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Microstructural changes of high-strength steel Fe-26Mn-10Al-1.2Si-2.2Cr-1.8Ni-0.15V-1C after ballistic damage

Voron M. M.[1], Candidate of Technical Sciences, Senior Researcher, Deputy Head of the Department, mykhailo.m.voron@gmail.com, ORCID: https://orcid.org/0000-0002-0804-9496

Tymoshenko A. M.[1], Candidate of Technical Sciences, Senior Researcher, Marschal@i.ua, ORCID: https://orcid.org/0000-0003-4038-1744

Semenko A. Yu.[1], Candidate of Technical Sciences, Senior Researcher, semenko.au@gmail.com, ORCID: https://orcid.org/0000-0002-0448-1636

Smirnov O. M.[1], Doctor of Technical Sciences, Professor, Head of the Department, stalevoz@i.ua, ORCID: https://orcid.org/0000-0001-5247-3908

Schwab S. L.[2], Candidate of Technical Sciences, Senior Researcher, serg.schwab@gmail.com, ORCID: https://orcid.org/0000-0002-4627-9786

Skorobagatko Yu. P.[1], Candidate of Technical Sciences, Senior Researcher, yulka.ukr@gmail.com, ORCID: https://orcid.org/0000-0002-1724-9895

[1]Physico-Technological Institute of Metals and Alloys of the National Academy of Sciences of Ukraine, Kyiv, Ukraine

[2]E. O. Paton Electric Welding Institute of the National Academy of Sciences of Ukraine, Kyiv, Ukraine

The manuscript is devoted to the study of structural changes in the penetration zone of Fe-26Mn-10Al-1.2Si-2.2Cr-1.8Ni-0.15V-1C steel, which was developed for ballistic damage protection. Obtaining thin plates from lightweight high-strength steels based on the Fe-Mn-Al-C system for such use is an extremely promising and unexplored task. The work analyzes the influence of mechanical properties and mechanisms of microstructural changes, according to which armor metallic materials and studied type steels can resist bullet damage most effectively. To provide the research, as-cast steel billet was obtained under conditions of open induction melting, and then it was subsequently subjected to hot plastic deformation and normalization. The resulting plate was ballistic tested and the structure of the near-bullet hollow area was investigated. Studies of mechanical properties revealed extremely high strength of steel, the required level of plasticity and average hardness: UTS=2240 MPa, YS=1784 MPa, δ=6.2 %, ψ=16.8 %, HRC=42-46.

Studies of the bullet hollow cross-section area showed the formation of two zones. The first zone repeated the shape of the bullet and demonstrated significant local plastic deformation of the steel near the strike surface. In the near-surface region, local recrystallization and a microhardness decrease were observed due to thermal influence. When moving closer to the central zone, surface *hardening is observed, which is expressed in a noticeable increase in microhardness. The second zone shows signs of brittle fracture. In its microstructure, adiabatic shear bands and places of stress accumulation that form microcracks are observed. In the zone of l plate failure, the formation of microdefects coalescence areas was detected, the shape of which repeats the load distribution inside the material.*

The obtained data indicate that the proposed steel is suitable for protection against ballistic damage, but at greater thicknesses. It is able to resist ballistic loads due to the effective absorption of impact energy, which is expressed in the implementation of several mechanisms simultaneously - local hardening, formation of ASB, accumulation of defects and stresses and the sequential formation of microcracks, as well as in their coalescence during the deformation of the material.

***Keywords**: Fe-Mn-Al-C, armor plate, ballistic damage, microstructure changes, microdefects, load sharing.*

Nowadays, there are many different materials for armor protection. These include ceramic and metal-ceramic compositions, non-ferrous light alloys, bimetallic products and some others, but steels remain the most widely used. For the production of steel armor, high-strength and ultra-high-strength special steels, with high hardness, plasticity and toughness are used most often [1].

Hardness determines the ability of the armor material to stop and destroy striking elements and their components, and the toughness of the material ensures the absorption of the energy of the blast wave and the energy of striking elements. The tensile strength for steel armor is a more significant value than the yield strength, according to data [2]. This is due to the almost instantaneous effect of a ballistic impact, which causes a local increase in temperature in the place of contact of the striking element with the armor due to adiabatic heating caused by a high rate of plastic deformation. Another important factor is the impossibility of uniform distribution of loads on the material, that can be balanced with high strength.

