

ИЗВЕСТИЯ ВЫСШИХ УЧЕБНЫХ ЗАВЕДЕНИЙ ЧЕРНАЯ МЕТАЛЛУРГИЯ

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МЕТАЛЛУРГИЧЕСКИЕ ТЕХНОЛОГИИ

Сравнительный анализ структуры и свойств
железнодорожных рельсов заэвтектоидного состава

МАТЕРИАЛОВЕДЕНИЕ

Структура и свойства сплавов специального назначения после отжига

Инженерия поверхности аморфного сплава $\text{Fe}_{53,3}\text{Ni}_{26,5}\text{B}_{20,2}$
эксимерным лазером

ИННОВАЦИИ В МЕТАЛЛУРГИЧЕСКОМ ПРОМЫШЛЕННОМ И ЛАБОРАТОРНОМ ОБОРУДОВАНИИ, ТЕХНОЛОГИЯХ И МАТЕРИАЛАХ

Анализ процесса обработки заготовки из мартенситной
нержавеющей стали 40X13, полученной методом проволочного
электронно-лучевого аддитивного производства

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CONTENTS

СОДЕРЖАНИЕ

METALLURGICAL TECHNOLOGIES

- Bessonova O.V., Polevoi E.V., Oskolkova T.N., Komarova T.A. Comparative analysis of the structure and properties of hypereutectoid railway rails 556
- Bogdanova N.A., Zhilin S.G., Predein V.V. Strength characteristics of investment patterns obtained by compaction of waxy material powders in the field of centrifugal forces 563

MATERIALS SCIENCE

- Popova N.A., Gromov V.E., Yur'ev A.B., Nikonenko E.L., Porfir'ev M.A. Hardening of surface layers of long rail heads during long-term operation 572
- Klimov S.A., Noskov F.M., Tokmin A.M., Masanskii O.A. Production of carbide steels based on high-speed steel by induction surfacing 581
- Aleutdinova M.I., Fadin V.V. Contact characteristics of C235 steel in dry sliding against C45 steel under high-density alternating current at different transformation coefficients of supply transformer 587
- Popova M.V., Polunin A.M. Structure and properties of special-purpose alloys after annealing 594
- Permyakova I.E., Ivanov A.A., Lukina I.N., Kostina M.V., Dyuzheva-Maltseva E.V. Surface engineering of $\text{Fe}_{53.3}\text{Ni}_{26.5}\text{B}_{20.2}$ amorphous alloy by excimer laser 598

PHYSICO-CHEMICAL BASICS
OF METALLURGICAL PROCESSES

- Drozin A.D., Dudorov M.V. Criteria for the existence of local equilibria in supercooled melts 607

INNOVATIONS IN METALLURGICAL INDUSTRIAL AND LABORATORY EQUIPMENT, TECHNOLOGIES AND MATERIALS

- Evstigneev A.I., Odionkov V.I., Chernyshova D.V., Evstigneeva A.A., Dmitriev E.A. Effect of annular seams on stress-strain state in cylindrical ceramic shell mold during solidification of a steel casting in it 613

МЕТАЛЛУРГИЧЕСКИЕ ТЕХНОЛОГИИ

- Бессонова О.В., Полевой Е.В., Осколкова Т.Н., Комарова Т.А. Сравнительный анализ структуры и свойств железнодорожных рельсов заэвтектоидного состава 556
- Богданова Н.А., Жилин С.Г., Предени В.В. Прочностные характеристики выплавляемых моделей, получаемых уплотнением порошков воскообразных материалов в поле действия центробежных сил 563

МАТЕРИАЛОВЕДЕНИЕ

- Попова Н.А., Громов В.Е., Никоненко Е.Л., Порфирьев М.А., Юрьев А.А. Упрочнение поверхностных слоев головки длинномерных рельсов при длительной эксплуатации 572
- Климов С.А., Носков Ф.М., Токмин А.М., Масанский О.А. Получение карбидосталей на основе быстрорежущей стали методом индукционной наплавки 581
- Алеутдинова М.И., Фадин В.В. Характеристики контакта стали Ст3 при сухом скольжении по стали 45 под переменным электрическим током высокой плотности при разных коэффициентах трансформации питающего трансформатора 587
- Попова М.В., Полунин А.М. Структура и свойства сплавов специального назначения после отжига 594
- Пермякова И.Е., Иванов А.А., Лукина И.Н., Костина М.В., Дюжева-Мальцева Е.В. Инженерия поверхности аморфного сплава $\text{Fe}_{53.3}\text{Ni}_{26.5}\text{B}_{20.2}$ эксимерным лазером 598

ФИЗИКО-ХИМИЧЕСКИЕ ОСНОВЫ
МЕТАЛЛУРГИЧЕСКИХ ПРОЦЕССОВ

- Дрозин А.Д., Дудоров М.В. Критерии существования локальных равновесий в переохлажденных расплавах 607

ИННОВАЦИИ В МЕТАЛЛУРГИЧЕСКОМ ПРОМЫШЛЕННОМ И ЛАБОРАТОРНОМ ОБОРУДОВАНИИ, ТЕХНОЛОГИЯХ И МАТЕРИАЛАХ

- Евстигнеев А.И., Одиноков В.И., Чернышова Д.В., Евстигнеева А.А., Дмитриев Э.А. Влияние кольцевых швов на напряженно-деформированное состояние в цилиндрической керамической оболочковой форме при затвердевании в ней стальной отливки 613

CONTENTS (Continuation)

СОДЕРЖАНИЕ (продолжение)

Lekhov O.S., Bilalov D.Kh. Calculation of temperature and thermoelastic stresses in strikers during production of hollow steel billets in a unit of combined casting and deformation. Part 2 621

Zhang C., Kozlov V.N., Chinakhov D.A., Klimenov V.A., Chernukhin R.V. Analysis of processing 40Kh13 martensitic stainless steel billet obtained by wire electron beam additive manufacturing 626

INFORMATION TECHNOLOGIES
AND AUTOMATIC CONTROL
IN FERROUS METALLURGY

Rybenko I.A., Protopopov E.V., Zapol'skaya E.M. General principles of modeling and optimization of electric steelmaking process 636

To the 50th Anniversary of O.V. Zayakin 646

Index of articles of "Izvestiya. Ferrous Metallurgy" for 2025, vol. 68 648

Лехов О.С., Билалов Д.Х. Расчет температуры и термоупругих напряжений в бойках при получении стальных полых заготовок на установке совмещенного литья и деформации. Часть 2 621

Чжан Ц., Козлов В.Н., Чинахов Д.А., Клименов В.А., Чернухин Р.В. Анализ процесса обработки заготовки из мартенситной нержавеющей стали 40Х13, полученной методом проволочного электронно-лучевого аддитивного производства 626

ИНФОРМАЦИОННЫЕ ТЕХНОЛОГИИ
И АВТОМАТИЗАЦИЯ
В ЧЕРНОЙ МЕТАЛЛУРГИИ

Рыбенко И.А., Протопопов Е.В., Запольская Е.М. Общие принципы моделирования и оптимизации электросталеплавильного процесса 636

Члену-корреспонденту РАН О.В. Заякину 50! 646

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COMPARATIVE ANALYSIS OF THE STRUCTURE AND PROPERTIES OF HYPEREUTECTOID RAILWAY RAILS

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Abstract. Due to the constant increase in axial loads and the weight of rolling stock intended for passenger and freight heavy traffic, more stringent requirements are being put forward by consumers of rail products. The pearlite-grade carbon steels currently used for the industrial production of rails have high consumer properties, namely, resistance to wear and the formation of contact and fatigue defects, which makes it possible to ensure operability in a wide range of operational and climatic conditions along the entire length of the railway network. An important technical task is to establish the patterns of formation of the microstructure of rails depending on the chemical composition of steel and influence of the structure on properties of the finished products. One of the main parameters determining the structure and consumer properties of pearlite steel railway rails is the size of the interlamellar spacing. Improving the operational stability of rails is one of the main tasks of the specialists of JSC EVRAZ ZSMK, the Kuzbass Metallurgical Plant, one of the five largest manufacturers of railway rails in the world. The results of a comparative analysis of chemical composition, size of the interlamellar spacing, as well as the strength and plastic properties of the metal of pearlite-class railway rails are presented. In accordance with GOST R 51685-2022, 90KhAF steel is hypereutectoid in terms of carbon content. As a result of the conducted research, the relationships between parameters of the structure and the physico-mechanical properties of rails, as well as between content of the main chemical elements in steel and the interlamellar spacing of pearlite, were established.

Keywords: chemical composition, mechanical properties, hardness, rail microstructure, microstructure parameters, interlamellar spacing, differentially heat-treated rails

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СРАВНИТЕЛЬНЫЙ АНАЛИЗ СТРУКТУРЫ И СВОЙСТВ ЖЕЛЕЗНОДОРОЖНЫХ РЕЛЬСОВ ЗАЭВТЕКТОИДНОГО СОСТАВА

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Аннотация. В связи с постоянным увеличением осевых нагрузок и массы подвижных составов, предназначенных для пассажирского и грузового тяжеловесного движения, выдвигаются более жесткие требования со стороны потребителей рельсовой продукции. Используемые в настоящее время для промышленного производства рельсов углеродистые стали перлитного класса обладают высокими показателями потребительских свойств, а именно, стойкостью к износу, образованию контактно-усталостных дефектов, что позволяет обеспечить работоспособность в широком диапазоне эксплуатационных и климатических условий на всей протяженности сети железных дорог. Важной технической задачей является установление закономерностей формирования микроструктуры рельсов в зависимости от химического состава стали и влияния структуры на свойства готовой продукции. Одним из основных параметров, определяющих структуру и потребительские свойства железнодорожных рельсов из перлитной стали, является величина межпластинчатого расстояния. Повышение эксплуатационной стойкости рельсов – одна из основных задач специалистов АО «ЕВРАЗ ЗСМК» – металлургического завода Кузбасса, входящего в пятерку крупнейших в мире производителей железнодорожных рельсов. В результате проведенных исследований

представлены результаты сравнительного анализа химического состава, величины межпластинчатого расстояния, а также прочностных и пластических свойств металла железнодорожных рельсов перлитного класса. В соответствии с ГОСТ Р 51685 – 2022 сталь марки 90ХАФ по содержанию углерода является заэвтектоидной. В результате проведенных исследований установлены связи между параметрами структуры и физико-механическими свойствами рельсов, а также между содержанием основных химических элементов в стали и межпластинчатым расстоянием перлита.

Ключевые слова: химический состав, механические свойства, твердость, микроструктура рельсов, параметры микроструктуры, межпластинчатое расстояние, дифференцированно термоупрочненные рельсы

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INTRODUCTION

Consumers of rail products are increasingly demanding longer intervals between maintenance operations and improved durability and reliability of rails, which remain the primary and most heavily loaded component of the railway superstructure. The operational durability of rails is governed to a large extent by the steel's structural and phase state. For this reason, a key technical objective is to clarify how the rail microstructure forms as a function of steel chemical composition and how the resulting structure affects the properties of the finished products.

Numerous studies [1 – 3] have identified the principal trends linking microstructure to rail operational durability. Worldwide, rails are most commonly manufactured from high-carbon pearlite-grade steels. Under accelerated cooling, these steels develop a fine lamellar ferrite – carbide mixture (pearlite), which provides a favorable combination of hardness, strength and ductility, impact toughness, fracture toughness, contact fatigue strength, and wear resistance [4 – 6]. Among the parameters that control both the microstructure and the consumer properties of pearlitic railway rails, a central role is played by the interlamellar spacing (ILS). In this work, ILS is understood as the sum of the thicknesses of the ferrite and cementite lamellae that form in rails after rolling and differential heat treatment [7 – 9]. Despite its importance, this aspect remains insufficiently explored [10 – 12]. Therefore, quantifying this structural parameter is essential when designing steel compositions, selecting heat-strengthening modes, and setting temperature – deformation rolling schedules aimed at achieving the required quality indicators of metal products.

This study presents the results of a comparative analysis of the chemical composition, interlamellar spacing, and strength and ductility properties of R65 railway rails manufactured by JSC EVRAZ United West Siberian Metallurgical Plant (JSC EVRAZ ZSMK) [13].

MATERIALS AND METHODS

Differentially heat-treated R65 railway rails, hardened using the residual heat from pre-rolling heating,

were selected as the object of investigation. According to GOST R 51685 – 2022, 90KhAF steel is classified as hypereutectoid with respect to carbon content.

To evaluate the metal quality of the rails, tensile tests were performed in accordance with GOST 1497 – 2023 using a Z-250 universal testing machine. Impact bending tests were carried out in accordance with GOST 9454 – 78 on an RKP-450 pendulum impact tester. Hardness measurements were conducted on the running surface of the rail head in accordance with GOST 9012 – 59 using a NEMESIS 9503 universal hardness tester.

Microstructural studies and measurements were performed using a Tescan MIRA3 scanning electron microscope (SEM). The interlamellar spacing and grain diameter were evaluated on transverse metallographic sections prepared from the fillet region after electropolishing and etching in a 4 % alcoholic solution of nitric acid.

The region selected for interlamellar spacing measurements was located at a distance of 2 – 4 mm from the running surface of the rail head. This location was chosen because it corresponds to the zone experiencing the maximum contact stresses arising from wheel – rail interaction during service. In accordance with the procedure described in [14], interlamellar spacing measurements were carried out within pearlite colonies exhibiting a clearly pronounced parallel arrangement of lamellae, regardless of their apparent spacing (visible interlamellar spacing). The obtained data were subsequently processed taking into account the asymmetry coefficient, followed by calculation of the average value of the true interlamellar spacing. Lamellar spacings showing evident signs of deformation (fractured, broken, or strongly curved lamellae) were excluded from the analysis.

Grain diameter in the rail steel of the investigated compositions was determined near the fillet surface using the ferrite/cementite network method in accordance with GOST 5639 – 82 “Steels and alloys. Methods for revealing and determining grain size”.

RESULTS

The chemical composition of the investigated rail steels, conventionally designated as compositions 1 – 5,

Table 1. Chemical composition of the studied rails metal

Таблица 1. Химический состав металла исследуемых рельсов

Composition	Element content, wt. %				(C + Cr), %
	C	Cr	V	Mo	
1	0.87	0.30	0.08	0.006	1.17
2	0.94	0.27	0.09	0.006	1.22
3	0.91	0.34	0.09	0.006	1.24
4	0.93	0.34	0.10	0.018	1.27
5	0.91	0.34	0.04	0.010	1.25
Requirements of GOST R 51685–2022 for 90KhAF steel	0.85 – 0.94	0.20 – 0.60	0.03 – 0.15	–	–

is presented in Table 1. Heating, rolling, and heat treatment of the rails were carried out under identical temperature – time processing conditions. For each nominal composition, several heats were selected and grouped, differing primarily in carbon, chromium, vanadium, and molybdenum contents. The contents of the remaining elements do not differ significantly between the heats, meet the requirements of GOST R 51685 – 2022, and are therefore not reported here.

The data show that all compositions meet the requirements of GOST R 51685 – 2022 with respect to carbon, chromium, and vanadium. Additional microalloying with molybdenum is permitted and is not regulated by the standard. A comparison of the chemical compositions indicates that composition 1 has the lowest carbon content, whereas composition 2 has the highest; compositions 3 – 5 fall in between. Vanadium contents are generally comparable, except for composition 5, where vanadium is lower by 0.04 – 0.06 % than in the other steels. The rail steels of compositions 4 and 5 contain more molybdenum, by approximately 0.012 and 0.004 %, respectively. Chromium is lowest in compositions 1 and 2; compositions 3 – 5 show slightly higher chromium levels, exceeding those of compositions 1 and 2 by 0.04 – 0.07 % on average.

It is well known [15; 16], that carbide-forming elements strongly affect the kinetics of austenite decomposition and, consequently, the shape of the isothermal transformation diagram. Specifically, the pearlitic and bainitic transformations become clearly separated, and an intermediate region with increased austenite stability appears. Studies [17 – 19] have shown that chromium alloying of pearlitic steels results in solid-solution strengthening of ferrite, chromium enrichment of cementite with formation of $(\text{Fe}, \text{Cr})_3\text{C}$, and possibly the formation of chromium carbides. In [20], the conditions for forming a pearlitic structure during rail heat strengthening were analyzed, and relationships were established between the total carbon and chromium content ((C + Cr), %), the pearlitic transformation temperature, and the inter-

lamellar spacing of pearlite in the steels under consideration (Table 1). In addition, dependencies of strength and ductility on carbon and chromium contents were identified; the combined (C + Cr) content should exceed 1.2 %, which promotes a finer pearlitic structure and provides an optimal balance between strength and ductility. In the present steels, the combined carbon and chromium content is broadly similar and exceeds 1.2 %, except for composition 1 (1.17 %). This is consistent with the relatively high strength values obtained while retaining high ductility (Table 2).

Mechanical testing showed that, in terms of tensile properties, impact toughness, and hardness measured on the running surface of the rail head, all investigated steels substantially exceed the standard requirements for hypereutectoid rails intended for DT370IK and DT400IK categories. Composition 3, in which carbon, chromium, and vanadium are increased simultaneously, exhibits the most favorable combination of strength, ductility, and impact toughness. At the same time, the running-surface hardness of the rail head in composition 3 is only slightly higher than that of composition 1, which contains the lowest levels of carbon and the above alloying elements. A threefold increase in molybdenum content in composition 4 is accompanied by a marked increase in impact toughness and by higher strength and ductility compared with the other steels. A reduction in the vanadium and molybdenum contents in experimental alloy 5 results in a pronounced decrease in impact toughness and ductility, while strength characteristics and hardness at the running surface of the finished rail head remain unchanged.

After rolling and differential heat treatment, the investigated rail steel exhibits a pearlitic microstructure, predominantly composed of pearlite lamellae with a regular morphology (Fig. 1).

At the same time, a considerable number of deformed pearlite colonies with broken and “wavy” lamellae are observed, as well as regions characterized by interwoven pearlite colonies (Fig. 2).

Table 2. Mechanical properties and hardness

Таблица 2. Механические свойства и твердость

Composition		Tensile properties				KCU (+20 °C), J/cm ²	Hardness at the running surface of the rail head, HB
		σ_y , N/mm ²	σ_u , N/mm ²	δ , %	ψ , %		
1		920	1370	10.5	27	26.5	393
2		980	1430	10.5	31	26.0	412
3		1030	1450	10.5	29	25.0	409
4		1028	1480	11.0	24	30.0	424
5		1049	1447	9.0	22	21.5	420
Requirements of GOST R 51685–2022 for rail category	DT370IK	no less					370 – 415
		870	1280	8.0	14.0	15	400 – 455
	DT400IK		1300				

As part of the quantitative analysis, key microstructural parameters – namely, the interlamellar spacing (pearlite dispersion) and grain diameter – were evaluated.

The results of the microstructural measurements, presented in Table 3, are given as arithmetic mean values with the corresponding standard deviations.

The data indicate that the average interlamellar spacing of pearlite and grain size measured in the rail steels of the investigated compositions are generally comparable. An exception is observed for compositions 1 and 2, in which the interlamellar spacing amounts to 0.116 and 0.112 μm , respectively, indicating comparatively lower

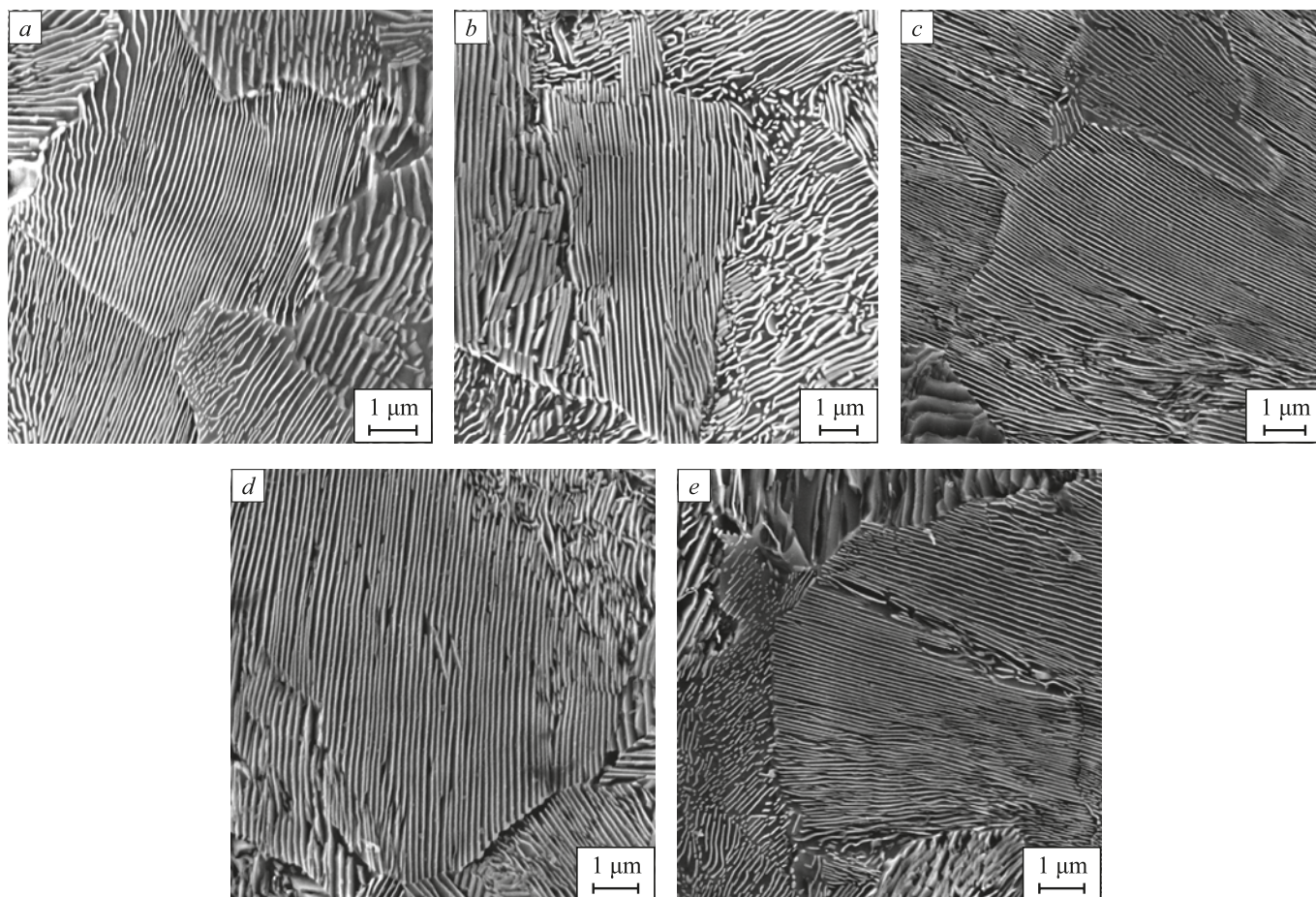


Fig. 1. Microstructure of the studied rails metal of compositions 1 – 5 (a – e)

Рис. 1. Микроструктура металла исследуемых рельсов составов 1 – 5 (a – e)

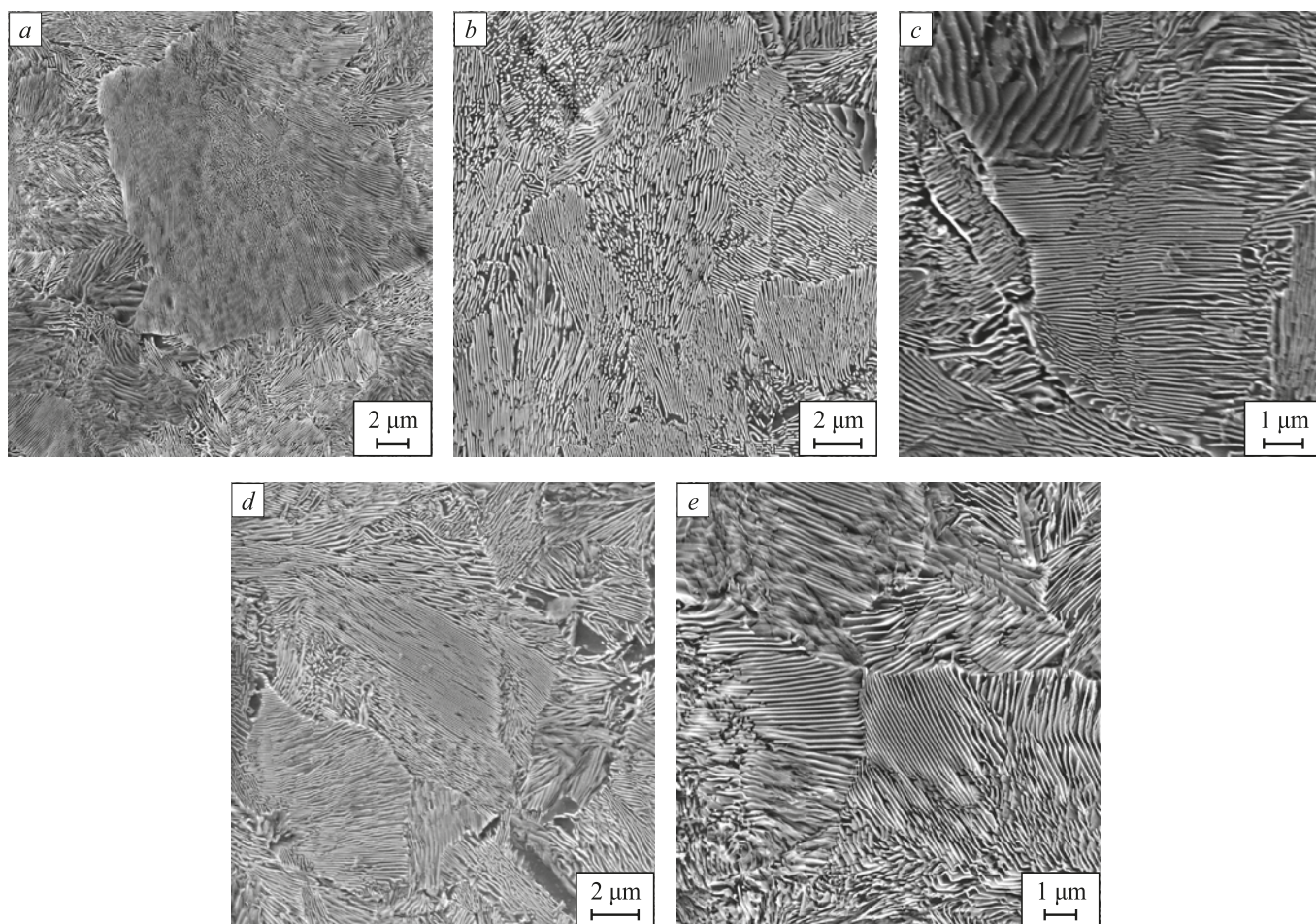


Fig. 2. Microstructure of the studied rails metal of compositions 1 – 5 with deformable plates (a – e)

Рис. 2. Микроструктура металла исследуемых рельсов составов 1 – 5 с деформируемыми пластинами (a – e)

strength levels. In contrast, the rail steel of composition 4 exhibits a significantly smaller average grain size – by approximately 3 – 5 μm – than the steels of the other investigated compositions (Table 3). The increased vanadium and molybdenum contents in composition 4 likely contributed to the enhancement of impact toughness as well as strength and ductility characteristics, which can generally be attributed to the grain growth – retarding effect of these elements during rail heat treatment [21].

Table 3. Microstructure parameters

Таблица 3. Параметры микроструктуры

Composition	Interlamellar spacing, μm	Grain diameter, μm
1	$0.116 \pm (0.019)$	$18.36 \pm (4.04)$
2	$0.112 \pm (0.020)$	$19.30 \pm (4.80)$
3	$0.100 \pm (0.013)$	$20.23 \pm (2.82)$
4	$0.110 \pm (0.011)$	$15.14 \pm (2.83)$
5	$0.109 \pm (0.011)$	$20.60 \pm (3.90)$

CONCLUSIONS

A comparative analysis was carried out covering the chemical composition, strength and ductility characteristics, and quantitative microstructural parameters of railway rails produced from hypereutectoid 90KhAF steel for the five investigated compositions.

The most favorable combination of strength and ductility characteristics – based on tensile testing, impact bending tests, and hardness measured on the running surface of the rail head – was obtained for rails produced from experimental composition 4, which is characterized by slightly increased contents of the microalloying elements vanadium and molybdenum, by approximately 0.01 – 0.06 and 0.008 – 0.012 wt. %, respectively.

Quantitative evaluation of the microstructural parameters showed that the average interlamellar spacing of pearlite in the investigated samples remains at a comparable level. At the same time, the rail steel of composition 4 exhibits a finer-grained structure, with the grain diameter reduced by 3 – 5 μm . This finer grain structure, in combination with a dispersed pearlitic microstructure,

contributed to achieving an optimal balance between strength and ductility in the investigated rails.

In service, a rail's resistance to contact fatigue defects and other damage mechanisms is influenced by many factors. Understanding how the microstructure forms and how it should be matched to the required mechanical properties enables the design of rails tailored to specific track sections, improving in-service performance and extending service life.

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STRENGTH CHARACTERISTICS OF INVESTMENT PATTERNS OBTAINED BY COMPACTION OF WAXY MATERIAL POWDERS IN THE FIELD OF CENTRIFUGAL FORCES

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Abstract. The use of investment casting process is aimed at producing castings of complex configuration with increased dimensional and geometric accuracy from a wide range of casting steels and non-ferrous alloys. A number of operations during the implementation of such a process are accompanied by the appearance of defects of a thermophysical nature (shrinkage of the pattern material and its thermal expansion during melting, leading to a violation of the ceramic mold integrity), which, to a certain extent, prevents the expansion of the casting nomenclature. The formation of experimental porous investment patterns by compaction of powders of waxy materials is aimed at eliminating such defects, but, due to the lack of information on the processes accompanying the compaction of waxy powders (in some cases manifested in the elastic response of the material or a change in the strength characteristics of the compacts), requires separate study. It was previously established that the distribution of density values in a paraffin powder compact is provided by directional loading of the compacted material, including in the field of centrifugal forces, which allows obtaining the surface configuration of a body of revolution with a predictable distribution of properties in each of its sections. In this paper, using the example of forming a section of a body of revolution, a comparison is given of the calculated and experimental dependencies of the relative density of compacts (obtained from different fractions of the PS50/50 material) on the stresses arising during their compaction in the field of centrifugal forces, as well as the average values of the density of compacts on the molds rotation speed. Pictures of the stress-strain state of compacts are presented when determining the values of their compressive strength characteristic of various waxy materials. The results of the experiment are aimed at solving the problems of increasing the efficiency of processes for obtaining investment patterns, the configuration of which is a body of revolution, formed by compaction of powdered waxy materials in the field of centrifugal forces.

Keywords: investment casting, ferrous and non-ferrous alloys, experimental modeling, mechanical engineering processes, field of centrifugal forces, stress-strain state, compact, porosity, elastic response, strength

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ПРОЧНОСТНЫЕ ХАРАКТЕРИСТИКИ ВЫПЛАВЛЯЕМЫХ МОДЕЛЕЙ, ПОЛУЧАЕМЫХ УПЛОТНЕНИЕМ ПОРОШКОВ ВОСКООБРАЗНЫХ МАТЕРИАЛОВ В ПОЛЕ ДЕЙСТВИЯ ЦЕНТРОБЕЖНЫХ СИЛ

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Аннотация. Использование процесса литья по выплавляемым моделям направлено на получение отливок сложной конфигурации с повышенной размерной и геометрической точностью из широкого диапазона литейных сталей и цветных сплавов. Ряд операций при осуществлении такого процесса сопровождается появлением дефектов теплофизической природы (усадка модельного материала и его температурное расширение при выплавлении, приводящее к нарушению целостности керамической формы), что, в известной мере, препятствует расширению номенклатуры литья. Формирование экспериментальных пористых выплавляемых моделей прессованием порошков воскообразных материалов направлено на устранение данных дефектов, но, в силу недостатка сведений о процессах, сопровождающих уплотнение воскообразных порошков (в ряде случаев, проявляющихся в упругом отклике материала или изменении прочностных характеристик прессовок), требует отдельного изучения. Ранее было установлено, что распределение значений плотности в прессовке, выполненной из порошка парафина, обеспечивается направленным нагружением уплотняемого материала, в том числе в поле действия центробежных сил, что позволяет получать конфигурацию поверхности тела вращения с прогнозируемым распределением свойств в каждом его участке. В настоящей работе, на примере формирования участка тела вращения, приведено сравнение расчетных и экспериментальных зависимостей относительной плотности прессовок (полученных из разных фракций материала ПС 50/50) от напряжений, возникающих при их уплотнении в поле действия центробежных сил, а также средних значений плотностей прессовок от скорости вращения форм. Представлены картины напряженно-деформированного состояния прессовок при определении значений их прочности на сжатие, характерные для различных воскообразных материалов. Результаты эксперимента направлены на решение задач повышения эффективности процессов получения выплавляемых моделей, конфигурация которых представляет собой тело вращения, образованное уплотнением порошковых воскообразных материалов в поле действия центробежных сил.

Ключевые слова: литье по выплавляемым моделям, черные и цветные сплавы, экспериментальное моделирование, машиностроительные процессы, поле действия центробежных сил, напряженно-деформированное состояние, прессовка, пористость, упругий отклик, прочность

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INTRODUCTION

Improving the efficiency of producing accurate billets from a wide range of ferrous and non-ferrous alloys by investment casting requires the development and modernization of technological processes that eliminate machining operations and expand the range of cast products for the aircraft, shipbuilding, and mechanical engineering industries. The manufacture of castings of complex spatial configuration [2; 3], characterized by high dimensional and geometric accuracy and low surface roughness, is predominantly carried out by investment casting [4 – 6].

The general sequence of traditional operations in the investment casting process involves the formation of investment patterns, their assembly into pattern clusters, sequential coating and drying of ceramic shell layers, melting out the pattern material from the ceramic mold, firing of the shell in a supporting filler, and pouring

the shell mold with liquid metal followed by machining of the cast billets [7 – 9].

Throughout the investment casting process, the technological stages mentioned above are accompanied by thermophysical phenomena that can lead to geometric distortion of the investment pattern due to shrinkage [10 – 13], deformation, and fracture of ceramic shell mold sections caused by thermal expansion of pattern materials both during dewaxing and at the stages of firing and metal pouring [14 – 17].

The traditional approach to improving the thermostability of investment patterns has been to select pattern material components with low thermal expansion [18], while increasing the crack resistance of ceramic molds is mainly achieved by using reinforcing elements or new binding agents [19; 20].

A promising way to address the complex of thermophysical problems is the use of porous investment patterns formed by compaction of fractions of waxy mate-

rials. Such investment patterns do not have shrinkage defects and do not exert expanding effects on ceramic molds [21]. A series of experiments has provided practical data on the stress–strain behavior of waxy materials, aimed at eliminating the elastic relaxation of compacts (whose magnitude is an order of magnitude smaller than the shrinkage value and amounts to about 1 % of the compact volume) formed during their compaction [22].

Greater control over the porosity of investment patterns can be achieved by compaction of powders of waxy materials in the field of centrifugal forces, which provides directional loading of the powder compact during compaction [23]. Compaction of powders of waxy materials makes it possible to obtain an investment pattern in the form of bodies of revolution, the configuration of the external surface of which is determined by the rotating mold. However, achieving a uniform distribution of properties within such compacts requires a substantially more complex mold design, whereas producing a body-of-revolution investment pattern by pressing a pasty pattern compound may lead to shrinkage defects upon cooling. Research in this area is motivated by the need to identify an energy-efficient compaction mode for forming compacts in the shape of bodies of revolution in the field of centrifugal forces, allowing the mold rotation speed to be reduced while maintaining acceptable strength of compacts made of various waxy materials.

Therefore, the purpose of this study is to determine the force parameters of the process of compaction in the field of centrifugal forces for fractions of powdered waxy pattern material PS 50/50 and to analyze the strength characteristics of such compacts in comparison with those obtained by compaction in a closed mold, including those produced from paraffin grade T1.

To achieve these objectives, the following tasks were set the study addressed the following tasks:

- selecting a computational method to obtain the stress dependencies (derived using the formulas of M. Yu. Balshin and G. N. Zhdanovich) arising during vertical uniaxial compaction of PS 50/50 powder compacts with particle sizes of 2.5 and 0.63 mm in the relative density range 0.8 – 1.0, best corresponding to the experimental data, and comparing the adapted calculated data with experimental results;

- comparing the calculated and experimental mold rotation speeds required to achieve average compact densities in the range 842 – 935 kg/m³ (corresponding to load-bearing capacity) for PS 50/50 fractions 2.5 and 0.63 mm;

- comparing the strength characteristics of compacts made of waxy powdered materials grades T1 and PS 50/50 obtained by uniaxial vertical compaction and compaction in the field of centrifugal forces.

METHODS

The waxy powder selected for this study was PS 50/50, the most widely used pattern material in investment casting. Particle size fractions of 0.63 and 2.5 mm were used, as these sizes are technologically convenient and ensure homogeneous compacts [24]. The fractions were obtained by sieving on model 026 sieves manufactured in accordance with GOST 29234.3–91 “Foundry sands. Method for determination of mean grain size and uniformity coefficient”. PS 50/50 is an alloy of equal mass fractions of paraffin and stearin and belongs to the first classification group of pattern materials [25]. As a reference material, paraffin grade T1 was used; its properties are regulated by GOST 23683–89 “Petroleum solid paraffins. Specifications”. The strength characteristics of compacts obtained from PS 50/50 were compared with those of compacts made from T1. Data on the average densities of compacts formed in the field of centrifugal forces at different mold rotation speeds were taken from the authors’ previous work [23]. The densities ρ of the materials in the experiment, corresponding to their cast state, were: PS 50/50 – $\rho = 935 \text{ kg/m}^3$, T1 – $\rho = 860 \text{ kg/m}^3$.

Earlier experiments refined the permissible porosity (P) range of $0 \% \leq P \leq 10 \%$, within which waxy compacts produced by uniaxial compaction of materials T1 and PS 50/50 retain their service strength [21]. Porosity was calculated as

$$P = \left(1 - \frac{\rho_{\text{comp}}}{\rho_{\text{cast}}} \right) \cdot 100 \%, \quad (1)$$

where ρ_{comp} is the density of the compacted sample, kg/m³; ρ_{cast} is the density of the cast material, kg/m³.

Thus, the density ranges ensuring satisfactory performance were 774 – 860 kg/m³ for T1 and 842 – 935 kg/m³ for PS 50/50.

