Tensile strength of five metals and alloys in the nanosecond load duration range at normal and elevated temperatures

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Abstract

The paper presents results of measurements of the resistance to tensile fracture at spallation in nickel, cobalt, stainless steel, AlMg6\% alloy, and Inconel IN 738 LC alloy. In the experiments carried out with a high-power ion beam as a shock-wave generator the load pulse duration was in the range of 50\,ns. The measurements were performed at peak stresses varying by a factor up to 2 which had no influence on the dynamic tensile strength of the materials tested. For cobalt and Inconel measurements were also done at elevated temperatures. Whereas the response of cobalt was practically insensitive to temperature, IN 738 LC demonstrated a transition from viscous to relatively brittle fracture accompanied by a significant increase of the spall strength at higher temperatures. © 2001 Elsevier Science Ltd. All rights reserved.

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1. Introduction

To predict fracture and fragmentation at high-velocity impact, information about strength properties of materials is needed. The dynamic tensile strength at load durations in the sub-microsecond range is studied by analyzing spall phenomena under shock wave loading. Spalling is the process of internal rupture of a body due to tensile stresses generated as a result of a
compression pulse reflected from a free surface. Comprehensive reviews of different aspects of the spall fracture phenomena have been collected by Davison et al. [1]. Among numerous recent publications in this field we may mention also the papers [2,3] dealing with spall tests in the nanosecond load duration range. Instrumental measurements of resistance to dynamic fracture are based on the analysis of wave profiles recorded during the shock-wave loading of a sample [1,4].

In this paper we present results of measurements of the dynamic tensile strength for several metals and alloys using the pulsed high-power proton beam of the Karlsruhe Light Ion Facility KALIF [5] as a shock wave generator. The ablative pressure pulse generated in the 10 mm diameter focal spot of the 0.15 TW/cm² pulsed proton beam by direct interaction with solid matter drives compression waves with peak stresses of up to several tens of GPa and pulse durations of approximately 40–60 ns into the test specimens. The bell-shaped power density distribution along the cross-section of the KALIF beam results in a planar ablation pressure wave with transversely varying amplitude. The ratio of the peak stresses at the beam axis and 5 mm away from the axis reaches 2. However, since the incident stress gradient in axial direction exceeds the radial stress gradient by almost two orders of magnitude, the distortion of planar geometry may be assumed as negligible. An optically recording velocity interferometer system (ORVIS) type line-imaging Doppler-velocimeter [6] provided the capability to record free surface velocity histories simultaneously for many points along a measuring line. With this technique, it is possible to make several measurements at different points of the sample in one single shot and to observe spatial variations of strength and the influence of peak stress in the compression wave on the spall strength if these phenomena take place.

The experimental procedure was similar to that described in detail earlier [5]. Part of the tests were performed at elevated temperatures. The samples were heated by resistive heaters placed on the irradiated surface. The temperature was controlled by a thermocouple which was placed between the sample and its holder ~ 5 mm away from the sample center. The power of the resistive heater was sufficient to heat the sample to ~ 600°C within approximately 10 min.

2. Materials

The measurements were done for nickel and cobalt of 99.9% purity, for 4301 stainless steel, for AlMg6% alloy, and for Inconel IN 738 LC alloy. The samples of cobalt, nickel, AlMg6% alloy, and stainless steel were cut out of rolled sheets of corresponding thickness; the IN 738 LC samples have been cut out of a massive cast block. The sample thickness was in the range of 0.4 to 1.5 mm.

The composition of 4301 stainless steel is 18% in weight Cr, 10 wt% Ni, 2 wt% Mn, 1 wt% Si, and iron the rest. IN 738 LC is a nickel-based refractory superalloy with 16 wt% Cr, 8.5 wt% Co, 1.75 wt% Mo, 3.5 wt% Ti, 3.5 wt% Al, 3 wt% W, and some small amounts of Ta, W, Nb, Zr, C, and B. The measured density of Inconel was 8.30 g/cm³.