The development of lightweight and highly effective armor materials is critical for reducing the weight of protective structures and increasing the mobility of equipment. Fe–Mn–Al–C steels demonstrate a reduced density compared to conventional armor steels due to the high Al content. At the same time, the high Mn and carbon content provide high strength and impact toughness due to complex deformation mechanisms (TWIP – twinning-induced plasticity, TRIP – transformation-induced plasticity) and the κ-carbides precipitation [3, 4]. Because of this, Fe–Mn–Al–C steels are considered as a promising alternative to traditional armor materials. However, the behavior of these steels under real impact loads is likely to differ from the statically tested characteristics and requires the analysis of dynamic microstructural transformations. For their correct assessment, it is necessary to consider the features of the structural and phase structure of the studied steels and the mechanisms of their change under the influence of loads.

In Fe–(15-30)Mn–(6-12)Al–(1-1.5)C (% wt.) steels, an austenitic matrix is usually formed at high temperatures, but after heat treatment it can contain many phases, among which it is important to note those that have the greatest influence on strengthening: nanosized κ-carbides $(Fe, Mn)_3AlC$ with a cubic lattice $L1_2$, ordered B2-phases or other intermetallic particles at a certain content of nickel and aluminum, as well as transition metal carbides [5-7]. Aluminum in Fe–Mn–Al–C steels not only reduces the density, but also promotes the formation of κ-carbides and the B2-phase. Manganese stabilizes austenite and changes the stacking-fault energy (SFE), which determines whether the twinning (TWIP) or martensite transformation (TRIP) mechanism will dominate. Carbon significantly reduces SFE and increases strength both by alloying austenite and by increasing the precipitation of strengthening κ-carbides [3, 8].

Low or moderate SFE values in Fe–Mn–Al–C steels stimulate the formation of twins during their deformation. If the SFE is further reduced, for example by increasing the carbon or nickel content, the $\gamma \rightarrow \varepsilon/\alpha'$ transformation (ε-martensite or α'-martensite) is possible, which will indicate the manifestation of the TRIP effect. In Fe–Mn–Al–C steels, the balance between these mechanisms depends on the ratio of the content of the main alloying elements and on the heat treatment, which will determine the structural-phase parameters [4].

Under impact loading (strain rates of ~10^3-10^6 s^{-1}), phenomena appear that are not observed in static tests – significant local heating, the formation of adiabatic shear bands (ASB), dynamic recrystallization in the ASB zone, local phase transformations (for example, dynamic austenite → martensite transformation or reverse reactions under the influence of temperature) and an increased probability of local brittle fracture in places of accumulation of strengthening phases. The combination of these phenomena significantly determines the resistance to bullet penetration [9].

Adiabatic shear bands can be formed under conditions of a large, rapidly increasing local shear and are accompanied by significant internal heating, which can lead to local softening of the material. In TWIP steels, ASB is often the dominant cause of breakdown during impact point deformation. The presence of κ-carbides in the matrix changes the local stress distribution and can either enhance or weaken the tendency to form ASB depending on the size and density of such particles. For example, small nanoscale inclusions can increase the resistance of steel to ultrafast impact loading by initiating dislocation scattering, while relatively large inclusions will create local stress concentrations and promote crack formation [6, 9]. Initially, the bullet energy is absorbed by the armor element, in our case Fe–Mn–Al–C steel, mainly through plastic deformation, and partly through the formation of ASB. When local critical stresses are reached, microcracks appear either in the matrix/precipitated phase interface zone, or in the zone of significant deformation gradient. So, to provide investigations of described problem, Fe-26Mn-10Al-1.2Si-2.2Cr-1.8Ni-0.15V-1C (% wt.) steel was prepared by using open induction furnace. C45-steel with 0.45 % wt. carbon content; Al-12Si aluminum alloy scrap; Al-10V master-alloy; metallic Mn, Al, Cr, Ni and V of ordinary purity were used as a charge materials. Graphite powder served as an additional source of carbon.

Alloy preparation was carried out in a 40-kW open induction furnace. Its heating zone was made of Al_2O_3:MgO=80:20 mixture. The charge materials were melted in the furnace gradually. After the melt reached the required temperature 1400-1500 °C, it was mechanically stirred and the slag was removed. After melting all the components and reaching a temperature of 1500 °C, the liquid metal was purged with argon three times for 10 seconds each and the remaining slag was removed. About 8kg of melt was poured from 1450 °C to a rectangle flat graphite mold. As a result, a plate-like casting with a 14 mm wall thickness was obtained.