Before placing a measured portion of the powdered pattern material into the mold, the inner surface of the mold was treated with kerosene to reduce friction between the compacted material and the wall. Uniform distribution of the powder within the mold prior to compaction was achieved by vibrating the mold on a vibrating table at 3.5 Hz for 5 min.

The compaction of waxy material powders in the field of centrifugal forces makes it possible to obtain investment patterns in the form of bodies of revolution, with the outer surface geometry defined by the shaping surface of the rotating mold. In this case, centrifugal forces generate an almost uniform pressure distribution throughout the compact from the center of rotation to the periphery, which results in a nearly uniform distribution of properties; in particular, the density is essentially the same at equal radial distances from the rotation axis.

Producing a compact in the form of a body of revolution imposes specific requirements on the equipment due to the high mold rotation speeds traditionally required (6,000 – 15,000 rpm), which is energy-inefficient [26].

Preliminary studies showed that the use of an attached mass acting as a punch is a feasible way to form compacts in the shape of bodies of revolution in the field of centrifugal forces. In this scheme, the compaction process includes the following steps:

- a measured portion of the powdered pattern material is placed into a mold mounted on the centrifuge shaft. The mold defines the geometry of the compact and is rotated at a speed sufficient to form a cavity at the center of the compact;

- a flat spiral spring containing the attached mass (for example, steel balls) is inserted into this cavity. The mold is then spun again, and the elements of the attached mass self-distribute along the inner surface of the steel spring.

Experiments showed that, for PS 50/50, this compaction scheme requires a mold rotation speed in the range of 3500 – 4000 rpm.

This design provides the necessary pressure directed from the centrifuge rotor axis toward the periphery at acceptable rotation speeds [23]. In the experiments, the attached mass was a washer made of steel 45 with a mass of 0.125 kg, radius $R = 0.023$ m, and height $l = 0.1$ m (Fig. 1).

To simplify the calculation of the parameters of the body-of-revolution forming process, the scheme shown in Fig. 1 was used. In Fig. 1, *a*, the schematic diagram shows compaction of powdered waxy pattern material in a mold (1) shaped as a cup of radius R , fixed to the centrifuge rotor holder (2) and at rest. In Fig. 1, *b*, the compaction stage corresponds to rotation of the centrifuge rotor at $\omega = 3500 - 4000$ rpm, when the z axis of the holder coincides with the mold axis. The coordinate z varies from z_0 to z_i and characterizes the movement of the powder surface during compaction, while Z is the distance from the rotor axis to the bottom of the cylindrical mold ($Z = 0.015$ m). To relieve stresses in the compacted powder and reduce the elastic response of the compact, the mold was rotated at angular velocity ω for 7 min.

During centrifugal compaction, the centrifuge rotor speed was continuously recorded. After compaction, the compacts were removed from the molds and their densities measured. The specimens were then subjected to compressive loading until fracture at a crosshead speed of 22 mm/min, which complies with GOST 4651–2014 “Plastics. Compression test method”. Experimental fracture data for compacts from PS 50/50 were compared with those obtained for compacts from T1. The stresses arising during loading were recorded using a Shimadzu AG-X plus testing machine.

Based on the computational and experimental results, the following relationships were obtained:

- comparison of calculated (according to the equations of M.Yu. Balshin and G.N. Zhdanovich) and experimental dependences of load on relative density for samples formed from 0.63 and 2.5 mm fractions;

- comparison of calculated and experimental average densities of compacts obtained from 0.63 and 2.5 mm fractions as a function of mold rotation speed;

- stress–strain relationships recorded during compressive loading of samples made of PS 50/50 and T1 materials produced by uniaxial pressing and by compaction in the field of centrifugal forces.

RESULTS AND DISCUSSION

The schematic diagram in Fig. 1, similar to that proposed in [27], is presented in Cartesian coordinates x , y , and z to enable numerical analysis of the compact formation process in a rotating cylindrical mold and to predict the property distribution within the compact. In this scheme, the z axis coincides with the cylinder axis, and the coordinate origin is located on the centrifuge rotor axis y .

The centrifugal force compacting the material is expressed as

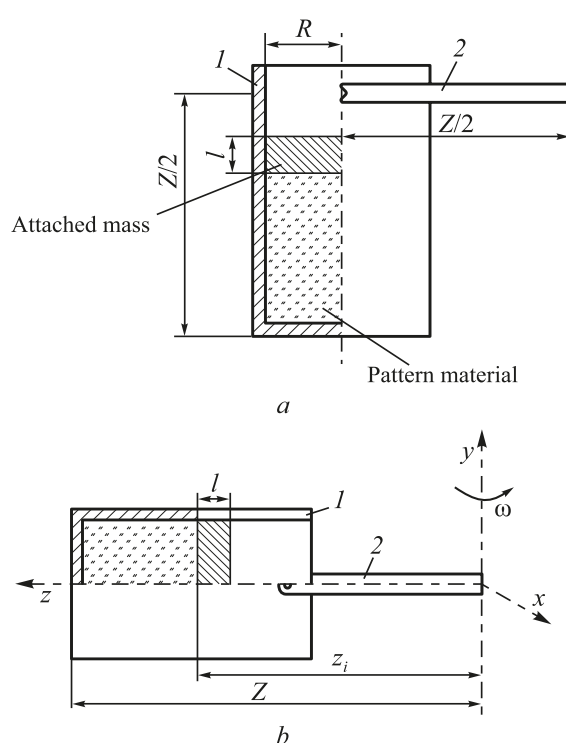


Fig. 1. Scheme of forming a compact from powder material in the field of centrifugal forces

Рис. 1. Схема процесса формирования прессовки из порошкового материала в поле действия центробежных сил

$$\vec{F} = m\vec{\omega}(\vec{r}\vec{\omega}). \quad (2)$$

The force vector has two nonzero components

$$F_x = mr_x\omega^2, F_z = mr_z\omega^2. \quad (3)$$

Under the action of the force $\vec{F}$, the F_z component causes the compacted powder material to move toward the bottom of the cylindrical mold, whereas the F_x component promotes material spreading from the mold center toward the walls along the x axis. The effect of F_x is small, and together with the friction force it can be neglected by setting both to zero, which simplifies the calculations. In this formulation, the centrifugal force depends only on the z coordinate, and the compaction process can be treated as layer-by-layer densification of a powder compact represented by n layers, each of height h_i^j (where $i = 1, \dots, n$ is the layer number, and j is the iteration number in the calculation process). The calculation starts from the upper layer ($z = z_0$) and proceed down to the layer adjacent to the mold bottom ($z = Z$). At the initial iteration ($j = 0$), all layer heights are equal

$$h_i^0 = \frac{Z - z_0}{n} = \frac{H}{n}. \quad (4)$$

The mass m can be expressed in terms of density as $m = \rho V = \rho HS$, where V is the compact volume, H is the compact height, and S is the cross-sectional area of the mold. Dividing the centrifugal force component F_z , which generates pressure, by the cross-sectional area $S = \pi R^2$ gives the stress in the i -th layer of the powder compact:

$$\sigma_i^j = \sum_{l=1}^i \rho_l^j h_l^j z_l^j \omega^2, z_i^j = Z - \sum_{l=n}^{i+1} h_l^j. \quad (5)$$

Since the density ρ_i^j depends on the stress σ_{i-1}^j , the calculation is carried out in two stages. First, the stress at the current iteration j is determined from the densities obtained at the previous iteration ($j = 1$). Then, the densities of each layer are recalculated until the difference between successive iterations becomes negligibly small. For this, a known functional dependence between the density of the material and the pressing stress $\rho = \rho(\sigma)$, is required. Fig. 2 presents the results of a series of preliminary experiments to determine the stresses arising during uniaxial vertical compaction of PS 50/50 powder fractions of 2.5 mm (Fig. 2, *a*) and 0.63 mm (Fig. 2, *b*) to relative densities within the range 0.8 – 1.0. Curves 1 and 2 in Fig. 2 show the calculated results based on the equations proposed by M.Yu. Balshin (Eq. (6)) and G.N. Zhdanovich (Eq. (7)) for describing the compaction behavior of powdered bodies [28; 29]:

$$\sigma = \sigma_{\max} \theta^m, m = 2 + \frac{\theta}{\theta - \theta_0}; \quad (6)$$

$$\sigma = \sigma_{\max} \frac{\theta^m - \theta_0^m}{1 - \theta_0^m}, m = 1 + \frac{2}{1 - \theta_0}, \quad (7)$$

where $\sigma_{\max}$ is the stress at which the density ρ reaches the cast density $\rho_{\max}$; $\theta = \rho/\rho_{\max}$ is the relative density; $\theta_0 = \rho_0/\rho_{\max}$ is the initial relative bulk density, and ρ_0 is bulk density of the loose powder.

As can be seen from Fig. 2, the stresses during compaction of the powder compact made from the 0.63 mm fraction are slightly lower than those for the 2.5 mm fraction.

Since the curve 2 obtained according to the Zhdanovich equation lies closer to the experimental exponential trend (curve 4), the original data were approximated by adjusting the exponent m for the best fit. The exponents m for the PS 50/50 material fractions were found to be 7.55 and 8.78, respectively.

The inverse dependence of relative density on stress, corresponding to Eq. (7), can be written as:

$$\rho = \rho_{\max} \theta = \rho_{\max} \left[\frac{\sigma}{\sigma_{\max}} (1 - \theta_0^m) + \theta_0^m \right]^{\frac{1}{m}}. \quad (8)$$

Using Eqs. (5) and (8), the density distribution in the powder compact along the height of the cylindrical mold can be determined for a given angular velocity ω . To reduce the angular velocity ω from 6,000 – 15,000 rpm to acceptable levels, an attached mass is placed on the surface of the compacted material. Its presence increases the uniformity of density distribution within the compact by exerting additional pressure on it.

For this purpose, by integrating the F_z component in Eq. (3) for a material with a uniform density distribution throughout its volume, the stress provided by the attached mass can be obtained. Dividing σ by the mold cross-sectional area S yields the value of the stress acting on the surface of the compacted material ($z = z_0$) as a result of the influence of the attached mass:

$$\sigma_{\text{add}} = \frac{m_{\text{add}}}{S} \left(z_0 - \frac{1}{2} l \right) \omega^2, \quad (9)$$

where m_{add} is the attached mass, l is the thickness of the attached mass made in the form of a washer. As compaction proceeds, the interface between the powder surface and the attached mass shifts from the mold center toward the periphery ($z = Z$).

Accordingly, for a given angular velocity, the stresses in the compact can be calculated from Eqs. (5) and (9), and the corresponding average compact densities can then be obtained using the adapted Zhdanovich equation.

For the PS 50/50 material, the angular velocity of the mold at which the average density of compacts

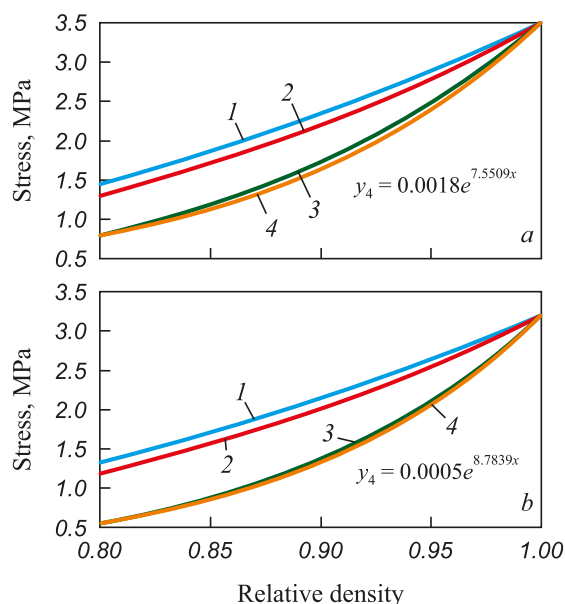


Fig. 2. Calculated and experimental dependences of stresses arising during uniaxial vertical compaction of PS50/50 material on the values of relative density of compacts: *a* – fraction 2.5 mm; *b* – fraction 0.63 mm

Рис. 2. Расчетные и экспериментальные зависимости напряжений, возникающих при одноосном вертикальном уплотнении материала ПС50/50, от значений относительной плотности прессовок: *a* – фракция 2,5 мм; *b* – фракция 0,63 мм

corresponds to the required range of 842 – 935 kg/m³ was determined. Fig. 3 shows the graphical dependences of the average density of compacts formed from PS 50/50 with particle sizes of 2.5 mm (Fig. 3, *a*) and 0.63 mm (Fig. 3, *b*) on the mold rotation speed. The calculated curves (1) lie slightly below the experimental exponential curves (2) in both cases. The experiment showed that a density of 935 kg/m³, corresponding to the cast state of the compact formed from the 0.63 mm fraction, is achieved at an angular velocity of 4050 rpm. For compacts produced from the 2.5 mm fraction, achieving the same density required a slightly higher velocity – approximately 4200 rpm. Overall, the use of an attached mass significantly reduced the necessary mold rotation speed to a range of 3500 – 4200 rpm.

If the density distribution in compacts produced by direct uniaxial vertical compaction and by compaction in the field of centrifugal forces differ, corresponding differences will also be expected in their behavior under uniaxial compression to failure. It is evident that the intrinsic properties of the materials (T1 and PS 50/50) will partly account for the observed differences in the compressive strength of the compacts.

Fig. 4 compares the stress–strain relationships for cylindrical compacts with porosity $P = 0\%$ obtained by direct pressing in a closed die (at a punch travel rate of 0.5 mm/s) and by compaction in the field of centrifugal

forces. Black curves represent samples produced by direct pressing of waxy powders in a closed cylindrical mold; red curves correspond to samples obtained by centrifugal compaction. Solid lines denote compacts made from 2.5 mm fractions; dashed lines correspond to 0.63 mm fractions.

Visual analysis of the “stress–strain” curves for samples made of T1 (Fig. 4, *a*) and PS 50/50 (Fig. 4, *b*) reveals notable differences in their deformation behavior under compressive loading. The photographs included in Fig. 4 show that T1 material (melting point $\approx 60^\circ\text{C}$ [21]) deforms in a brittle manner, whereas PS 50/50 material (melting point $\approx 52^\circ\text{C}$) exhibits greater plasticity. Overall, the resistance to compressive loading of compacts formed from T1 powder is slightly higher than that of compacts produced from PS 50/50, which, in all likelihood, can also be explained by the greater plasticity of the latter. It should be noted that a common feature characterizing all variants of experimental compacts is the dominance of the peak compressive stress values obtained during strength testing of samples made from the 2.5 mm fraction over the strength values of samples made from the smaller fraction – on average by 7 % for the case of single-action uniaxial compaction and by 12 % for compaction in the field of centrifugal forces.

Analysis of the relationships presented in Fig. 4 also showed that the compressive strength of samples formed in the field of centrifugal forces is, on average, 15 % lower than that of compacts obtained by single-action compaction, which, however, is sufficient to ensure their further technological applicability.

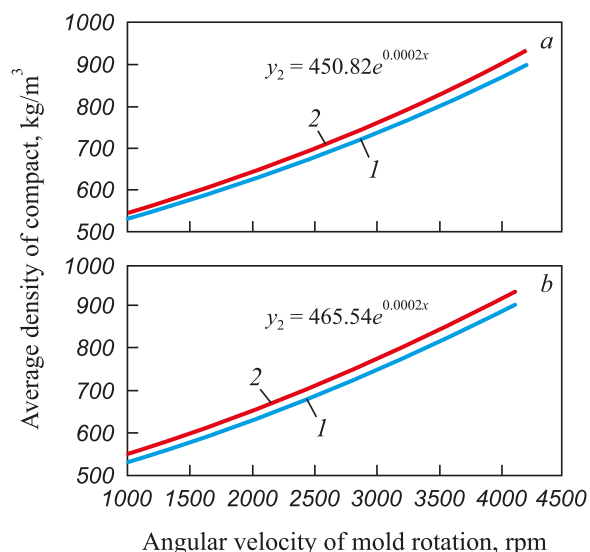


Fig. 3. Dependences of the average density of compacts made of PS50/50 material on angular velocity of the mold rotation: *a* – fraction 2.5 mm; *b* – fraction 0.63 mm

Рис. 3. Зависимости средней плотности прессовок из материала ПС50/50 от угловой скорости вращения формы: *a* – фракция 2,5 мм; *b* – фракция 0,63 мм

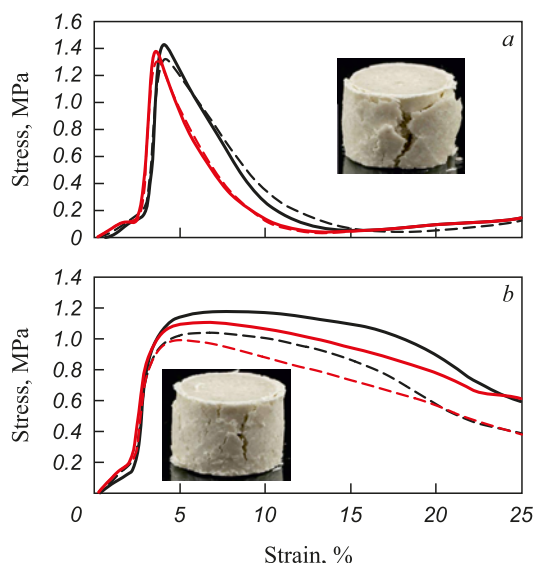


Fig. 4. Dependence of stress on deformation under vertical uniaxial compressive loading of the samples made of materials of the following grades:
a – T1; b – PS50/50

Рис. 4. Зависимость напряжения от деформации при вертикальном одноосном сжимающем нагружении образцов, выполненных из материалов марок:
a – Т1; b – ПС50/50

CONCLUSIONS

As a result of a series of computational and experimental studies, the force parameters of the compaction process under centrifugal forces were determined for powdered bodies made from fractions of the waxy pattern material PS 50/50. A comparative analysis was performed of the compressive strength values of cylindrical compacts and those of samples obtained by compaction of paraffin grade T1 in a closed mold.

The calculation method proposed by G.N. Zhdanovich, adapted to determine the stress relationships arising during vertical uniaxial compaction of PS 50/50 powder compacts in the technologically acceptable range of relative densities (0.8 – 1.0), was found to be the most suitable.

The experimental results showed that density values corresponding to compacts with porosity of $0\% \leq P \leq 10\%$, formed from PS 50/50 fractions using an attached mass, are achieved at mold rotation speeds ranging from 3500 to 4200 rpm. The points on the calculated dependences of the average density of PS 50/50 compacts on the mold rotation speed were found to be, on average, 5 % lower than the experimental data, which may be attributed to the neglect of frictional forces between the compacted material and the mold walls in the calculations.

Analysis of the experimental data also established that the compressive strength of samples formed under centrifugal forces is, on average, 15 % lower than that

of compacts obtained by single-action compaction, which, however, remains sufficient to ensure their further technological use.

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V. V. Predein – setting up experiments, setting up the calculation part.

Н. А. Богданова – проведение экспериментов, обработка экспериментальных данных, обсуждение результатов, написание статьи.

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Original article

Оригинальная статья

HARDENING OF SURFACE LAYERS OF LONG RAIL HEAD DURING LONG-TERM OPERATION

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Abstract. Starting in 2018, JSC EVRAZ United West Siberian Metallurgical Plant (EVRAZ ZSMK) produced rails of the DT400IK category with increased wear resistance and cyclic crack resistance for heavy traffic and difficult sections of track with steep curves with a radius of less than 650 m. The method of transmission diffraction electron microscopy was used to study the structural and phase states and defect substructure at different distances (0, 2, 10 mm) from the “wheel – rail” contact surface along the central axis of symmetry of the rail head (tread surface) and along the radius of rounding of the rail head (fillet) of differentially hardened long rails of the DT400IK category made of hypereutectoid steel after long-term operation on the experimental ring of Russian Railways (passed tonnage of 187 million tons). Based on the obtained structure parameters, the quantitative estimates were made of the dislocation substructure and main strengthening mechanisms (strengthening of the pearlite component, incoherent cementite particles, grain boundaries and subboundaries, dislocation substructure and internal stress fields) in various morphological components and in the material as a whole, forming the additive yield strength in the studied steel. A comparison of the quantitative parameters of the fine structure and contributions to strengthening on the tread surface and fillet was carried out. It was established that near the “wheel – rail” contact on the tread surface the prevailing morphological component is the subgrain structure, in the fillet – a ferrite-carbide mixture (completely destroyed pearlite). Strength of the rail head metal depends on the distance to the “wheel – rail” contact surface. It is shown that the main strengthening mechanisms on the tread surface are strengthening by internal stress fields, in the fillet – strengthening by incoherent particles.

Keywords: rails, fillet, tread surface, electron microscopy, hypereutectoid steel, phase composition, fine structure parameters, strengthening mechanisms

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УПРОЧНЕНИЕ ПОВЕРХНОСТНЫХ СЛОЕВ ГОЛОВКИ ДЛИННОМЕРНЫХ РЕЛЬСОВ ПРИ ДЛИТЕЛЬНОЙ ЭКСПЛУАТАЦИИ

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Аннотация. Начиная с 2018 года на АО «ЕВРАЗ Объединенный Западно-Сибирский металлургический комбинат» производятся рельсы категории ДТ400ИК повышенной износостойкости и циклической трещиностойкости для тяжеловесного движения и сложных участков пути с крутыми кривыми радиусом менее 650 м. Методом просвечивающей дифракционной электронной микроскопии изучены структурно-фазовые состояния и дефектная субструктура на разных расстояниях (0, 2, 10 мм) от поверхности контакта «колесо – рельс» вдоль центральной оси симметрии головки рельса («поверхность катания») и по радиусу скругления головки рельса («выкружка») дифференцированно закаленных длинномерных рельсов категории ДТ400ИК из заэвтектоидной стали после длительной эксплуатации на экспериментальном кольце РЖД (пропущенный тоннаж 187 млн т). На основании полученных параметров структуры выполнены

количественные оценки дислокационной субструктуры и основных механизмов упрочнения (упрочнение перлитной составляющей, некогерентными частицами цементита, границами зерен и субграницами, дислокационной субструктурой и внутренними полями напряжений), в различных морфологических составляющих и в целом по материалу, формирующих аддитивный предел текучести в исследуемой стали. Проведено сравнение количественных параметров тонкой структуры и вкладов в упрочнение на «поверхности катания» и «выкружке». Вблизи контакта «колесо – рельс» на поверхности катания преобладающей морфологической составляющей является субзеренная структура, в выкружке – феррито-карбидная смесь (полностью разрушенный перлит). Прочность металла головки рельсов зависит от расстояния до поверхности контакта «колесо – рельс». Основным механизмом упрочнения на поверхности катания является упрочнение полями внутренних напряжений, в выкружке – упрочнение некогерентными частицами.

Ключевые слова: рельсы, выкружка, поверхность катания, электронная микроскопия, заэвтектоидная сталь, фазовый состав, параметры тонкой структуры, механизмы упрочнения

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INTRODUCTION

According to data from the Russian Railways (RZhD), the primary causes of rail failure and subsequent withdrawal from service are contact fatigue damage and surface wear. These defects markedly reduce rail service life and directly affect traffic safety [1]. In recent years, a pronounced increase in axle loads and operating speeds in railway transport has been observed, which makes the development of rails with enhanced operational performance particularly relevant [2 – 5]. During service, the surface layers of rails undergo substantial structural and phase transformations [5 – 7]. These changes are accompanied by high microhardness values, decarburization [8 – 10], and the progression of relaxation, recrystallization, and related processes, which ultimately lead to the degradation of mechanical properties [11 – 13].

Increasing the carbon content in rails beyond 0.8 wt. % reduces the interlamellar spacing and promotes the formation, in the surface layers, of a subgrain structure with a high fraction of low-angle boundaries [14 – 17]. This approach can therefore be regarded as one of the promising strategies for mitigating contact fatigue [18 – 21].

Since 2018, Russia has been producing long, differentially hardened special-purpose rails with enhanced wear resistance and contact fatigue endurance – DT400IK category rails made of hypereutectoid steel – at EVRAZ ZSMK. These rails are intended for operation at speeds of up to 200 km/h on both straight and curved track sections without restrictions on traffic load intensity [22 – 24].

The scientific literature contains very few studies addressing rails made of hypereutectoid steel; existing works mainly report qualitative changes [25 – 30]. It is known that for rails with a carbon content below 0.8 wt. %, long-term operation leads to more pronounced evolution of the proportions of various morphological structure types, fine structure parameters, and cementite content at the fillet than at the tread surface [31; 32].

The aim of the present study is to provide a comparative assessment of the quantitative fine structure parameters and deformation strengthening mechanisms in the surface layers of the rail head – specifically, the tread surface and the fillet – of hypereutectoid steel rails after long-term operation (passed tonnage: 187 million tons, gross).

MATERIALS AND METHODS

The samples were taken from differentially hardened DT400IK category rails made of E90KhAF steel manufactured by EVRAZ ZSMK. The rails had accumulated a passed tonnage of 187 million tons (gross) in service on the experimental ring of Russian Railways (RZhD) in Shcherbinka. The internal structure and phase composition were examined in these samples. According to GOST 5185–2013 and TU 24.10.75111-298-057576.2017, the chemical composition of E90KhAF rail steel was (wt. %): 0.92 C, 0.4 Si, 1.0 Mn, 0.3 Cr, 0.14 V, balance Fe.

Two sets of the samples, *A* and *B*, were cut from the rails (Fig. 1). Set *A* was sampled along the central axis of symmetry of the rail head, corresponding to the tread surface (Fig. 1). Set *B* was taken along the radius of rounding of the rail head, i.e., from the fillet (Fig. 1). The samples from both sets were produced by electrical discharge cutting at identical distances from the wheel – rail contact surface, namely 0 mm (the uppermost layer of the contact surface), 2 and 10 mm below the surface. The studies were carried out using transmission diffraction electron microscopy (TEM) on thin foils, employing a JEM-2100 transmission electron microscope (JEOL, Japan) at operating magnifications ranging from 15,000 to 500,000[×].

For each sample, the morphological components of the microstructure were classified, the phase composition was established, and the key fine structure parameters were quantified, including the volume fractions

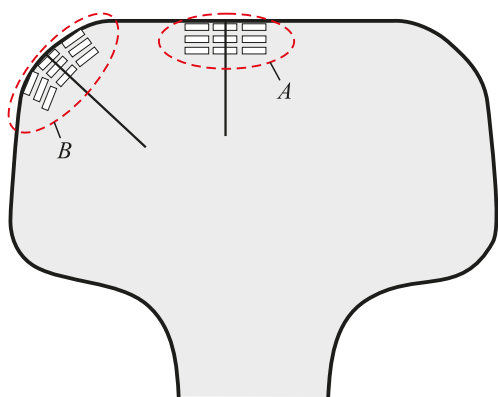


Fig. 1. Scheme of foil preparation during examination by electron diffraction microscopy at distances of 0, 2, 10 mm from the surface along the central axis (A) and the radius of rounding of the fillet (B)

Рис. 1. Схема подготовки фольг при исследовании методом электронной дифракционной микроскопии на расстояниях 0, 2 и 10 мм от поверхности по центральной оси (A) и радиусу скругления выкружки (B)

of the identified components. The distribution of the carbide phase (cementite) was also mapped, and for each location the particle shape, particle size, interparticle spacing, and volume fraction were determined. Dislocation characteristics were quantified as the scalar dislocation density (ρ) and the excess dislocation density ($\rho_{\pm}$), together with the amplitudes of internal stresses: (σ_F the shear (“forest”) stress generated by the dislocation structure, and σ_L the long-range (local) stress associated with regions of excess dislocation density. All quantitative fine structure parameters were determined both for each morphological component and for the material as a whole, followed by statistical processing. The methodology for determining these quantitative parameters is described in detail in [33; 34]. On the basis of the obtained data, and according to the authors of [33; 35; 36], the principal strengthening mechanisms contributing to the formation of the yield strength of the investigated steel were evaluated for each sample set.

RESULTS AND DISCUSSION

Previous studies [33; 34; 36] showed that, at a depth of 10 mm from the wheel – rail contact surface along the central axis of symmetry (the tread surface), the microstructure of the steel after long-term operation is dominated by pearlite with different morphologies. This includes lamellar (ideal) pearlite with nearly parallel α -phase and cementite lamellae, partially destroyed (defective) pearlite with bent and locally broken cementite lamellae, and globular pearlite. Together, these pearlitic constituents account for about 80 % of the structure. The remaining 20 % is represented by fragmented lamellar pearlite, in which dislocation walls form transverse to the α -phase lamellae; the average fragment size

is approximately 90×420 nm. Representative micrographs of these morphological components are given in [33; 34; 36].

With decreasing distance to the wheel – rail contact surface, pearlite undergoes progressive degradation, while the fragmented structure becomes more refined and its characteristic dimensions decrease. Under these conditions, a subgrain structure forms and develops rapidly (Fig. 2, a). At the contact surface, this structure consists predominantly of dislocation-free subgrains with an average size of about 80 nm, occupying nearly 90 % of the material volume.

Service exposure is accompanied by the breakup and redistribution of cementite particles and by a slight increase in both the scalar and excess dislocation densities, followed by their decrease as the dislocation-free subgrain structure begins to develop intensively. The fragmenta-

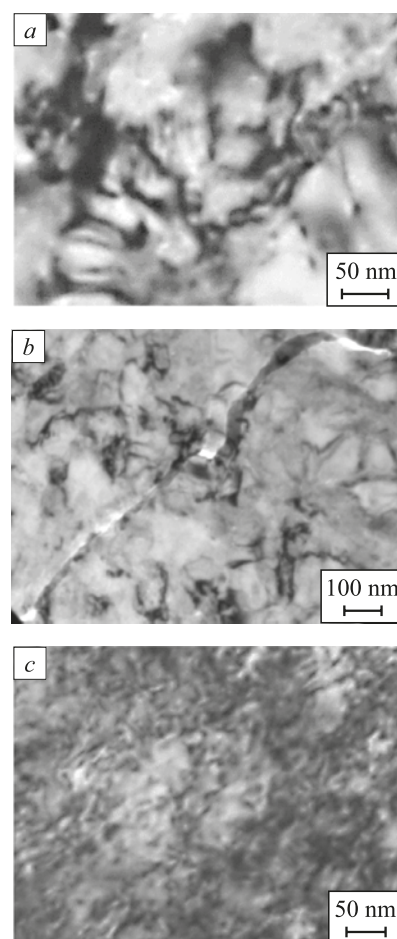


Fig. 2. TEM images of the subgrain structure (a) and microcracks in the subgrain structure (b) on the tread surface and ferrite-carbide mixture (c) in the fillet in E90KhAF steel after long-term operation (wheel – rail contact surface)

Рис. 2. ПЭМ-изображения субзеренной структуры (a) и микротрещины в субзеренной структуре (b) на «поверхности катания» и феррито-карбидной смеси (c) в «выкружке» в стали Э90ХАФ после длительной эксплуатации (поверхность контакта «колесо – рельс»)

tion of lamellar pearlite and the formation of the subgrain structure lead to elastic distortion of the α -phase crystal lattice. As a result, the amplitude of the long-range internal stresses σ_L , associated with regions of excess dislocation density, becomes approximately 4.5 times higher than that of the internal shear stresses σ_F generated by the dislocation structure. At the contact surface, the elastic component of these long-range stresses exceeds the plastic component by more than an order of magnitude, which

provides the primary driving force for microcrack initiation within the subgrain structure (Fig. 2, b). The evolution of the average quantitative fine structure parameters with decreasing distance from the wheel–rail contact surface is summarized in Fig. 3, a – d.

In contrast to the central axis of symmetry (the tread surface), the steel structure along the radius of rounding of the rail head (the fillet) at the same depth from the wheel–rail contact surface (10 mm) contains, in addi-

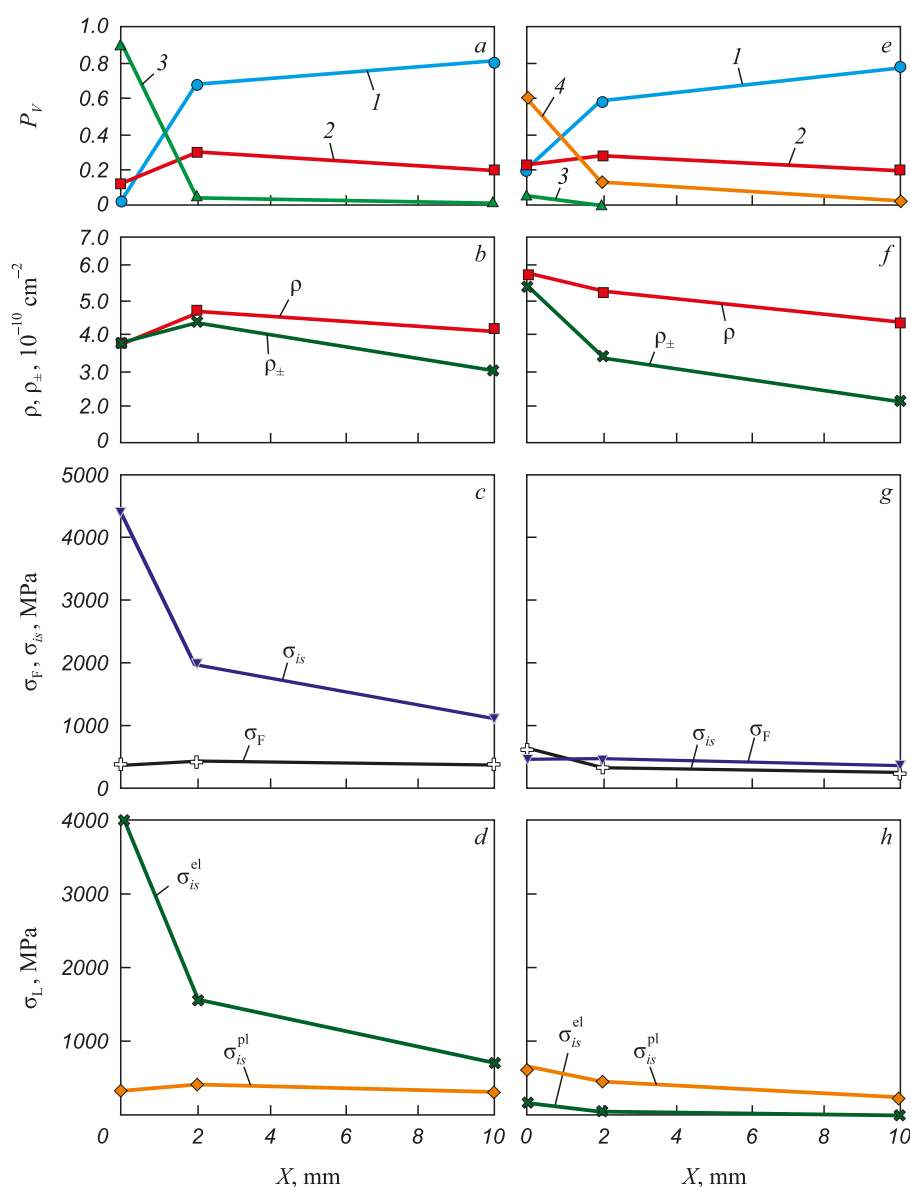


Fig. 3. Dependences of quantitative parameters of fine structure on the distance to wheel – rail contact surface on the tread surface (a – d) and fillet (e – h):

a, e – volume fractions of morphological components P_v (1 – pearlite of different morphology; 2 – fragmented lamellar pearlite; 3 – subgrain structure; 4 – ferrite-carbide mixture); b, f – scalar ρ and excess $\rho_{\perp}$ dislocation density; c, g – amplitudes of internal stresses (shear σ_F and long-range σ_{is}); d, h – components of long-range stresses (elastic σ_{is}^{el} and plastic σ_{is}^{pl})

Рис. 3. Зависимости количественных параметров тонкой структуры от расстояния до поверхности контакта «колесо – рельс» на «поверхности катания» (a – d) и «выкружке» (e – h):

a, e – объемные доли морфологических составляющих P_v (1 – перлит разной морфологии; 2 – фрагментированный пластинчатый перлит; 3 – субзеренная структура; 4 – феррито-карбидная смесь); b, f – скалярная ρ и избыточная $\rho_{\perp}$ плотность дислокаций; c, g – амплитуды внутренних напряжений (сдвига σ_F и дальнодействующих σ_{is}); d, h – составляющие дальнодействующих напряжений (упругой σ_{is}^{el} и пластической σ_{is}^{pl})

tion to pearlite of various morphologies and fragmented lamellar pearlite, an additional morphological component. The volume fractions of pearlite and fragmented lamellar pearlite at this depth are close to those observed at the tread surface (78 and 20 %, respectively). The additional component, present in a small amount (approximately 2 %), is a ferrite–carbide mixture, representing regions with completely destroyed pearlite colonies (Fig. 2, *c*). According to diffraction analysis data [33], these regions contain fine needle-shaped cementite particles with an average size of 10×25 nm and exhibit a high scalar dislocation density. With decreasing distance to the wheel – rail contact surface, microstructural evolution at the fillet follows trends similar to those observed at the tread surface. Pearlite undergoes intensive destruction, the fragmented structure becomes increasingly refined with a reduction in fragment size, and the ferrite – carbide mixture progressively expands. At the contact surface, this component accounts for approximately 60 % of the material volume. Pearlite (lamellar, destroyed, and globular) remains present, while a subgrain structure appears in a small fraction (approximately 2 %). The evolution of the volume fractions of the morphological components in the fillet region as the contact surface is approached is shown in Fig. 3, *d*.

As at the tread surface, the dislocation structure within all morphological components consists of either randomly distributed dislocations or dislocation networks. The scalar dislocation density (ρ) increases in all components as the wheel – rail contact surface is approached. The highest values of ρ are observed in the ferrite – carbide mixture (the completely destroyed structure), whereas the lowest values correspond to the subgrain structure. Because, at the contact surface, the ferrite – carbide mixture occupies 60 % of the material volume and the subgrain structure only 2 %, the average scalar dislocation density of the material is governed primarily by the ferrite – carbide mixture. As a result, unlike the tread surface, the average value of ρ increases toward the contact surface (Fig. 3, *f*). The curvature – torsion of the α -phase crystal lattice also increases toward the contact surface. Accordingly, the excess dislocation density rises at an even higher rate and rapidly approaches the value of ρ (Fig. 3, *f*). This behavior is attributed to the appearance of an elastic component in lattice bending – torsion within the ferrite – carbide mixture, fragmented lamellar pearlite, and subgrain structure. As a consequence, the amplitude of the long-range internal stresses σ_{is} exceeds the shear stress σ_F (Fig. 3, *h*), with the elastic component of σ_{is} dominating over the plastic component, as also observed at the tread surface.