IN 738 LC has been designed for use as a material of industrial gas turbine blades at working temperature up to 90% of the melting temperature. Because of this, measurements with IN 738 LC were done both at room temperature and at elevated temperatures. Experiments with cobalt also have been done at room temperature and at elevated temperature of 400–450°C. It was expected that the h.c.p. → f.c.p. polymorphous transition in cobalt may exhibit itself in the free surface velocity history. It is known [7] the temperature of the transition, which is 450°C at zero pressure,
grows with increasing pressure at a rate of \(60\text{–}90 \text{K/GPa}\), so under tension the transition may occur at a lower temperature.

3. Experimental data

Fig. 1 shows an example of a line-imaging ORVIS interferogram obtained by recording the fringe shift with a streak camera. The experiment was done with a 1.5 mm thick sample of 4301 stainless steel. In this shot recorded peak stresses varied from 11 GPa near the beam axis down to 6.2 GPa at the beam edge. The gradient of peak stresses along the ~ 4 mm long measuring line on the sample surface results in different propagation velocities of the shock wave that explains the tilt of the shock wave front relative to the sample plane. The radial stress gradient should not cause problems for the measurements since the resulting tilt of the spall plane relative to the surface of ~ 1 degree is small enough to approximately meet the assumption of planar geometry used in the data evaluation. With stress wave lengths of ~ 0.3 mm, the axial stress gradient of ~ 30 GPa/mm in the unloading wave much exceeds the radial stress gradient of ~ 1 GPa/mm. In view of spall thicknesses of ~ 0.1 mm a contribution of the radial gradient is considered practically negligible.

The deflection of the interference fringes from the horizontal line (upwards shift means velocity increase in this illustration) is proportional to the local velocity. But the ordinate also represents a space co-ordinate. In order to obtain the exact local velocity history the intensity variation along a corresponding horizontal line \(y = \text{const.}\) in the interferogram has to be analyzed accounting for the direction of shift of the fringes. Because the intensity distribution is influenced by the target surface, the speckle structure of the interference pattern, and local sensitivity variations as well as electronic noise of the streak camera, this straight forward analysis does not provide accurate velocity data. This is why we determined the velocity history, \(u(t)\), of a fixed point on the sample surface (that corresponds to fixed value \(y\) in the interferogram) by the linear approximation

\[
u(t) = n \times u_0 \times u_0 \times \Delta y(t)/\Delta y_r(t)
\]

(1)
where \( u_0 \) is the velocity-per-fringe constant of the interferometer or the velocity increment corresponding to a local interference phase shift by one period; \( n \) is the number of fringes crossing a horizontal line of a given \( y \); \( \Delta y \) is the deflection of the fringe closest to the given \( y \) in the direction of the velocity increase; \( \Delta y_f \) is the distance between two neighboring fringes enclosing the given \( y \) in the interferogram. Note that in the case of line-imaging interferometry the inter-fringe distance \( \Delta y_f \) is not constant but depends on \( u_0 \) and the velocity histories along the measuring line. In particular, the inter-fringe distance becomes smaller than the initial value if there is a velocity gradient in the direction opposite to the fringe deflection. Obviously, Eq. (1) may be used for interpreting the line-imaging interferograms when the velocity gradients \( du/dy \) are constant or almost constant. In particular, this condition is not satisfied when the compression wave front is tilted relative to the sample surface plane. In order to decrease the error caused by varying velocity gradients as a result of a temporal shift of the local velocity histories, we used in Eq. (1) the values of \( \Delta y \) and \( \Delta y_f \) along directions of smallest local gradients in the interferogram. In our treatment the lines of smallest gradients were constructed parallel to the \( y_s(t) \) trajectory in the interferogram that describes the arrival of the tilted shock wave front on the sample surface.

The velocity histories \( u_{s_i}(t) \) evaluated from experimental interferograms are shown in Figs. 2–4. We made 1 shot for each material. Every shot provided several measurements of spall strength at different peak stresses. However, for simplicity the figures displays only one typical velocity history for each material.