As-cast billet was intended for hot plastic deformation. First, it was subjected to homogenization annealing at 1050 °C for 90 minutes in a resistance furnace. After that, the billet was deformed in a rolling mill with a compression rate 1 mm in one pass. After every two passes, the workpiece was returned to the furnace, reheated to the annealing temperature, held for 10 min, and the deformation was repeated. Finished plate-like part had thickness 10 mm and was air cooled.

The samples for mechanical properties determination were cut out from plate part, close to peripheral area. Plate was ballistically tested by shooting 5.45x39 mm enhanced impact bullet on speed 910 m/s.

Bullet hollow was cross-cut to be investigated. Sample was prepared by using standard grinding and polishing methods with subsequent etching in a 3 % alcohol solution of nitric acid. Metallographic analysis, determination of structural-phase components local chemical composition and the distribution of elements (EDX-method) were performed by using a Tescan Vega 3 electron microscope with a Bruker EDX analyzer. Microhardness test was provided by using LHVS-1000Z testing machine with load 100 g. Tensile test was provided on UTM-100 testing machine. It showed outstanding mechanical properties of the steel: UTS=2240 MPa, YS=1784 MPa, δ = 6.2 %, ψ = 16.8%, HRC= 42-46.

A cross-sectional image of the bullet penetration hole is presented in Fig. 1. It can be described as two-zone. The profile closer to the contact zone shows plastic deformation with the displacement of the alloys part to the surface, which indicates ductile fracture. It is well seen, that near-contact deformed area reaches almost 1 mm under plate`s surface. It indicates a significant work of the material, that was occurred due to its resistant to ballistic impact. At the same time, this part demonstrates local heating and recrystallisation, what can be noticed by comparing microstructures at the top of Fig. 1.

The hollow profile shows the high ability of the plate to resist the bullet penetration by its sort of compression with materials` volume. However, lower part of the profile shows signs of brittle fracture, indicating a significant accumulation of stresses and defects in the zone after hollow starts narrowing. It also happens at the moment, when the residual thickness remains low.

Microhardness of different zones of the armor element allows us to assess the degree of load distribution due to the degree of local strengthening, caused by deformation. It can also be seen that the zone of first contact with the bullet, which receives the greatest temperature effect and load, has a recrystallized structure that differs from the microstructure of the alloy in the center of the plate and has lower microhardness, caused by temperature influence.

