Use of High-Current Nanosecond Relativistic Electron Beam as a Shock-Wave Generator for Investigation of High Strain Rate and Spall Fracture of Hadfield Steel


Abstract – The high-current nanosecond relativistic electron beam (20 kA, 45 ns, 1.35 MeV, $3.4 \cdot 10^{10}$ W/cm$^2$) formed by accelerator “SINUS-7” was used as a shock-wave generator for investigation of the high-rate deformation and spall fracture of Hadfield steel. According to the numerical simulation performed for stainless steel, the peak pressure and duration of shock wave are 31 GPa and ~ 0.2 μs, respectively, and strain rate is ~ $2 \cdot 10^6$ s$^{-1}$. It was found that spallation is carried out by mixed ductile–brittle intergranular fracture, while in the quasi-static tensile and Charpy impact tests the ductile transgranular fracture was observed. It was shown that the cause of intergranular spallation is the strain localization near the grain boundaries, containing carbide precipitations.

1. Introduction

Hadfield manganese austenitic steel, due to high strain hardening capacity, is an attractive candidate for study of strain and fracture behavior of fcc metals and alloys. Most of the experiments were carried out at quasi-static loading. Effect of dynamic loading was studied in Ref. [1], where the samples were subjected to shock compression at strain rates of $\dot{\varepsilon} = (1+8) \cdot 10^3$ s$^{-1}$. Of particular interest is the study of those phenomena at $\dot{\varepsilon} \geq 10^5$ s$^{-1}$ realized under shock-wave loading in the modes of rear spall [2]. The aim of this paper was to investigate the features of the strain and spall fracture phenomena in the polycrystalline Hadfield steel under shock-wave loading at $\dot{\varepsilon} \geq 10^6$ s$^{-1}$ and to compare with the quasi-static tensile and Charpy impact tests.

2. Experimental

Shock-wave experiments were carried out using the high-current electron accelerator SINUS-7. The electron beam was generated in a vacuum diode with a hemispherical explosive emission cathode of 6-mm diameter (stainless steel 304) placed into external longitudinal magnetic field of 17 kOe. At peak voltage of 1.35 MeV and 7-mm cathode to anode distance, the peak diode current reached 20 kA. The target was placed behind the grounded graphite aperture of 8-mm diameter and 3-mm thick. The rear surface of the target was free. The maximum current density at the target was 25 kA/cm$^2$, which corresponds to peak power density of $3.4 \cdot 10^{10}$ W/cm$^2$; the pulse duration was 45 ns (FWHM). The beam current density distribution on the target was close to uniform one.

The samples of 15 mm diameter and thickness in the range of 1.5÷4 mm were used as targets. The samples were cut from a Hadfield steel blank, previously subjected to hot forging, solution treatment at 1050 °C (2 h) followed by quenching into water. The resulting microstructure was fully austenitic with average grain size of 118 μm, the microhardness of 2.7 GPa, yield stress $\sigma_y = 410$ MPa, tensile stress $\sigma_T = 940$ MPa, elongation $\delta = 33\%$.

3. Results and discussion

Due to the volume character of beam energy deposition (electron penetration depth for Fe is ~ 1.2 mm), a single e-beam pulse led to the ablation and formation of the hole of 0.6+0.8 mm depth and 6+7 diameter at the target front surface. The ablative pressure pulse generated a quasi-plane compressive shock wave, which propagated with the attenuation into the depth and reached a back surface. Due to reflection from a back surface, the tensile shock wave was formed, which led to the rear spall.

Simulation of the shock wave was carried out using a mathematical model that took into account the dynamics of dislocations [3]. The 304 stainless steel was used in simulation because it is very similar to Hadfield steel in physical properties. Fig. 1, a shows the evolution of the shock wave profile during its propagation into the target. Near the front surface the amplitude and duration of the shock wave are ~ 31 GPa and ~ 0.2 μs, respectively. At the beginning, the shock front shortens, which accompanies by increasing strain rate from ~ $4 \cdot 10^5$ s$^{-1}$ for 100 ns to...
9 \times 10^6 \text{s}^{-1} \text{ for } 0.6 \mu \text{s}. As a shock wave propagates into the target, its front spreads out and strain rate drops to \( \sim 2 \times 10^6 \text{s}^{-1} \text{ for } 1.6 \mu \text{s}. \) Fig. 1, \( b \) shows the evolution of a reflected shock wave profile in time (0.9÷1.1 \mu) for a target of 4 mm in thickness. Under reflected shock wave loading the strain rate decreases from \( 2 \times 10^6 \) to \( 10^6 \text{s}^{-1} \) for the same time.