Using the obtained quantitative fine structure parameters, the principal strengthening mechanisms of hypereutectoid rail steel were analyzed and compared at different distances from the contact surface after long-term

operation, both along the central axis of symmetry of the rail head (the tread surface) and along the radius of rounding of the rail head (the fillet). The following contributions were considered: strengthening due to pearlite $\Delta\sigma_{pearl}$ (barrier braking within pearlite colonies); strengthening by incoherent cementite particles $\Delta\sigma_{or}$ (Orowan bypass mechanism); strengthening by grain boundaries and subboundaries $\Delta\sigma_{sub}$ (substructural strengthening associated with intraphase boundaries); strengthening by the dislocation substructure $\Delta\sigma_F$ (forest dislocation strengthening, i.e., internal shear stress); and strengthening by internal stress fields $\Delta\sigma_L$ (long-range stress-field strengthening). Quantitative evaluation of these contributions was performed using the relationships given in [33 – 36]. The results are summarized in Fig. 4.

The analysis shows that, irrespective of the sampling direction, the strength of the rail metal is governed by the distance from the wheel – rail contact surface (Fig. 3). As the contact surface is approached, all major strength characteristics increase, with the most pronounced strengthening confined to a near-surface layer no thicker than 2 mm. At greater depths, the strength properties of the steel remain close to those of the initial-state steel. Along the tread surface, strengthening is dominated by internal long-range (local) stresses, primarily of an elastic nature, together with substructural strengthening and strengthening by incoherent particles. This reflects the fact that, at the contact surface, the subgrain structure occupies nearly the entire material volume (90 %). The nanometer-scale size of the subgrains (approximately 80 nm) results in a high density of subboundaries and junctions, predominantly triple junctions, which act as sources of bending extinction contours. These are mainly elastic in nature and give rise to high long-range internal stresses, whose elastic component exceeds the plastic component by more than an order of magnitude.

At the contact surface of the fillet, strengthening is governed primarily by incoherent particles, together with contributions from internal long-range (local) stresses and internal shear stresses associated with forest dislocations. This behavior arises because the main morphological components controlling strengthening are the ferrite–carbide mixture, which occupies 60 % of the material volume and has a low boundary density, and fragmented lamellar pearlite, accounting for 20 %. Although the subgrain structure formed at the contact surface increases the number of grain junctions and, consequently, the contribution of $\Delta\sigma_L$, its overall effect on strengthening remains limited because of its small volume fraction (2 %).

CONCLUSIONS

A quantitative assessment of the fine structure and strengthening mechanisms of hypereutectoid rail steel

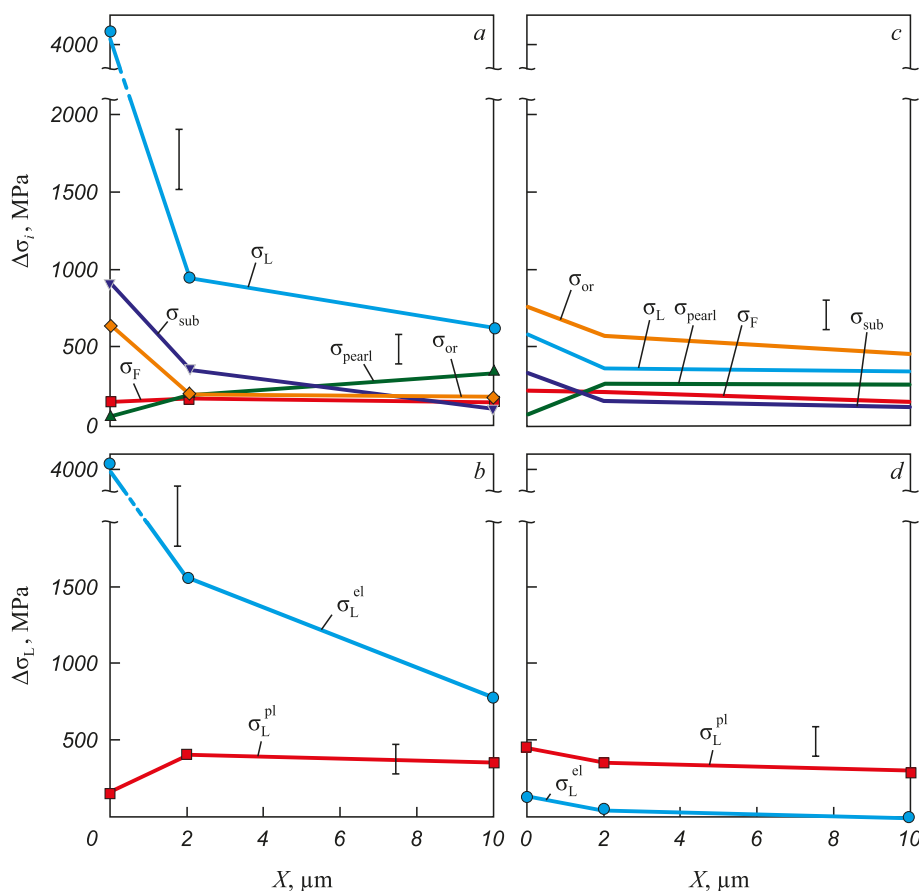


Fig. 4. Contributions of the main strengthening mechanisms $\Delta\sigma_i$ to yield strength of E90KhAF steel at different distances from the wheel – rail contact surface on the tread surface (a) and fillet (c) and contributions of the plastic $\Delta\sigma_{\text{pl}}$ and elastic $\Delta\sigma_{\text{el}}$ components of the internal long-range stresses $\Delta\sigma_L$ on the tread surface (b) and fillet (d)

Рис. 4. Вклады основных механизмов упрочнения $\Delta\sigma_i$ в предел текучести стали Э90ХАФ на различных расстояниях от поверхности контакта «колесо – рельс» на «поверхности катания» (a) и «выкружке» (c) и вклады пластической $\Delta\sigma_{\text{pl}}$ и упругой $\Delta\sigma_{\text{el}}$ составляющих внутренних дальнедействующих напряжений $\Delta\sigma_L$ на «поверхности катания» (b) и «выкружке» (d)

was performed for individual morphological components and for the material overall at various distances from the wheel – rail contact surface, considering both the tread surface (along the central axis of symmetry of the rail head) and the fillet (along the radius of rounding of the rail head), after a passed tonnage of 187 million tons (gross).

A clear microstructural contrast was observed depending on where the gradient-affected layers develop within the rail head – at the tread surface versus the fillet. Near the wheel – rail contact surface, the tread surface is dominated by a subgrain structure, whereas the fillet is characterized primarily by a ferrite – carbide mixture. This difference leads to distinct prevailing strengthening mechanisms: at the tread surface, strengthening is governed mainly by internal stress fields, while at the fillet it is driven predominantly by incoherent particles.

Regardless of the sampling direction, the strength of the rail metal varies systematically with distance from the wheel – rail contact surface. The strongest strengthening response is confined to the near-surface layer no

thicker than 2 mm. Deeper into the rail head, the steel largely retains strength levels comparable to those of the initial state steel.

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Original article

Оригинальная статья

PRODUCTION OF CARBIDE STEELS BASED ON HIGH-SPEED STEEL BY INDUCTION SURFACING

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Abstract. The work is devoted to the study of the possibility of obtaining carbide steel based on powdered high-speed steel 10R6M5 with additives of tungsten (WC) and titanium (TiC) carbides by induction surfacing. The selected compositions of the deposited charge and the proposed composition of the flux based on fused borax with additives of boric acid and a number of oxides satisfy the technology. The developed technology includes a flux, a method of briquetting charge using a piston device that minimizes the movement of ferromagnetic components of the charge under the influence of inductor electromagnetic field during surfacing. Deposited layers of carbide steel based on high-speed steel reinforced with tungsten and titanium carbides were produced and studied. The obtained layers were analyzed using optical and electron microscopy (using a microanalyzer), phase composition of the deposited layers was controlled by the X-ray phase method, and hardness of the layers was measured by the Rockwell method. Addition of tungsten carbide to powdered high-speed steel leads to the formation of ledeburite structure during surfacing, which is characteristic of high-tungsten high-speed steels. An increase in the amount of tungsten carbide in the carbide steel leads only to its partial melting in liquid steel, which helps to preserve the particles of introduced carbides in the microstructure. Titanium carbide added to the carbide steel composition significantly changes the morphology of ledeburite precipitates. According to X-ray phase analysis data, a number of carbides of $Me_{12}C$, Me_6C , Me_2C and MeC types were observed in the composition of carbide steels, which are characteristic of carbide steels obtained by various methods (plasma surfacing, sintering, impregnation of a carbide frame, etc.). It is shown that hardness of the samples of carbide steels with additives of tungsten and titanium carbides varies from 59 to 63 HRC, depending on the composition and technological modes of surfacing.

Keywords: carbide steel, induction surfacing, structure, high-speed steel, tungsten carbide, titanium carbide

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ПОЛУЧЕНИЕ КАРБИДОСТАЛЕЙ НА ОСНОВЕ БЫСТРОРЕЖУЩЕЙ СТАЛИ МЕТОДОМ ИНДУКЦИОННОЙ НАПЛАВКИ

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Аннотация. Работа посвящена исследованию возможности получения карбидостали на основе порошковой быстрорежущей стали 10Р6М5 с добавками карбидов вольфрама (WC) и титана (TiC) методом индукционной наплавки. Подобранные составы наплавляемой шихты и предложенный состав флюса на основе плавящейся буре с добавками борной кислоты и ряда оксидов удовлетворяют технологии. Разработанная технология включает в себя флюс, способ брикетирования шихты с помощью поршневого устройства, минимизирующего перемещение ферромагнитных компонентов шихты под влиянием электромагнитного поля индуктора в процессе наплавки. Получены и исследованы наплавленные слои карбидостали на основе быстрорежущей стали с упрочнением карбидами вольфрама и титана. Исследование полученных слоев проводилось методами оптической и электронной микроскопии (с применением микроанализатора), рентгенофазовым методом контролировался фазовый состав наплавленных слоев, твердость слоев измерялась методом Роквелла. Добавление карбида вольфрама в порошковую быстрорежущую сталь приводит при наплавке к формированию ледебуритной структуры, характерной для высоковольфрамистых быстрорежущих сталей. Повышение количества карбида вольфрама в составе карбидостали приводит только к частичному его расплавлению в жидкой стали, что способствует сохранению в микроструктуре частиц внесенных карбидов. Карбид титана, добавленный в состав карбидостали, существенно меняет морфологию ледебуритных выделений. По данным рентгенофазового анализа в составе карбидосталей наблюдается ряд карбидов типа $Me_{12}C$, Me_6C , Me_2C и MeC , свойственных карбидосталям, полученным различными методами (плазменной наплавкой, спеканием, пропиткой карбидного каркаса и др.). Показано, что твердость образцов карбидосталей с добавками карбидов вольфрама и титана варьируется от 59 до 63 HRC, в зависимости от состава и технологических режимов наплавки.

Ключевые слова: карбидосталь, индукционная наплавка, структура, быстрорежущая сталь, карбид вольфрама, карбид титана

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INTRODUCTION

The development of new structural and tool materials with improved physicomechanical and operational properties has become increasingly important. This need arises from the limited potential to further enhance the characteristics of well-known materials such as wear-resistant [1 – 3] and tool steels (including high-speed steels), as well as hard alloys [4 – 6].

Carbide steels, consisting of refractory carbides (most often tungsten and titanium carbides) and steels, represent a new class of promising materials. They occupy an intermediate position between hard alloys and steels, combining the properties of both the carbide reinforcing phase and the steel matrix [7 – 9].

Carbide steels are most commonly produced by powder metallurgy methods, including sintering of compacted powders, impregnation of a carbide frame with steel, hot pressing, or hot extrusion [7]. However, these processes involve numerous complex technological operations, which limit their practical use. Alternative approaches such as plasma [10 – 13] or laser [14; 15] have also been investigated, but these methods are constrained by the high cost of laser systems, loss of powder due to plasma jet scattering, and consumption of expensive gases. The key factor limiting the widespread application of carbide steels remains the complexity of conventional production technologies, which require sophisticated equipment and long process cycles.

Induction surfacing of metallic layers [16 – 20], based on high-frequency current heating, has recently become a promising alternative. Under the multifactor influence of the inductor electromagnetic field on the metal substrate, flux, and initial charge, a multilayer composite is formed, in which a surface layer may be formed that exhibits a set of improved properties, including wear resistance, acid resistance, heat resistance, and so on. This method is characterized by relatively low equipment costs, simplicity of implementation, and short processing times.

Moreover, the surfacing process can be partially combined with heat treatment of the deposited layer if necessary. The present study aimed to investigate the possibility of producing carbide steel by induction surfacing.

The main objectives were to:

- select suitable compositions of the deposited charge based on high-speed steel with tungsten and titanium carbide additives for induction surfacing;

- develop a flux composition for induction surfacing of carbide steel;

- produce deposited layers of carbide steel based on high-speed steel reinforced with tungsten and titanium carbides on steel substrates;

- analyze the microstructure and properties of the obtained samples.

MATERIALS AND METHODS

Powdered high-speed steel 10R6M5 was used as the main component of the charge for producing the deposited layers. To obtain carbide steel, the powdered steel was mixed on an organic binder with tungsten carbide (WC) and titanium carbide (TiC) powders in various proportions (5 – 20 wt. % relative to steel). The upper limit was deliberately set below the conventional range (20 – 70 wt. %) [7], because induction surfacing is characterized by considerably shorter heating times compared with traditional processes, and, consequently, the available time for interaction between the matrix and the reinforcing phase is also reduced. Consequently, excessive amounts of reinforcing carbides were avoided to ensure effective component interaction and a defect-free structure.

Flux plays an essential role in induction surfacing, protecting both the molten metal and the surface of the steel substrate from oxidation by atmospheric oxygen [9]. The flux mixture consisted of powdered fused borax, boric acid, and oxide additives of silicon, magnesium, calcium, and sodium.

During flux selection, the influence of the magnetic field generated in the surfacing zone on the charge was taken into account. One of the challenges in producing carbide steel based on high-speed steel is the ferromagnetism of powdered steel, which interacts strongly with the electromagnetic field at the initial heating stage (before the transition to a paramagnetic state). To prevent displacement of the powder charge under these conditions, the mixture was compacted into briquettes. The flux also acts as a kind of binder, holding the charge particles together within the temperature range in which the flux was already molten while the metallic part of the charge had not yet fused.

A piston-type compaction system proved most effective for briquetting. The mixed charge components were placed into a container pre-lubricated with a plasticizer based on an organic compound, which reduced adhesion

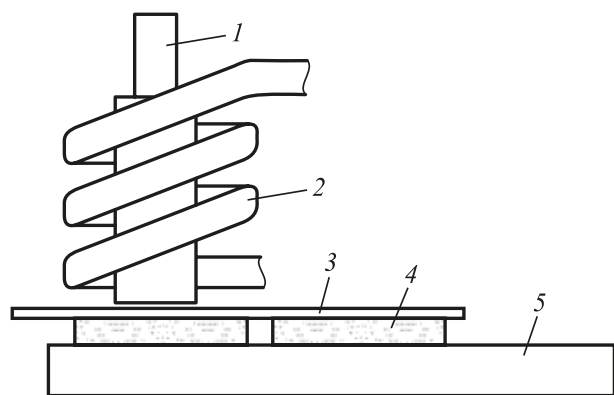


Fig. 1. Scheme of induction surfacing:

1 – ferrite core; 2 – coil inductor; 3 – asbestos gasket;
4 – deposited briquette; 5 – metal substrate in the form of a plate

Рис. 1. Схема индукционной наплавки:

1 – ферритный сердечник; 2 – витковый индуктор;
3 – асбестовая прокладка; 4 – наплавляемый брикет;
5 – металлическая подложка в форме пластины

of the components to the container walls and the piston. Compaction was performed by applying pressure with the piston, during which excess binder and plasticizer could be released. The briquettes were then dried for at least 2 h at 80 °C.

Plates of medium-carbon structural steel 45 were used as substrates.

Surfacing of the plates (Fig. 1) was carried out using a high-frequency induction unit of the UVG 2-25 type equipped with a GNOM-25M1 generator, with a power output of up to 20 kW and an operating frequency of 44 to 66 kHz. A coil inductor with a water-cooled ferrite core was used. To fix the briquettes at the initial stage of surfacing and protect the inductor, an asbestos gasket was placed between the briquette and the substrate (Fig. 1).

Microstructural studies were conducted using a Carl Zeiss Axio Observer.D1 optical microscope and a Hitachi TM4000 scanning electron microscope with a micro-analyzer. The phase composition was examined by X-ray phase analysis using a Bruker diffractometer with copper radiation. The hardness of the deposited layers was measured by the Rockwell method.

RESULTS AND DISCUSSION

Microstructural analysis of the deposited layers revealed that the overall structure corresponds to the characteristic features of cast high-speed steel. All samples contained ledeburite eutectic of varying morphology depending on the composition and carbide content in the charge. A solid solution based on austenite was also present, its composition varying widely depending on the alloy composition.

When powdered high-speed steel 10R6M5 was surfaced without additives (Fig. 2, *a*), a typical cellular structure was observed, with small inclusions of ledeburite eutectic of feather-like morphology. The addition of a small amount of tungsten carbide led to its almost complete dissolution in the melt; the resulting solidified structure resembled that of high-speed steel without additives (Fig. 2, *b*), but with a higher amount of feather-like ledeburite eutectic.

A further increase in the amount of added carbides leads to the formation of ledeburite eutectic with the so-called skeletal morphology, characteristic of high-tungsten high-speed steel of the R18 type (Fig. 2, *c*). This is explained by the dissolution of the added carbides in the liquid steel during surfacing. However, because of the short duration of the process, the solubility limit is apparently governed not so much by the constraints of the phase diagram as by the insufficient time for complete interaction. As a result, a structure may be identified in which, along with the skeletal eutectic, clusters of undissolved tungsten carbide particles with a characteristic angular shape are observed (Fig. 2, *d*).

According to X-ray phase analysis, the structure contained austenite, martensite, cementite, and ledeburite with carbides of the $\text{Fe}_6\text{W}_6\text{C}$ and $\text{Fe}_3\text{W}_3\text{C}$ types. In addition, X-ray phase analysis revealed carbide inclusions of the W_2C and WC types, the latter representing partially undissolved particles of the introduced carbide phase remaining in the solid solution in samples with a relatively high content of added carbides.

Hardness tests showed that deposited 10R6M5 layers without reinforcement exhibited 60 – 61 HRC, while those with tungsten carbide additions reached 61 – 63 HRC.

Producing carbide steel based on high-speed steel reinforced with titanium carbide (TiC) proved more difficult. This was due to the specific nature of the wetting of titanium carbide by molten steel, as well as mutual dissolution between the matrix and the reinforcing phase, among other factors [7]. Nevertheless, deposited layers of carbide steel with TiC were successfully obtained. Their microstructures (Fig. 3) varied depending on TiC content.

At low TiC concentrations (Fig. 3, *a*), the structure consisted of austenitic cells with a carbide network and visible martensite. No skeletal ledeburite was observed. As the TiC content increased, the deposition quality decreased and the tendency to pore formation became more pronounced; inclusions of ledeburite appeared with an intricate “Arabic script” morphology (Fig. 3, *b*). These features are apparently due to the ability of molten steel to dissolve only a limited amount of the introduced carbides and from the specific solidification behavior of the melt.

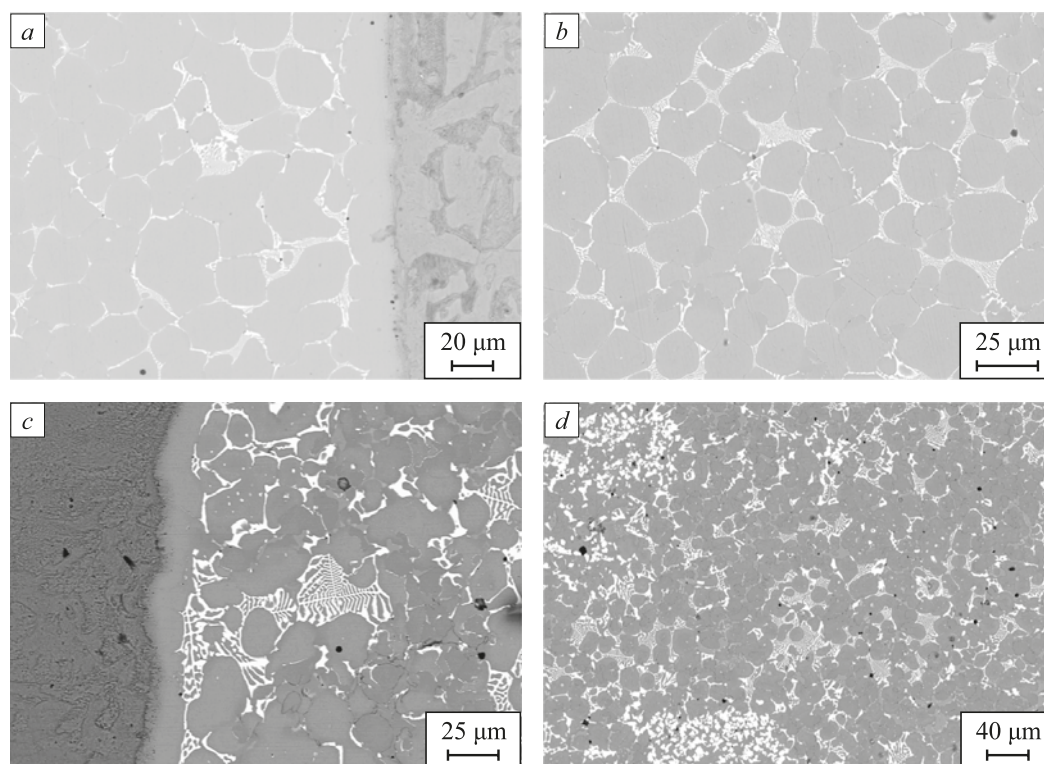


Fig. 2. Microstructure of deposited layers:

a – high-speed steel 10R6M5 without additives; *b* – *d* – carbide steel with 5, 10, 20 wt. % WC

Рис. 2. Микроструктура наплавленных слоев:

a – быстрорежущая сталь 10Р6М5 без добавок; *b* – *d* – карбидосталь с 5, 10 и 20 мас. % WC

X-ray phase analysis of TiC-containing samples revealed austenite, martensite, cementite, and ledeburite with carbides of $\text{Fe}_6\text{W}_6\text{C}$ and $\text{Fe}_3\text{W}_3\text{C}$ types. Alongside titanium carbide TiC, carbide inclusions of W_2C were also identified.

The hardness of these samples ranged from 59 to 63 HRC.

CONCLUSIONS

Compositions of the deposited charge suitable for induction surfacing based on high-speed steel 10R6M5 with up to 20 wt. % tungsten carbide (WC) and titanium carbide (TiC) were selected. A flux composition for induction surfacing of carbide steel was developed,

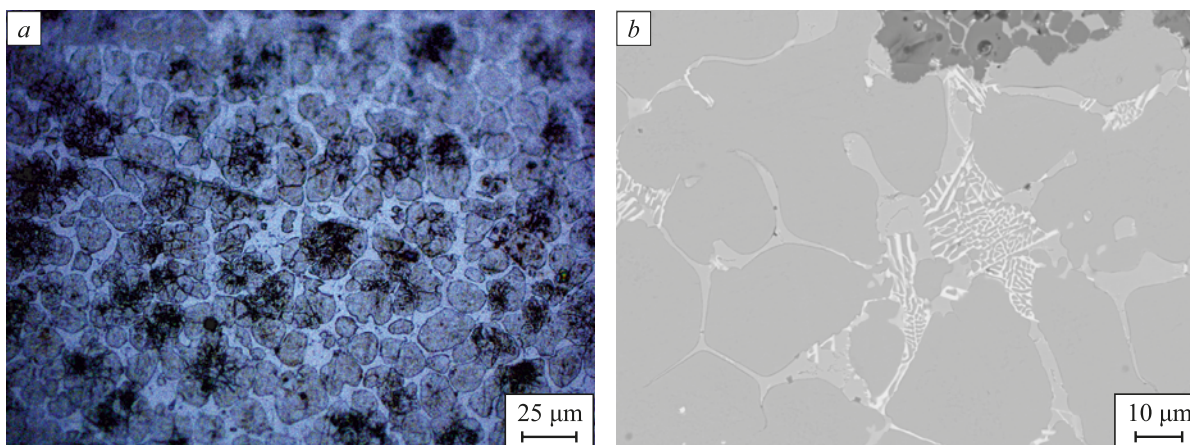


Fig. 3. Structure of carbide steel 10R6M5 – TiC:

a – 10 wt. % TiC (optical microscope); *b* – 20 wt. % TiC (electron microscope)

Рис. 3. Структура карбидостали: сталь 10Р6М5 – TiC:

a – 10 мас. % TiC (оптический микроскоп); *b* – 20 мас. % TiC (электронный микроскоп)

comprising borax and boric acid as the base components, with oxide additives of silicon, magnesium, calcium, and sodium. A complete induction surfacing technology was designed and successfully applied to produce deposited layers of carbide steel based on high-speed steel reinforced with tungsten and titanium carbides on steel 45 substrates. The microstructure of the obtained samples contained austenite, martensite, cementite, and a number of special carbides, including those of the Me_6C , Me_2C , and MeC types. The hardness of the deposited carbide steel layers varied from 59 to 63 HRC, depending on the composition of the initial charge.

Thus, the feasibility of producing carbide steel based on powdered high-speed steel 10R6M5 with tungsten and titanium carbide additives by induction surfacing has been demonstrated.

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A. M. Tokmin – formulation of the problem, analysis of the samples, analysis of results, formulation of conclusions.

O. A. Masanskii – production, preparation and analysis of the samples; formulation of conclusions.

Вклад авторов

С. А. Климов – разработка технологии наплавки; получение, подготовка и исследование образцов; формулирование выводов.

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Original article

Оригинальная статья

CONTACT CHARACTERISTICS OF C235 STEEL IN DRY SLIDING AGAINST C45 STEEL UNDER HIGH-DENSITY ALTERNATING CURRENT AT DIFFERENT TRANSFORMATION COEFFICIENTS OF SUPPLY TRANSFORMER

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Abstract. The authors studied the tribotechnical behavior of C235 steel under conditions of dry sliding electrical contact with a current density of more than 100 A/cm² at different transformation coefficients of the supply transformer. A decrease in the transformation coefficient leads to a decrease in the wear resistance and electrical conductivity of the contact. Metallographic methods revealed the formation of transfer layers on the contact surfaces. Thickness of the transfer layers does not exceed 20 μm. Morphological patterns of worn contact surfaces on the scale of the nominal (geometric) contact area consist of two sectors, where one sector has signs of melting. X-ray phase analysis has shown that the transfer layers contain more than 70 vol. % FeO. That is why the transfer layers could be represented as a quasi-dielectric medium, where FeO acts as a dielectric. The authors assume that strong self-induction pulses occur in the contact zone, which cause high-density displacement currents. These currents act directly on FeO ions and convert them into a melt. These concepts allow us to assert that the melt consists of atoms or ions of iron and oxygen. A decrease in the transformation coefficient (that is, an increase in the inductance of the secondary winding of the supply transformer) causes an increase in self-induction pulses and displacement currents, which leads to an increase in the amount of FeO melt, its easy removal from the contact area, and a corresponding decrease in the wear resistance and electrical conductivity of the contact. The data obtained can serve as guidelines when choosing wear-resistant materials for high-current sliding contact and, in particular, when defining its design.

Keywords: sliding electrical contact, contact electrical conductivity, wear intensity, adhesion, oxidation of friction zone, friction coefficient, melt on sliding surface, displacement current

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ХАРАКТЕРИСТИКИ КОНТАКТА СТАЛИ СТ3 ПРИ СУХОМ СКОЛЬЖЕНИИ ПО СТАЛИ 45 ПОД ПЕРЕМЕННЫМ ЭЛЕКТРИЧЕСКИМ ТОКОМ ВЫСОКОЙ ПЛОТНОСТИ ПРИ РАЗНЫХ КОЭФФИЦИЕНТАХ ТРАНСФОРМАЦИИ ПИТАЮЩЕГО ТРАНСФОРМАТОРА

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Аннотация. Авторы изучили триботехническое поведение стали Ст3 в условиях сухого скользящего электрического контакта с плотностью тока более 100 А/см² при разных коэффициентах трансформации питающего трансформатора. Снижение коэффициента трансформации приводит к снижению износостойкости и электропроводности контакта. Методами металлографии было обнаружено образование слоев переноса на контактных поверхностях. Толщины слоев переноса не превышают 20 мкм. Морфологические картины изношенных контактных поверхностей в масштабе номинальной (геометрической) площади контакта состоят из двух секторов, где один сектор имеет признаки расплава. Рентгеновский фазовый анализ показал, что слои переноса содержат более 70 об. % FeO. Именно поэтому слои пере-

носа можно представить как квази-диэлектрическую среду, где FeO выступает как диэлектрик. Авторы делают предположение, что в зоне контакта возникают сильные импульсы самоиндукции, которые вызывают токи смещения высокой плотности. Эти токи воздействуют непосредственно на ионы FeO и переводят их в расплав. Данные представления позволяют утверждать, что расплав состоит из атомов или ионов железа и кислорода. Снижение коэффициента трансформации (то есть увеличение индуктивности вторичной обмотки питающего трансформатора) вызывает усиление импульсов самоиндукции и токов смещения, что приводит к увеличению количества расплава FeO, его легкому удалению из зоны контакта и к соответствующему уменьшению износостойкости и электропроводности контакта. Полученные данные могут служить ориентирами при выборе износостойких материалов для сильноточного скользящего контакта и, в частности, при задании его конструкции.

Ключевые слова: скользящий электрический контакт, электрическая проводимость контакта, интенсивность изнашивания, адгезия, окисление зоны трения, коэффициент трения, расплав на поверхности скольжения, ток смещения

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INTRODUCTION

One of the primary objectives of tribology is to ensure satisfactory wear resistance of friction pairs under severe operating conditions. This can be achieved through the appropriate design of the friction unit, the desired environmental conditions, or the selection of appropriate friction pair materials. Dry sliding under high-density electric current is one of the extreme types of external influences on contact zone materials. Known current collector materials are typically operated under current densities of 15 – 40 A/cm² during dry sliding against a copper counterbody. Generally, known current collector materials are not used for dry sliding at current densities exceeding 60 A/cm² [1], as such sliding leads to rapid deterioration of the friction pair contact layers.

Material sliding under high-density currents (higher than 100 A/cm²) is of scientific and practical interest. A tribosystem with current collection must have contact layers whose condition ensures high wear resistance and high electrical conductivity of the contact. It is known [2] that high electrical conductivity of a metal/steel sliding contact typically corresponds to high wear resistance under high current density. Therefore, increasing the electrical conductivity of a sliding electrical contact can simultaneously increase its wear resistance. Changing the design parameters of the current collector unit can lead to improved contact characteristics.

A sliding electrical contact can be implemented by incorporating a friction unit into the secondary power winding circuit of a transformer. One of the fundamental equations of an ideal transformer can be written as

$$i_{1-xx}n_1 = i_1n_1 + i_2n_2$$

or

$$i_2 = \frac{(i_{1-xx} - i_1)n_1}{n_2} = (i_{1-xx} - i_1)k,$$

where n_1 and n_2 are the number of turns in the primary and secondary windings, respectively; i_{1-xx} is the current in the primary winding when the transformer is no-load ($i_2 = 0$); i_1 and i_2 are the currents in the primary and secondary coils when the secondary winding is loaded ($i_2 > 0$); $k = n_1/n_2$ is the turn ratio.

It is clear from this that the value of $i_2 = (i_{1-xx} - i_1)$ can formally be increased by increasing k under certain conditions. The current i_2 is the contact current ($i_2 = i_c$) and its increase at a low contact voltage drop will correspond to an increase in the contact conductivity. Therefore, the assumption that the contact current $i_2 = i_c$ increases with increasing $k = n_1/n_2$ should be verified experimentally. Some metals (tungsten, molybdenum, etc.) are not capable of sliding against steel with high contact conductivity, so they cannot serve as model materials for these experiments. C235 steel is the most convenient model material.

The aim of this study is to determine the regularities of change in the electrical conductivity of a dry sliding steel/steel electrical contact and its wear resistance at different turn ratios of the power transformer.

EXPERIMENTAL MATERIALS AND METHODS

Low-carbon steel C235 (0.2 % C) served as the material for the cold-worked samples with a diameter of 3.5 mm and a height of 8 mm. The sliding surfaces were examined using an optical microscope (OM, Axiovert 200 M). The hardness of the samples ($H_u = 2.1$ GPa) was determined using a Micro-Vickers TVM-5215-A micro-hardness tester under a load of 1 N. X-ray phase analysis of the sample contact layers was performed using a DRON-7 diffractometer in CoK_α radiation. The volume content of phases in the contact layer was determined according to a known method [3; 4], where the intensity of the X-ray wave I_{HKL-j} , scattered from the reflecting plane (HKL) of a certain crystalline j-th phase, was written as

$$I_{HKL-j} = I_0 k_0 K_{HKL-j} c_{v-j}, \quad (1)$$

where I_0 is the intensity of the X -ray wave incident on the multiphase surface; k_0 is a coefficient that takes into account the geometric parameters of the X -ray apparatus; K_{HKL-j} is a complex proportionality coefficient for the j -th phase; c_{v-j} is the volume concentration of the given j -th phase in the multiphase medium.

The qualitative phase composition and the integrated intensities of the K_{HKL-j} peaks should be found from the X -ray diffraction patterns (Fig. 1, *b*), the necessary reference data can be found in [4]. Taking into account that $\sum c_{v-j} = 1$, the volume concentrations of the phases can be found.

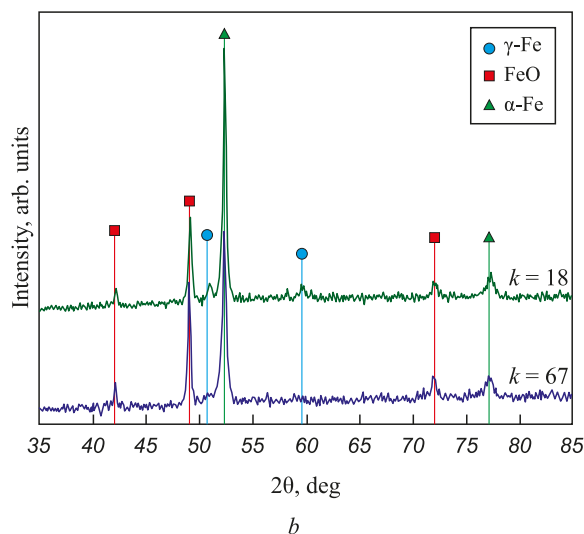
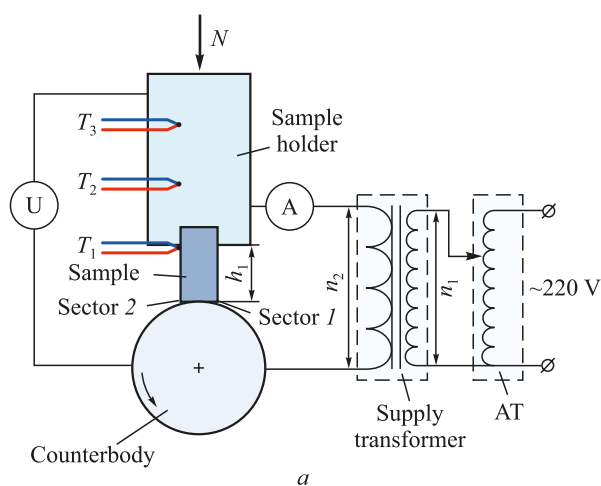


Fig. 1. Scheme of sliding electrical contact of *pin-on-ring* configuration (*a*) (AT – autotransformer; n_1 and n_2 – number of turns in the primary and secondary windings) and X -ray diffraction patterns of contact layers of C235 steel samples under friction with $k = 67$ and $k = 18$ (*b*)

Рис. 1. Схема скользящего электроконтакта типа *pin-on-ring* (*a*) (АТ – автотрансформатор; n_1 и n_2 – количество витков в первичной и вторичной обмотках) и рентгенограммы контактных слоев образцов стали Ст3 при трении при $k = 67$ и $k = 18$ (*b*)

The materials were loaded by dry friction under alternating current (50 Hz) at a contact pressure of $p = 0.13$ MPa and a sliding velocity of $v = 5$ m/s using the well known pin-on-ring configuration (Fig. 1, *a*). Chromel-Copel thermocouples T_1 , T_2 , T_3 were fixed to the sample holder with screws. The linear wear intensity was defined as $I_h = h/D$ (where h is the change in sample height per a sliding distance D). The contact current density was defined as $j = i_2/A_a$ (where i_2 is the contact current; A_a is the nominal contact area). The specific surface electrical conductivity of the contact was defined as $\sigma_A = j/U$ (where U is the contact voltage drop). The friction coefficient was determined using a ZET7111 strain gauge. Before testing, the samples were lapped against a counterbody (C45 steel ($H_\mu = 5.8$ GPa)). Each test was performed three times.

RESULTS

It is evident that the initial structure of the surface layers of C235 steel specimens before friction contains the α -Fe phase predominantly. Peak α -Fe, high-intensity FeO peaks, and low-intensity γ -Fe peaks are observed in the X -ray diffraction patterns of the contact layers of the steel specimens after friction (Fig. 1, *b*). The FeO and γ -Fe phases appeared on the surface of the specimens under the influence of friction and current. The intensities of the strongest peaks I_{200} (FeO), I_{111} (γ -Fe), and I_{110} (α -Fe) were inserted into equation (1) and the volume concentrations c_{v-j} of these phases in the contact layers of the specimens after friction were calculated for any k value (see Table). It is evident that FeO is the main phase in the contact layers. The γ -Fe concentration is low for any k value and is not of interest for discussion. The lattice parameters of the α -Fe, γ -Fe and FeO phases are generally close to the lattice parameters of the same phases from the ASTM database.

