastic wave traveled back and forth between the impact and the fixed end, thereby increasing the stress level. After some of these reflections of the elastic wave, the yielding starts either at the fixed or the impact end. Then, as a first approximation, we can treat the wave propagation after the yielding occurs in the situation demonstrated in Figs. 2 and 3. When a metal like iron starts to yield, the first thing to occur is the formation of a Lüders band. Let the propagation velocity of the band be C_L and the difference of the material velocity in the deformed and the undeformed parts be V_L . The mass conservation law and the momentum conservation law give

$$C_L = \sqrt{\frac{1}{\rho} \frac{(\sigma_1 - \sigma_0)}{(\varepsilon_1 - \varepsilon_0)}}, \quad (4)$$

$$V_L = \sqrt{\frac{1}{\rho} (\sigma_1 - \sigma_0)(\varepsilon_1 - \varepsilon_0)}, \quad (5)$$

where σ_1 and ε_1 are, respectively, the stress and the strain in the Lüders band and σ_0 and ε_0 are the corresponding quantities in the

Fig. 5 Emergence of impact wave with a sharp wave front during propagation along a specimen. (Schematic)

undeformed region, see Fig. 6. As the impact velocity increases, the stress and strain in the Lüders band increase according to (5). When V_L exceeds V_t given by

$$V_t = \sqrt{\frac{1}{\rho}(\sigma_t - \sigma_0)(\varepsilon_t - \varepsilon_0)}, \quad (6)$$

where σ_t and ε_t are the stress and strain at the point at which the straight line from $(\varepsilon_0, \sigma_0)$ touches the smooth part of the stress-strain curve (Fig. 7), the normal plastic wave follows the Lüders band. This situation is envisaged in Fig. 8. In region II of Fig. 8, the material velocity V is related to the stress and strain by

Fig. 6 Schematic diagram of Lüders band propagation.

Fig. 7 Schematic stress-strain diagram of iron and steel.

Fig. 8 Distributions of stress and strain in a specimen which has been deformed dynamically with a higher velocity than V_t in eq. (6).

$$V_t = \sqrt{1/\rho(\sigma_t - \sigma_0)(\varepsilon_t - \varepsilon_0)} \quad (6)$$

$$\frac{dV}{d\varepsilon} = \sqrt{\frac{1}{\rho} \frac{d\sigma}{d\varepsilon}}. \quad (7)$$

Combining the solution of (7) with (6), one can write the impact velocity as

$$V = \sqrt{\frac{1}{\rho}(\sigma_t - \sigma_0)(\varepsilon_t - \varepsilon_0)} + \int_{\varepsilon_t}^{\varepsilon_V} \sqrt{\frac{1}{\rho} \frac{d\sigma}{d\varepsilon}} d\varepsilon, \quad (8)$$

where ε_V is the strain at the impact end. The remaining discussion parallels that given in section III.1. That is, when the impact velocity exceeds $V_{cr.q}$ given by

$$V_{cr.q} = \sqrt{\frac{1}{\rho}(\sigma_t - \sigma_0)(\varepsilon_t - \varepsilon_0)} + \int_{\varepsilon_t}^{\varepsilon_m} \sqrt{\frac{1}{\rho} \frac{d\sigma}{d\varepsilon}} d\varepsilon, \quad (9)$$

the specimen fails at the impact end, without the plastic deformation propagating along the specimen⁽³⁾. The fracture mode is still ductile in a localized manner. We call $V_{cr.q}$ the quasi-critical impact velocity.

3. Grain size effect of $V_{cr.q}$

We interpret the results of Figs. 2 and 3 in such a manner that the impact velocity exceeds $V_{cr.q}$ of the small grained specimens while it is below $V_{cr.q}$ of the large grained specimens. This requires that the larger the grain size, the larger $V_{cr.q}$.

It is known that the effect of the grain size on the stress-strain curve of iron is large at yielding and diminishes as the amount of deformation increases, because the work hardening plays a more important role in the later

stage of deformation; specifically, the yield stress and the yield elongation are larger for the smaller grained specimen and the work hardening region eventually coincides in the large and small grained specimens. This situation is schematically shown in Fig. 9. It is immediately evident from Fig. 9 that the large grained specimen has the larger V_{cr-q} . In fact, this has been rigorously proved⁽³⁾.

I would like to point out that the above mentioned fact is a consequence of the material properties; one of the two aspects of an impact test. The impact tests of iron and steel are examples of the cooperative interaction of material properties and wave propagation mechanics.