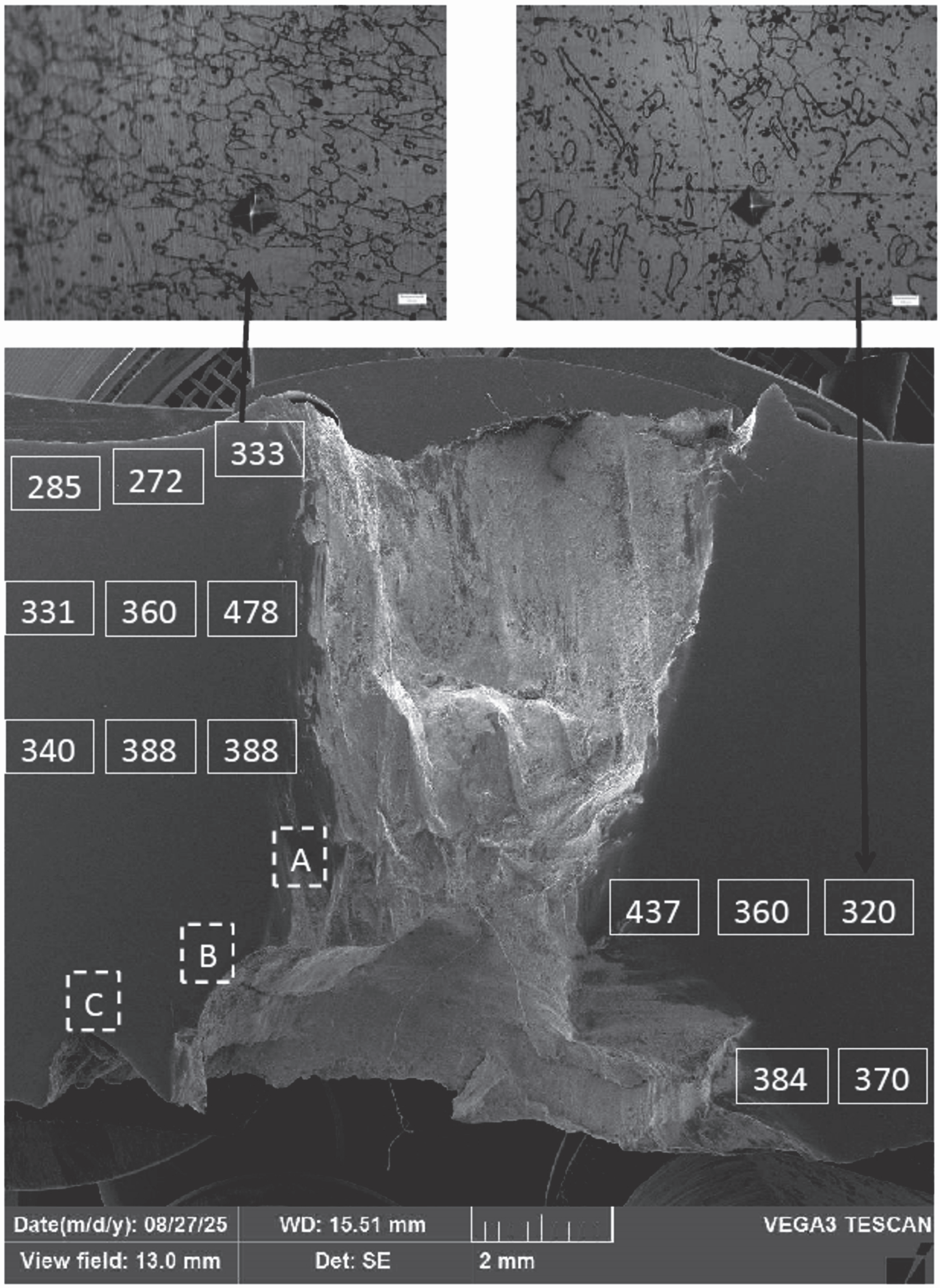

Fig. 1. Cross-section sample of bullet hollow, made after ballistic impact in Fe-26Mn-10Al-1.2Si-2.2Cr-1.8Ni-0.15V-1C steel and marked areas with their microhardness (HV100) and most hardened and rough defected area (A-C).

Microhardness studies in different areas of the cross-section showed that there is a strengthening of the metal in the contact zone, which is associated with thermal effects and local recrystallization. Further, after 1.5 mm in depth, one can observe the absence of thermal effects and maximum contact surface strengthening, which gradually decreases for 2-3 mm in the depth of the plate. Near the middle of the thickness, one can note the narrowing zone of the hole, where the microhardness is increased and does not change for a distance of about 2 mm, which indicates a significant accumulation of stresses and defects in this zone. Below this level, the tearing of the material along a certain profile can be noticed. It shows signs of plastic deformation and surface hardening from it in the fracture zone.

Local study of the zones A-C microstructures, marked in Fig. 1, showed that at the initial stages of armor element destruction, beyond the zone of maximum strengthening, caused by local deformation hardening, ASB preserving is possible. At the same time, the formation of local microcracks is observed. They do not have signs of intergranular destruction, but rather appear in places of highest stresses and defects accumulation. The place and nature of their formation do not show regularities in relation to the structure, also do not coincide with the boundaries of the ferrite-austenite interface and do not have a significant extent, which is limited by sizes up to 50-60 microns. It is characteristic that the ferrite and austenite of the resulting steel are both quite alloyed and have a small difference in microhardness – 360 and 320 (HV^{100}), respectively. Thus, cracks do not form at the phase interface boundaries, as it shown in Fig. 2 a, b, d. Thus, the chemical composition of the alloy leads to desirable phase formation features.

Figure 2 c shows the implementation of several operational mechanisms of the proposed armor steel [13]. In the final fracture zone, it is clearly seen, how the long line of the ASB gradually forms a crack. Along this area, a microcrack coalescence profile is formed, which partially repeats the profile of the ASB line.

Conclusions

The obtained data indicates that the proposed steel is potentially suitable to apply as a protection against ballistic damage. It is able to significantly resist point-like high dynamic loads due to the effective absorption of impact energy, which is expressed in the implementation of several mechanisms simultaneously. The first one consists of a local hardening, caused by deformation load. The second one bases on stress collecting, which obviously leads to formation of adiabatic shear bonds (ASB). Then accumulation of defects and stresses leads to the sequential formation of microcracks, as well as to their coalescence, while the intensive deformation of the material happens. According to this, it can be concluded, that obtained steel demonstrates several different consequent mechanisms to resist ballistic impact stress. Also, it is shown, that developed high strength steel is almost fully austenitic and its phases have a high level of stacking fault energy SFE, what causes defects coalescence instead of twinning, what is better for ballistic resistance.