It is obvious that FeO and γ -Fe phases appeared under the influence of current, temperature and plastic deformation of the contact layers of the samples. Deformation and deterioration of the contact layers occur under frictional fatigue conditions. Current density is the main factor determining fatigue deterioration (wear) of the electrical contact zone. An increase in current density j in the contact causes an increase in the wear intensity I_h for any value of k (Fig. 2, *a, b*). The current dependence of the electrical conductivity σ_A of the contact has positive slope in the range of $j < 300$ A/cm² at $k = 67$ and in the range of $j < 100$ A/cm² at $k = 18$. At $j > 100$ A/cm² and $j > 300$ A/cm² (Fig. 2, *a, b*), a sharp increase in I_h occurs, which indicates the onset of catastrophic wear. At the same time, the slopes of the $\sigma_A(j)$ curves become negative. It is also evident that $\sigma_A(j)$ for the contact at $k = 67$ is significantly higher than for $k = 18$. But I_h is

Volumetric concentrations of phases in the contact layer of C235 steel and contact characteristics (σ_{Ac} , I_{hc} , j_c) in sliding under current with different transformation coefficients k of supply transformer (all parameters correspond to the catastrophic wear onset)

Объемные концентрации фаз в контактном слое стали Ст3 и характеристики контакта (σ_{Ac} , I_{hc} , j_c) при скольжении под током при разных коэффициентах трансформации k питающего трансформатора (все параметры соответствуют началу катастрофического изнашивания)

$k = 67$			$k = 18$		
$c_v(\text{FeO})$	$c_v(\gamma\text{-Fe})$	$c_v(\alpha\text{-Fe})$	$c_v(\text{FeO})$	$c_v(\gamma\text{-Fe})$	$c_v(\alpha\text{-Fe})$
0.77	0.02	0.21	0.80	0.03	0.17
σ_{Ac} , S/cm	I_{hc} , $\mu\text{m/km}$	j_c , A/cm ²	σ_{Ac} , S/cm ²	I_{hc} , $\mu\text{m/km}$	j_c , A/cm ²
176	44	300	19	67	100

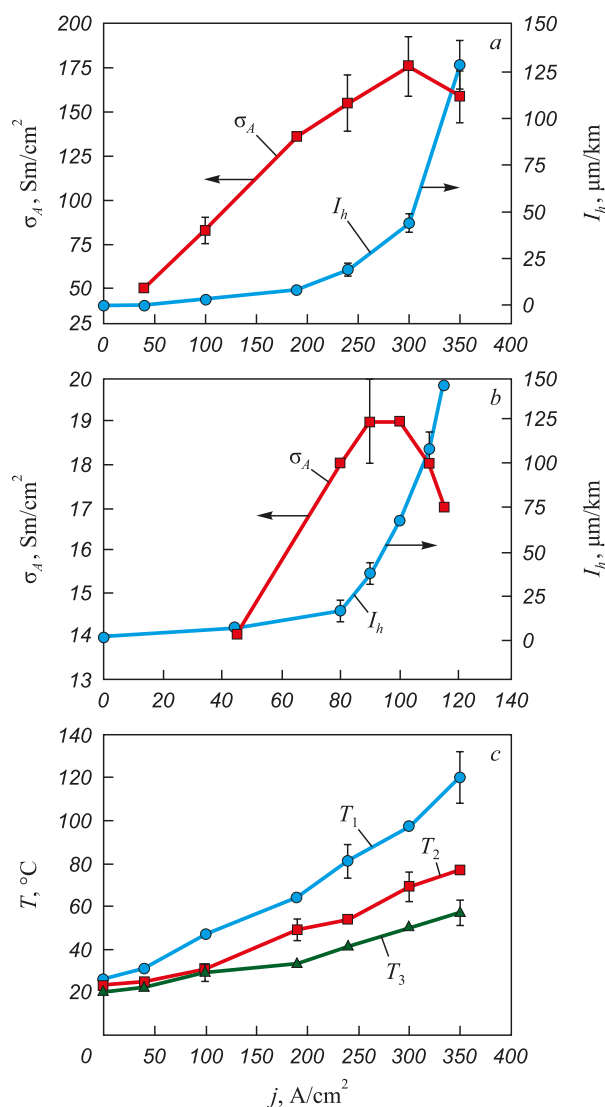


Fig. 2. Current dependences of wear intensity (I_h) and contact specific electrical conductivity (σ_A) with $k = 67$ (a), $k = 18$ (b), and temperatures of the sample holder side surface at different distances from the contact zone in sliding with $k = 67$ (c)

Рис. 2. Токовые зависимости интенсивности изнашивания (I_h) и удельной электрической проводимости (σ_A) контакта при $k = 67$ (a) и $k = 18$ (b), а также температуры боковой поверхности образца на разных расстояниях от зоны контакта при скольжении при $k = 67$ (c)

significantly lower for the contact at $k = 67$ than that for $k = 18$.

The friction coefficient depends weakly on k ; during sliding without current $f \approx 0.7$ and decreases to $f \approx 0.4$ with increasing j . The temperatures (T_1 , T_2 , T_3) of the side surface of the specimen holder are indicators of the thermal state of the specimen and the specimen holder. The $T(j)$ dependences are nonlinear (e.g., Fig. 2, c). Changing of k does not significantly affect the nature of the $T(j)$ curves or the numerical values of the temperatures, which can exceed 100 °C. This may indicate the similarity of the thermal states of the specimen contact layers during sliding at different k in the normal wear regime, that is, before the catastrophic wear onset.

The worn surfaces of the specimens have approximately the same images at any k value; namely, the contact surface is divided into two sectors with different morphological details (Fig. 3, a). Sector 1 (light area in Fig. 3, a) is formed on the frontal part of the nominal contact area of the samples, i.e. sector 1 faces the oncoming sliding surface of the counterbody (Fig. 1, a). Plastic deformation and wear in sector 1 occur due to adhesion and plowing by the counterbody asperities (Fig. 3, b), which is described for normal friction without current, for example, in [5; 6]. The contact layers in sector 2 are deformed mainly by the viscous fluid mechanism, which is shown in detail in Fig. 3, c. This should facilitate a sufficiently rapid stress relaxation. There is a certain transition zone between these sectors, more than 10 μm long (for these friction pairs), where both deformation mechanisms considered are realized simultaneously. It should be noted that the appearance of a melt in the contact zone is not accompanied by its glow. This means that the temperature of the contact zone is lower than 600 °C and the nature of the melt must be established.

DISCUSSION

It was noted above that the morphological features of the worn surfaces are identical, and the failure mechanisms of the contact surfaces do not differ sig-

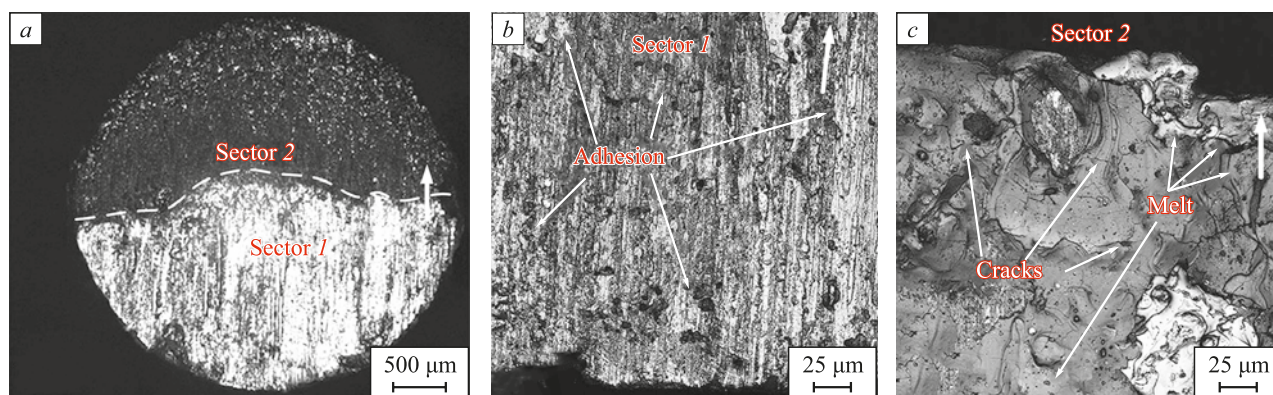


Fig. 3. Nominal contact area (a) and morphological images of worn surfaces of C235 steel samples in sector 1 (b) and in sector 2 (c) when sliding at $k = 67$ under a current density of 300 A/cm²

Рис. 3. Номинальная площадь контакта (a) и морфологические изображения изношенных поверхностей образцов стали Ст3 в секторе 1 (b) и в секторе 2 (c) при скольжении при $k = 67$ под током плотностью 300 A/cm²

nificantly. The phase compositions of the contact layers are also approximately the same (see Table). The values of the thermal power ($f_p v + jU$) of the external impact, corresponding to the catastrophic wear onset, are approximately the same, which can be calculated from Fig. 2, a, b. Obviously, the slight difference in these output parameters of a tribosystem with current collection cannot serve as a satisfactory basis for understanding the difference in the rate of contact layer failure at different values of the turn ratio k . It should be noted that sliding without current in the presence of oxides (e.g., [7 – 10]) and in the absence of oxides (e.g., [11 – 13]) in the contact zone does not lead to the formation of melt on the contact surfaces. The formation of melt was also not observed during sliding under low-density current [14 – 16] or under high-density current [17]. The steel/steel contact surfaces in the present study showed no signs of melting at $j > 700$ A/cm² in stationary contact ($v = 0$ m/s). These data and the presented observations (Fig. 2 and Table) suggest that melting occurs at a certain sliding velocity ($v > 0$ m/s), at a certain current density ($j > 0$ A/cm²), and at a certain FeO concentration ($c_{\text{FeO}} > 0$).

In general, the total current density j_0 (in any conducting circuit) and, in particular, the total current density in the contact can be written as $j_0 = j_f + j_D$ (where j_f is the current density of free charges; j_D is the bias current density (i.e., the current density of bound charges)). The transfer layers contain a dielectric (FeO) with ionic polarization; here, the bound charges are ions in FeO crystals. Obviously, an increase in j_D should cause an increase in the energies of oxygen and iron ions in FeO crystals. It should be taken into account that adhesion and roughness in any dry contact always determine the intermittent nature of sliding in the stick-slip mode. This leads to current oscillations in the contact and to corresponding self-induction pulses. Usually, the self-induction EMF (electric moving force) is written as $\mathcal{E} = -L di/dt$ (where L

is the inductance of the conducting circuit; i is the current in the conducting circuit). The design of the friction unit (Fig. 1, a) contains an inductance L in the secondary winding of the transformer supplying power to the sliding contact (where $L \sim n^2$; n is the number of turns in the winding). The appearance of an EMF pulse ($\mathcal{E}$) in the contact sets the electric field strength E in the contact, so we can approximately write $\mathcal{E} = -L di/dt \approx |E| h_0$ (where h_0 is a parameter that can characterize the interval of gradient of the electric field in the contact, m). Now, knowledge of the parameter h_0 is not important, since it is necessary to show an increase in E with an increase in L . An increase in self-induction pulses with an increase in L should cause an increase in E , $\partial E/\partial t$ and, accordingly, j_D . It should be noted that the turn ratio of the power transformer decreases with an increase in L . In addition, the voltage in the contact during self-induction pulses can significantly exceed the average voltage between the contact surfaces. These pulsed voltages determine high values of E , $\partial E/\partial t$, and corresponding j_D , capable of deteriorating the FeO crystal lattice and transferring FeO ions into the melt (Fig. 3). The highest values of j_D should be in the vicinity of the contact spots; therefore, melt should appear only at the contact spots and their vicinity, and only during the existence of the self-induction pulse. Obviously, an increase in j_D by increasing E (in particular, by increasing L) will lead to an increase in the energy of the self-induction pulse, to higher loads at the contact spots, and to more intense deterioration of the transfer layer. It is possible that relatively strong self-induction pulses, corresponding to a high inductance L , cause the formation of relatively large volumes of melt with low viscosity. The last two factors (large melt volume and its low viscosity) contribute to the acceleration of the deterioration of the transfer layer. For this reason, the melt should not be considered as a good lubricant. This means that an increase in L (i.e., a decrease in k) leads to a higher I_h (Fig. 2, a, b).

It is to be expected that the melt layer thickness is smaller than the transfer layer thickness. The presence of melt predominantly in sector 2 suggests that the FeO concentration is higher in this sector than in sector 1. This indicates a general unevenness of FeO distribution in the transfer layer. The steel/steel contact parameters presented here correspond to a circular nominal contact area. The values of these parameters are close to the values corresponding to rectangular nominal contact areas [18]. It should be noted that melt can appear at a low oxide concentration in a contact layer having two sectors (e.g. W/Mo or W/steel contacts [19], as well as steel/steel, where there is only melt [20]). Quite similar morphological types and phase compositions of transfer layers containing more than 70 vol. % FeO allow us to expect the manifestation of these features in many metal/steel contacts during sliding under current.

CONCLUSIONS

Dry sliding of C235 steel against hardened grade C45 steel was performed under alternating electric current with a density higher than 100 A/cm² while varying the turn ratio of the power transformer in this study. The turn ratio was reduced by increasing the inductance of the transformer's secondary supply winding.

A decrease in the turn ratio resulted in a decrease in contact conductivity, an increase in wear intensity and a decrease in current density, corresponding to the catastrophic wear onset.

In the sliding contact zone under current, transfer layers are formed, which have two morphologically different sectors on the worn surfaces at different turn ratios: one sector shows signs of deformation due to adhesion, while the other sector is deformed with melt formation.

It was found that the transfer layers contain more than 70 vol. % FeO.

An explanation for melt formation has been proposed: high bias currents arise from strong self-induction pulses in the contact, which remove Fe²⁺ and O²⁻ ions from the FeO crystal lattice nodes and cause melting of the contact layer.

A decrease in the turn ratio causes high self-induction pulses and correspondingly high bias current densities. This results in a relatively strong energy impact on the contact layer and its high wear.

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Contribution of the Authors

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Вклад авторов

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В. В. Фадин – написание и редактирование текста, проведение исследований.

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Short report

Краткое сообщение

STRUCTURE AND PROPERTIES OF SPECIAL-PURPOSE ALLOYS AFTER ANNEALING

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Abstract. Special-purpose alloys based on the Al–Si–Cu system are widely used in various industries, including engine and instrument engineering. The effect of annealing in the range of 100 – 900 °C on microstructure, density, and microhardness of Al – 30 % Si – 50 % Cu alloy was studied. Scanning electron microscopy showed that as the annealing temperature increases, the form of eutectic silicon particles changes and their coagulation occurs. According to the results of microrentgenospectral analysis, change in the eutectic structure is accompanied by segregation of copper in its individual sections. After annealing, there is a slight decrease in density and microhardness of the alloy.

Keywords: special alloys, aluminum, silicon, copper, annealing, microstructure, eutectic, microhardness, density

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СТРУКТУРА И СВОЙСТВА СПЛАВОВ СПЕЦИАЛЬНОГО НАЗНАЧЕНИЯ ПОСЛЕ ОТЖИГА

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Аннотация. Сплавы специального назначения на основе системы Al–Si–Cu в современных условиях широко используются в различных отраслях промышленности, включая двигателе- и приборостроение. В работе изучено влияние отжига в интервале 100 – 900 °C на микроструктуру, плотность и микротвердость сплава Al – 30 % Si – 50 % Cu. С помощью сканирующей электронной микроскопии установлено, что с повышением температуры отжига происходят изменение формы частиц эвтектического кремния и их коагуляция. Согласно результатам микрорентгеноспектрального анализа изменение строения эвтектики сопровождается сегрегацией меди в ее отдельных участках. После отжига происходит небольшое уменьшение плотности и микротвердости сплава.

Ключевые слова: специальные сплавы, алюминий, кремний, медь, отжиг, микроструктура, эвтектика, микротвердость, плотность

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INTRODUCTION

Special-purpose Al–Si–based alloys doped with copper, nickel, and other elements are finding broad application across contemporary industries, particularly in engine building, instrument engineering, electronics, and aerospace [1 – 4]. In many functional units, these alloys are often in contact with components made of various steels or ceramics; therefore, their properties must match those of contacting materials in the coefficient of linear

thermal expansion (CLTE), ensuring dimensional stability and, when required, vacuum-tight joints. In addition to a regulated CLTE value, special-purpose alloys must exhibit high wear resistance and low density, while their specific mechanical properties are comparable to those of medium-carbon structural steels [5; 6]. Because many precision components operate over a wide temperature range, special-purpose alloys must retain stable properties even at the upper operating limits. To ensure this stability, aluminum–silicon alloys of hypereutectic compo-

sition are alloyed with refractory elements, modified, and subjected to heat treatment [7–9].

Previous studies have shown that hypereutectic Al–Si alloys containing copper in amounts comparable to or exceeding the silicon content exhibit a low and stable CLTE across the wide temperature range [10].

In this context, the present study investigated the effect of isochronous annealing at 100–900 °C (holding time 11 h, air cooling) on the microstructure, density, and microhardness of the Al–30 % Si–50 % Cu alloy.

MATERIALS AND METHODS

The study used silumins containing 30 % Si and 50 % Cu. The starting materials for alloy preparation were aluminum grade A7, silicon grade Kr0, and copper grade M1.

Aluminum A7 was melted first, followed by sequential additions of silicon and copper in amounts of 30 and 50 %, respectively. Once the alloying elements were fully dissolved, the melt was treated with fine-fraction wet dolomite at 880 °C. The melt was then held, and casting was carried out at 1100 °C into a cold aluminum chill mold.

Heat treatment at 100, 250, and 350 °C was performed in SNOL–3.5.3.5.3.5/3.5–I2M resistance furnaces with a working chamber of 350×350×350 mm and a temperature variation of ± 5 °C. Heat treatment at higher temperatures (500–900 °C) was carried out in SNOL–1.6.2.5.1/9–I3 resistance furnaces with a chamber size of 160×250×100 mm and a temperature deviation of ± 5 °C within the working range.

Structural analysis of the Al–30 % Si–50 % Cu alloy samples was performed using a KYKY EM6900 Std scanning electron microscope (SEM) at the Laboratory of Electron Microscopy and Image Processing, SibGIU, in secondary and backscattered electron modes (SE + BSE) at an accelerating voltage of 25–30 kV, working distance of 15–18 mm, and magnifications ranging from 200 to 1000 $\times$. To analyze elemental distribution among structural components, micro-X-ray spectral analysis (MXSA) was performed using an energy-dispersive spectroscopy (EDS) module.

Density was determined by hydrostatic weighing on WA-21 analytical balances with an accuracy of 0.0001 g. Microhardness was measured on an HVS-1000 digital microhardness tester under a load of 0.245 N (25 gf).

RESULTS AND DISCUSSION

SEM is widely used to address specific research and technological problems because of its high resolution and the reliability of the results obtained [2–5]. Owing

to their high depth of field, SEM enables detailed examination of the structure of heterophase alloys with pronounced surface microrelief at high magnifications and, importantly, allows precise observation of eutectic structures (see Figure).

Microstructural analysis of the highly alloyed Al–30 % Si–50 % Cu alloy at different magnifications showed that its structure is primarily defined by plate-shaped primary silicon crystals (PSC). Between the PSC regions are areas of ternary eutectic (α + Si + CuAl₂) with a fine needle-like structure formed at the final stage of solidification. Elemental mapping of the polished surface was performed to examine how elements were distributed among the structural components. The results showed that silicon is mainly concentrated in the PSC, with a smaller fraction found in the eutectic. Copper is predominantly concentrated in the ternary eutectic of the Al–30 % Si–50 % Cu alloy, while aluminum is evenly distributed throughout the eutectic. A slight increase in iron concentration was observed in the eutectic as needle-shaped phases.

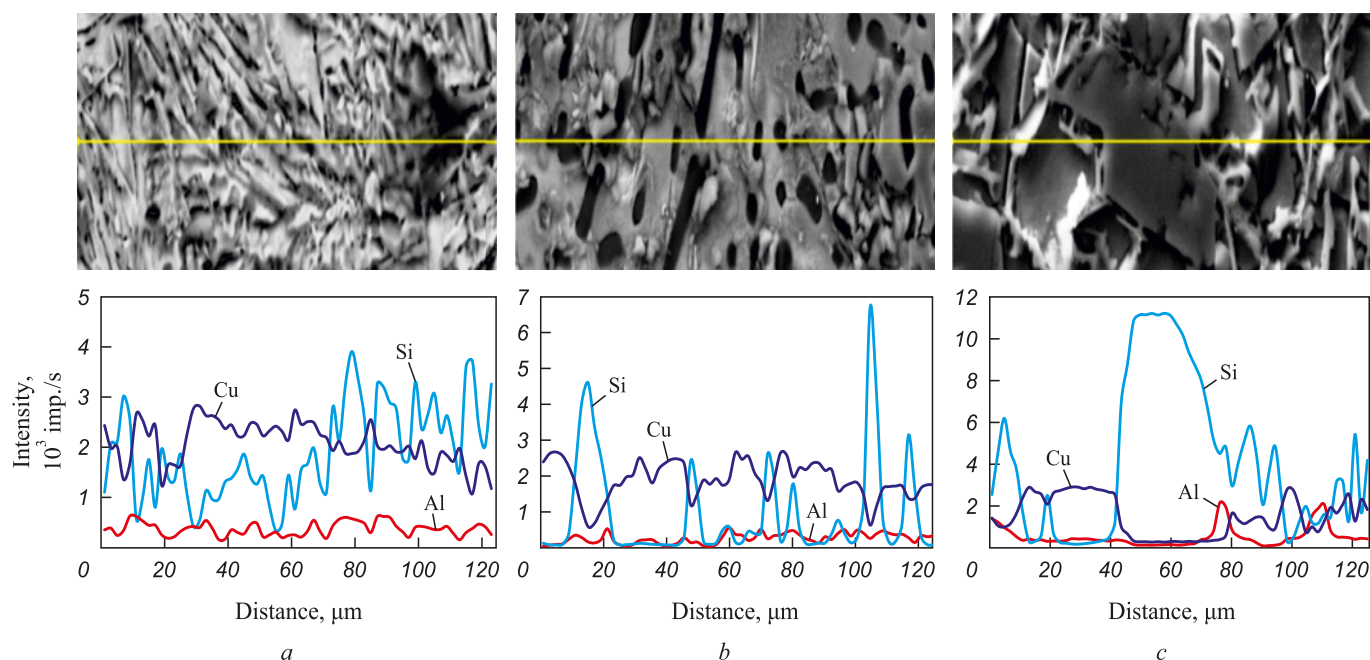
According to MXSA performed at various points within the eutectic and along the scan line (see Figure, *a*), copper is the predominant element (57–80 %), while the silicon content does not exceed 35 %, and aluminum accounts for 8–13 %. The highest copper content (80 %) was observed in dark regions, whereas the lowest (57 %) was found in needle-like crystals up to 1 μ m in size. The dark regions correspond to equilibrium (CuAl₂) and nonequilibrium (Cu₄Al₉ and CuAl) intermetallic phases.

It was established that the distinctive feature of the high-copper Al–30 % Si–50 % Cu alloy is its high thermal stability, which allows for prolonged annealing not only at 400–500 °C (as in binary silumins) but also at 700–900 °C.

Electron microscopy revealed that prolonged annealing at 710 °C transforms the eutectic morphology from fine-needle to finely dispersed, while the eutectic silicon particles acquire a rounded shape (see Figure, *b*). The topographic contrast highlights the surface relief of the sample. The dark rounded crystals correspond to silicon, while the remaining eutectic matrix mainly consists of copper (53–68 %) and aluminum (8–18 %).

After high-temperature annealing at 900 °C, relatively large silicon crystals (up to 10 μ m) with well-defined facets are formed in the eutectic (see Figure, *c*). In the bright intercrystalline regions, copper segregates (up to 75 %), accompanied by 17–20 % aluminum. In some areas, all alloying elements are detected, confirming the presence of the ternary eutectic (α + Si + CuAl₂).

The effect of annealing temperature on the density (ρ) and microhardness (μ) of the Al–30 % Si–50 % Cu



Al – 30 % Si – 50 % Cu alloy microstructure (1000 $\times$) and chemical elements distribution diagram (scanning along the line):
a – in initial state; b and c – after annealing at 710 and 900 $^{\circ}$ C

Микроструктура сплава Al – 30 % Si – 50 % Cu (1000 $\times$) и диаграмма распределения химических элементов (сканирование вдоль линии):
a – в исходном состоянии; б и в – после отжига при 710 и 900 $^{\circ}$ C

alloy was investigated (see Table). Microhardness was measured on the eutectic, and the results were averaged over at least four measurements.

The data show that increasing annealing temperature results in a decrease in both density and microhardness. The initial density and microhardness of the alloy are 4.4113 g/cm 3 and 413.6 HV, respectively, while after annealing at 900 $^{\circ}$ C, these values decrease to 4.2067 g/cm 3 and 345.5 HV. The slight (≤ 5 %) decrease in density can be attributed to greater hydrogen uptake from the furnace atmosphere and the acceleration of diffusion processes. The decrease in microhardness (up to 16 %) is associated with eutectic silicon coagulation and increased heterogeneity in the distribution of alloying elements.

Effect of annealing temperature on density and microhardness of high-copper silumin

Влияние температуры отжига на плотность и микротвердость высокомедистого силумина

Annealing temperature, $^{\circ}$ C	ρ , g/cm 3	Measurement error $\Delta\rho \cdot 10^{-5}$	μ , HV
initial state	4.4113	4.3332	413.6
100	4.3594	4.3332	341.0
250	4.3270	4.2620	386.0
560	4.3182	4.2427	354.6
710	4.2152	4.0210	408.5
900	4.2067	4.0030	345.5

CONCLUSIONS

The distinctive feature of the high-copper Al – 30 % Si – 50 % Cu alloy is its high thermal stability, which enables prolonged annealing in the 700 – 900 $^{\circ}$ C range, a condition unacceptable for binary silumins.

Electron microscopy showed that, as the annealing temperature increases, the eutectic silicon particles change shape and undergo coagulation, which is accompanied by redistribution and segregation of copper in localized regions of the eutectic. After annealing at 710 and 900 $^{\circ}$ C, upward diffusion processes further intensify the alloy's heterogeneity. In addition, increasing the annealing temperature causes a slight decrease in density and microhardness, which is also attributed to the enhanced diffusion rate.

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Вклад авторов

М. В. Попова – научное руководство, формирование основной концепции, частичное написание текста статьи, формулирование выводов.

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Original article

Оригинальная статья

SURFACE ENGINEERING OF $\text{Fe}_{53,3}\text{Ni}_{26,5}\text{B}_{20,2}$ AMORPHOUS ALLOY BY EXCIMER LASER

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Abstract. The problem of differences in surface quality exists in the production of amorphous alloy (AA) ribbons by ultra-fast single-roll melt spinning. Structural inhomogeneities that can disrupt the isotropy of properties occur on the side of the ribbons adjacent to the quenching drum. In this regard, there is a need to develop a promising surface modification technology of AA which will not only eliminate roughness, but also controllably manage the structure along the ribbon depth, as well as selective processing of its individual sections to improve mechanical, magnetic and catalytic characteristics. Application of short-pulse laser systems has great potential for achieving these goals. In this research work, the effect of an excimer ultraviolet laser operating in nanometer wavelength range on the structural evolution, mechanical behavior and morphological changes of the surface of $\text{Fe}_{53,3}\text{Ni}_{26,5}\text{B}_{20,2}$ AA with varying the pulse number and their frequency were comprehensively studied using profilometry, indentation, optical and transmission electron microscopy methods. It is shown that laser irradiation of the contact matte side of the studied AA ribbon according to the selected mode (100 pulses, $f = 20$ Hz, $E = 150$ mJ, $W = 0,6$ J/cm²) effectively acts upon the surface relief and smoothes out production irregularities (pores, gas lines, scratches, etc.). In addition, the laser processing parameters are established that contribute to the AA structure softening, and therefore improve workability for possible forming, as well as the mode of transfer AA to an amorphous-nanocrystalline state with increased hardness and preservation of the ability to flow shear.

Keywords: amorphous alloy, laser irradiation, surface, hardness, structure, amorphous-nanocrystalline state, nanocrystal, hardening, softening

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ИНЖЕНЕРИЯ ПОВЕРХНОСТИ АМОРФНОГО СПЛАВА

$\text{Fe}_{53,3}\text{Ni}_{26,5}\text{B}_{20,2}$ ЭКСИМЕРНЫМ ЛАЗЕРОМ

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Аннотация. При изготовлении лент аморфных сплавов (АС) методом сверхбыстрого одновалкового спиннингования существует проблема различия качества их поверхностей. На стороне лент, прилегающей к закалочному барабану, возникают структурные неоднородности, которые могут нарушать изотропию свойств. В связи с этим существует потребность в разработке перспективной технологии поверхностного модифицирования АС, которая позволит не только устранять шероховатости, но и контролируемо управлять структурой по глубине ленты, а также селективно обрабатывать отдельные ее участки для повышения механических, магнитных и каталитических характеристик. Применение короткоимпульсных лазерных установок имеет большой потенциал для реализации этих целей. В работе с привлечением профилометрии, индентирования, оптической и просвечивающей электронной микроскопии комплексно изучено влияние воздействия эксимерного ультрафиолетового лазера, работающего в нанометровом диапазоне длин волн на структурную эволюцию, механическое поведение и морфологические видоизменения поверхности АС $\text{Fe}_{53,3}\text{Ni}_{26,5}\text{B}_{20,2}$ при варьировании числа

импульсов и их частоты. Показано, что лазерное облучение контактной, матовой стороны исследуемой ленты АС по подобранному режиму (100 импульсов, $f = 20$ Гц, $E = 150$ мДж, $W = 0,6$ Дж/см²) эффективно воздействует на рельеф поверхности, сглаживает производственные неровности (поры, газовые строчки, царапины и т. д.). Кроме того, установлены параметры лазерной обработки, способствующие размягчению структуры АС, а значит улучшению обрабатываемости для возможного формования, а также режим перевода АС в аморфно-нанокристаллическое состояние с повышенной твердостью и сохранением способности к пластическому сдвигу.

Ключевые слова: аморфный сплав, лазерное облучение, поверхность, твердость, структура, аморфно-нанокристаллическое состояние, нанокристалл, упрочнение, размягчение

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INTRODUCTION

Amorphous alloys (AAs) are deeply supercooled materials characterized by a glass transition temperature, below which an energetically unstable non-crystalline state prevails [1; 2]. This structural feature provides a combination of ductility with excellent strength, hardness, and an elastic limit of up to 2 % due to the absence of long-range order [3 – 5]. In addition to their outstanding mechanical characteristics, a number of AAs exhibit high magnetic properties, corrosion resistance, and biocompatibility, which makes them attractive for various applications [6 – 9]. Existing fabrication methods for AAs, such as copper mold casting and melt spinning, are effective in retaining the glassy state but are subject to significant limitations in terms of scalability, critical dimensions, and geometric complexity. Moreover, AAs are difficult-to-machine materials with a narrow range of thermal stability and a tendency to become brittle at elevated temperatures [10; 11]. These challenges drive the search for more advanced technologies for the production and processing of AAs in order to expand their engineering applications. In recent years, research has increasingly focused on the fundamental study of structural modification, phase formation, and property response in AAs under laser irradiation [12 – 14]. The introduction of selective laser melting with ultra-high cooling rates is a highly promising method for producing bulk AAs [15 – 18]. Laser-induced periodic surface structuring of AAs makes it possible to:

- produce color through the formation of oxide films of different thicknesses [19];
- tune hydrophobic/hydrophilic behavior in wettability tests [20];
- control domain structure and magnetic behavior [21];
- reduce friction and wear in tribological applications [22];
- fabricate precision diffraction gratings for sensor devices and related applications [23].

Surface functionalization of AAs therefore has the potential to broaden their application range and to introduce new functionalities into AA-based components.

Pulsed laser processing offers such advantages as high peak power and energy density, controlled thermal effects, rapid heating and cooling, high precision, and minimal deformation of the material compared with continuous-wave lasers [14; 24]. The pulse duration determines the degree of thermal diffusion, which plays an important role for nanosecond lasers, in contrast to femtosecond lasers that primarily induce phonon relaxation [25; 26]. Nanosecond laser treatment, characterized by larger affected zones and greater penetration depth, makes it possible to tune the magnetic behavior of AAs and to modify their mechanical properties by changing the surface microstructure [27 – 30]. However, several important questions remain to be addressed, for example:

- how to develop adequate physical models describing the interaction of short-pulse laser irradiation with AAs in the absence of a detailed understanding of the underlying mechanisms;
- whether improved material properties can be achieved by forming gradient amorphous–crystalline composite structures through pulsed laser treatment;
- how to design the structure of AAs in a controlled and efficient way and which laser parameters should be considered optimal.

At present, numerous studies are underway on pulsed laser processing of bulk AAs based on zirconium, titanium, and copper. In contrast, for rapidly quenched iron-based amorphous ribbons, the available data are limited, fragmented, and call for further exploratory investigation. It should be emphasized that iron-based AAs deserve particular attention due to the low cost of raw materials, their outstanding mechanical and soft magnetic characteristics, and excellent catalytic activity. Optimization of laser irradiation parameters to improve the functional properties of such alloys and to control their structure therefore remains an important research objective.

The purpose of this study is to investigate the effect of excimer ultraviolet laser irradiation operating in the nanometer wavelength range on structural and phase transformations, mechanical response, and morphological changes at the surface of Fe–Ni–B amorphous alloy under variation of pulse number and frequency.

MATERIALS AND METHODS

The object of study was a rapidly quenched $\text{Fe}_{53.3}\text{Ni}_{26.5}\text{B}_{20.2}$ amorphous alloy produced by the melt-spinning method in the form of a ribbon 10 mm wide and 25 μm thick.

Laser irradiation of the AA samples was performed using an excimer ultraviolet (UV) KrF laser of the CL-7100 series (Optosystems, Russia) with a wavelength of $\lambda = 248$ nm and a pulse duration of $\tau = 20$ ns. The irradiation was applied through a circular diaphragm with an area of $S = 7$ mm² at two pulse modes 100 and 500 pulses – while varying the repetition frequency f from 2 to 50 Hz. In both cases, the pulse energy E was 150 mJ, and the energy density W was 0.6 J/cm². The contact matte side of the AA ribbon, i.e., the side that was in contact with the copper quenching drum during fabrication, was subjected to laser treatment.

Hardness H_{IT} was measured using a DUH-211S dynamic ultramicrohardness tester (Shimadzu, Japan). Indentation was carried out in accordance with ISO 14577, employing a Vickers diamond indenter under a load of 10 mN in a loading–unloading mode at a rate of 70 mN/s.

Structural characterization of the AA was performed by transmission electron microscopy (TEM) using a JEM-2100 microscope (JEOL, Japan).

The surface morphology of the laser-irradiated areas was examined with a GX51 inverted metallographic microscope (Olympus, Japan). Surface rough-

ness was evaluated according to GOST 2789 – 73 using a NewView 7300 profilometer (Zygo, USA).

RESULTS AND DISCUSSION

Figure 1 shows the appearance of both sides of the $\text{Fe}_{53.3}\text{Ni}_{26.5}\text{B}_{20.2}$ amorphous alloy ribbon in its as-fabricated state. Unlike the non-contact side, which has a mirror-like surface (Fig. 1, *a*), the contact side exhibits extended surface irregularities such as dimples, cavities, and gas streaks elongated along the ribbon axis (Fig. 1, *b*). The formation of this microrelief is associated with the interaction of the melt puddle with the boundary gas layer on the non-ideal surface of the copper quenching drum [31]. These defects create internal stresses that adversely affect the magnetic response of the alloy [32 – 34]. Therefore, developing a laser modification technique and selecting an effective mode for smoothing the rough surface of amorphous ribbons, while improving their wear resistance and maintaining the amorphous structure, are essential research tasks.

After excimer UV laser irradiation under the first mode, surface transformations of the AA were examined (Fig. 2).

The results show that laser processing contributes to the elimination of surface roughness on the contact side of the ribbon. However, at a frequency of 2 Hz, the process proceeds unevenly. In the center of the irradiated circular area, a distinct “healing” of irregularities is observed (Fig. 2, *a*), while toward the periphery, large surface defects often remain unaffected by the laser (Fig. 2, *b*).