The wave profiles in Figs. 2–4 show the elastic–plastic compression waves followed by unloading waves. Tensile stresses developing in the body after reflection of the compression pulse by the free surface increase with the distance from the surface in the initial stage of reflection. When the peak tensile stress reaches a critical magnitude, nucleation and growth of fracture are initiated. While

![Fig. 2. Free surface velocity profile of a 1.5-mm-thick sample of 4301 stainless steel (the original interferogram of this shot is shown in Fig. 1) and the velocity history at the interface with a water window of a 0.4-mm-thick Ni sample. The measurements on Ni were done at the proton beam axis whereas with 4301 samples the velocity history represents the wave process ~ 2.5 mm away of the axis. Larger thickness of the steel sample resulted in a greater decay of the shock pulse.](image-url)
fracture develops, the tensile stresses relax to zero. As a result, a compressive disturbance called “spall pulse” is produced in the free surface velocity profile. Thereafter, a wave reverberation is observed in some cases within the spalled layer. The period of the velocity oscillation is a measure of the thickness of the spall plate. The velocity pullback from the peak value to the value right ahead of the spall pulse, $\Delta u_{fs}$, is a measure of the incipient fracture strength of the material. In these experiments we did not either observe any notable spatial variation of the spall fracture nor any influence of the peak shock stress on the strength.

The materials tested demonstrate quite different fracture behavior. Note that the spall signal should be steep when the fracture proceeds very rapidly, whereas in the case of relatively slow fracture the spall signal should become less steep and, in principle, may appear even as just a small

Fig. 3. Experimental data for cobalt at normal and elevated temperatures.

Fig. 4. Free surface velocity profiles measured on IN 738 LC at normal and elevated temperatures.
deflection from the extrapolation of the preceding part of the $u_p(t)$ profile. If the fracture accelerates the spall signal has an increasing steepness that appears as a smooth minimum in the free surface velocity profile instead of a distinct kink. In this sense, cobalt and nickel demonstrate a brittle-like rapidly accelerating fracture behavior with a quite distinct spall signal. 4301 stainless steel and, especially, IN 738 LC alloy exhibit at room temperature a very viscous fracture behavior: the spalled layer remains bonded with the main body of the sample which causes a deceleration of the spall plate during relatively long time after appearance of the spall signal. In the shot with the nickel sample the deceleration is caused by the reverse pressure of the water window. Spall fracture of cobalt was largely completed within 20 ns; after that the deceleration of the spall plate was practically stopped.

Within the experimental accuracy we did not record any peculiarities in the wave structure for cobalt which could be associated with the expected polymorphous transformation under tension. The behavior of IN 738 LC changed radically with temperature increase: instead of viscous fracture observed at room temperature the free surface velocity history at $586^\circ C$ demonstrates a relatively rapid and distinctly expressed spall fracture process. The increase of the velocity pullback amplitude marks a larger tensile strength at elevated temperature. The free surface velocity history of the AlMg6% sample is similar to that of cobalt and does not contain any peculiarities.

4. Data evaluation and results

In order to calculate the fracture stress the peak free surface velocity, $u_{pk}$, and the minimum free surface velocity just ahead of the spall pulse, $u_m$, are determined directly from the free surface velocity profile. The tensile stress value just before spalling, $\sigma^*$, is then determined from an analysis invoking the method of characteristics. From the acoustic approach, the following linear approximation [4]:

$$\sigma^* = \frac{1}{2} \rho_0 c_0 \Delta u_{ts}$$

(2)