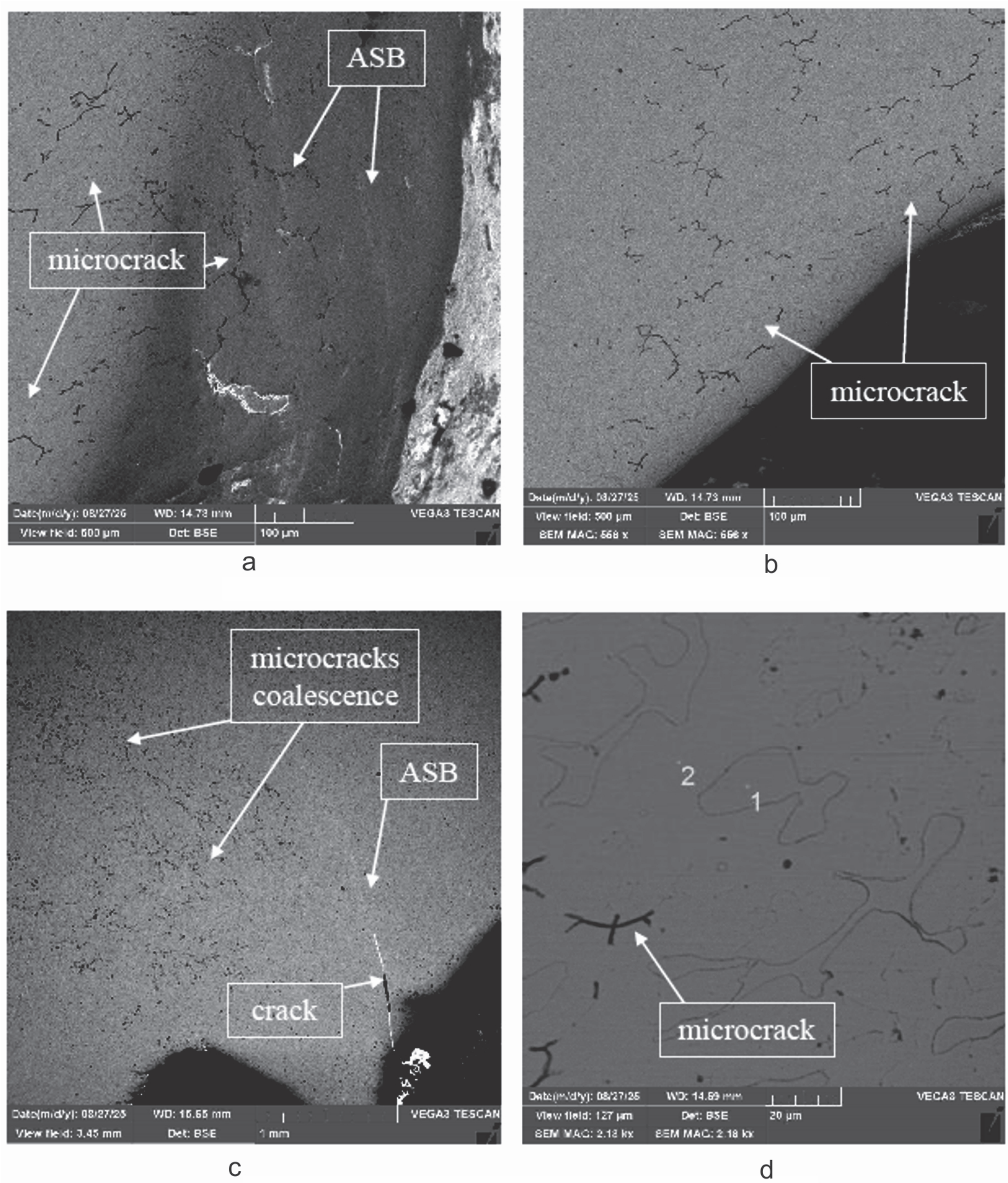

Fig. 2. Microstructure of sections A (a), B (b), C (c), marked at the different profiles of bullet hollow, according to figure 1; Microstructure and local chemical analysis of the alloy (d), where chemical composition of ferrite (1) and austenite (2) are: 1 – Fe-20Mn-9Al-0.9Si-1.35Cr-2.75Ni-0.17V (wt. %), 2 – Fe-28Mn-6.7Al-1Si-2.1Cr-2.2Ni-0.21V (wt. %).