At frequencies ranging from 10 to 50 Hz, the laser impact becomes more uniform across the entire irradiated zone (Fig. 2, *c* – *e*), although the degree of reduction in surface height irregularities (parameter R_z) varies. Fig. 3 presents profilometry results for the ribbon in both

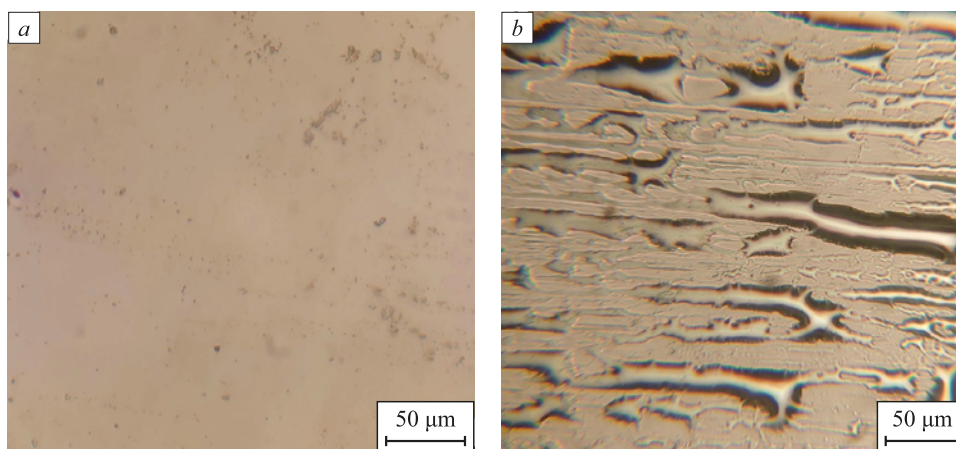


Fig. 1. Morphology of non-contact (*a*) and contact (*b*) sides of the ribbon of $\text{Fe}_{53.3}\text{Ni}_{26.5}\text{B}_{20.2}$ amorphous alloy before irradiation

Рис. 1. Морфология неконтактной (*a*) и контактной (*b*) сторон ленты АС $\text{Fe}_{53.3}\text{Ni}_{26.5}\text{B}_{20.2}$ до облучения

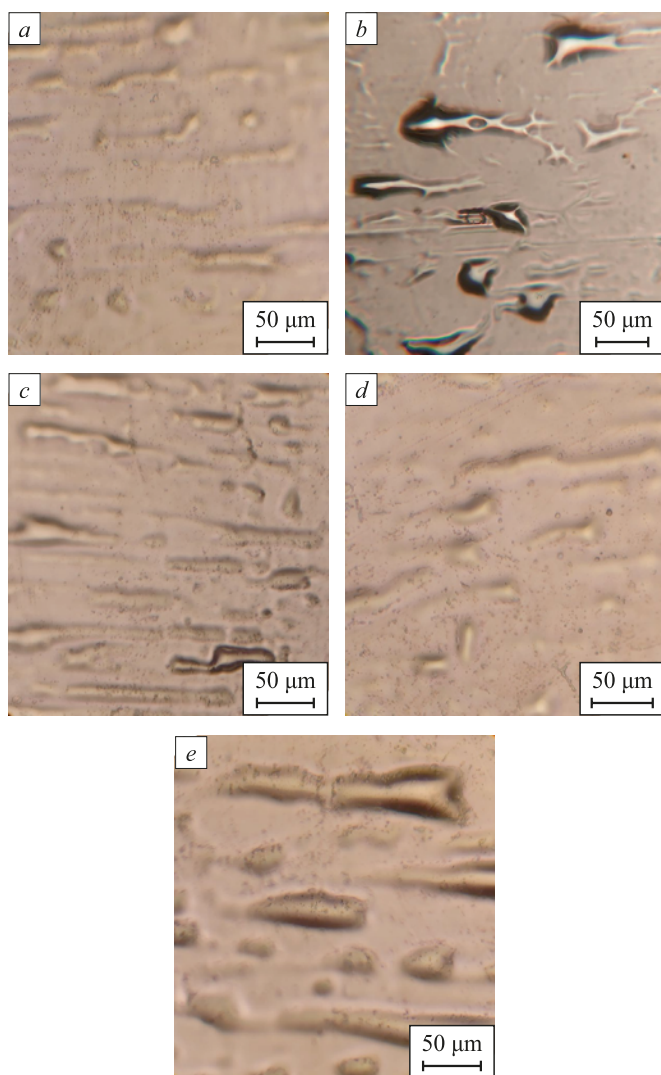


Fig. 2. Dependence of morphology of irradiated surface of Fe_{53.3}Ni_{26.5}B_{20.2} amorphous alloy at 100 pulses on the pulse frequency: *a* – 2 Hz (center of irradiation zone), *b* – 2 Hz (edge of irradiation zone); *c* – 10 Hz, *d* – 20 Hz, *e* – 50 Hz

Рис. 2. Морфология облученной поверхности АС Fe_{53.3}Ni_{26.5}B_{20.2} при 100 импульсах в зависимости от изменения частоты их следования:
a – 2 Гц (центр зоны облучения); *b* – 2 Гц (край зоны);
c – 10 Гц; *d* – 20 Гц; *e* – 50 Гц

the initial state and the laser-irradiated regions shown in Figs. 1 and 2. The surface topography was scanned perpendicular to the ribbon axis.

The best smoothing effect, including the collapse of volumetric accumulations of gas streaks, was achieved at a pulse repetition frequency of 20 Hz (Fig. 2, *d*; curve 6 in Fig. 3).

The calculated surface-relief data are summarized in the Table, confirming that 20 Hz is the optimal frequency for laser modification aimed at leveling the contact side of the ribbon and bringing its quality closer to that of the ideally smooth non-contact side, which corresponds to the highest (13th) surface roughness class.

Subsequently, the mechanical response of the alloy was evaluated by measuring hardness after laser irradiation. Under 100-pulse irradiation, an increase in pulse frequency leads to material softening (Fig. 4, curve 1). At $f = 20$ Hz, the hardness H_{IT} decreases by approximately 35 % relative to the initial value ($H_{IT0} = 6$ GPa) of the untreated alloy. At $f = 50$ Hz, H_{IT} slightly increases to 5.4 GPa.

When the number of pulses increases to 500, the dependence $H_{IT}(f)$ exhibits a more complex behavior, with two pronounced maxima at 2 and 20 Hz (Fig. 4, curve 2). At $f = 50$ Hz, a softening effect is observed – the hardness decreases to 5.8 GPa, approaching the value obtained for 100-pulse irradiation at the same frequency.

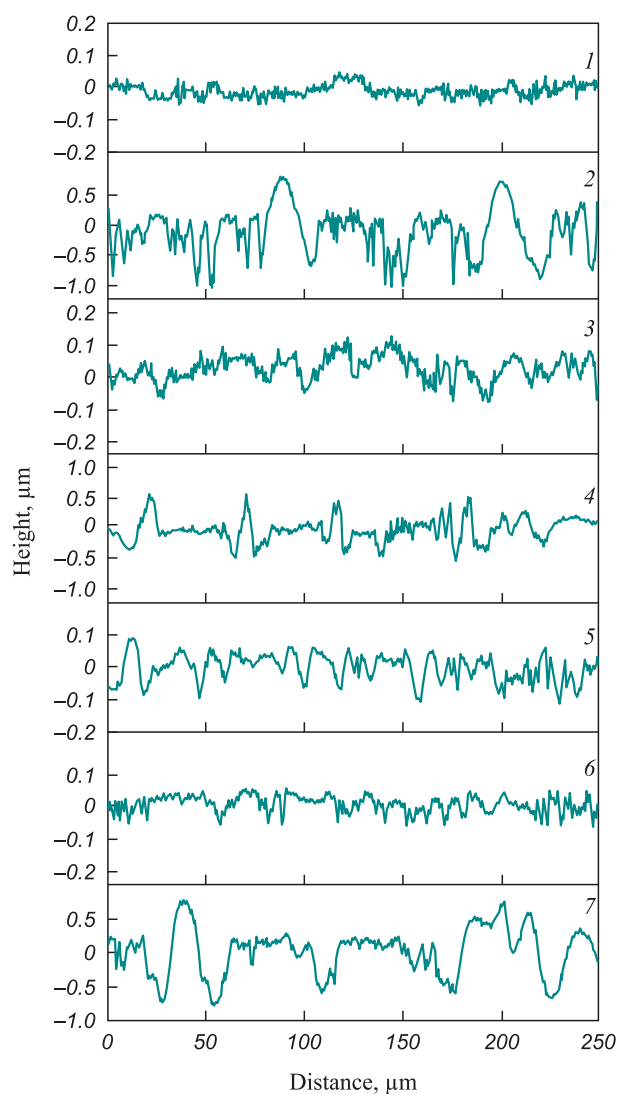


Fig. 3. Surface profilograms of the ribbon of Fe_{53.3}Ni_{26.5}B_{20.2} amorphous alloy before and after laser irradiation:
 1 and 2 – non-contact and contact sides (without treatment);
 3 – 2 Hz (center); 4 – 2 Hz (edge); 5 – 10 Hz; 6 – 20 Hz; 7 – 50 Hz

Рис. 3. Профилограммы поверхности ленты АС Fe_{53.3}Ni_{26.5}B_{20.2} до и после лазерного облучения:
 1 и 2 – неконтактная и контактная стороны (без обработки);
 3 – 2 Гц (центр); 4 – 2 Гц (край); 5 – 10 Гц; 6 – 20 Гц; 7 – 50 Гц

Surface profilometry parameters of Fe_{53,3}Ni_{26,5}B_{20,2} amorphous alloy before and after laser irradiation at 100 pulses, $E = 150$ mJ, $W = 0.6$ J/cm² and varying frequency, *I* and *II* – non-contact and contact sides of the ribbon

Параметры профилометрии поверхности АС Fe_{53,3}Ni_{26,5}B_{20,2} до и после лазерного облучения 100 импульсами, $E = 150$ мДж, $W = 0,6$ Дж/см² при варьировании частоты, *I* и *II* – неконтактная и контактная стороны ленты

Surface characteristic	Before irradiation		After irradiation				
	<i>I</i>	<i>II</i>	<i>f</i> , Hz				
			2		10	20	50
			center	edge			
Height of irregularities R_z , μm	0.08	1.60	0.19	1.00	0.39	0.10	0.78
Surface roughness class	13	9	12	9	11	13	10

Thus, it can be concluded that excimer laser treatment at 100 pulses, $f = 20$ Hz, $E = 150$ mJ, $W = 0.6$ J/cm² is an effective way to enhance the plasticity of the amorphous alloy. The mechanism underlying the hardness decrease observed for curve *I* in Fig. 4 can be interpreted as follows. During irradiation, short laser pulses with high energy density propagate from the surface into the material, generating an intense shock wave [12; 13]. When the peak pressure of the shock wave exceeds the yield strength, the amorphous alloy undergoes plastic deformation. The formation of a residual stress regions containing shear bands and free volume form within the ribbon is facilitated, resulting in improved plasticity [14].

In contrast, irradiation under the 500-pulse mode (curve 2 in Fig. 4) at $E = 150$ mJ and $W = 0.6$ J/cm² leads to optimal hardening within the amorphous state at $f = 2$ Hz and to maximum hardening within the amor-

phous–nanocrystalline state at $f = 20$ Hz, as confirmed by TEM observations.

Figs. 5 and 6 present TEM images illustrating the structural evolution of the studied alloy as a function of pulse number and frequency.

Under the first laser processing mode (100 pulses), the amorphous structure is preserved within the frequency range $f = 2 - 20$ Hz (Fig. 5, *a - d*). The selected area electron diffraction patterns show two diffuse halos characteristic of the amorphous phase, while the TEM images display a characteristic fine “salt-and-pepper” contrast that remains essentially unchanged when switching from bright-field to dark-field imaging. However, at $f = 20$ Hz, a slight disturbance of the homogeneous contrast is observed (Fig. 5, *c*), accompanied by halo broadening, which indicates the onset of structural rearrangement preceding crystallization. The amorphous matrix locally reorganizes, facilitating plastic deformation processes. This structural relaxation manifests itself as a decrease in the hardness parameter H_{IT} . At this irradiation frequency, the free volume undergoes rearrangement and coalescence, promoting the formation and propagation of shear bands.

At $f = 50$ Hz, nanocrystals of α -Fe with a BCC lattice and γ -Fe with an FCC lattice precipitate within the amorphous matrix, with a total volume fraction of about 40 % (Fig. 5, *e*). The nanocrystal size ranges from 20 to 70 nm. The coexistence of amorphous and nanocrystalline phases leads to an increase in hardness.

Under the second irradiation mode (500 pulses), the transition from the amorphous (Fig. 6, *a*) to the amorphous – nanocrystalline state occurs at lower frequencies, i.e., at $f = 20$ Hz (Fig. 6, *b*). The coexistence of two structural components, together with high hardness, provides conditions for the development of plastic shear. In Fig. 6, *b*, the propagation of shear bands is visible, including their looping, branching, and braking on nanocrystals.

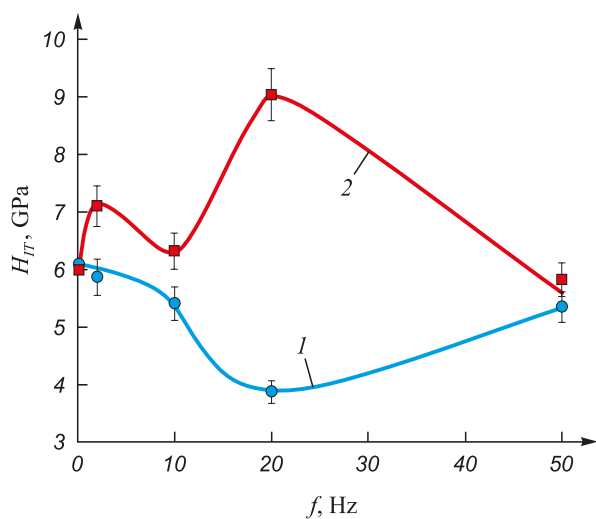


Fig. 4. Laser frequency dependence of hardness of Fe_{53,3}Ni_{26,5}B_{20,2} amorphous alloy:
I – 100 pulses; *2* – 500 pulses

Рис. 4. Зависимость твердости АС Fe_{53,3}Ni_{26,5}B_{20,2} от частоты лазера:
I – 100 импульсов; *2* – 500 импульсов

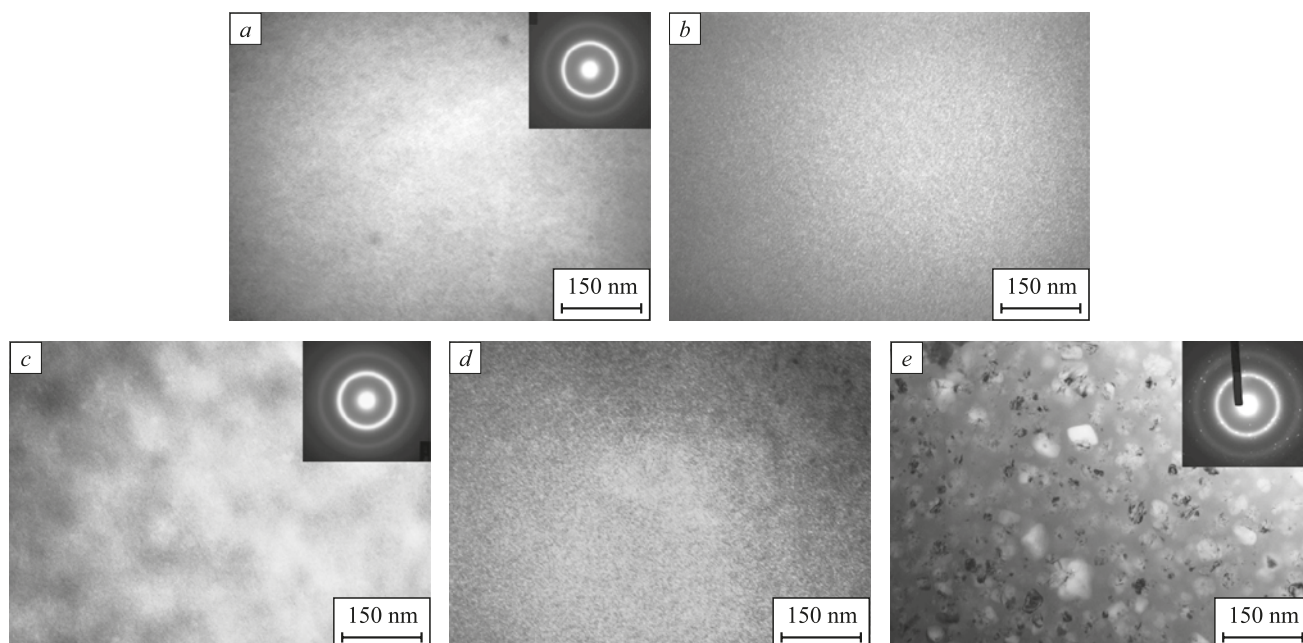


Fig. 5. TEM images and the corresponding selected area electron diffraction patterns of Fe_{53.3}Ni_{26.5}B_{20.2} amorphous alloy structure after laser treatment at 100 pulses:

a – 2 Hz, bright-field image; *b* – 2 Hz, dark-field image;
c – 20 Hz, bright-field image; *d* – 20 Hz, dark-field image; *e* – 50 Hz, bright-field image

Рис. 5. ПЭМ-изображения и микроэлектроннограммы структуры АС Fe_{53.3}Ni_{26.5}B_{20.2} после лазерной обработки 100 импульсами:

a – 2 Гц, светлое поле; *b* – 2 Гц, темное поле;
c – 20 Гц, светлое поле; *d* – 20 Гц, темное поле; *e* – 50 Гц, светлое поле

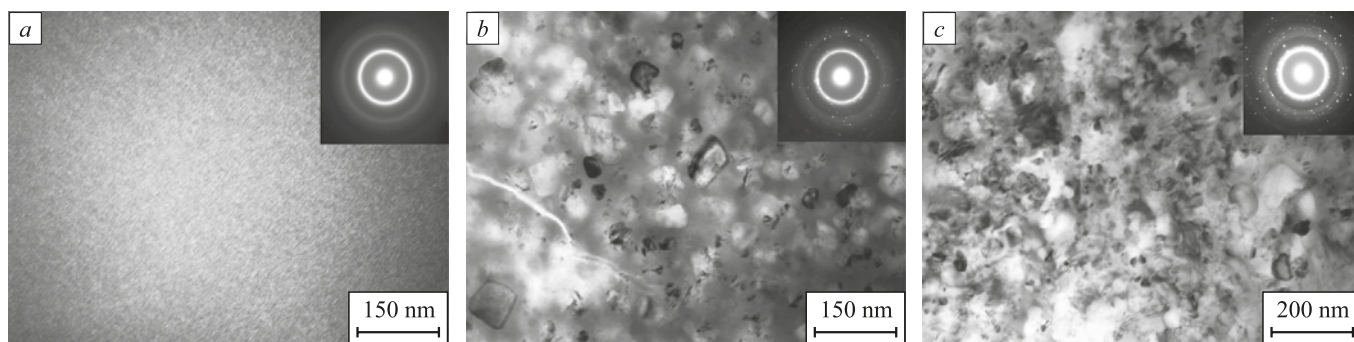


Fig. 6. TEM images and the corresponding selected area electron diffraction patterns of Fe_{53.3}Ni_{26.5}B_{20.2} amorphous alloy structure after laser treatment at 500 pulses:

a – 2 Hz, dark-field image; *b* – 20 Hz, bright-field image; *c* – 50 Hz, bright-field image

Рис. 6. ПЭМ-изображения и микроэлектроннограммы структуры АС Fe_{53.3}Ni_{26.5}B_{20.2} после лазерной обработки 500 импульсами:

a – 2 Гц, темное поле; *b* – 20 Гц, светлое поле; *c* – 50 Гц, светлое поле

An increase in pulse repetition frequency to $f = 50$ Hz stimulates crystallization processes in the alloy. Along with the formation of crystalline α , γ -(Fe,Ni), and eutectic γ -(Fe,Ni) + Fe₃B phases, grain growth is observed (Fig. 6, *c*), which leads to softening, i.e., a decrease in H_{IT} (curve 2 in Fig. 4).

CONCLUSIONS

Nanosecond excimer UV lasers offer significant advantages for low-cost, environmentally friendly, high-

precision, and selective processing of amorphous alloys with minimal material loss. They ensure efficient energy transfer and can achieve high laser energy densities. With properly selected parameters, such treatment can activate either structural rejuvenation of the amorphous alloy – accompanied by loosening and softening that improve workability and formability – or partial crystallization aimed at achieving optimal strength combined with satisfactory plasticity.

It has been demonstrated that excimer UV laser irradiation of the contact (matte) side of the rapidly melt-

quenched Fe_{53,3}Ni_{26,5}B_{20,2} amorphous alloy ribbon under the selected mode (100 pulses, $f = 20$ Hz, $E = 150$ mJ, $W = 0.6$ J/cm²) most effectively modifies the surface relief, reduces roughness, and eliminates manufacturing defects such as pores and gas streaks formed during melt spinning.

A nonmonotonic dependence of the H_{IT} hardness of Fe_{53,3}Ni_{26,5}B_{20,2} on the laser pulse repetition frequency was established. Irradiation with 100 pulses at 20 Hz produces a pronounced softening effect while preserving the amorphous structure. Atomic rearrangements occur without long-range diffusion. Increasing the pulse number to 500 leads to energy accumulation within the material and intensified laser heating, resulting in a two-stage hardening at 2 and 20 Hz followed by a hardness drop at 50 Hz. This behavior is associated with a transition from structural relaxation in the amorphous alloy (involving local topological and compositional ordering) to crystallization processes characterized by nucleation and emergence of crystalline phases and grain growth.

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Original article

Оригинальная статья

CRITERIA FOR THE EXISTENCE OF LOCAL EQUILIBRIA IN SUPERCOOLED MELTS

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Abstract. The article is intended for researchers working in the field of supercooled metallic melts. It addresses an important theoretical problem: the possibility of establishing local thermodynamic equilibrium at the phase boundary during crystallization from a supercooled melt. Such processes play a crucial role in determining the microstructure of materials during solidification, particularly under rapid cooling conditions characteristic of modern metallurgical and powder technologies. During supercooling, nuclei of a new, solid phase begin to form in the melt. To mathematically describe the growth of such nuclei, it is necessary to specify boundary conditions that define the composition of the adjacent liquid phase. Traditional models assume that local equilibrium can be established near the nucleus and that its parameters can be derived from the equilibrium phase diagram. However, as demonstrated by our study of binary systems, local equilibrium may, in some cases, be fundamentally unattainable. This article presents a theoretical analysis of the conditions under which equilibrium may or may not be established. The analysis considers chemical potentials of the components in both the solid nucleus and the liquid melt. Based on the equilibrium phase diagram of the corresponding macrosystem, one can infer the relative chemical potentials of the components in each phase. It is shown that when the nucleus consists of a single component, local equilibrium is always possible in principle. However, when the nucleus is a solution, equilibrium may only be realized under specific thermodynamic conditions. In such cases, the application of first-kind boundary conditions becomes invalid, and it is necessary to take into account the rates of chemical reactions involved in the interphase transfer of each component.

Keywords: local equilibrium, supercooled melt, crystallization, rapid solidification, phase equilibrium, phase diagram

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КРИТЕРИИ СУЩЕСТВОВАНИЯ ЛОКАЛЬНЫХ РАВНОВЕСИЙ В ПЕРЕОХЛАЖДЕННЫХ РАСПЛАВАХ

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Аннотация. Статья предназначена для исследователей, работающих в области переохлажденных расплавов. В ней рассматривается важный теоретический вопрос, связанный с возможностью установления локального термодинамического равновесия на границе раздела фаз при кристаллизации из переохлажденных металлических расплавов. Такие процессы играют ключевую роль в формировании структуры материалов при их затвердевании, особенно в условиях быстрого охлаждения, характерных для современных технологий металлургии и порошковой металлургии. При переохлаждении в расплаве начинают формироваться зародыши новой твердой фазы. Для математического описания роста зародыша необходимо задать граничные условия, определяющие состав прилегающей жидкой фазы. В традиционных подходах предполагается, что вблизи зародыша может быть установлено локальное равновесие, параметры которого извлекаются из равновесной диаграммы состояния. Однако, как показали исследования авторов для двухкомпонентных систем, в некоторых случаях локальное равновесие невозможно в принципе. В данной работе проведен теоретический анализ условий равновесия. Для этого рассматривались химические потенциалы компонентов обеих фаз: твердого зародыша и жидкого расплава. По равновесной диаграмме соответствующей макросистемы можно составить представление о химических потенциалах их компонентов, в частности, в какой фазе химический потенциал каждого компонента ниже. Показано, что, когда зародыш новой фазы состоит из одного компонента, такое локальное равновесие, в принципе, всегда возможно. Однако, когда зародыш является раствором, такое возможно лишь при определенных условиях. В этих случаях применение граничных условий первого рода становится некорректным, и необходимо учитывать скорости химических реакций перехода каждого компонента из одной фазы в другую.

Ключевые слова: локальное равновесие, переохлажденный расплав, кристаллизация, сверхбыстрое переохлаждение, фазовое равновесие, диаграмма состояния

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INTRODUCTION

The study of the kinetics of crystal nucleation and growth during rapid solidification of the melt has become particularly relevant in recent years due to the development of nanotechnologies and the wide use of amorphous alloys. To calculate the growth rates of nanocrystals [1–3] in a melt or, for instance, in an amorphous alloy during annealing [4–5], it is necessary to know the concentrations of the melt components at their surfaces.

The aim of this study is to provide a theoretical analysis of the conditions under which local equilibrium may (or may not) be established at the boundary between a solid nucleus and a supercooled melt.

Let us consider a case in which a binary metallic melt is instantaneously cooled to a temperature below the solidus. Nuclei of a new, solid phase begin to form within the melt. To formulate the differential equations describing the growth of such a nucleus, it is necessary to specify boundary conditions – that is, the concentrations of the liquid-phase components at the phase interface. Traditionally, it is assumed that any system tends to reach an equilibrium state, and although true equilibrium cannot exist in a supercooled liquid melt, it can be assumed that, at nanoscale distances from the nucleus, the composition of the liquid phase will be nearly equilibrium, i.e., local equilibrium is established. The parameters of this quasi-equilibrium are usually taken from the equilibrium phase diagram. However, our analysis shows that in many cases this assumption is not valid.

Our conclusions regarding deviations from local equilibrium are supported by the results of studies on the “solute trapping” effect caused by the high rate of crystal growth in the melt [6–9]. Theoretical studies by several authors [10–13], as well as our own research [14–15], have made it possible to identify certain regularities in these processes. Experimental studies of various systems [16–17], particularly supercooled metallic melts [18–20], confirm the manifestation of such effects in practice. Meanwhile, no general thermodynamic criteria have been formulated so far to assess the conditions of local equilibrium during crystal growth. This study focuses on analyzing these conditions.

To evaluate the equilibrium conditions, let us consider the processes associated with phase equilibrium through the chemical potentials of their components:

$$\mu_i = \mu_i^0 + RT \ln a_i, \quad (1)$$

where μ_i^0 is the part of the chemical potential of the i -th component that does not depend on concentration, R is the universal gas constant, T is the temperature, and a_i is the activity of the i -th component. For the sake of generality, we will use only the following properties of chemical potentials:

– the dependence of the activity of a component in the phase on the concentration of that component is a monotonically increasing continuous function;

– as the concentration of a component approaches zero, its activity likewise tends to zero; hence, by Eq. (1), its chemical potential tends to $-\infty$, and the molar contribution of this component to the phase’s Gibbs free energy of mixing tends to zero.

BINARY SYSTEM WITH EQUILIBRIUM

IN ONE COMPONENT

Consider a model case of a liquid solution (melt) of components A and B with an equilibrium phase diagram shown in Fig. 1 at temperature T_1 . From the diagram, several conclusions can be drawn about the chemical potentials of the phase components. Since, at the fraction of component B corresponding to the pure component A , the system is at equilibrium in the solid state, the chemical potential of component A in the solid phase must be lower than in the liquid phase: $\mu_A^{S_0} < \mu_A^{L_0}$. In accordance with the above considerations, let us draw the corresponding curve $\mu_A^L = \mu_A^L(x)$.

Since, according to Eq. (1), as $x \rightarrow 1$ the chemical potential $\mu_A^L \rightarrow -\infty$ and $\mu_A^L(0)$ larger $\mu_A^{S_0}(0)$, then by the Bolzano–Cauchy theorem for continuous functions [21] there exists a point x such that $\mu_A^L(x) = \mu_A^{S_0}$. In the case shown in Fig. 1 at temperature T_1 , this concentration is x_1 .

Note that the same reasoning also applies to temperatures below the eutectic. Local equilibrium will be established between the solid-phase nucleus and the continuation of the liquidus line. This statement is applicable provided that an approximation (or an analytical expression) of the liquidus line below the solidus is available.

The analysis has been limited to component A because component B is absent in the nucleus and its chemical potential does not affect phase equilibrium. Similar reasoning can be carried out for the growth of a crystal of phase B .

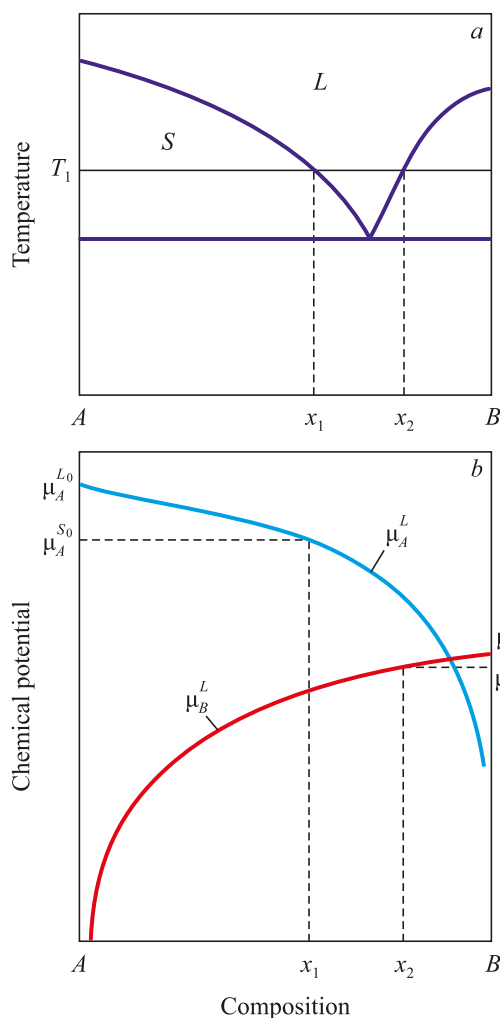


Fig. 1. Chemical potentials of the components in a system with a eutectic phase diagram (a) at temperature T_1 (b)

Рис. 1. Диаграмма состояния (a) и химические потенциалы компонентов системы с эвтектической диаграммой состояния при температуре T_1 (b)

According to Fig. 1, at temperature T_1 nuclei of the pure component B are in local equilibrium with a solution of composition x_2 .

BINARY SYSTEM WITH EQUILIBRIUM IN BOTH COMPONENTS

Consider a more complex case. Let components A and B exhibit unlimited mutual solubility in both the liquid and solid states, as represented by the phase diagram in Fig. 2.

Based on the diagram (Fig. 2, a), we draw the required conclusions about the chemical potentials of the components at temperature T_1 . Since, at the fraction of component B corresponding to the pure component A: $x = 0$, the system is at equilibrium in the solid state, the chemical potential of component A in the solid phase $\mu_A^{S_0}$ must be lower than that in the liquid phase $\mu_A^{L_0}$, as shown sche-

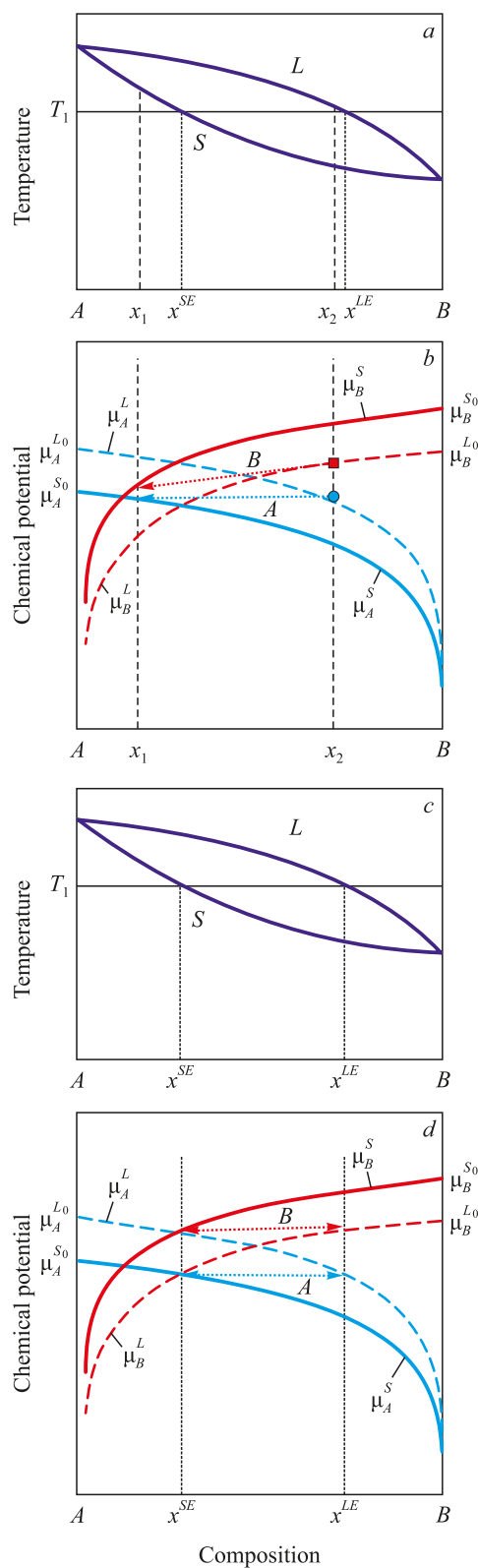


Fig. 2. Chemical potentials of the components (a, c) in a system with unlimited mutual solubility in liquid and solid states at temperature T_1 : (b) partial equilibrium and (d) complete equilibrium between the liquid and solid phases

Рис. 2. Диаграммы состояния (a, c) и химические потенциалы компонентов системы с неограниченной взаимной растворимостью компонентов в жидком и твердом состояниях при температуре T_1 при неполном равновесии (b) и при полном равновесии (d) между жидкой и твердой фазами

matically in Fig. 2, *b*. In accordance with Eq. (1), we plot the curves $\mu_A^L = \mu_A^L(x)$ and $\mu_A^S = \mu_A^S(x)$. In the general case, the curve μ_A^L should lie above and to the right of μ_A^S . We perform analogous constructions for component *B*. Since at $x = 1$ (pure *B*) the system is at equilibrium in the liquid state, the chemical potential of *B* in the liquid phase $\mu_B^{L_0}$ must be lower than in the solid phase $\mu_B^{S_0}$, as also shown in Fig. 2, *b*. According to Eq. (1), we then plot $\mu_B^L = \mu_B^L(x)$ and $\mu_B^S = \mu_B^S(x)$. In the general case, μ_B^S should lie above and to the left of μ_B^L .

Note also that the mutual placement of the chemical-potential curves for *A* and *B* is immaterial for the present argument.

For the system to be in equilibrium with respect to component *A*, the compositions of the liquid and solid phases must satisfy $\mu_A^L(c_A^L) = \mu_A^S(c_A^S)$. Geometrically, this means that the compositions defined by the intersection points of the curves $\mu_A^L(x)$ and $\mu_A^S(x)$ with any horizontal line correspond to *A* – equilibrium. There is an infinite set of such composition pairs.

However, for the system to be in equilibrium, equilibrium with respect to component *B* must also be achieved at the same concentrations. Thus, if the composition of the solid phase corresponds to point x_1 in Fig. 2, *b*, then, despite the fact that the chemical potentials of component *A* in both phases are equal, equilibrium with respect to component *B* will not be attained. The chemical potential of component *B* in the liquid phase is higher than in the solid phase, and therefore a transfer of component *B* from the liquid phase into the solid phase will occur. This disrupts *A* – equilibrium and shifts the solid phase composition. In the case shown in Fig. 2, *b*, concentration redistribution enriches both phases in component *B*. The process terminates only when point x_1 moves to position x^{SE} and x_2 moves to x^{LE} .

Let us determine when true equilibrium – simultaneously in both components – is possible. At equilibrium the equalities and: $\mu_A^L(x^L) = \mu_A^S(x^S)$ and $\mu_B^L(x^L) = \mu_B^S(x^S)$ must hold at the same time. As seen in Fig. 2, *c*, *d*, this is necessary and sufficient for the chemical affinities of the transitions from the liquid to the solid phase for component *A*: $A_A = \mu_A^{L_0} - \mu_A^{S_0}$ and for component *B*: $A_B = \mu_B^{L_0} - \mu_B^{S_0}$ to have opposite signs, that is, either $(\mu_A^{L_0} < \mu_A^{S_0}, \mu_B^{L_0} > \mu_B^{S_0})$, or $(\mu_A^{L_0} > \mu_A^{S_0}, \mu_B^{L_0} < \mu_B^{S_0})$.

We now prove this mathematically. For simplicity, consider the case where the component activities are equal to the mole fractions. At equilibrium, the equalities must hold simultaneously: $\mu_A^L(x^L) = \mu_A^S(x^S)$ and $\mu_B^L(x^L) = \mu_B^S(x^S)$, where x^L and x^S are the mole fractions of component *B* in the liquid and solid phases, respectively. Using Eq. (1), we can write these relations in the form:

$$\begin{cases} \mu_A^{L_0} + RT \ln(1 - x^L) = \mu_A^{S_0} + RT \ln(1 - x^S), \\ \mu_B^{L_0} + RT \ln x^L = \mu_B^{S_0} + RT \ln x^S \end{cases} \Rightarrow \begin{cases} \frac{1 - x^L}{1 - x^S} = \exp\left(\frac{\mu_A^{S_0} - \mu_A^{L_0}}{RT}\right), \\ \frac{x^L}{x^S} = \exp\left(\frac{\mu_B^{S_0} - \mu_B^{L_0}}{RT}\right). \end{cases} \quad (2)$$

Introduce the notation

$$\begin{aligned} A_A^0 &= \mu_A^{L_0} - \mu_A^{S_0}, \quad A_B^0 = \mu_B^{L_0} - \mu_B^{S_0}, \\ K_A &= \exp\left(-\frac{A_A^0}{RT}\right), \quad K_B = \exp\left(-\frac{A_B^0}{RT}\right). \end{aligned} \quad (3)$$

From system (2) it follows that it is impossible to have $K_A = 1$ or $K_B = 1$, or $K_A = K_B$, since in any of these cases one obtains $K_A = K_B = 1$ and $x^L = x^S$, that does not occur in this case.

Except for these cases, from Eq. (2) we obtain

$$x^S = \frac{K_A - 1}{K_A - K_B}, \quad x^L = \frac{K_A - 1}{K_A - K_B} K_B. \quad (4)$$

However, Eq. (4) may also yield physically meaningless results (negative values or values exceeding unity). Let us determine for which values of K_A and K_B the constraints $0 \leq x^S \leq 1$, $0 \leq x^L \leq 1$, are satisfied, i.e.,

$$0 \leq \frac{K_A - 1}{K_A - K_B} \leq 1, \quad 0 \leq \frac{K_A - 1}{K_A - K_B} K_B \leq 1. \quad (5)$$

An analysis of the system of inequalities (5) shows that they hold when $(K_A < 1, K_B > 1)$, or, conversely, $(K_A > 1, K_B < 1)$. Using the notation of Eq. (3), we obtain that equilibrium in the system under consideration is possible if the chemical affinities A_A^0 and A_B^0 have opposite signs: if for one component, say *A*: $\mu_A^{L_0} > \mu_A^{S_0}$, then for the other it must necessarily be the opposite: $\mu_B^{L_0} < \mu_B^{S_0}$. Thus, the criterion for the possibility of local equilibrium is the set of conditions

$$(\mu_A^{L_0} < \mu_A^{S_0}, \mu_B^{L_0} > \mu_B^{S_0}) \text{ or } (\mu_A^{L_0} > \mu_A^{S_0}, \mu_B^{L_0} < \mu_B^{S_0}). \quad (6)$$

Let us consider the case where these conditions are not satisfied – namely, the same system at a temperature T_2 below the solidus line (Fig. 3).

At temperatures below the solidus, both components in the pure state are solid:

$$(\mu_A^{L_0} > \mu_A^{S_0}, \mu_B^{L_0} > \mu_B^{S_0})$$

so the necessary conditions for phase equilibrium are not fulfilled. This is evident from Fig. 3, since the μ_A^L curve

lies above and to the right of the μ_A^S curve; at equilibrium with respect to component A one has $x^S < x^L$.