is used, where $\Delta u_{ts} = u_{pk} - u_m$, $c_0$ is the sound velocity and $\rho_0$ is the initial density. Nonlinear compressibility of matter has to be accounted for in the stress range realized in our experiments. This can be easily done by extrapolating the Hugoniot in the pressure-particle velocity plane down to the negative pressure region. However, some problem with the sound velocity arises for elastic-plastic solids. In the case of a one-dimensional process, compressive and rarefaction waves propagate with longitudinal sound velocity, $c_l = \sqrt{(K + 4G/3)/\rho}$, if the deformation is elastic, and with bulk sound velocity $c_b = \sqrt{K/\rho}$, $c_b < c_l$ in the plastic deformation region [8]. Here $K$ and $G$ are the bulk modulus and shear modulus, respectively, and $\rho$ is the density. We have to decide which sound velocity to use in Eq. (2) when calculating the tensile stress in the spall plane. Since the spall pulse is a compression wave which propagates through the stretched material, the spall pulse front should propagate at longitudinal elastic wave velocity $c_l$ whereas the incident rarefaction plastic wave ahead of it propagates at bulk sound velocity $c_b$. Since the spall pulse front is considered an elastic wave, it was concluded [9] that the relationship by which to calculate the
Table 1
Results of measurements of dynamic tensile strength

<table>
<thead>
<tr>
<th>Material</th>
<th>(\dot{V}/V_0) (s(^{-1}))</th>
<th>(h_s) (μm)</th>
<th>(\Delta u_{fs}) (m/s)</th>
<th>(\sigma^*, \text{Eq. (2)},) (GPa)</th>
<th>(\sigma^*, \text{Eq. (3)},) (GPa)</th>
</tr>
</thead>
<tbody>
<tr>
<td>Ni</td>
<td>(2 \times 10^6)</td>
<td>75</td>
<td>430 ± 20(^a)</td>
<td>5.9</td>
<td>6.6</td>
</tr>
<tr>
<td>Steel 4301</td>
<td>(5 \times 10^5)</td>
<td>150</td>
<td>230 ± 10(^a)</td>
<td>4.0</td>
<td>4.45</td>
</tr>
<tr>
<td>AlMg6%</td>
<td>(1.2 \times 10^6)</td>
<td>66</td>
<td>270 ± 10(^a)</td>
<td>1.9</td>
<td>2.06</td>
</tr>
<tr>
<td>Co, 20°C</td>
<td>(1 \times 10^6)</td>
<td>78 ± 4</td>
<td>260 ± 10(^a)</td>
<td>5.2</td>
<td>5.75</td>
</tr>
<tr>
<td>Co, 400°C</td>
<td>(1 \times 10^6)</td>
<td>85 ± 4</td>
<td>240 ± 10(^a)</td>
<td>4.85</td>
<td>5.35</td>
</tr>
<tr>
<td>Co, 450°C</td>
<td>(1 \times 10^6)</td>
<td>78 ± 4</td>
<td>250 ± 15(^a)</td>
<td>5.05</td>
<td>5.55</td>
</tr>
<tr>
<td>IN 738 LC, 20°C</td>
<td>(6 \times 10^5)</td>
<td>110</td>
<td>220 ± 20(^a)</td>
<td>4.0</td>
<td>4.5</td>
</tr>
<tr>
<td>IN 738 LC, 586°C</td>
<td>(1.2 \times 10^6)</td>
<td>90 ± 4</td>
<td>315 ± 10(^a)</td>
<td>5.75</td>
<td>6.4</td>
</tr>
</tbody>
</table>

\(^a\)Velocity pullback at the interface between sample and water window.

The evaluated spall strength data are summarized in Table 1. The results presented are average values from 4 to 5 measurements at different points on the sample and at different peak stresses. The total scatter of the spall strength, including both inaccuracy of the measurements and the strength variations, in most tests did not exceed ±5%. In the case of Inconel it is difficult to determine accurately the value of the velocity ahead of the spall signal which causes a larger scatter of \(\Delta u_{fs}\). The spall strength data obtained applying both Eqs. (2) and (3) are presented. To account for nonlinear compressibility of the materials, we used the extrapolation of the Hugoniot in the pressure-particle velocity plane into the negative pressure region. For nickel and cobalt we used the Hugoniots and elastic modules given in Ref. [12]. The evaluation of 4301 steel data was performed using Hugoniot and elastic modulus of stainless steel 304. The Hugoniot of IN 738 LC being not known, we used \(U_s = 4.58 + 1.47u_p\) as a relationship between shock velocity \(U_s\) and particle velocity \(u_p\) which is an average between Hugoniots of nickel and stainless steel which actually are very close to each other. In order to characterize the load rate conditions, the table presents the value of fracture stress \(\sigma^*\) is

\[
\sigma^* = \rho_0 c_b \Delta u_{fs} \frac{1}{1 + c_b/c_1}.
\] (3)