I would like to mention another contribution of material properties to the occurrence of the above mentioned phenomenon. In reference (2), we have reported that even the specimens annealed at lower temperatures to produce smaller grain size show good elongation in impact tests if they are quenched. This can be easily understood; quenching eliminates the sharp yielding and the Lüders elongation and lowers the yield stress.

Pseudo-brittle fracture does not occur solely in iron and mild steel. Some high-strength steels have the yield ratio of ~ 0.95 and the low work hardening rate. Thus, we should expect the identical fracture mode in such steels. In

Fig. 9 Stress-strain diagram for a fine grained iron specimen with high yield point and large yield elongation, and for a coarse grained specimen with lower yield point and smaller yield elongation.

Fig. 10 Changes of stress-strain curves of high-strength steel on static tensile tests with various strain rates and slow impact test.

fact, a high-strength steel with $0.09\% \text{C}$, $0.02\% \text{Si}$, $0.36\% \text{Mn}$, $0.12\% \text{P}$ and $0.014\% \text{S}$ has shown elongation less than 1% in an impact test of $42.3/\text{s}$. This steel has fine grains of $\sim 20\text{ }\mu\text{m}$ and a stress strain curve of Fig. 10. It should be noted that the elongation at the lower strain rate tests is $10\sim 18\%$ ⁽⁸⁾.

4. Pseudo-brittle fracture of Al

I would like to point out that one of the metallurgical factors relating to the occurrence of the pseudo-brittle fracture is work hardening. Low work hardening favors pseudo-brittle fracture. Thus, if we produce a situation where low work hardening is forced into effect we can observe such fracture even in static tests. Figure 11 manifests this⁽⁹⁾. Here, high purity Al specimens were prestrained at -196°C in dynamic tests of $41/\text{s}$ and were

Fig. 11 Stress-strain curves of static tensile test of aluminum specimen at R.T. after straining with dynamic loading at -196°C (—). For comparison the stress-strain curves of dynamic (—) and static (---) test at R.T. are shown.

followed by static tests at room temperature. Note that the elongation at the static tests is very small. This is due to the so-called work softening. The work softening is induced when specimens are strongly work hardened at lower temperatures and/or at higher strain rates and are deformed again at higher temperatures or at lower strain rates. The deformation due to the work hardening is localized and does not propagate.

When the amount of prestrain is small, the above mentioned pseudo-brittle fracture is not observed. This is because the work softening is not significant and the subsequent work hardening is large. Figure 12 demonstrates in Al single crystals that the larger prestrains induce the pseudo-brittle fracture⁽¹⁰⁾.

These observations give warning to the metal working processes of Al. If Al is highly strained at higher working rates, we should avoid the subsequent working with low strain rates in order to keep good workability.

Fig. 12 Behaviors of Al single crystals prestrained at -196°C statically and subsequently tested at room temperature statically.

IV. Summary

To demonstrate the two aspects of an

impact test, material property and wave propagation, I have shown the pseudo-brittle fracture observed in iron and mild steel in a dynamic test. The small grained specimens fracture with little elongation while the large grained ones show large elongation.

The almost zero elongation of the small grained specimens has been attributed to the fact that plastic deformation does not propagate but localizes, and the critical impact velocity, over which this type of fracture occurs, has been derived. It has been shown that the larger the grain size, the larger the critical impact velocity. The implication of the pseudo-brittle fracture has also been discussed in Al.

Acknowledgements

I am expressing my appreciation to many organizations and institutions that offered me opportunities to conduct research on dynamic tests of metals. Also, I vividly remember the pleasant happy days spent in research with my mentors and students, one of whom has given me editorial assistance on this occasion.

REFERENCES

- (1) M. Mashima: J. Appl. Phys. Japan, **5** (1936) 275.
- (2) S. Sakui and M. Ohmori: J. Iron Steel Inst. Japan, **45** (1959) 415.
- (3) S. Sakui and T. Mori: J. Japan Inst. Metals, **28** (1964) 537.
- (4) J. R. Low: *Relation of Properties to Microstructure*, ASM (1954).
- (5) T. von Karman and P. D. Duwez: J. Appl. Phys. **21** (1950) 987.
- (6) G. I. Taylor: J. Inst. Civil Engrs. (London), **26** (1946) 486.
- (7) A. H. Cottrell: Char. Mech. Engrs., **4** (1957) 448.
- (8) Unpublished results.
- (9) S. Sakui, T. Mori, T. Kakuma and K. Sato: J. Japan Inst. Metals, **30** (1966) 493.
- (10) S. Sakui and K. Sato: J. Japan Inst. Metals, **33** (1969) 1268.