The experiments showed that the rear spall is realized for all of the samples. The elongation of spall layer, estimated from the arc length of spall cup, is \( 10\div20\% \), i.e., by a factor of 1.7÷3.5 less than that under quasi-static tensile loading.

With increase in the target thickness from 1.5 to 4 mm the thickness of spall layer increases almost linearly from 0.4 to 0.6 mm. Such dependence is typical for spall fracture of metal targets under shock wave loading using nanosecond laser [4], ion [5] and electron [6, 7] beams. This is due to the fact that with increase in target thickness, the shock wave amplitude near the back surface decreases due to attenuation; thus, the distance, at which the amplitude of the reflected wave reaches the dynamic strength (spall strength) increases.

Figure 2 shows a typical microhardness profile in shock-loaded sample in range from a bottom of the ablation hole to a spall surface. It can be seen that the shock-wave loading leads to strain hardening, likely at the quasi-static loading. All of the microhardness profiles have a maximum located at a depth of \( \sim 0.7 \text{ mm} \) for targets of 2.5÷4 mm in thickness. The maximum microhardness is \( 3.5 \div 4.7 \text{ GPa} \), which is close to that for the quasi-static loading. Monotonic decrease in the microhardness on the right of the maximum agrees with an attenuation of a shock wave during its propagation into a target (Fig. 1, \( a \)). The softening near the front surface is associated with intense annealing of the shock-loaded material in the heat affected zone.

Fig. 2. Microhardness depth profile (Vickers) of the 4-mm-thick shock-loaded Hadfield steel sample

Fig. 3 shows the SEM cross-section images of shock-loaded samples. Judging by the weak intensity of traces of plastic deformation, its degree is much lower than that in the quasi-static tensile and Charpy impact tests. The strain is extremely non-uniform within a single grain, as well as from grain to grain.

Fig. 3. SEM cross-section images of shock-loaded Hadfield steel samples. In (a) the spall fracture surface is at the top. In (b) the arrow shows the longitudinal microcrack in the grain-boundary carbide

Metallographic examinations of shock-loaded samples indicated that in the grains with appropriate orientation and size of \( \delta \leq 80 \mu \text{m} \), the strain within the whole grain volume usually takes place. In larger
grains, the strain localization near the grain boundaries, located on a certain angle to the sample axis, takes place (Fig. 3, b). Obviously, the strain non-uniformity is associated with different lattice orientation in each grain.

As follows from Fig. 3, a, the spallation of shock-loaded samples is realized by intergranular fracture, in contrast to transgranular fracture that observed under quasi-static tensile loading (\(\dot{\varepsilon} = 1.4 \cdot 10^{-3} \text{s}^{-1}\)) and Charpy impact tests (\(\dot{\varepsilon} \sim 10^2 \text{s}^{-1}\)). The intergranular spall fracture is associated with presence of network of carbide precipitates (M₃C) along the grain boundaries as well as with the strain localization near these boundaries. Primary microvoids are formed near the interface between the carbide precipitate and \(\gamma\)-phase grain in sites of strain localization. Hereinafter, microvoids grow and coalescence along the carbide layer, that results in extended microcracks at the grain boundary. Sometimes a brittle fracture of carbide layer is observed (Fig. 3, b).

Figure 4 shows the SEM images of the fracture surfaces of the Hadfield steel. Fig. 4, a presents the polyhedral form of several neighboring grains separated by grain-boundary cracks. At the macroscopic level, the average polyhedron size is equal to average grain size in the initial condition. This is additional evidence of intergranular mode of spall fracture. At the microscopic level the fracture is either ductile (Fig. 4, b, c) or mixed ductile-brittle (Figs. 4, d).

Three types of spall fractographs are observed. The most common type has a periodic stripe-like structure showing in Fig. 4, b, that is associated with the ductile shear. The second type showing in Fig. 4, c presents the typical ductile dimple fracture. The average dimple size is \(~ 1 \mu\text{m}\), that is three times less than that in the quasi-static ductile fracture. Inside the dimples a globular second phase particles (supposedly, M₃C carbide) are observed.