This relationship does not account for the spall plate thickness. However, the distortion of the wave profile obviously should increase with increasing distance as a result of different propagation velocities. An attempt to introduce an additional correction into Eq. (3) has not been sufficiently proved [10]. A more detailed analysis of the interaction of the elastic compression wave with the plastic rarefaction wave, both of which propagate in the same direction, has shown that if spall fracture occurs as a result of the reflection of a triangular compression pulse the spall pulse front propagates at velocity \(c_1\) independent of its gradient [11]. In this case the spall strength may be calculated using the approach given by Eq. (3). In the more common case of trapezoidal compression pulse some leading characteristics of the spall pulse will disappear with time, the propagation velocity of the spall pulse front is less than \(c_1\), and the evaluation of the fracture stress becomes somewhat more complicated.
decompression strain rates $\dot{V}/V_0 = -\dot{u}_1/c_b$ where $\dot{u}_1$ is the free surface velocity derivative in the unloading part of the incident shock compression pulse ahead of the spall signal, and the thickness of the spall layer $h_s$.

5. Discussion

The experimental velocity profiles for cobalt, Inconel and nickel demonstrate no evidence of an elastic front of the spall pulse. In these materials, the period of the velocity oscillations after spall practically coincides with the duration of first velocity pulse. In case of a distinct elastic–plastic response, the period of the oscillations should be $t_2 = 2h_s/c_1$, whereas the duration of the first velocity pulse is $t_1 = h_s/c_b + h_s/c_1 > t_2$ where $h_s$ is the thickness of the spall plate. In other words, the recorded spall pulse front obviously indeed propagates with bulk sound velocity. Therefore, spall strength data calculated using relationship (2) are supposed to be better suited for these materials. For steel and AlMg6% the data in the last column of the table apply because the free surface velocity profiles for these materials clearly show different durations $t_1$ and $t_2$, their ratio being close to the expected value.

The data for nickel and AlMg6% alloy may be compared with spall strength measurements at lower strain rates. Fig. 5 summarizes the results presented in Ref. [13] and the results of this study as a function of the decompression strain rate. In general, the materials exhibit a similar dependence of the resistance to rupture on the strain rate: the spall strength in the nanosecond load duration range is by a factor of $\sim 3$ higher than in the microsecond range.

Experiments with cobalt did not demonstrate any essential influence neither of the temperature nor of the temperature-induced polymorphous transformation, on the dynamic tensile strength. In contrast, the dynamic tensile strength of IN 738 LC increased by 30% at high temperature. Moreover, the free surface velocity profiles of this material show a transition from viscous to

![Fig. 5. The spall strength of nickel and AlMg6% as a function of the decompression strain rate. The results of this work are two points at ultimate strain rates. The other data were published earlier [13].](image-url)
relatively brittle fracture at elevated temperature. While at room temperature the alloy demonstrates a very viscous fracture behavior and the scabbed layer remains bonded with the bulk of the sample during relatively long time, the time of debonding at 586°C obviously did not exceed 15 ns. Such evolution of the shape of the spall signal could be explained in terms of a relationship between fracture time and load time [1]. However, the experiment at elevated temperature was done with even shorter load duration than the room temperature experiment. At each test temperature and with peak stresses varying by a factor of 1.5 the free surface velocity histories look very similar. We believe this distinct temperature dependence of the fracture process to be another manifestation of the well known complex behavior of Ni-base superalloys at mid-range temperatures [14].

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