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Мікроструктурні зміни високоміцної сталі Fe-26Mn-10Al-1.2Si-2.2Cr-1.8Ni-0.15V-1C після балістичного ураження

М. М. Ворон[1], кандидат технічних наук, старший дослідник, заступник завідувача відділу, mykhailo.m.voron@gmail.com, ORCID: https://orcid.org/0000-0002-0804-9496
А. М. Тимошенко[1], кандидат технічних наук, старший дослідник, старший науковий співробітник, Marschal@i.ua, ORCID: https://orcid.org/0000-0003-4038-1744
А. Ю. Семенко[1], кандидат технічних наук, старший дослідник, старший науковий співробітник, semenko.au@gmail.com, ORCID: https://orcid.org/0000-0002-0448-1636
О. М. Смірнов[1], доктор технічних наук, професор, завідувач відділу, stalevoz@i.ua, ORCID: https://orcid.org/0000-0001-5247-3908
С. Л. Шваб[2], кандидат технічних наук, старший дослідник, старший науковий співробітник, serg.schwab@gmail.com, ORCID: https://orcid.org/0000-0002-4627-9786
Ю. П. Скоробагатько[1], кандидат технічних наук, старший дослідник, старший науковий співробітник, yulka.ukr@gmail.com, ORCID: https://orcid.org/0000-0002-1724-9895

[1]Фізико-технологічний інститут металів та сплавів НАН України, Київ, Україна
[2]Інститут електрозварювання ім. Є. О. Патона НАН Україна, Київ, Україна

Summary

Роботу присвячено дослідженню структурних змін у зоні пробиття сталі Fe-26Mn-10Al-1.2Si-2.2Cr-1.8Ni-0.15V-1C, яку розробляли для захисту від балістичного ураження. Одержання для такого застосування тонких пластин з легких надміцних сталей на основі системи Fe-Mn-Al-C є надзвичайно перспективною та невивченою задачею. В роботі проведено аналіз впливу механічних властивостей та механізмів мікроструктурних змін, за якими броньові металеві матеріали та сталі досліджуваного типу чинять опір кульовому балістичному ураженню найбільш ефективно. Дя проведення досліджень, в умовах відкритої індукційної плавки, було одержану литу сталеву заготовку, яку надалі піддавали гарячій пластичній деформації і нормалізації. Одержану пластину випробовували на пробстріл та досліджували будову отвору балістичного пробиття. Дослідження механічних властивостей виявили надзвичайно високу міцність сталі, потрібний рівень пластичності та посередню твердість: $\sigma_в$ = 2240 MPa, $\sigma_т$=1784 MPa, δ = 6.2 %, ψ = 16.8 %, HRC = 42-46. Дослідження поперечного профілю отвору прострілу пластини показали формування двох зон. Перша зона повторювала контур кулі та демонструвала значні ознаки локальної пластичної деформації сталі біля поверхні ураження. В приповерхневій області спостерігається локальна рекристалізація і зниження твердості через термічний вплив. При віддаленні від поверхні, ближче до центральної зони отвору спостерігається поверхневе наклепування, яке виражається у помітному зростанні мікротвердості.

Друга зона проявляє ознаки крихкого руйнування. В її мікроструктурі спостерігаються смуги адіабатичного зсуву та місця накопичення напружень, які утворюють мікротріщини. В зоні остаточного руйнування пластини виявлено формування областей коалесценції мікродефектів, форма яких повторює розподіл навантаження всередині матеріалу.

Отримані дані свідчать про те, що запропонована сталь придатна для захисту від балістичних пошкоджень, але при більшій товщині. Вона здатна протистояти балістичним навантаженням завдяки ефективному поглинанню енергії удару, що виражається в реалізації кількох механізмів одночасно − локальному зміцненні, утворенні адіабатичних смуг зсуву, накопиченні дефектів і напружень та послідовному утворенні мікротріщин, а також в їх коалесценції під час деформації матеріалу.

Ключові слова: Fe-Mn-Al-C, броньова пластина, балістичне ураження, зміни мікроструктури, мікродефекти, розподіл навантаження.