For component B , the μ_B^L curve lies above and to the left of the μ_B^S curve, and equilibrium with respect to component B is possible only if $x^L < x^S$. These conditions cannot be satisfied simultaneously – equilibrium is impossible.

Thus, at temperatures below the solidus, when both components are in the solid state, the equilibrium conditions for both components cannot be fulfilled simultaneously. The analysis shows that, in this case, any local equilibrium between the liquid and solid phases is impossible because it would require mutually contradictory values of the chemical potentials. Consequently, any analytic continuation of the equilibrium lines into the region of supercooled melts below the solidus has no thermodynamic justification and cannot be used to provide a correct description of phase transformations on supercooling.

RESULTS AND DISCUSSION

The aim of this work was to use equilibrium phase diagrams to study the dynamics of nucleation and growth of nuclei of a new phase. It is evident that under rapid cooling the system does not attain global equilibrium; however, in the vicinity of the solid – liquid interface a quasi-equilibrium local state may be established, which makes it possible to employ thermodynamic approaches. The use of local equilibria is a key principle of nonequilibrium thermodynamics.

The analysis shows that such local equilibria are not always possible. If the nucleus consists of a single component, a quasi-equilibrium description is applicable. For two or more components, the feasibility of local equilibrium is determined by the relative signs of the chemical affinities for the interphase transitions. In particular, if these affinities have the same sign, local equilibrium is impossible, and the phase transformations must be described exclusively by kinetic models.

The proposed approach can be extended to more complex systems, provided that functional dependences of phase activities on composition are available. For more accurate modelling of nucleus growth, the model should be supplemented with diffusion equations describing mass transport to the phase boundary and with rate equations governing the interphase chemical reactions that transfer components between the phases.

CONCLUSIONS

We have developed a method for determining the possibility of local equilibrium between a solid nucleus and the liquid (parent) phase based on an analysis of the chemical potentials of the components. With the necessary modi-

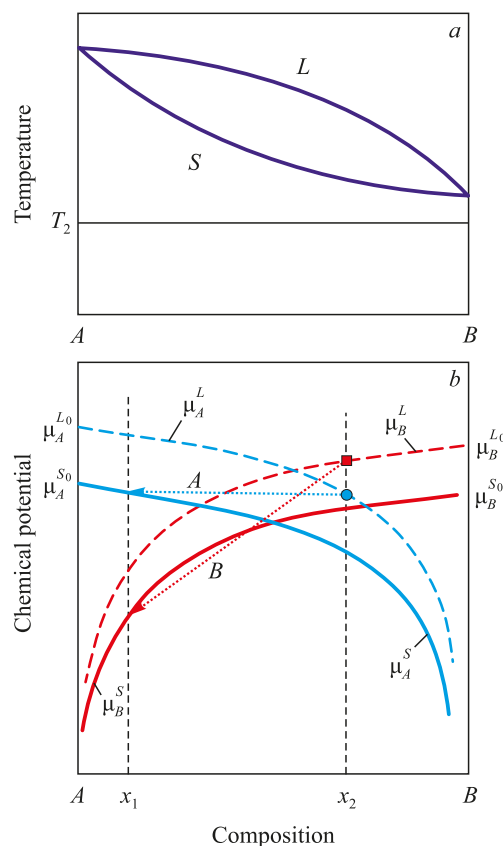


Fig. 3. Chemical potentials of the components (a) in a system with unlimited mutual solubility in liquid and solid states at temperature T_2 at incomplete equilibrium (b)

Рис. 3. Диаграмма состояния (a) и химические потенциалы компонентов системы с неограниченной взаимной растворимостью компонентов в жидком и твердом состояниях при температуре T_2 при неполном равновесии (b)

fications, this approach can be applied to real multicomponent systems. In cases where the establishment of local equilibrium is impossible, a mathematical description of crystallization must include not only diffusion equations but also equations for the rates of interfacial chemical reactions for each component.

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INNOVATION IN METALLURGICAL
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ИННОВАЦИИ В МЕТАЛЛУРГИЧЕСКОМ
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ОБОРУДОВАНИИ, ТЕХНОЛОГИЯХ И МАТЕРИАЛАХ



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Original article

Оригинальная статья

EFFECT OF ANNULAR SEAMS ON STRESS-STRAIN STATE IN CYLINDRICAL CERAMIC SHELL MOLD DURING SOLIDIFICATION OF A STEEL CASTING IN IT

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Abstract. During the crystallization of liquid metal in a shell casting mold, significant normal stresses occur on its surfaces. On the inner – compressive, on the outer – tensile. They are especially pronounced at the initial moment of cooling time. This can lead to damage to the casting mold, and hence damage to the crystallizing metal casting. It is possible to reduce the level of stress-strain state in the surface layers by applying special annular (temperature) recesses (seams) to the outer and inner surfaces. In this paper, the problem of the influence of temperature seams in inner and outer layers of a shell mold (SM) on the level of its stress-strain state (SSS) during crystallization of a steel casting was formulated and solved. The normal stresses σ_{22} , σ_{33} , which occur both on the inner and outer surfaces of SM at the initial moment of metal casting and cooling of the steel casting, are accepted as a parameter of SM resistance to cracking. An axisymmetric problem for a cylindrical ceramic SM is considered. Based on the formulated objective function, the paper presents an algorithm for solving the problem using the equations of the linear theory of elasticity, the equation of thermal conductivity and the proven numerical method. As a result of solving the problem, the minimum number and locations of recesses on the inner and outer surfaces of SM, ensuring a decrease in normal stresses, were determined. The results of solving the problem are presented in the form of stress plots across the sections of the considered area. The authors analyzed the obtained results of SM resistance to cracking and gave recommendations on the use of the obtained results in various scientific and technical fields.

Keywords: investment casting, shell mold, stress state, temperature seam, crack resistance, stresses detection algorithm

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ВЛИЯНИЕ КОЛЬЦЕВЫХ ШВОВ НА НАПРЯЖЕННО-ДЕФОРМИРОВАННОЕ СОСТОЯНИЕ В ЦИЛИНДРИЧЕСКОЙ КЕРАМИЧЕСКОЙ ОБОЛОЧКОВОЙ ФОРМЕ ПРИ ЗАТВЕРДЕВАНИИ В НЕЙ СТАЛЬНОЙ ОТЛИВКИ

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Аннотация. При кристаллизации жидкого металла в оболочковой литейной форме на ее поверхностях возникают значительные нормальные напряжения: на внутренней – сжимающие, на наружной – растягивающие. Особенно ярко они проявляются в начальный момент времени охлаждения. Это может привести к повреждению литейной формы, а значит, и к повреждению кристаллизующейся металлической отливки. Снизить уровень напряженно-деформированного состояния в поверхностных слоях можно с помощью нанесения на внешнюю и внутреннюю поверхности специальных кольцевых (температурных) выточек (швов). В настоящей работе сформулирована и решена задача по влиянию температурных швов во внутренних и внешних слоях оболочковой формы (ОФ) на уровень её напряженно-деформированного состояния (НДС) при кристаллизации стальной отливки. В качестве параметра стойкости ОФ к трещинообразованию приняты нормальные напряжения σ_{22} , σ_{33} , возникающие как на внутренней, так и внешней поверхностях ОФ в начальный момент заливки металла и охлаждения стальной отливки. Рассматривается осесимметричная задача для цилиндрической керамической ОФ. На основе сформулированной целевой функции приведен алгоритм решения задачи с использованием уравнений линейной теории упругости, уравнения теплопроводности и апробированного численного метода. В результате решения задачи определено минимальное количество и места расположения выточек на внутренней и внешней поверхности ОФ, обеспечивающих уменьшение нормальных напряжений. Результаты решения задачи представлены в виде эпюр напряжений по сечениям рассматриваемой области. Дан анализ полученных результатов стойкости ОФ к трещинообразованию. Даны рекомендации по использованию результатов работы в различных научно-технических областях.

Ключевые слова: литье по выплавляемым моделям, оболочковая форма, напряженное состояние, температурный шов, трещиностойкость, алгоритм определения напряжений

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INTRODUCTION

Investment casting is widely used to produce geometrically complex castings while maintaining dimensional accuracy.

However, this process is characterized by a relatively high rejection rate of shell molds (SMs), primarily due to micro- and macrocracking as well as partial or complete mold failure during shape formation and, most critically, during key technological operations such as firing and metal pouring, especially at the early stage of cooling. These failures result from nonuniform heating across the thickness of the shell mold. Consequently, the limited durability of SMs is mainly associated with elevated stress–strain state (SSS) levels within the mold. To mitigate these effects, a range of technological solutions is applied in industrial practice.

The stress–strain state of multilayer shell casting molds has been extensively studied by both domestic and international researchers. In particular, studies [1; 2] examined the influence of shell mold shape and geometry; works [3; 4] focused on wall thickness; investiga-

tions [5; 6] addressed the properties of mold materials; and publications [7 – 9] analyzed the influence of casting geometry. Domestic studies dedicated to this problem are presented in [10 – 13]. Similar issues have also been investigated in relation to permanent mold casting [14; 15].

The present study continues the authors' research on the crack resistance of ceramic SMs used in investment casting. In earlier work, the SSS of cylindrical SMs during pouring with liquid metal was investigated using mathematical modeling. The theoretical analysis identified optimal physical parameters of the SM material and its morphological structure as key factors governing crack resistance. These results formed the basis for the development of new SM designs and configurations, for which several Russian Federation patents for inventions have been granted (including Nos. 2743439 and 2763359).

The theoretical investigations are based on a numerical method [16] applied to the following problem: liquid metal is poured into a multilayer SM, where it solidifies to form a casting; during subsequent cooling, the temperature field and stress–strain state in the cross-sections of the SM are determined.

At the initial stage of the study, the analysis was performed for a casting in the form of a cylinder with a spherical rounding at the lower part, which simulates a riser-type casting model within a shell mold.

Further theoretical investigations of cylindrical SMs focused on quantifying how the force interaction with the supporting filler (SF) and the interlayer friction parameters within the shell mold affect its stress–strain state (SSS) [17; 18]. These results also led to the granting of Russian Federation patents (Nos. 2769192 and 2788296).

As shown by industrial monitoring of casting-mold durability, the most unpredictable casting geometry is spherical (ball-shaped). For spherical SMs, the optimal supporting-filler (SF) coverage angle and its effect on the stress–strain state (SSS) of the SM were determined in [19].

Foreign studies [20 – 22] present numerical-method-based mathematical modeling of these processes, whereas works [23; 24] address modeling of the SSS in a solidifying casting.

The search for technological solutions to reduce critical SSS levels in SMs led to a new ceramic SM design [25]. The proposed design builds on a well-known approach for reducing thermal stresses in castings by introducing so-called stiffening ribs [26].

It was shown that the durability of spherical SMs increases when annular (temperature) seams, or recesses, are introduced on the inner surface of the mold. A similar durability increase associated with such seams has also been reported for metal casting molds.

When steel is poured into a spherical shell mold, the internal normal stresses across the cross-section are fully compressive and reach relatively high values. Accordingly, the theoretical analysis focused on conditions that reduce the absolute magnitude of these stresses. The most effective solution was a shell mold design incorporating annular seams in the inner layer (lining) [25]. In contrast, for cylindrical shell molds, the greatest risk is associated with tensile normal stresses at the mold–SF contact surface.

In the present study, we consider a ceramic shell mold with a cylindrical section and theoretically analyze the effect of temperature seams not only in the outer layer of the shell mold but also on its inner surface. A “rigid” mold configuration is examined, namely a single-layer ceramic mold with a constant shear modulus.

MATHEMATICAL FORMULATION OF THE PROBLEM

An axisymmetric body of revolution is considered (Fig. 1), comprising a liquid phase (metal) (*I*), a solidifying shell (*II*), a shell mold (*III*), a supporting filler (*IV*), circular recesses a_i on the surface of the lining (surface S_2), and circular recesses b_i on the contact surface between the mold (*III*) and the supporting filler (*IV*) (surface S_3).

Let A be a finite set of circular recesses a_i on surface S_2 ; $A = \{a_i, i = 1, \dots, n\}$; and let B be a finite set of circu-

lar recesses b_i on surface S_3 ; $B = \{b_k, k = 1, \dots, m\}$. Let $C = A \cup B$. As follows from the authors’ numerous studies, the critical stresses during steel pouring into a shell mold are σ_{22} and σ_{33} . Moreover, during steel cooling in a SM with cylindrical sections, the hazardous stresses are the tensile stresses σ_{22} on surface S_3 , whereas during steel cooling in a shell mold of spherical configuration, the hazardous stresses are the compressive stresses σ_{33} on surface S_2 .

Problem statement. Determine the minimum number and placement of recesses A on surface S_2 of the shell mold and the number of recesses B on surface S_3 , as well as their geometric arrangement, such that during metal cooling in the casting mold the maximum (absolute) stresses in the domain Q at $\tau = \tau^*$ remain within prescribed limits for the axisymmetric problem:

$$\begin{aligned} |\sigma_{33}| &< 55 \text{ MPa}; \\ \sigma_{22} &< 20 \text{ MPa}, \end{aligned} \quad (1)$$

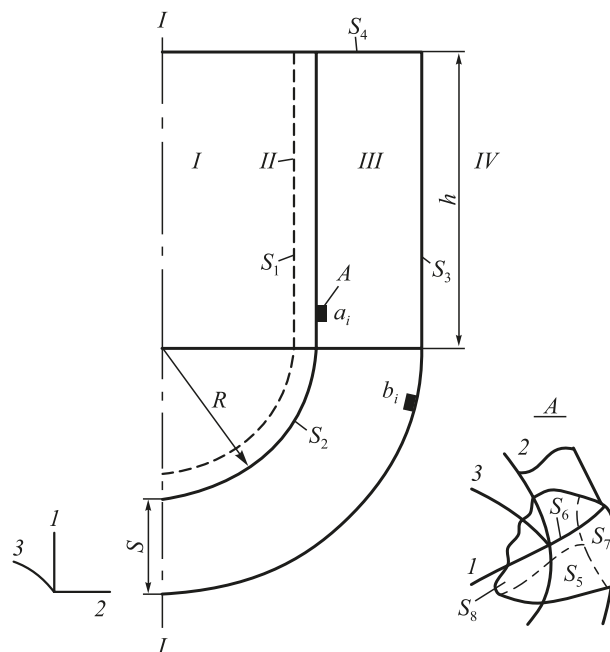


Fig. 1. Calculation scheme of the system with indication of the surface to the problem boundary conditions:

- S_1 – inner contact surface of liquid and solidified metal;
- S_2 – inner contact surface of solidified metal and ceramic mold;
- S_3 – outer contact surface of ceramic mold and supporting filler;
- I* – liquid metal; *II* – crust of solidified metal;
- III* – shell mold (SM); *IV* – supporting filler;
- R – radius of SM spherical part; h – height of SM cylindrical part

Рис. 1. Расчетная схема системы с указанием поверхности к граничным условиям задачи:

- S_1 – поверхность контакта жидкого и затвердевшего металла;
- S_2 – внутренняя поверхность контакта затвердевшего металла и керамической формы;
- S_3 – внешняя поверхность контакта керамической формы с опорным наполнителем;
- I* – жидкий металл; *II* – корочка затвердевшего металла;
- III* – оболочковая форма; *IV* – опорный наполнитель; R – радиус сферической части ОФ; h – высота цилиндрической части ОФ

here Q is the meridional cross-section domain; τ^* is the maximum cooling time after which the temperature over the domain Q begins to equalize and the normal stresses σ_{22} and σ_{33} start to decrease in absolute value.

The value of τ^* is determined from the function

$$F = \max |\sigma_{33}(\tau, Q)|, \quad (2)$$

under the constraint $\tau \leq 60$ s.

To compute F , we write the governing equations in a Cartesian coordinate system for each subdomain (Fig. 1) using linear elasticity theory:

– domain I :

$$\begin{aligned} \sigma_{11} = \sigma_{22} = \sigma_{33} = \sigma = P; \\ P_1 = -\gamma h; \quad \dot{\theta} = a_1 \Delta \theta; \end{aligned} \quad (3)$$

– domain II, III :

$$\begin{cases} \sigma_{ij,j} = 0, \quad i, j = 1, 2, 3; \\ s_{ij} - \sigma \delta_{ij} = 2G_p \varepsilon_{ij}^*; \quad \varepsilon_{ij}^* = \varepsilon_{ij} - \frac{1}{3} \varepsilon \delta_{ij}; \quad \varepsilon = \varepsilon_{ii}; \\ \varepsilon_{ii} = 3k_p \sigma + 3\alpha_p (\theta - \theta_p^*); \quad \varepsilon_{ij} = 0,5(U_{i,j} + U_{j,i}); \\ C_p \gamma \frac{\partial \theta}{\partial \tau} = \text{div}(\lambda \text{grad} \theta), \end{cases} \quad (4)$$

where σ_{ij} are the components of the stress tensor; σ is the hydrostatic stress; ε_{ij} are the components of the elastic strain tensor; h is the height of the liquid metal column;

$k_p = \frac{1-2\mu}{E}$ is the bulk compression coefficient; μ is Poisson's ratio; E is Young's modulus; $G_p(\theta)$ is the shear modulus in domain p ($p = II, III$); α_p is the linear thermal expansion coefficient; a_1 is the thermal diffusivity coefficient in domain (I); τ is time; θ is temperature; C_p is the specific heat capacity in domain (p); γ is the specific weight; θ_p^* is the initial temperature in domain (p); $\lambda = \lambda(\theta)$ is the thermal conductivity coefficient; summation over repeated indices is implied.

During the cooling of liquid metal, under the condition that the metal temperature $\theta_m \leq \theta_c$ (where θ_c is the crystallization temperature), the thickness of the solidified layer Δ_i is determined from the solution of the phase-transition equation.

Initial conditions:

$$\begin{aligned} \Delta|_{\tau=0} &= 0 \text{ – absence of solid phase;} \\ \theta_I^*|_{\tau=0} &= \theta_m^* \text{ – temperature of the poured liquid metal;} \\ \theta_{III}^*|_{\tau=0} &= \theta^* \text{ – initial mold temperature.} \end{aligned}$$

Boundary conditions of the problem in orthogonal coordinates (Fig. 1):

– for the axisymmetric problem

$$\begin{aligned} U_3 = 0; \quad \sigma_{31} = \sigma_{32} = 0; \quad \varepsilon_{13} = \varepsilon_{23} = 0; \\ \frac{\partial u_i}{\partial x_3} = 0; \quad \frac{\partial \sigma_{3i}}{\partial x_3} = 0, \quad i = 1, 2, 3; \end{aligned} \quad (5)$$

– on the axis of symmetry

$$U_2 = 0; \quad \sigma_{21} = 0; \quad q_n = 0; \quad \theta = \theta_m;$$

– on surfaces S_1, S_3, S_4

$$\begin{aligned} \sigma_{11}|_{S_1} = -P; \quad \sigma_{12}|_{S_1} = 0; \quad U_1|_{S_3} = 0; \quad \sigma_{21}|_{S_4} = 0; \\ \sigma_{22}|_{S_4} = 0; \quad \sigma_{11}|_{S_3} = 0; \\ \sigma_{12}|_{S_3} = -\psi \frac{U_{sl}}{U^*} \tau_s \cos(n_1 x_1); \\ \theta|_{S_3} = \theta^*; \quad \theta|_{S_2} = \theta_m, \end{aligned} \quad (6)$$

where U_{sl} is the displacement of the mold material during sliding relative to the SF (sand), U^* is the normalizing displacement; ψ is the friction coefficient between the mold and the supporting filler; τ_s is the conditional yield limit in shear; q_n is the heat flux.

For the thermal analysis, we used Dirichlet (first-kind) boundary conditions. To determine $\theta_m(\tau)$ and $\theta^*(\tau)$, the data from Ref. [27] were adopted:

$$\begin{aligned} \theta_m = \theta_m^* - \frac{\tau}{\tau_1} \theta_1; \\ 0 < \tau < 60 \text{ s}; \\ \theta^* = \theta_0 \left(1 + \sqrt{\frac{\tau}{\tau_2}} \right), \end{aligned} \quad (7)$$

here τ is the cooling time, s; $\theta_m^* = 1550$ °C; $\theta_1 = 100$ °C; $\theta_0 = 20$ °C; θ^* is the temperature on surface S_3 ; $\tau_1 = 60$ s; $\tau_2 = 1$ s.

The time τ does not exceed 60 s, since for $\tau \geq 60$ s the stresses in the SM do not pose a risk of failure.

The shear modulus of the SM is taken as

$$G_{\text{mold}} = 2960 \text{ kg/mm}^2. \quad (8)$$

The algorithm for solving system (4) under boundary conditions (5) – (7) is described in detail in Ref. [27].

The calculation yielded the following results:

$$F = -65.6 \text{ MPa}; \quad \tau^* = 21.65 \text{ s}. \quad (9)$$

The solution results are presented in Fig. 2 in the form of stress plots along the cross-section of the shell under consideration. The stresses σ_{22} and σ_{33} are highly significant. In the lining, σ_{22} and σ_{33} are negative and reach large magnitudes in the cylindrical part of the SM, with $|\sigma_{33}|$ being approximately 1.5 times greater than $|\sigma_{22}|$. In the spherical part, the difference between σ_{22} and σ_{33} is smaller; however, as the stresses approach the cylindrical region, the difference becomes pronounced. In the outer layer (at the contact with the supporting filler), the stresses σ_{22} are positive. In the spherical part, they are approximately equal in magnitude, whereas in the cylindrical part they increase

toward the upper portion of the mold. The stresses σ_{33} on this surface (surface S_3) are comparable to σ_{22} in the spherical region and are practically zero in the cylindrical region. Fig. 2 shows that the stresses σ_{33} and σ_{22} arising during steel pouring into the ceramic shell mold significantly exceed (in absolute value) the limits specified in (1).

Having determined the value of τ^* from (9), we proceed to solving the formulated problem. The process of steel cooling in a ceramic SM with temperature seams (annular recesses) is considered. In contrast to the previous problem, the cross-section Q represents a multiply connected domain. The initial and boundary conditions largely coincide with those of the previous problem. Boundary conditions (6) are supplemented as shown in Fig. 1):

$$(\sigma_{22} = \sigma_{21})|_{S_i} = 0, i = 5, 6; (\sigma_{11} = \sigma_{12})|_{S_i} = 0, i = 7, 8; (10)$$

Relationship (7) is also satisfied for the adopted value of the shear modulus given in (8).

SOLUTION ALGORITHM

1. The geometric dimensions of the domain, the final cooling time τ^* , the geometric dimensions of the recesses,

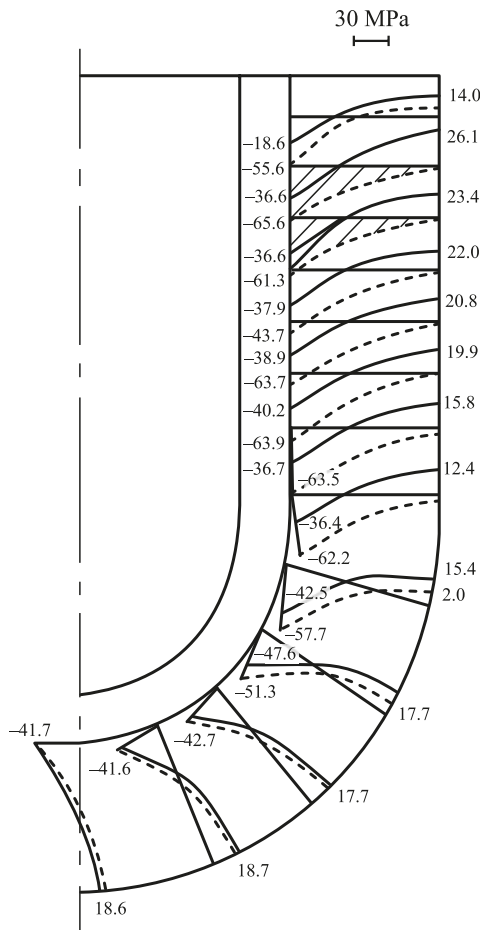


Fig. 2. Stress plots σ_{22} (—) and σ_{33} (---) along SM section

Рис. 2. Эпюры напряжений σ_{22} (—) и σ_{33} (---) по сечению ОФ

and their initial coordinates on surfaces S_2 and S_3 are specified: $a_i(0)$, $b_i(0)$. The cooling time τ^* is divided into a finite number of time steps: $\tau^* = \sum \Delta\tau_n$ (where n is the time-step index).

2. The investigated domain is divided into a finite number of elements by a system of orthogonal surfaces.

3. The arc lengths of the elements are calculated $S_{ij}^k(i, k = 1, 2, 3; i \neq k; j = 1, 2)$.

4. Initial and boundary conditions are specified for the elements forming the considered domain ((5), (6), (10)), as well as the constants of the physical and mechanical properties of the materials.

5. The temperature field at the time step $\Delta\tau_n$ is determined by a numerical solution of the heat conduction equation using an iterative scheme [27], taking into account the initial and boundary conditions at the given time step. The presence of recesses was not taken into account when solving the thermal problem.

6. If the condition $\theta|_{S_2} \leq \theta_c$ for domain (I) at surface S_2 , is satisfied, the thickness of the crystallized shell Δ_n is calculated.

7. System of equations (3) and (4) is solved taking in account boundary conditions (6) and (10), finite-difference analogs, and the developed methodology using the *Odyssey* software package¹. The stress fields σ_{ij} and displacement fields $U_i(i, j = 1, 2)$ are determined.

8. On surface S_3 , the mold-supporting contact is assessed for each element: if $\sigma_{11}|_{S_3} > 0 \Rightarrow \sigma_{11} = 0$, $\sigma_{12} = 0$, the boundary conditions are reassigned and operation 7 is performed.

9. A time step is performed. According to relations (7) the boundary conditions for solving the thermal problem are refined. If $\sum \Delta\tau_n < \tau^*$, operation 5 is performed. If $\sum \Delta\tau_n = \tau^*$, operation 10 is performed.

10. Over the domain Q , the values of σ_{33} and σ_{22} are analyzed, and the maximum values (in absolute magnitude) exceeding the constraints (1) are selected. Corresponding matrices are formed $[\bar{\sigma}_{22}]$, $[\bar{\sigma}_{33}]$. If constraints (1) are satisfied, operation 12 is performed.

11. From the matrices $[\bar{\sigma}_{22}]$, $[\bar{\sigma}_{33}]$ the maximum values are selected, and recesses are introduced in the corresponding cross-sections. Operation 7 is then performed.

12. The calculation process is completed.

RESULTS OF THE STUDY

Geometric parameters: $S = 5$ mm; $R = 20$ mm; $h = 50$ mm.

Time intervals $\Delta\tau_n$: 0.01, 0.02, 0.03, 0.04, 0.05, 0.1, 0.2, 0.3, 0.4, 0.5, 2, 5, 5, 5, 3, 3, 5, 5, 5, 5 s.

The following physical parameters of the poured steel at $\theta > 1000$ °C ($\theta_m^* = 1550$ °C) [27]:

¹ Onikov V.I., Prokudin A.N., Sergeeva A.M., Sevastyanov G.M. ODYSSEY. Certificate of State Registration of a Computer Program No. 2012111389. Registered in the Register of Computer Programs on December 13, 2012.

$$G = 1000 \text{ kg/mm}^2; \alpha = 12 \cdot 10^{-6} \text{ deg}^{-1};$$

$$\lambda = 0.0298 \text{ W/(mm} \cdot ^\circ\text{C)};$$

$$L = 270 \cdot 10^3 \text{ J/kg (latent heat of fusion)};$$

$$C = 444 \text{ J/(kg} \cdot ^\circ\text{C)}; \gamma = 7.80 \cdot 10^{-6} \text{ kg/mm}^3; \theta_c = 1450 \text{ }^\circ\text{C}.$$

Physical properties of the ceramic shell mold:

$$G = 2960 \text{ kg/mm}^2; \alpha = 0.51 \cdot 10^{-6} \text{ deg}^{-1};$$

$$\lambda = 0.000812 \text{ W/(mm} \cdot ^\circ\text{C)};$$

$$C = 840 \text{ J/(kg} \cdot ^\circ\text{C)}; \gamma = 2.0 \cdot 10^{-6} \text{ kg/mm}^3.$$

Recess dimensions: $a_i = 1 \times 2 \text{ mm}$, $b_i = 1 \times 3 \text{ mm}$.

Calculations performed using the above algorithm yielded the following results: $F = 5$; $a_i = 2$; $b_i = 3$. The geometric locations of the recesses (a_i , b_i) and the temperature distribution in the cross-section ($x_2 = 0$) are shown in Fig. 3. The resulting stresses σ_{33} and σ_{22} are presented in Fig. 4.

As shown in Fig. 4, all maximum values of the stresses σ_{33} (in absolute magnitude) and the tensile stresses σ_{22} satisfy the specified constraints (1), although in some cross-sections they are very close to the limiting values.

CONCLUSIONS

An axisymmetric problem was formulated and solved to optimize the cooling of a steel casting in a ceramic

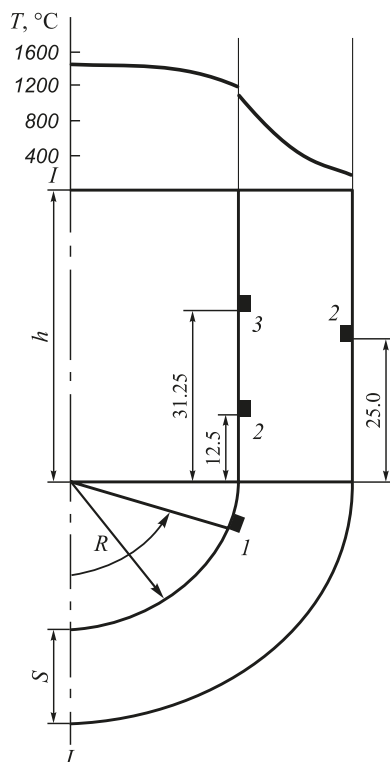


Fig. 3. Calculation scheme of the system with a group of recesses on the lining and outer layers of SM

Рис. 3. Расчетная схема системы с группой выточек на облицовочном и наружном слоях ОФ

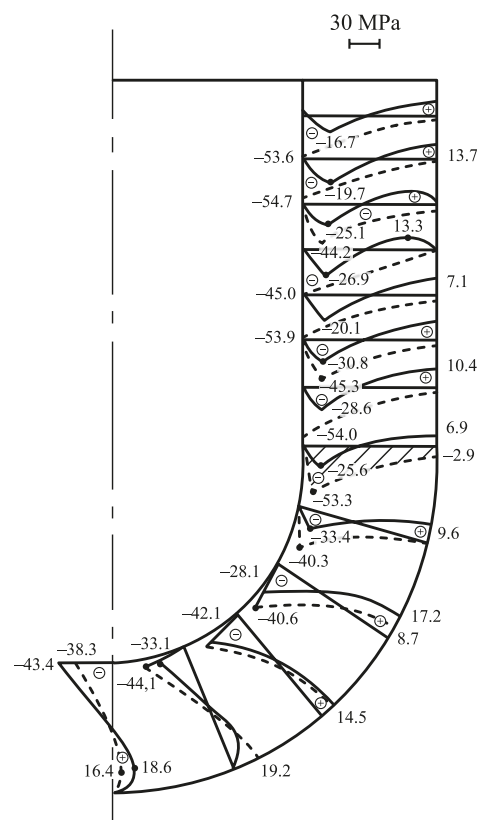


Fig. 4. Stress plots σ_{22} (—) and σ_{33} (---) along SM section

Рис. 4. Эпюры напряжений σ_{22} (—) и σ_{33} (---) по сечению ОФ

shell mold with cylindrical and spherical sections and annular temperature recesses.

The effectiveness of introducing annular recesses on the outer and inner mold surfaces in contact with the cooling metal was demonstrated.

The results can be used to analyze related processes, perform strength calculations, and support optimization studies.

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Оригинальная статья

CALCULATION OF TEMPERATURE AND THERMOELASTIC STRESSES IN STRIKERS DURING PRODUCTION OF HOLLOW STEEL BILLETS IN A UNIT OF COMBINED CASTING AND DEFORMATION. PART 2

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Abstract. The article solves the problem of determining thermoelastic stresses in calibrated strikers in a unit of combined casting and deformation during production of hollow steel billets using the calculation methodology developed by the authors. The authors substantiate the relevance of determining thermoelastic stresses in insulated strikers when compressing the wall of a hollow billet and at idle run when cooling the strikers with water, and describe the striker geometry to produce a hollow billet in one pass. The paper considers the initial data and temperature boundary conditions for calculating the temperature field of the striker during production of hollow billets in a unit of combined casting and deformation. The boundary conditions are given to determine the striker temperature as well as the values of heat flow and effective heat transfer coefficient. The results of calculating the temperature fields are performed in four sections and are presented for characteristic lines and points located on the striker contact surface and in the contact layer at a depth of 5 mm from the working surface. Dimensions of the finite element grid are given to calculate thermoelastic stresses in calibrated strikers using the finite element method with ANSYS package. The authors determined the magnitudes and patterns of distribution of thermoelastic stresses in a calibrated striker when compressing the wall of a hollow billet and at idle run when such a billet is produced in one pass in a unit of combined casting and deformation. Based on the calculated temperature values and the magnitude of the maximum compressive thermoelastic stresses on the contact surface of the strikers, it is proposed to use a pipe billet as a material for making strikers.

Keywords: unit, calibrated striker, casting, deformation, mold, hollow billet, temperature field, thermoelastic stresses, finite element

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РАСЧЕТ ТЕМПЕРАТУРЫ И ТЕРМОУПРУГИХ НАПРЯЖЕНИЙ В БОЙКАХ ПРИ ПОЛУЧЕНИИ СТАЛЬНЫХ ПОЛЫХ ЗАГОТОВОК НА УСТАНОВКЕ СОВМЕЩЕННОГО ЛИТЯ И ДЕФОРМАЦИИ. ЧАСТЬ 2

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Аннотация. В статье решается задача определения термоупругих напряжений в калиброванных бойках установки совмещенного литя и деформации при получении стальных полых заготовок с использованием разработанной авторами методики расчета. Авторы обосновывают актуальность определения термоупругих напряжений в калиброванных бойках при обжатии стенки полый заготовки и на холостом ходу при охлаждении бойков водой и представляют геометрию бойка для получения полый заготовки за один проход. Представлены исходные данные для расчета термоупругих напряжений в калиброванных бойках установки совмещенного литя и деформации при получении полых заготовок и температурные граничные условия для их расчета. Для определения температуры бойка приводятся граничные условия и значения теплового потока и эффективного коэффициента теплоотдачи. Результатом расчета термоупругих напряжений в четырех сечениях являются характерные линии и точки, расположенные на контактной поверхности бойка и в прикон-

тактом слое на глубине 5 мм от рабочей поверхности. Для расчета термоупругих напряжений в калиброванных бойках с использованием метода конечных элементов с применением пакета ANSYS приводятся размеры сетки конечных элементов. Авторы определили величины и закономерности распределения термоупругих напряжений в калиброванном бойке при обжатии стенки полой заготовки и на холостом ходу при получении за один проход такой заготовки на установке совмещенного литья и деформации. На основании расчетных значений температуры и величины максимальных сжимающих термоупругих напряжений на контактной поверхности в качестве материала для изготовления бойков предложено использовать трубную заготовку.

Ключевые слова: установка, калиброванные бойки, литье, деформация, кристаллизатор, полая заготовка, температурное поле, термоупругие напряжения, конечный элемент

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INTRODUCTION

The main and most heavily loaded elements of a unit of combined casting and deformation are the strikers [1; 2]. During the working stroke, they simultaneously compress the wall of a steel hollow shell and move it in the casting direction. In this process, combined stresses arise in the strikers during the working stroke due to the compression force and the thermal load [3 – 5]. Therefore, a relevant problem is to substantiate the selection of the striker material that ensures the ability to withstand cyclic loading under the combined action of mechanical and thermoelastic stresses [6 – 8].

To calculate the stress state of the striker in the unit, its temperature field was first determined based on solving the unsteady thermal conductivity equation with the corresponding initial and boundary conditions [9 – 11].

The outer and inner diameters of the hollow billet made of 09G2S steel are 100 and 60 mm, respectively. The angular speed of the eccentric shafts was taken as 40 min^{-1} . At this speed, the contact time of the striker with the wall of the hollow billet during the working stroke is 0.375 s, and the pause time is 1.125 s. The temperature of the hollow billet in the striker entry zone is 1200°C [12 – 14].

The geometry of the striker with dimensions is given in [9]. Iterative calculations showed that, to ensure a temperature of 350°C on the striker contact surface at a heat transfer coefficient of $2000 \text{ W}/(\text{m}^2 \cdot \text{K})$, the heat flux density is $4.6 \text{ MW}/\text{m}^2$ [15].

It was established that, under the action of heat flow during compression of the hollow billet wall by the strikers, the temperature on the striker contact surface lies in the range of $370 - 451^\circ\text{C}$. Subsequently, during idle operation with water cooling, the temperature of the striker contact surface decreases and falls within the range of $289 - 370^\circ\text{C}$.

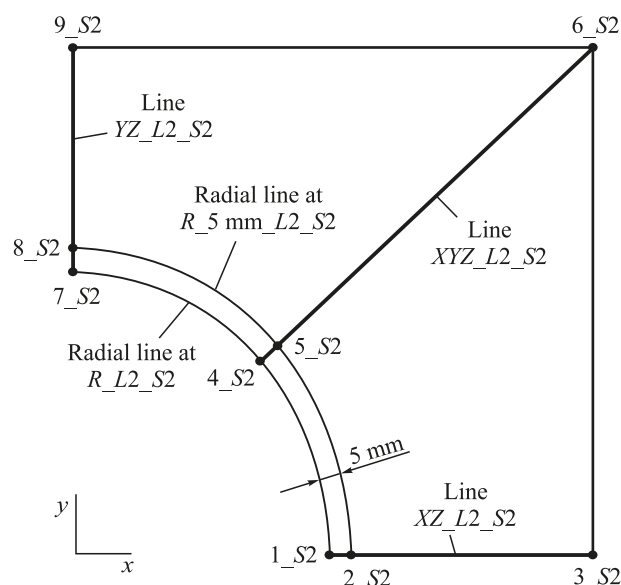
Based on the obtained temperature fields, thermoelastic stresses in the strikers of the unit were determined by the finite element method using the ANSYS package [16] (with elements of the calculation methodologies described in [16 – 18]). The finite element size in the deformation

zone and in the calibrating sections of the strikers was taken as 1 mm. By symmetry, the calculation was performed for one half of the striker shown in Fig. 1 [9]. The calculation results for radial (SR), tangential (S θ), axial (SZ), and equivalent (SEQV) thermoelastic stresses are presented only for the characteristic lines XZ_L2_S2, XYZ_L2_S2, and R_L2_S2 and for the points of section 2 associated with the striker contact surface (Fig. 1).