Both types of ductile fracture are observed mainly in the central region of the spall cup. Supposedly, they arise near the grain boundaries due to a ductile fracture of \(\gamma\)-phase under shear and normal stresses, respectively. The third type of the fractographs is typical for a brittle fracture (Fig. 4, d). This type is observed at the periphery, and its amount decreases rapidly from the edge toward the center of the spall cup. The cleavage steps most likely are formed as a result of brittle fracture (cleavage) of the carbide precipitate. Thus, the spall fracture of Hadfield steel is realized in an intergranular mixed ductile-brittle manner, but more in a ductile manner, in contrast to transgranular ductile fracture observed under quasi-static tensile as well as impact loading.

The change in the mechanisms of deformation and fracture under a shock wave loading compared to a quasi-static loading can be explained as follows. In Ref. [8] using Ni samples was shown, that at strain rate in range of \(10^3\sim10^4 \text{s}^{-1}\), the dramatic increase in dynamic strength and decrease in ductility are observed. Similar results were obtained for Cu [9] and Hadfield steel [1] samples.

![Fig. 4. SEM fractographs of spall fracture of Hadfield steel samples. Second-phase particles are shown by arrows in (c). Cleavage steps are indicated by arrows in (d).](image)

A theoretical model, which describes adequately the change in the strain behavior of fcc metals at increasing a strain rate is proposed in Ref. [10]. Accord-
ing to this model, in range of $\dot{\varepsilon} \sim 10^3$ to $10^4$ $s^{-1}$ the mechanism of strain changes from a thermally acti-
vated dislocation glide, which is typical for slow load-
ing, to an unstable plastic flow at $\dot{\varepsilon} \geq 10^4$ $s^{-1}$ due to the
accelerated generation of dislocations.

Regardless of the strain rate, the plastic flow starts
from the grain boundaries that are stress raisers. The
size of region of strain localization depends on the
dislocation speed and loading time $\tau$. Assuming, ac-
cording to Ref. [11], that the maximum dislocation
speed in fcc metals (Cu, Al) is $10\cdot 10^3$ m/s, we find
that in quasi-static ($\tau = 2.3 \cdot 10^2$ s) and impact
($\tau \sim 10^{-3}$ s) loading, the strain is developed within
the whole grain volume, that is observed experimentally.
It is a reason of the transgranular character of the frac-
ture in these modes of loading.

In the case of shock-wave loading of the target of
4 mm in thickness (see Fig. 1, a), the shock wave am-
plitude (31 GPa) drops several times near the back
surface due to strong attenuation, and its duration in-
creases from $1.7 \cdot 10^{-7}$ to $3.4 \cdot 10^{-7}$ s. It follows that
the size of the strain localization region is $\sim 40$ $\mu$m,
that agrees with the characteristic length of plastic
deforination traces (Fig. 3, b). This means that if grain
has a favorable orientation and its size $d \leq 80$ $\mu$m, the
deforination traces will cross the whole grain. For
larger grains the strain will be localized near the grain
boundaries. These estimates are in qualitative agree-
ment with an experiment.

The dramatic decrease in the strain localization size is associated with the fact that the accelerated
generation of dislocations near the grain boundaries
occurs under conditions of its limited outflow in the
grain volume due to a short-term loading. The growth
of the density of twins [1] also contributes to the ac-
cumulation of dislocations near grain boundaries.

The strain localization results in the nucleation of
microcracks, its growth and coalescence along grain
boundaries in a shock-loaded samples (Fig. 3). This
ultimately leads to intergranular spall fracture. The
targets fractured in a mixed ductile-brittle manner
(Fig. 4). The decreasing a dimple size compared to
quasi-static loading agrees with the change of the
mechanism of plastic flow at high strain rates men-
tioned above. The presence of cleavage steps in the
fractographs (Fig. 4, d) is associated with a brittle
fracture of the carbide precipitates. Consequently,
effect of the brittle fracture on the spall strength is
determined by the purity of the material and micro-
structure of the second phase precipitates located at the
grain boundaries.

As it follows from the simulation of shock wave
dynamics (Fig. 1, b), the spall strength of Hadfield steel
is $\sim 8$ GPa. This agrees with the experimental
data for Ni and high-strength alloy targets subjected to
shock wave loading at strain rate of $\dot{\varepsilon} \geq 10^6$ $s^{-1}$ [2, 5].

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