The Table presents the values of thermoelastic stresses at the points of the characteristic lines of section 2 under the action of heat flow (HF) and during water cooling (WC).

Fig. 2 shows the patterns of thermoelastic stress distribution in the strikers along the characteristic line R_L2_S2.

Let us consider the distribution of thermoelastic stresses in the strikers along the characteristic lines of section 2 (Fig. 1), where the maximum compressive stresses arise during compression of the wall of the steel hollow billet. Thus, at point 1_S2 of the XZ_L2_S2 line, which passes along the axis of symmetry of the striker



Values of radial, tangential, axial and equivalent stresses at points of section 2 lines due to the effect of heat flow (HF) and cooling with water (WC) on the striker

Значения радиальных, тангенциальных, осевых и эквивалентных напряжений в точках линий сечения 2 от воздействия на боек теплового потока (HF) и охлаждения водой (WC)

Point	Stress, MPa							
	SR	S0	SZ	SEQV	SR	S0	SZ	SEQV
	At the end of the pause (WC)				At the end of contact (HF)			
1_S2	–9	–58	–191	178	–39	–589	–703	671
2_S2	–27	–243	–361	317	–33	–218	–348	297
3_S2	–56	–114	–62	56	–38	–113	–42	61
4_S2	–4	148	–142	258	–34	–381	–652	592
5_S2	–13	–91	–312	289	–18	–58	–295	278
6_S2	–542	–454	141	459	–561	–470	130	459
7_S2	21	25	341	342	–19	–59	–63	71
8_S2	53	–1	355	340	–32	–1	340	372
9_S2	0	1	857	856	0	1	859	858

caliber, at the contact surface, the maximum compressive axial stresses SZ reach -703 MPa (Fig. 2, Table).

During idle operation with water cooling, compressive thermoelastic stresses $S0$ and SR at the striker contact surface at point 1_S2 are significantly lower and amount to -58 and -191 MPa, respectively (Fig. 2, Table).

It should be noted that along the XYZ_L2_S2 line, the compressive thermoelastic stresses $S0$ and SZ at the striker contact surface during compression of the hollow billet wall are quite high; at point 4_S2, they are -381 and -652 MPa, respectively (Fig. 2, Table).

During idle operation, at the striker contact surface at point 4_S2, the tangential stresses $S0$ are tensile, with a maximum value of 148 MPa, while the axial stresses SZ are compressive, with a value of -142 MPa. However, through the striker thickness, the axial stresses initially

increase to -312 MPa, then decrease, change sign, and become tensile, reaching a maximum value of 141 MPa (see Table).

Let us consider the distribution of thermoelastic stresses along the radius of the striker contact surface (line R_L2_S2 , Fig. 1).

During compression of the hollow billet wall, at point 1_S2 of the striker contact surface, the thermoelastic stresses $S0$ and SZ are compressive, amounting to -589 and -703 MPa, respectively. Along the radius, they decrease and at point 4_S2 become -381 and -652 MPa, while at point 7_S2 they are -59 and -63 MPa, respectively (Fig. 2, Table).

During idle operation, at point 1_S2 of the striker contact surface, compressive thermoelastic stresses $S0$ and SZ arise with values of -58 and -191 MPa, respec-

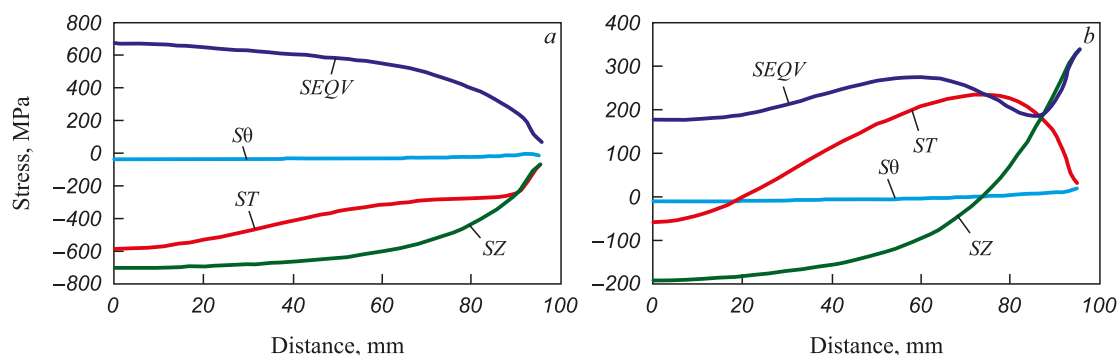


Fig. 2. Nature of thermoelastic stresses along the R_L2_S2 line:
a – due to the effect of heat flow (TP) on the striker (at the end of the contact);
b – due to the effect of cooling with water (OXLI) (at the end of the pause)

Рис. 2. Характер термоупругих напряжений вдоль линии R_L2_S2 :
a – от воздействия на боек теплового потока (ТР) (в конце контакта);
b – от охлаждения водой (ОХЛ) (в конце паузы)

tively. Further along the radius of the striker contact surface, at point 7_S2, they become tensile, reaching 25 and 341 MPa, respectively (Fig. 1, Table).

As a result of the study, it was established that the temperature of the striker contact surface during compression of the steel hollow billet wall can reach 451 °C. At this temperature, the compressive thermoelastic stresses at the striker contact surface reach 703 MPa.

The calculated values of thermoelastic stresses correlate well with the results obtained in similar studies using the QForm analysis software system [19].

Considering the above, the strikers of the unit of combined casting and deformation for producing steel hollow billets should be manufactured from 4Kh4VMFS steel. This is a die steel with increased resistance to heat checking and high wear resistance. The conditional yield strength of the steel at a temperature of 500 °C is 1309 MPa, which significantly exceeds the maximum thermoelastic stresses caused by the thermal load [20]. This steel is also used for manufacturing tooling for high-speed machine forging and upsetting on horizontal forging machines.

CONCLUSIONS

A procedure has been developed for calculating thermoelastic stresses in calibrated strikers of a unit of combined casting and deformation during the production of steel hollow pipe billets. The magnitudes and patterns of thermoelastic stress distribution in three sections of the calibrated striker during the production of steel hollow billets in a unit of combined casting and deformation have been determined. It has been established that the maximum compressive thermoelastic stresses at the striker contact surface reach 703 MPa. It is proposed that the strikers of the unit of combined casting and deformation for producing steel hollow billets be manufactured from 4Kh4VMFS steel, whose yield strength at a temperature of 500 °C is 1309 MPa.

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ИННОВАЦИИ В МЕТАЛЛУРГИЧЕСКОМ
ПРОМЫШЛЕННОМ И ЛАБОРАТОРНОМ
ОБОРУДОВАНИИ, ТЕХНОЛОГИЯХ И МАТЕРИАЛАХ



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Original article

Оригинальная статья

ANALYSIS OF PROCESSING 40Kh13 MARTENSITIC STAINLESS STEEL BILLET OBTAINED BY WIRE ELECTRON BEAM ADDITIVE MANUFACTURING

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Abstract. The authors investigated the microstructure and mechanical properties of the samples obtained by the method of wire electron beam additive manufacturing (WEBAM), and their machinability by milling forces using the Taguchi method. Grains of previous austenite and annealed martensite were observed in the samples in various directions. The grains of the previous austenite grow along the surfacing direction and exhibit a pronounced orientation. On the lateral surface of the sample, the grains of the previous austenite are columnar, their hardness is approximately 505 HV_{0.1}. On the upper surface of the sample, the grains of the previous austenite are isometric, their hardness is approximately 539 HV_{0.1}. The degree of transformation into martensite varies in different parts of the sample. In the part close to the lateral surface, martensite is shallower and the previous austenitic grain boundaries are not observed. Its hardness is approximately 514 HV_{0.1}. In the lower part of the sample, due to multiple thermal cycles, martensite decomposes, while its hardness is low and is approximately 480 HV_{0.1}. In the upper part of the sample, martensite and the previous austenitic grain boundaries are observed, the hardness is approximately 513 HV_{0.1}. Due to the high hardness of the sample during climb milling, a stronger impact of the cutting edge on the sample leads to an increase in cutting force. Due to the low plasticity of the sample during conventional milling, a decrease in the volume of material pressed into the back surface of the tool leads to a decrease in cutting force. As the feed rate to the tooth increases, deformation of the material increases, the temperature increases, which leads to a decrease in the material strength. Reducing the material strength slows down the growth of cutting force.

Keywords: electron beam additive manufacturing, microstructure, hardness, machinability, martensitic stainless steel, Taguchi method

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АНАЛИЗ ПРОЦЕССА ОБРАБОТКИ ЗАГОТОВКИ ИЗ МАРТЕНСИТНОЙ НЕРЖАВЕЮЩЕЙ СТАЛИ 40X13, ПОЛУЧЕННОЙ МЕТОДОМ ПРОВОЛОЧНОГО ЭЛЕКТРОННО-ЛУЧЕВОГО АДДИТИВНОГО ПРОИЗВОДСТВА

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Аннотация. Авторы исследовали микроструктуру и механические свойства образцов, полученных методом проволочного электронно-лучевого аддитивного производства (WEBAM), и их обрабатываемость по силам фрезерования с использованием метода Тагучи. В образцах в различных направлениях наблюдались зерна предыдущего аустенита и отожженный мартенсит. Зерна предыдущего аустенита растут вдоль направления наплавки и демонстрируют выраженную ориентацию. На боковой поверхности образца зерна предыдущего аустенита являются столбчатыми, их твердость составляет примерно 505 HV_{0,1}. На верхней поверхности образца зерна предыдущего аустенита являются изометрическими, их твердость составляет примерно 539 HV_{0,1}. В разных частях образца степень превращения в мартенсит различается. В части, близкой к боковой поверхности, мартенсит более мелкий и предыдущие аустенитные межзеренные границы не наблюдаются. Его твердость составляет примерно 514 HV_{0,1}. В нижней части образца, вследствие множественных термоциклов, происходит разложение мартенсита, при этом его твердость низкая и составляет примерно 480 HV_{0,1}. В верхней части образца наблюдаются мартенсит и предыдущие аустенитные межзеренные границы, твердость составляет примерно 513 HV_{0,1}. Из-за высокой твердости образца при попутном фрезеровании более сильный удар режущей кромки о образец приводит к увеличению силы резания. Вследствие низкой пластичности образца при встречном фрезеровании уменьшение объема материала, вдавливаемого в заднюю поверхность инструмента, приводит к снижению силы резания. При увеличении скорости подачи на зуб деформация материала увеличивается, температура повышается, что приводит к снижению прочности материала и, соответственно, замедляет рост силы резания.

Ключевые слова: проволочное электронно-лучевое аддитивное производство, микроструктура, твердость, обрабатываемость, мартенситная нержавеющая сталь, метод Тагучи

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INTRODUCTION

In recent years, wire electron beam additive manufacturing (*Wire Electron Beam Additive Manufacturing* – WEBAM) has attracted growing interest because it combines a high deposition rate (up to 2500 cm³/h) with the ability to produce parts that show high strength and fatigue resistance [1 – 6]. The process is also flexible from a manufacturing standpoint: it can operate with wire diameters down to 0.5 mm and can be used to obtain materials with a prescribed phase content, for example in nickel–aluminum alloy systems [7].

However, in contrast to conventional casting and forging, thermal processes in additive manufacturing are more complex, resulting in uncertainty in the mechanical properties of printed parts. For example, studies on 10Kh12N10T stainless steel reported that increased strength is associated with a high density of dislocations and intermetallic compounds concentrated at interlayer boundaries [8]. A study of heat dissipation conditions during the deposition of 308LSi steel showed that the use of copper as a cooling medium results in hardness values

approximately 5 % higher than those obtained with air, while the hardness of the upper part of the sample is about 8 % higher than that of the lower part [9]. When thin-walled parts are produced, a columnar crystalline structure tends to develop, which leads to pronounced anisotropy; the strength difference between longitudinal and transverse directions can reach 70 MPa [10]. Due to the high heat input, WEBAM-produced parts exhibit inferior as-built surface quality, which adversely affects subsequent conventional machining operations such as milling and turning [11; 12]. 40Kh13 martensitic stainless steel (an analogue of AISI 420) is widely used for large, complex-shaped parts because it is relatively inexpensive while offering moderate corrosion resistance and high strength. At the same time, its high hardness limits machinability and accelerates tool wear [13]. In addition, because martensitic steels are highly sensitive to temperature variations and because the thermal gradient during deposition is strongly directional, WEBAM-built samples often exhibit anisotropy in both microstructure and mechanical properties [14; 15]. As a result, machining of such martensitic steels becomes even more uncertain.

The *Taguchi method* applied in this study is an approach to experimental optimization that uses the *signal-to-noise ratio* (SNR) together with *orthogonal arrays* to identify an optimal combination of parameters [16 – 20].

The present work therefore focuses on the microstructure and microhardness of WEBAM-deposited 40Kh13 martensitic stainless steel in different directions and regions, and evaluates machinability using the Taguchi method.

MATERIALS AND METHODS

The samples were produced by wire electron beam additive manufacturing on equipment designed and manufactured at Tomsk Polytechnic University. A filler wire with a diameter of 1.2 mm made of 40Kh13 martensitic stainless steel was used as the feedstock. The chemical composition of the wire was as follows (wt. %: 0.41 C, 13.2 Cr, 0.53 Si, 0.51 Ni, 0.49 Mn, 0.017 S, 0.021 P, with iron as the balance. The substrate was manufactured from the same material (40Kh13 steel). The dimensions of the samples were 70×15×14 mm (length×width×height). The printing parameters were

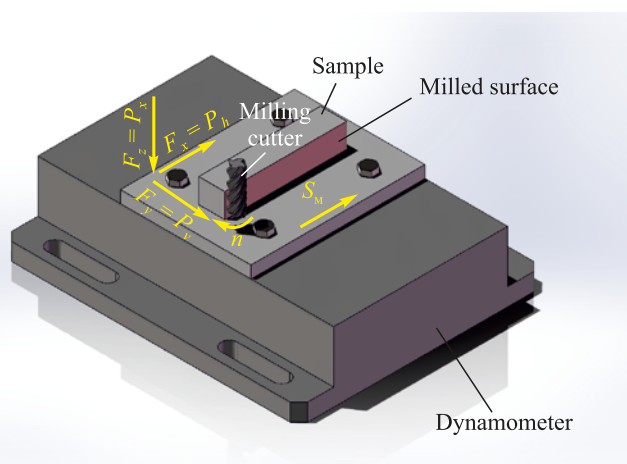
as follows: accelerating voltage of 40 kV, beam current of 21 mA, scanning beam diameter of 3 – 5 mm, wire feed rate of 1050 mm/min, and wire feeding angle of 45°. The deposition process was carried out in a vacuum at a pressure of $5 \cdot 10^{-3}$ Pa.

Microstructural analysis was performed using a BIOMED MMP-1 metallographic microscope and by scanning electron microscopy with a JEOL JSM-6000 microscope. Microhardness measurements were conducted using an EMCO-TEST DuraScan-10 hardness tester, with a load holding time of 10 s.

The machinability of the samples was evaluated based on milling forces. Machining experiments were carried out on a CNC EMCO CONCEPT Mill 155 machine. Cutting forces were measured using a Kistler 9257 V dynamometer (Switzerland). In the dynamometer software (originally intended for turning-force measurements), the force components are displayed as F_x , F_y and F_z . In the present milling tests, these components were interpreted as the milling-force components P_h , P_v и P_x , respectively (Fig. 1). For machining, an end milling cutter with a diameter of 8 mm and four teeth, manufactured by GESAC, was used. The helix angle (ω) of the milling cutter was 35°, while the rake angle (γ) and clearance



a



b

Fig. 1. Appearance (a) and model (b) of the dynamometer, milling cutter and sample installation

Рис. 1. Внешний вид (a) и модель (b) установки динамометра, фрезы и образца

Table 1. Experimental levels of the factors

Таблица 1. Экспериментальные уровни факторов

Factor	Level		
	1	2	3
A: milling scheme	Climb milling (CI)	Conventional milling (Con)	
B: feed rate s_m , mm/min	56	28	5,6
C: milling cutter rotational speed n , rpm	2000	1000	500

angle (α) were 7° and 5° , respectively. The base material of the milling cutter was VK8 cemented carbide (92 % tungsten carbide and 8 % cobalt as a binder). The surface of the milling cutter was coated with a wear-resistant AlCrSiN coating.

The experimental levels of the factors are presented in Table 1.

To determine the minimum milling force, the signal-to-noise ratio $S/N(\eta)$ was used:

$$S/N(\eta) = -10 \lg \left(\frac{1}{j} \sum_{i=1}^j P_i^2 \right), \quad (1)$$

where P_i is the force value measured during the i -th milling pass.

The mean milling force P_{avg} was calculated using the following expression:

$$P_{\text{avg}} = \frac{1}{x} \sum_{i=1}^x P_i, \quad (2)$$

where x is the number of experimental repetitions.

RESULTS AND DISCUSSION

Analysis of the microstructure and mechanical properties of the sample in different directions

As shown in Fig. 2, the OZ axis corresponds to the deposition build direction, the OY axis is oriented along the weld bead, and the OX axis represents the transverse direction relative to the weld bead.

At low magnification, the XOZ and YOZ planes reveal austenite grains prior to the martensitic transformation (previous austenite grains), which exhibit a columnar

morphology with their long axes aligned along the OZ direction. This behavior is associated with the thermal conditions during deposition: the lower layers experience multiple thermal cycles and accumulate heat, resulting in the formation of a dominant heat flow opposite to the deposition direction (i.e., opposite to the OZ axis). As a consequence, a pronounced preferential growth orientation of the previous austenite grains develops along the OZ axis. In contrast, grains observed on the XOY plane display an equiaxed morphology. This can be attributed to the high deposition rate combined with relatively low heat input, which suppresses grain growth along the OY direction and equalizes heat dissipation conditions along the OX and OY directions. This promotes equal grain growth rates along the OX and OY directions, resulting in the formation of equiaxed grains. In addition, while the degree of corrosion within individual previous austenite grains is nearly uniform, noticeable differences are observed between different grains. This behavior may be related to variations in the extent and morphology of martensitic transformations caused by elemental segregation under the high cooling rates characteristic of the deposition process [15].

Fig. 2, *b* clearly shows that a large number of needle-like or plate-like martensitic structures formed within the previous austenite grains as a result of a diffusionless phase transformation.

The martensite within these grains is distributed in the form of an interwoven network. The orientation of martensite varies significantly from one austenite grain to another. Furthermore, a pronounced transgranular phenomenon is observed at the boundaries of previous austenite grains, which may be associated with local stress concentrations or energy gradients. In the YOZ and XOZ planes, the sizes of the previous austenite grains are nearly identical, further confirming the similarity of temperature

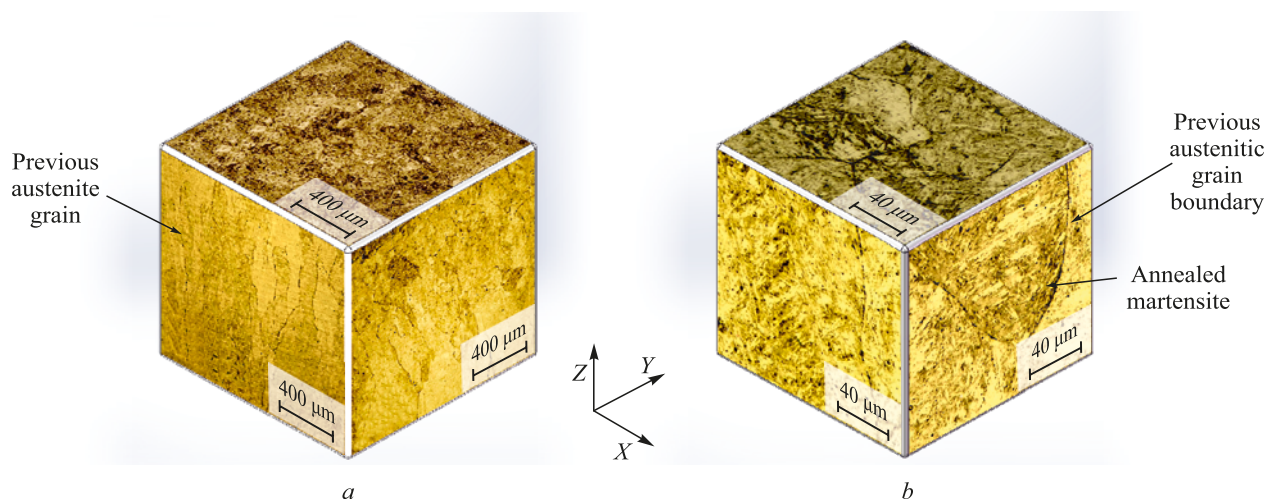


Fig. 2. Microstructure of the samples deposited by WEBAM in different planes at low (*a*) and high (*b*) magnification

Рис. 2. Микроструктура образцов, наплавленных методом WEBAM, в разных плоскостях при малом (*a*) и большом (*b*) увеличении

gradients along the OX and OY directions. In contrast, the smaller size of previous austenite grains in the XOY plane leads to a denser martensitic network with a shorter average length of martensitic features. This difference can be explained by faster cooling along the OX and OY directions or by the formation of more pronounced compositional gradients in this plane due to elemental segregation, which suppresses grain coarsening [21]. As a result, the frequency of transgranular phenomena at grain boundaries in the XOY plane is also reduced.

A scanning electron microscopy image of the sample microstructure in the XOZ plane is presented in Fig. 3. Plate-like martensite and carbides are clearly visible. The measured hardness of the sample in the XOZ plane is $504.67 \text{ HV}_{0.1}$, which is significantly lower than that of quenched martensite (750 HV) [21]. However, the thickness of the plate-like martensitic layer is relatively small, amounting to $1.23 \pm 0.56 \mu\text{m}$. This reduction in hardness can be explained by the fact that the sample underwent several thermal cycles, during which martensite experienced thermal activation and partial decomposition, leading to an increased fraction of retained austenite. During the deposition of the first layer, the austenitic phase rapidly transforms into quenched martensite owing to the high cooling rate, resulting in the maximum martensite content. However, when the second or third layer is deposited, the first layer remains within the heat-affected zone. This promotes carbon diffusion at martensite – martensite or martensite – retained austenite interfaces, causing partial reversion of martensite to austenite. At the same time, chromium atoms impede carbon diffusion, thereby limiting martensite decomposition and restricting it to a partial transformation. Consequently, despite repeated thermal cycles, the hardness of the mate-

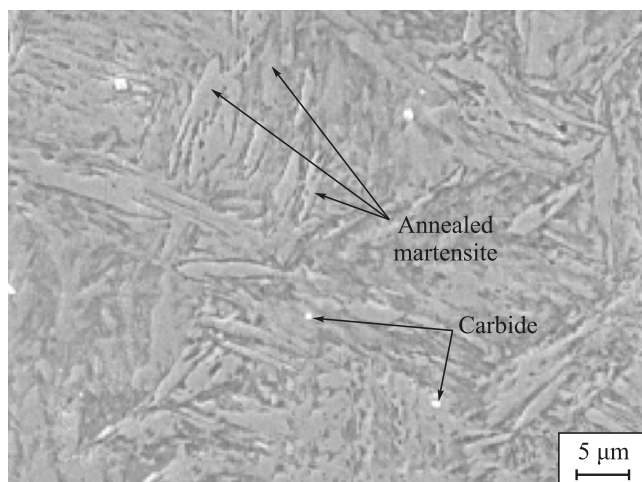


Fig. 3. Microstructure of the sample deposited by WEBAM in XOZ plane observed using the scanning electron microscope

Рис. 3. Микроструктура образца, наплавленного методом WEBAM в плоскости XOZ , наблюдаемая с помощью сканирующего электронного микроскопа

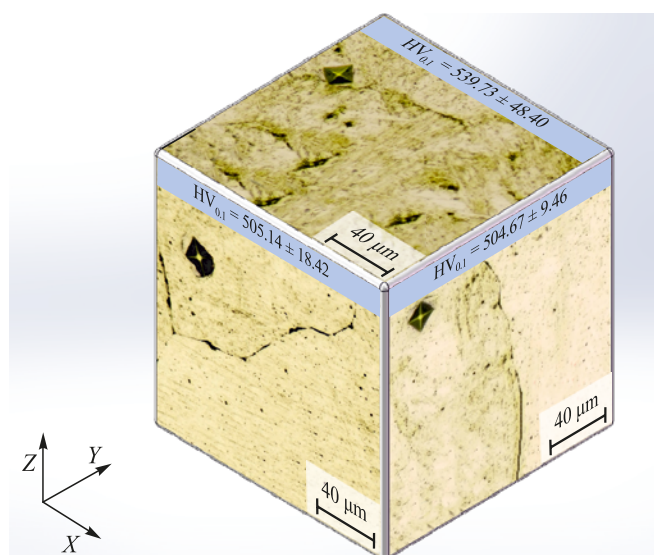


Fig. 4. Microhardness of the sample deposited by WEBAM in different directions

Рис. 4. Микротвердость образца, наплавленного методом WEBAM, в разных направлениях

rial remains significantly higher than that of austenitic steel [8].

The hardness of the sample in different directions is shown in Fig. 4. As noted above, the higher cooling rates along the OX and OY directions result in smaller previous austenite grain sizes in the XOY plane and a higher degree of martensitic transformation. This promotes the formation of a more continuous and homogeneous martensitic network in the XOY plane, which effectively impedes dislocation motion and increases the hardness to $539.73 \text{ HV}_{0.1}$. In contrast, the cooling rate along the OZ direction is lower, while temperature gradients along the OX and OY directions are similar. This leads to a lower degree of martensitic transformation in the XOZ and YOZ planes, facilitating dislocation motion and multiplication and promoting plastic deformation. As a result, the hardness values on these two planes are nearly identical, amounting to 505.14 and $504.67 \text{ HV}_{0.1}$, respectively.

Analysis of the microstructure in different parts of the sample

Microstructural images obtained from different regions of the sample are shown in Fig. 5. In the region close to the lateral surface of the sample (Fig. 5, *a*), the previous austenitic grain boundaries are indistinct. The precipitated carbide inclusions are small, and the clusters of annealed martensite are fine and uniformly distributed. This behavior is attributed to more favorable heat dissipation conditions near the lateral surface, where the cooling rate is higher. The increased cooling rate promotes the formation of a large number of fine martensitic fea-

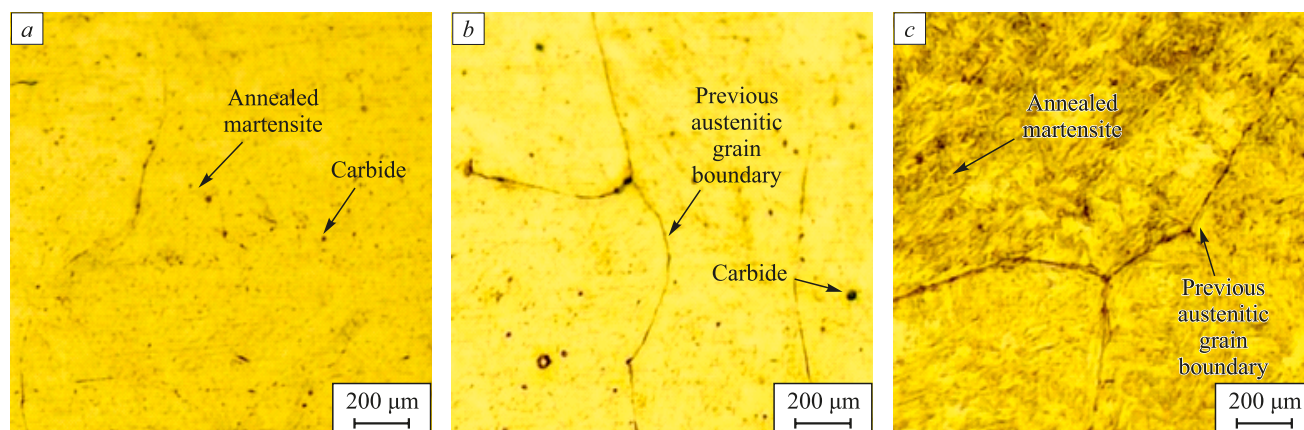


Fig. 5. Microstructure of the sample deposited by WEBAM in YOZ plane in the part close to the lateral surface (a), in the lower part (b), and in the upper part (c)

Рис. 5. Микроструктура образца, наплавленного методом WEBAM, в плоскости YOZ в части, близкой к боковой поверхности (a), в нижней части (b) и в верхней части (c)

tures. At the same time, rapid cooling enhances the manifestation of a transgranular phenomenon during the martensitic transformation, as a result of which the previous austenitic grain boundaries become blurred. In the lower part of the sample (Fig. 5, b), the previous austenitic grain boundaries are clearly visible, together with large carbide precipitates and a relatively small amount of martensite. This is explained by the lower cooling rate in the lower region compared with the area close to the lateral surface, which leads to a less pronounced transgranular phenomenon and, consequently, to clearly defined previous austenitic grain boundaries. In addition, because the lower part of the sample experienced a relatively larger number of thermal cycles during deposition, the amount of precipitated carbides and decomposed martensite increased. In the upper part of the sample (Fig. 5, c), the microstructure differs from both the lower region and the area close to the lateral surface. The previous austenitic grain boundaries

and martensite are clearly visible, and the thickness of the martensitic layer is greater. This is associated with the smaller number of thermal cycles experienced by the upper part of the sample, which results in only limited martensite decomposition and allows its structure to remain well defined. At the same time, compared with the region near the lateral surface, the cooling rate in the upper part was lower, leading to the formation of a thicker martensitic layer.

Indentations obtained during microhardness measurements in different regions of the sample are shown in Fig. 6. The microhardness is lowest in the lower part of the sample due to martensite decomposition. In the region close to the lateral surface, the microhardness is slightly higher than that measured in the upper part of the sample; however, both values exceed the microhardness measured in the central region of the YOZ plane shown in Fig. 4. This can be explained by the higher

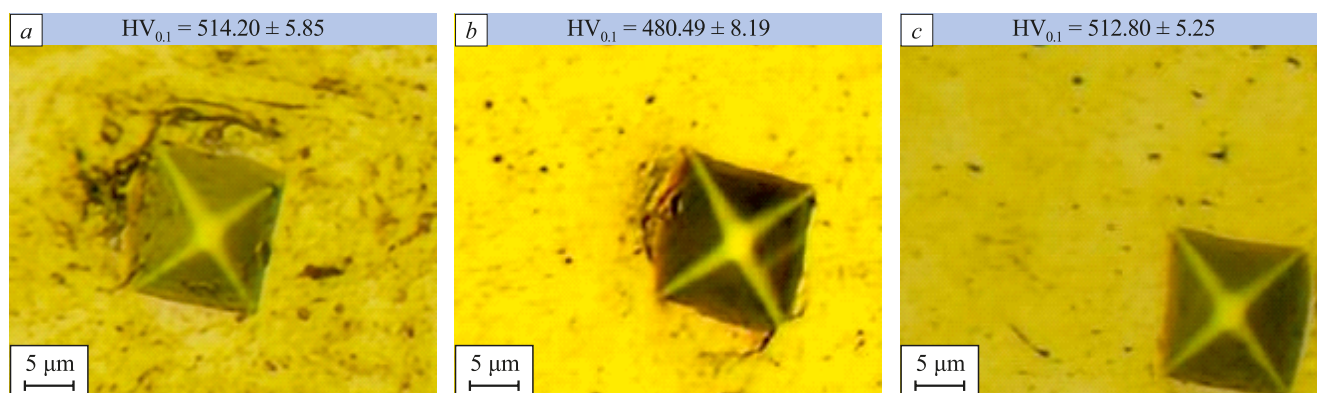


Fig. 6. Microhardness of the sample deposited by WEBAM in YOZ plane in the part close to the lateral surface (a), in the lower part (b), and in the upper part (c)

Рис. 6. Измерение микротвердости образца, наплавленного методом WEBAM, в плоскости YOZ в части, близкой к боковой поверхности (a), в нижней части (b) и в верхней части (c)

martensite content in the upper part of the sample and by the smaller size of martensitic regions near the lateral surface, both of which contribute to increased hardness. In addition, owing to the smaller size of the previous austenitic grains, the microhardness on the XOY plane is higher than that on the YOZ plane in the different parts of the sample.

Analysis of machinability based on milling forces

Using the Taguchi method, the present study investigated the machinability of a 40Kh13 martensitic stainless steel sample produced by the WEBAM process during milling along the OX direction on the XOZ surface. Based on calculations using Eqs. (1) and (2), the mean values of the experimental results and the corresponding signal-to-noise ratios are presented in Table 2.

The expression used to calculate the mean signal-to-noise ratio is given by:

$$S/N_{\text{avg}} = \frac{1}{m} \sum_{i=1}^m S/N_j, \quad (3)$$

where m is the number of parameter combinations at the same level of the given factor [19].

As follows from Tables 2 and 3, during climb milling the milling force P_h , acting along the feed direction, is lower than that obtained during conventional milling, whereas the force P_v , acting perpendicular to the feed direction, is higher than in conventional milling (Fig. 7). These differences arise because, when machining with a new milling cutter, the dominant tangential force P_z (acting along the cutting speed v) during conventional milling is oriented almost along the feed direction, while during climb milling it is oriented nearly perpendicular to the feed direction [14]. In addition, owing to the high hardness of martensitic stainless steel, the impact force acting on the cutting edge during climb milling is higher, whereas during conventional milling the low plasticity of the material reduces the volume of material pressed

Table 2. Mean values of experimental results and signal-to-noise ratio

Таблица 2. Средние значения экспериментальных результатов и отношение сигнал/шум

Machining parameters			Mean values of experimental results, N			Signal-to-noise ratio S/N , dB		
A : milling strategy	B : s_M , mm/min	C : n , rpm	$P_{h\text{avg}}$	$P_{v\text{avg}}$	$P_{x\text{avg}}$	S/N_{P_h}	S/N_{P_v}	S/N_{P_x}
Climb milling	56	2000	52.11	146.10	22.87	−34.33	−43.29	−27.18
Climb milling	28	1000	74.25	159.77	19.87	−37.41	−44.07	−25.96
Climb milling	5.6	500	43.09	82.03	18.22	−32.76	−38.28	−25.21
Conventional milling	56	500	212.12	181.89	26.09	−46.53	−45.19	−28.33
Conventional milling	28	1000	122.06	115.01	17.45	−41.73	−41.21	−24.84
Conventional milling	5.6	2000	63.24	22.09	8.25	−36.02	−26.88	−18.35

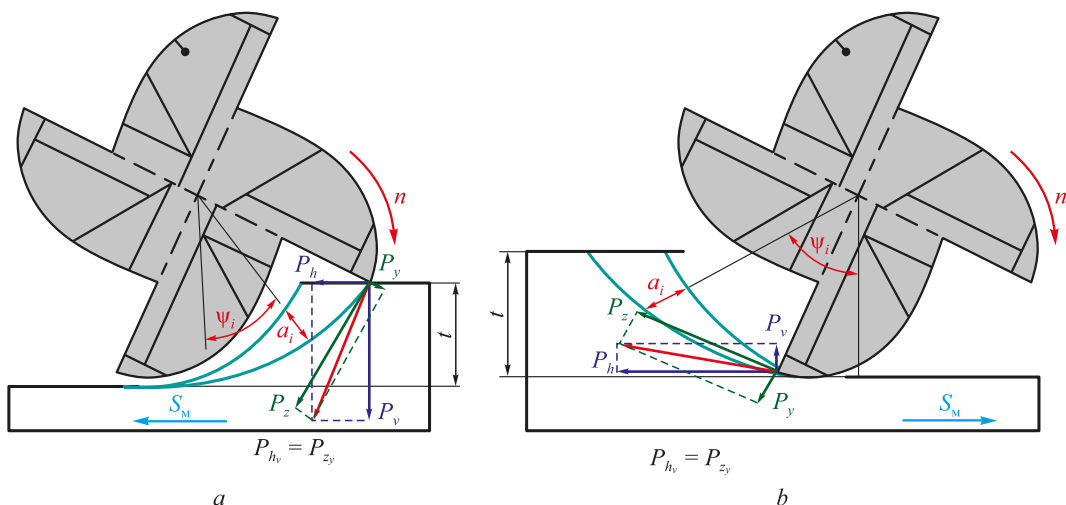


Fig. 7. Direction of forces P_h , P_v , P_z and P_y during climb milling (a) and conventional milling (b)

Рис. 7. Направление сил P_h , P_v , P_z и P_y при попутном фрезеровании (a) и при встречном фрезеровании (b)

Table 3. Results of analysis of the factors effects on signal-to-noise ratio

Таблица 3. Результаты анализа влияния факторов на отношение сигнал/шум

Factor	Mean signal-to-noise ratio level $S/N_{P_{avg}}$, dB			Factor–signal-to-noise ratio plot for P_h
	1	2	3	
A: milling scheme	–34.83	–41.43	–	
B: feed rate s_m , mm/min	–40.43	–39.57	–34.14	
C: rotational speed n , rpm	–35.17	–39.57	–39.65	
Factor	Mean signal-to-noise ratio level $S/N_{P_{avg}}$, dB			Factor–signal-to-noise ratio plot for P_v
	1	2	3	
A: milling scheme	–41.88	–37.76	–	
B: feed rate s_m , mm/min	–40.43	–42.64	–32.58	
C: rotational speed n , rpm	–35.08	–42.64	–41.73	
Factor	Mean signal-to-noise ratio level $S/N_{P_{avg}}$, dB			Factor–signal-to-noise ratio plot for P_x
	1	2	3	
A: milling scheme	–26.11	–37.76	–	
B: feed rate s_m , mm/min	–27.75	–25.40	–32.58	
C: rotational speed n , rpm	–22.76	–25.40	–25.77	

into the flank surface of the milling cutter, which leads to a decrease in the milling force.

As the feed rate s_m increases, the chip thickness a_i and material deformation increase, and the cutting temperature also rises. This results in an increase in the milling forces P_h and P_v , however, the rate of their growth decreases. Moreover, an increase in cutting temperature may lead to martensite decomposition, which further reduces the material strength and additionally slows the growth of P_h and P_v . When the rotational speed n decreases, the feed per tooth s_z , increases, and consequently the chip thickness a and material deformation increase; therefore, the milling forces P_h and P_v increase. According to the data in Table 2, the axial force P_x shows little sensitivity to the feed rate s_m and the rotational

speed n , because is only weakly affected by chip thickness. However, due to temperature variations, the axial force P_x may change slightly.

CONCLUSIONS

In the present study, the microstructure and mechanical properties of the sample were investigated in different directions. On the lateral surface of the sample, columnar grains of previous austenite were observed, with a hardness of approximately 505 HV_{0.1}, whereas the upper surface exhibited equiaxed grains with a higher hardness of 539.73 HV_{0.1}. Overall, the microstructure of the sample corresponds predominantly to annealed martensite, within which a pronounced transgranular phe-

nomenon can be observed. This microstructural state is associated with the multiple thermal cycles experienced during deposition.

The microstructure and mechanical properties were also examined in different regions of the sample. In the region close to the lateral surface, the higher cooling rate results in finer martensitic features and increased hardness, reaching $514.2 \pm 5.85 \text{ HV}_{0.1}$. In both the lower and upper regions of the sample, the previous austenitic grain boundaries are clearly visible. Because the lower part of the sample experienced a greater number of thermal cycles, martensite decomposition occurred, leading to a reduction in hardness to $480.49 \pm 8.19 \text{ HV}_{0.1}$. In contrast, the upper part of the sample was subjected to fewer thermal cycles and retained a substantial amount of martensite; as a result, the hardness remained relatively high at $512.80 \pm 5.25 \text{ HV}_{0.1}$.

The machinability of the sample was evaluated using the Taguchi method. Owing to the high hardness of the material, the impact of the cutting edge on the machined surface during climb milling leads to an increase in milling forces. Conversely, during conventional milling, the low plasticity of the material reduces the volume of material pressed into the flank surface of the milling cutter, resulting in lower milling forces. Furthermore, as the feed per tooth increases, the rise in cutting temperature reduces the material strength, which slows the growth of milling forces.

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D. A. Chinakhov – final editing of the article, data processing.

V. A. Klimenov – writing the text, analysis and data processing.

R. V. Chernukhin – systematization and processing of experimental data.

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GENERAL PRINCIPLES OF MODELING AND OPTIMIZATION OF ELECTRIC STEELMAKING PROCESS

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Abstract. Development of metallurgy provides for further increase and improvement of steel production volumes through the introduction of various advanced resource- and energy-saving technologies. The main and most universal control actions that affect the course and technical-economic indicators of the process are inextricably linked to the optimization of technology parameters which is focused on achieving the best results in the field of productivity, product quality and reduction of resource costs. This is achieved through the regular monitoring and analysis of key indicators, as well as making necessary adjustments to process management. A successful combination of these factors contributes to maximizing the production efficiency and increasing the competitiveness of products on the market. To calculate the process static parameters, it is advisable to use the resources of mathematical modeling and development of an instrumental system. When creating a static calculation model, the electric steelmaking process was considered as a complex thermodynamic system into which condensed and gaseous input media enter, and the final products are metal, slag and gas. Calculation of the static modes of the electric steelmaking process is carried out on the basis of calculations of material and thermal balances based on the laws of mass and energy conservation relative to the components of a heterogeneous system. The solution of the optimization problem based on formal methods involves selection of various criteria and setting a system of restrictions (requirements for metal composition; ranges of change in the cost of components of charge materials and system state parameters; compliance with the law of mass conservation at the level of fluxes, substances and elements; compliance with the law of energy conservation). A feature of the developed method of mathematical modeling and optimization of the electric steelmaking process is the systematic solution of a set of interrelated optimization problems to determine the optimal conditions for the processes in the metallurgical system and the optimal solutions for implementation of electric smelting technology.

Keywords: electric steelmaking, charge material, metal temperature, modeling, optimization

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ОБЩИЕ ПРИНЦИПЫ МОДЕЛИРОВАНИЯ И ОПТИМИЗАЦИИ ЭЛЕКТРОСТАЛЕПЛАВИЛЬНОГО ПРОЦЕССА

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Аннотация. Развитие металлургии предусматривает дальнейшее увеличение объемов и совершенствование процессов выплавки стали за счет внедрения различных прогрессивных ресурсо- и энергосберегающих технологий. Основные и наиболее универсальные управляющие воздействия, влияющие на ход и технико-экономические показатели процесса, неразрывно связаны с оптимизацией параметров технологии. Оптимизация параметров технологического процесса ориентирована на достижение наилучших результатов в области производительности, качества продукции и снижения затрат ресурсов. Это достигается посредством регулярного мониторинга и анализа ключевых показателей, а также внесения необходимых корректировок в управление процессом. Удачное сочетание указанных факторов способствует максимизации производственной эффективности и повышению конкурентоспособности продукции на рынке. Для расчета статических режимов процесса целесообразно использование ресурсов математического моделирования и разработки инструментальной системы. При создании статической модели расчета электросталеплавильного процесса рассматривали как сложную термодинамическую систему, в которую поступают конденсированные и газообразные входные среды, а конечными продуктами являются металл, шлак и газ. Расчет статических режимов электросталеплавильного процесса, осуществляемый на основе материального и теплового балансов, бази-

руется на законах сохранения массы и энергии относительно составляющих гетерогенной системы. Решение оптимизационной задачи на основе формальных методов предусматривает выбор различных критериев и задание системы ограничений (требования к составу металла, диапазоны изменения расходов компонентов шихтовых материалов и параметров состояния системы, соблюдение закона сохранения массы на уровне потоков, веществ и элементов, соблюдение закона сохранения энергии). Особенностью разработанного метода математического моделирования и оптимизации электросталеплавильного процесса является системное решение комплекса взаимосвязанных оптимизационных задач по определению оптимальных условий протекания процессов в металлургической системе и оптимальных режимов реализации технологии электроплавки.

Ключевые слова: электросталеплавильное производство, шихтовый материал, температура металла, моделирование, оптимизация

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INTRODUCTION

Federal programs for the technological development of metallurgy currently call for further growth in steel output and for continued improvement of steel-making processes through the adoption of advanced resource- and energy-saving technologies. A central objective of modern steelmaking is to produce a molten metallic semiproduct with a specified chemical composition and temperature while minimizing material and energy consumption [1 – 5]. The most universal control actions affecting both process evolution and its technical and economic indicators are typically inseparable from the optimization of technology parameters [6 – 8]. This, in turn, necessitates solving problems related to establishing optimal operating modes for electric arc furnaces and forecasting the key indicators of electric melting [5 – 8].

Addressing these tasks benefits from mathematical modeling and from developing an instrumental system for calculating the static modes of the process, i.e., for analyzing the process at the level of input–output flows [9 – 12].

Static-mode calculation methods enable, among other things, determination of the material and thermal balances¹ [13 – 15]; evaluation of limiting energy capabilities, identification of thermodynamic feasibility limits and interaction conditions for components within a metallurgical unit, and selection of the most effective components of the input flow to obtain metal with specified properties while achieving optimal technical and economic indicators [13 – 16].

METHODS

When developing the static calculation model, the electric steelmaking process was considered as a complex thermodynamic system into which condensed

and gaseous input media are supplied, while the final products are metal, slag, and gas [17 – 20]. Depending on the selected technological route, the condensed input flows include: metal charge (liquid or pig iron, steel scrap); solid oxidizers (sinter, iron ore, pellets); slag-forming materials (lime, limestone, fluorspar); carburizers (metallurgical coke, coke breeze, dust from dry coke quenching units, electrode scrap, etc.); and ferroalloys. The gaseous input flows are oxygen and natural gas [21 – 23].

Accordingly, the defining parameters of the input flows are: K – the total number of flows; K^f , K^g and G_k^f , G_k^g – the number and mass of condensed and gaseous flows, respectively (the subscripts f and g denote condensed and gaseous flows); N_k^f and N_k^g – the number of substances in the k -th flow; t_k^f , t_k^g and ρ_k^f , ρ_k^g – the temperature ($^{\circ}\text{C}$) and density (kg/m^3) of the flows; $|R_m|_k$, $\{R_m\}_k$ – the content of substance R_m in the k -th flow, %.

The output flows are characterized by the following parameters: G_m , G_{sl} , G_g – the mass of metal, slag, and gas; t_m , t_{sl} , t_g – the temperature of metal, slag, and gas, $^{\circ}\text{C}$; N_m , N_{sl} , N_g – the number of substances in the corresponding phases; $[R_n]$, (R_n) , $\{R_n\}$ – the content of substance R_n in the metallic, slag, and gas phases, respectively.

In defining the system parameters, it was assumed that substances R_m and R_n are elements of the set $\mathcal{R}$ and may be present in different phases in the form of various compounds $E_{ix}E_{jy}$, composed of elements E_i and E_j from the elemental set x . With the input flows, the following compounds may enter the reactor: condensed phase – Fe, C, Mn, Si, S, P, Al, Ni, Mo, W, V, Ti, Cr, B, Cu, FeO, Fe_2O_3 , Al_2O_3 , CaO, SiO_2 , MgO, MnO, P_2O_5 , CaS, CaF_2 , NiO, V_2O_5 , Cr_2O_3 , TiO_2 , MoO_2 , WO_2 , CuO, B_2O_3 ; gaseous phase – H_2O , CO_2 , O_2 , CH_4 , CO, N_2 , H_2 . The elements forming these substances are: Fe, C, O, Mn, Si, S, P, Al, Ca, Mg, H, F, Ni, Ti, V, Cr, N, B, Cu, Mo, W.

Calculation of the static modes of the electric steelmaking process involves determining the consumption rates of charge materials required to achieve the specified metal parameters and metal temperature. The calculations are based on balance models and thermodynamic equations. The calculation scheme is shown in Fig. 1.

¹ Simulation model of the electric steelmaking shop of the Chelyabinsk Metallurgical Plant. URL: <https://www.anylogic.ru/resources/case-studies/chelyabinsk-metallurgical-plant-uses-a-simulation-model-electric-furnace-melting-shop/>

Specification of the initial data includes entering the following information: the input flow parameters; the process parameters; and the element distribution coefficients among phases.

The adopted process parameters are: t_m , t_{sl} , t_g – phase temperatures, °C; Q_{loss} – heat losses to the environment; Q_{drop} – metal losses with metal droplets; η_{CO} – the degree of carbon oxidation to CO; η_{FeO} – the degree of iron oxidation to FeO; α – the oxygen utilization degree.

The element distribution coefficients among phases are specified on the basis of experimental data within the prescribed ranges of variation of the input and output flow parameters.

The following technical and economic indicators are used: E – process energy intensity, kW·h/t; C – production cost, RUB/t; PU – unit productivity, t/h; g_k – specific material consumption, kg/t.

Once all required parameters have been specified, an initial data table for the input flows is generated. The input information includes the temperature, density,

and flow rate of the k -th input flow. The procedure also requires the chemical composition of the charge materials to be specified. For all calculation variants, the software provides a baseline composition for the condensed and gaseous input flows, as well as for the volatile components of the charge; if necessary, this composition can be modified by the user.

RESULTS AND DISCUSSION

The static modes of the electric steelmaking process are calculated from the material and thermal balances, formulated on the basis of the laws of conservation of mass and energy for the components of a heterogeneous system. The calculation procedure for the material and thermal balances is outlined in Fig. 2.

At the level of input–output flows, the overall material balance for the system can be written as:

$$\sum_{k=1}^K G_k = \sum_{l=1}^L G_l.$$

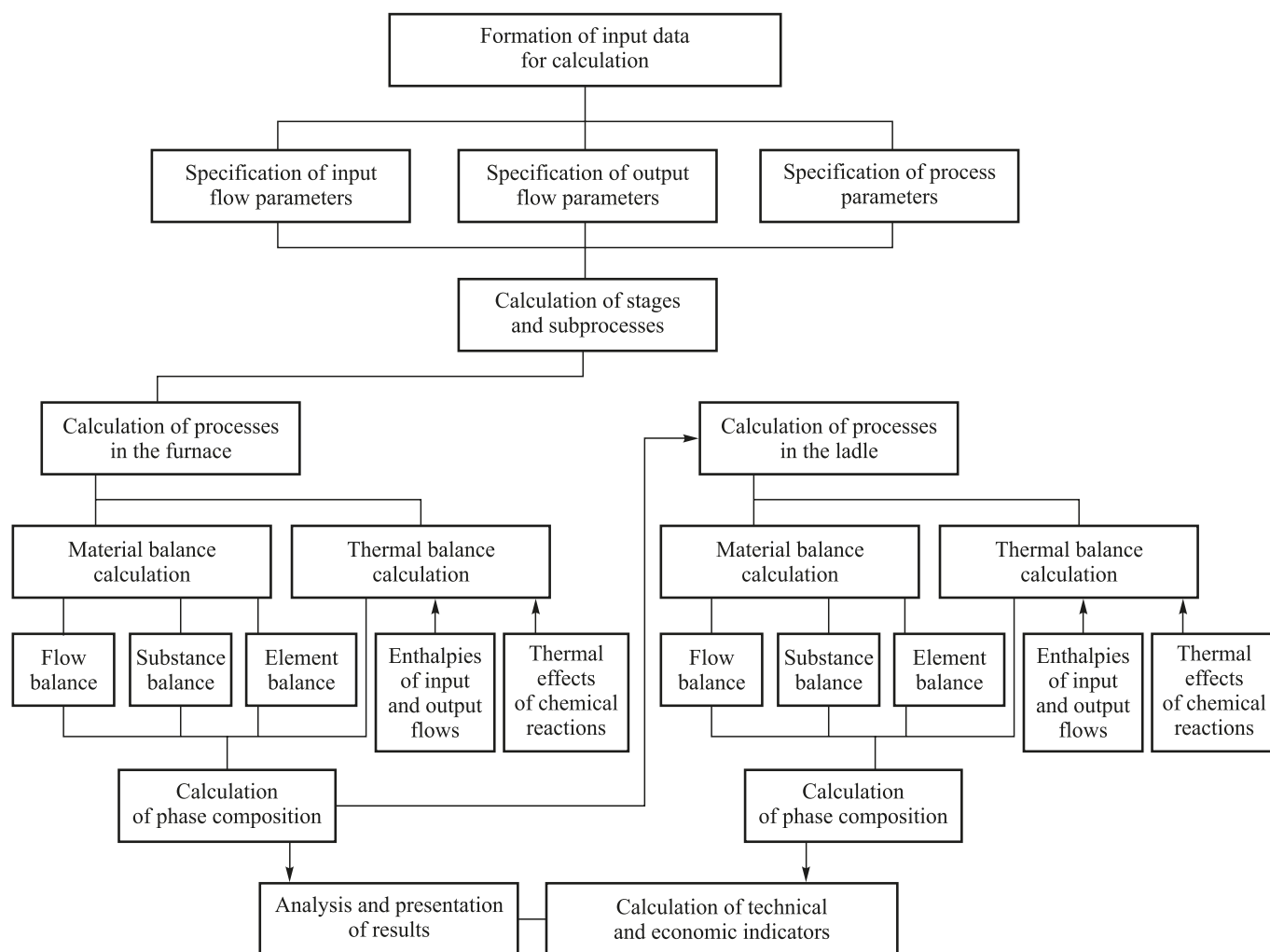


Fig. 1. Calculation scheme of electric steelmaking process

Рис. 1. Схема расчета электросталеплавления процесса

When phase separation is taken into account, the material-balance relationship can be expressed in the following form:

$$\sum_{k=1}^{K^f} G_k^f + \sum_{k=1}^{K^g} G_k^g = G_m + G_{sl} + G_g.$$

The total mass of substances entering the furnace with the input flows is:

$$\sum_{k=1}^K G_k = \sum_{k=1}^{K^f} \sum_{m=1}^{N_k^f} \frac{G_k^f |R_m|_k}{100} + \sum_{k=1}^{K^g} \sum_{m=1}^{N_k^g} \frac{G_k^g \{R_m\}_k}{100}.$$

The mass of substances leaving the system with the output flows is defined as the sum of the masses contained in the metallic, slag, and gas phases:

$$\sum_{l=1}^L G_l = \sum_{n=1}^{N_m} \frac{G_m [R_n]}{100} + \sum_{n=1}^{N_{sl}} \frac{G_{sl} (R_n)}{100} + \sum_{n=1}^{N_g} \frac{G_g \{R_n\}}{100}.$$

Accordingly, the material-balance equation at the level of substance flows takes the form:

$$\begin{aligned} & \sum_{k=1}^{K^f} \sum_{m=1}^{N_k^f} \frac{G_k^f |R_m|_k}{100} + \sum_{k=1}^{K^g} \sum_{m=1}^{N_k^g} \frac{G_k^g \{R_m\}_k}{100} = \\ & = \sum_{n=1}^{N_m} \frac{G_m [R_n]}{100} + \sum_{n=1}^{N_{sl}} \frac{G_{sl} (R_n)}{100} + \sum_{n=1}^{N_g} \frac{G_g \{R_n\}}{100}. \end{aligned}$$

To determine the phase composition of the output flows, element balance equations are written for each element E_i , since a given element may be present in different phases as a range of compounds:

$$\begin{aligned} & \sum_{k=1}^{K^f} \sum_{m=1}^{N_k^{E_i^f}} \frac{G_k^f |E_{ix_m} E_{jy_m}|_k}{100} \frac{x_m M_{E_i}}{M_{E_{ix_m} E_{jy_m}}} + \\ & + \sum_{k=1}^{K^g} \sum_{m=1}^{N_k^{E_i^g}} \frac{G_k^g \{E_{ix_m} E_{jy_m}\}_k}{100} \frac{x_m M_{E_i}}{M_{E_{ix_m} E_{jy_m}}} = \\ & = \frac{G_m [E_i]}{100} + \sum_{n=1}^{N_{sl}^{E_i}} \frac{G_{sl} (E_{ix_n} E_{jy_n})}{100} \frac{x_n M_{E_i}}{M_{E_{ix_n} E_{jy_n}}} + \\ & + \sum_{n=1}^{N_g^{E_i}} \frac{G_g \{E_{ix_n} E_{jy_n}\}}{100} \frac{x_n M_{E_i}}{M_{E_{ix_n} E_{jy_n}}}, \end{aligned}$$

where $|E_{ix_m} E_{jy_m}|_k$, $\{E_{ix_m} E_{jy_m}\}_k$ – the concentrations of the m -th substance containing element E_i in k -th condensed and gaseous input flows, respectively, %; $[E_i]$ – the concentration of the substance consisting of element E_i , in the metal, %; $(E_{ix_n} E_{jy_n})$, $\{E_{ix_n} E_{jy_n}\}$ – the concentrations of the n -th substance containing element E_i in the slag and gas phases of the output flow, respectively, %; $N_k^{E_i^f}$, $N_{sl}^{E_i}$, $N_g^{E_i}$ – the numbers of substances containing element E_i in

the k -th input flow, slag, and gas; x_m , y_m – the stoichiometric coefficients of the m -th compound of element E_i ; M_{E_i} , $M_{E_{ix_m} E_{jy_m}}$ – the molar masses of element E_i and its compounds, kg/mol.

The total mass of the condensed substance R entering the system with the input flows is given by:

$$G_{|R|}^{\text{in}} = \sum_{k=1}^{K^f} \sum_{m=1}^{N_k^f} \frac{G_k^f |R_m|_k}{100}.$$

The mass of the gaseous substance R entering the system with the input flows is also determined taking into account its content in the gaseous input flows:

$$G_{\{R\}}^{\text{in}} = \sum_{k=1}^{K^g} \sum_{m=1}^{N_k^g} \frac{G_k^g \{R_m\}_k}{100}.$$

The phase composition of the output flow is determined as follows. In accordance with the element distribution coefficients among phases, the amount of element E_i in the metal is determined by the following expression:

$$G_{[E_i]} = G_{E_{ii}}^{\text{in}} L_{[E_i]},$$

here $L_{[E_i]}$ – the distribution coefficient of element E_i in the metal.

The amount of the oxide of element E_i in the slag is determined accordingly as

$$G_{(E_{ix_n} O_{jy_n})} = G_{E_i}^{\text{in}} L_{(E_i)} \frac{M_{E_{ix_n} O_{jy_n}}}{x_n M_{E_i}}.$$

Next, the amount of oxygen remaining in the system after oxidation of all components and consumed for iron oxidation is determined as follows:

$$\begin{aligned} G_O &= \sum_{k=1}^{K^f} \sum_{m=1}^{N_k^{E_i^f}} \frac{G_k^f |E_{ix_m} O_{y_m}|_k}{100} \frac{0.016 y_m}{M_{E_{ix_m} O_{y_m}}} - \\ & - \sum_{n=1}^{N_{sl}^{E_i}} \frac{G_{(E_{ix_n} O_{jy_n})}}{100} \frac{0.016 y_n}{M_{E_{ix_n} O_{jy_n}}}. \end{aligned}$$

The amount of iron oxides in the slag is then determined as:

$$G_{(\text{FeO})} = G_O \eta_{\text{FeO}};$$

$$G_{(\text{Fe}_2\text{O}_3)} = G_O (1 - \eta_{\text{FeO}}).$$

Accordingly, the amount of iron in the metal is calculated as follows:

$$G_{[\text{Fe}]} = G_{\text{Fe}}^{\text{in}} - G_{(\text{FeO})} \frac{0.056}{0.072} - G_{\text{Fe}_2\text{O}_3}^{\text{in}} \frac{0.112}{0.160}.$$

Consequently, the masses of metal and slag are determined as:

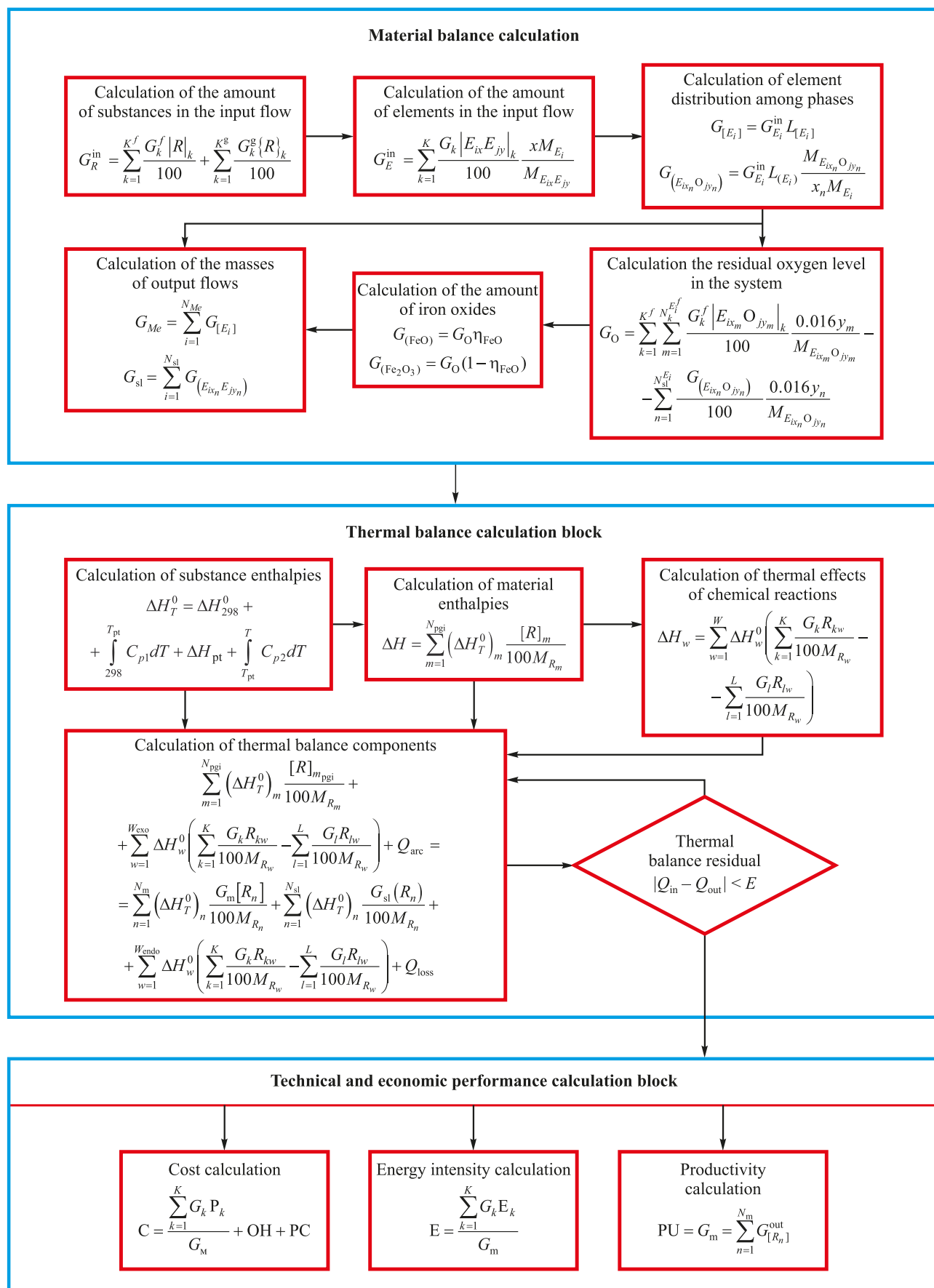


Fig. 2. Algorithm for calculating material and thermal balances

Рис. 2. Алгоритм расчета материального и теплового балансов

$$G_{Me} = \sum_{i=1}^{N_{Me}} G_{[E_i]};$$

$$G_{sl} = \sum_{i=1}^{N_{sl}} G_{(E_{i,sl} E_{i,ym})}.$$

The calculation of the phase composition and masses in the ladle is performed in a similar manner; however, in this case the input flows consist of the metal produced in the furnace and slag that partially enters the ladle during tapping.

To calculate the composition of the gas-phase components, the following mechanism of off-gas formation in the electric furnace is proposed. For technologies involving the use of natural gas, complete combustion is assumed, which results in the chemical composition of the gases in the output flow corresponding to that obtained under classical decomposition conditions.

During complete combustion of natural gas, CO_2 and H_2O are formed:

$$G_{\text{CO}_2}^{\text{comb}} = G_{\text{CH}_4}^{\text{in}} (1 - \eta_{\text{conv}}) \frac{0.044}{0.016};$$

$$G_{\text{H}_2\text{O}}^{\text{comb}} = G_{\text{CH}_4}^{\text{in}} (1 - \eta_{\text{conv}}) \frac{0.018}{0.016}.$$

The amount of CO formed during combustion of the carbon contained in the charge can be determined as follows:

$$G_{\text{CO}} = \eta_{\text{CO}} \left(\sum_{k=1}^{K^f} \frac{G_k^f [C]_k}{100} - \frac{G_m [C]}{100} \right) \frac{0.028}{0.012}.$$

Taking into account the combustion of natural gas, the amount of CO_2 in the off-gases is given by:

$$G_{\text{CO}_2} = (1 - \eta_{\text{CO}}) \times \left(\sum_{k=1}^{K^f} \frac{G_k^f [C]_k}{100} - \frac{G_m [C]}{100} \right) \frac{0.044}{0.012} + G_{\text{CO}_2}^{\text{comb}}.$$

The total amount of hydrogen in the off-gases is:

$$G_{\text{H}_2} = G_{\text{H}_2}^{\text{in}}.$$

The amount of water vapor is determined as

$$G_{\text{H}_2\text{O}} = G_{\text{H}_2\text{O}}^{\text{in}} + G_{\text{H}_2\text{O}}^{\text{comb}}.$$

The amount of nitrogen in the off-gases is equal to the amount of nitrogen entering the system with the input flows per unit time:

$$G_{\text{N}_2} = G_{\text{N}_2}^{\text{in}}.$$

The amount of SO_2 in the off-gases is determined in accordance with the sulfur transfer coefficient to the gas phase:

$$G_{\text{SO}_2} = G_{\text{S}}^{\text{in}} \alpha_{\text{S}} \frac{0.064}{0.032}.$$

The total mass of gas is determined as follows:

$$G_{\text{r}} = G_{\text{CO}}^{\text{out}} + G_{\text{CO}_2}^{\text{out}} + G_{\text{H}_2}^{\text{out}} + G_{\text{H}_2\text{O}}^{\text{out}} + G_{\text{SO}_2}^{\text{out}} + G_{\text{N}_2}^{\text{out}}.$$

When deriving the thermal balance equations, it is assumed that the principal processes governing the thermal state of the system are: the sensible heat of the charge materials; heat exchange with the surrounding environment; heat released by electric arc combustion; and chemical reactions with their associated thermal effects. The thermal balance equations are formulated on the basis of the law of conservation of energy, expressed as $Q_{\text{in}} = Q_{\text{out}}$.

Taking into account the enthalpy of the input and output flows, the total heat of chemical reactions occurring in the system, the heat supplied by the electric arc, as well as heat exchange with the environment, the thermal-balance equation can be written in the following form:

$$\sum_{k=1}^{K^f} G_k^f \Delta H_k^f + \sum_{k=1}^{K^g} G_k^g \Delta H_k^g + Q_{\text{arc}} = \sum_{l=1}^L G_l \Delta H_l + \Delta H_{\text{cr}} + Q_{\text{loss}},$$

where ΔH_k^f , ΔH_k^g , ΔH_l – the enthalpies per unit mass of the k -th condensed or gaseous input flow and the l -th output flow, respectively, referred to standard conditions and a temperature of 298 K; Q_{arc} – the heat input to the system from electric arc combustion; Q_{loss} – heat losses to the environment; ΔH_{cr} – the total thermal effect of the chemical reactions.

Only liquid pig iron has a nonzero enthalpy; therefore, the enthalpy per unit mass of liquid pig iron relative to $T = 298$ K is given by:

$$\Delta H_{\text{pgi}} = \sum_{m=1}^{N_{\text{pgi}}} (\Delta H_T^0)_m \frac{[R]_{m_{\text{pgi}}}}{100 M_{R_m}},$$

where $(\Delta H_T^0)_m$ – the enthalpy change of the m -th substance of liquid pig iron upon heating from 298 K to T_k , kJ/mol; $[R]_{m_{\text{pgi}}}$ – the concentration of substance R_m , %; M_{R_m} – is the molar mass of substance R_m , kg/mol.

The enthalpies per unit mass of metal, slag, and gas are determined, respectively, as follows:

$$\Delta H_m = \sum_{n=1}^{N_m} (\Delta H_T^0)_n \frac{[R_n]}{100 M_{R_n}};$$

$$\Delta H_{\text{sl}} = \sum_{n=1}^{N_{\text{sl}}} (\Delta H_T^0)_n \frac{(R_n)}{100 M_{R_n}};$$

$$\Delta H_g = \sum_{n=1}^{N_g} (\Delta H_T^0)_n \frac{\{R_n\}}{100 M_{R_n}}.$$

The thermal effect of chemical reactions is defined as the sum of the thermal effects of independent reactions describing the transition of the system from the initial to the final state:

$$\Delta H_{cr} = \sum_{w=1}^W \Delta H_w,$$

where W – the number of independent reactions; $\Delta H_w = \Delta H_w^0 \Delta n_{R_w}$ – the thermal effect of the w -th chemical reaction, kJ/mol; ΔH_w^0 – the standard thermal effect of the w -th chemical reaction; Δn_{R_w} – the number of moles of substance R that reacted in the w -th reaction.

The number of moles of the reacting substance R_w is determined as the difference between its amounts in the input and output flows:

$$\Delta n_{R_w} = \sum_{k=1}^K \frac{G_k R_{kw}}{100 M_{R_w}} - \sum_{l=1}^L \frac{G_l R_{lw}}{100 M_{R_w}}.$$

The heat released by arc combustion is determined taking into account the electric energy consumption:

$$Q_{arc} = 3600E,$$

where E – the electric energy consumption, kW·h.

Taking the above into account, the thermal-balance equation can be written in the following form:

$$\begin{aligned} & \sum_{m=1}^{N_{pgi}} \left(\Delta H_T^0 \right)_m \frac{[R]_{m_{pgi}}}{100 M_{R_m}} + \\ & + \sum_{w=1}^{W_{exo}} \Delta H_w^0 \left(\sum_{k=1}^K \frac{G_k R_{kw}}{100 M_{R_w}} - \sum_{l=1}^L \frac{G_l R_{lw}}{100 M_{R_w}} \right) + Q_{arc} = \\ & = \sum_{n=1}^{N_m} \left(\Delta H_T^0 \right)_n \frac{G_m [R_n]}{100 M_{R_n}} + \sum_{n=1}^{N_{sl}} \left(\Delta H_T^0 \right)_n \frac{G_{sl} (R_n)}{100 M_{R_n}} + \\ & + \sum_{w=1}^{W_{endo}} \Delta H_w^0 \left(\sum_{k=1}^K \frac{G_k R_{kw}}{100 M_{R_w}} - \sum_{l=1}^L \frac{G_l R_{lw}}{100 M_{R_w}} \right) + Q_{loss}. \end{aligned}$$

The principal heat inputs in the thermal balance include the heat of liquid pig iron, the heat of exothermic oxidation reactions of charge components with gaseous oxygen, the heat of slag-forming reactions, and the heat released by arc combustion. The heat expenditures include heat required for heating the metal, slag, and gas; heat consumed for decomposition of carbonates and iron oxides; heat required for moisture evaporation; and heat losses to the environment.

The enthalpies of the input and output flows were calculated using the following reference data: coefficients of the approximation equation for the heat capacity of a substance ($c_0, c_{-1}, c_1, c_2, c_3$), values of the enthalpy of formation and entropy at standard temperature ($\Delta_f H_{298}^0, S_{298}^0$), and data on phase transitions ($T_{pt}, \Delta H_{pt}^0$) for individual substances.

The solution of the optimization problem using formal methods involves selecting appropriate criteria and defining a system of constraints, including requirements for metal composition, allowable ranges of variation in the consumption rates of charge-material components and system state parameters, compliance with the law of conservation of mass at the levels of flows, substances, and elements, and compliance with the law of conservation of energy.

The following indicators may be used as optimization criteria:

– total consumption (TC) of charge materials per unit of product, kg/t,

$$TC = \sum_{k=1}^K g_k = \frac{\sum_{k=1}^K G_k}{G_m} \cdot 1000 \rightarrow \min;$$

– production cost (C), RUB/t,

$$C = \frac{\sum_{k=1}^K G_k P_k}{G_m} + OH + PC \rightarrow \min;$$

– energy consumption (E) per unit of product, GJ/t,

$$E = \frac{\sum_{k=1}^K G_k E_k}{G_m} \rightarrow \min;$$

– unit productivity (PU)

$$PU = G_m = \sum_{n=1}^{N_m} G_{[R_n]}^{out} \rightarrow \max,$$

where P_k – the price of the k -th input material, RUB/t; OH – general plant overhead costs, RUB/t; PC – processing costs, RUB/t; E_k – is the specific energy intensity of the k -th material, GJ/t.

The formulation of the optimization problem consists in finding an extremum of one of the selected criteria subject to the following constraints:

– on the ranges of variation in charge-material consumption rates:

$$G_{k \min} < G_k < G_{k \max};$$

– on metal and slag parameters:

$$[R_n]_{\min} < [R_n] < [R_n]_{\max} \quad (n = 1 \div N_m);$$

$$(R_n)_{\min} < (R_n) < (R_n)_{\max} \quad (n = 1 \div N_{sl});$$

$$T_{l \min} < T_l < T_{l \max} \quad (l = 1 \div L);$$

– compliance with the law of conservation of mass at the levels of flows, substances, and elements:

$$\begin{aligned} \sum_{k=1}^{K^f} G_k^f + \sum_{k=1}^{K^g} G_k^g &= G_m + G_{sl} + G_g; \\ \sum_{k=1}^{K^f} \sum_{m=1}^{N_m^f} \frac{G_k^f |R_m|_k}{100} + \sum_{k=1}^{K^g} \sum_{m=1}^{N_m^g} \frac{G_k^g \{R_m\}_k}{100} &= \\ = \sum_{n=1}^{N_m} \frac{G_m[R_n]}{100} + \sum_{n=1}^{N_{sl}} \frac{G_{sl}(R_n)}{100} + \sum_{n=1}^{N_g} \frac{G_g\{R_n\}}{100}; \\ \sum_{k=1}^{K^f} \sum_{m=1}^{N_m^f} \frac{G_k^f |E_{ixm} E_{jym}|_k}{100} \frac{x_m M_{E_i}}{M_{E_{ixm} E_{jym}}} + \\ + \sum_{k=1}^{K^g} \sum_{m=1}^{N_m^g} \frac{G_k^g \{E_{ixm} E_{jym}\}_k}{100} \frac{x_m M_{E_i}}{M_{E_{ixm} E_{jym}}} &= \\ = \frac{G_m[E_i]}{100} + \sum_{n=1}^{N_{sl}} \frac{G_{sl}(E_{ixn} E_{jyn})}{100} \frac{x_n M_{E_i}}{M_{E_{ixn} E_{jyn}}} + \\ + \sum_{n=1}^{N_g} \frac{G_g\{E_{ixn} E_{jyn}\}}{100} \frac{x_n M_{E_i}}{M_{E_{ixn} E_{jyn}}}; \end{aligned}$$

– compliance with the law of conservation of energy:

$$\begin{aligned} \sum_{m=1}^{N_{pgi}} (\Delta H_T^0)_m \frac{[R]_{m_{pgi}}}{100 M_{R_m}} + \\ + \sum_{w=1}^{W_{exo}} \Delta H_w^0 \left(\sum_{k=1}^K \frac{G_k R_{kw}}{100 M_{R_w}} - \sum_{l=1}^L \frac{G_l R_{lw}}{100 M_{R_w}} \right) + Q_{arc} = \\ = \sum_{n=1}^{N_m} (\Delta H_T^0)_n \frac{G_m[R_n]}{100 M_{R_n}} + \sum_{n=1}^{N_{sl}} (\Delta H_T^0)_n \frac{G_{sl}(R_n)}{100 M_{R_n}} + \\ + \sum_{w=1}^{W_{endo}} \Delta H_w^0 \left(\sum_{k=1}^K \frac{G_k R_{kw}}{100 M_{R_w}} - \sum_{l=1}^L \frac{G_l R_{lw}}{100 M_{R_w}} \right) + Q_{loss}; \end{aligned}$$

– fulfillment of at least one of the following target conditions:

$$\begin{aligned} TC_{\min} < TC < TC_{\max}; \\ C_{\min} < C < C_{\max}; \\ E_{\min} < E < E_{\max}; \\ PU_{\min} < PU < PU_{\max}. \end{aligned}$$

Thus, a method for mathematical modeling and optimization of the electric melting process has been developed, providing a systematic solution to a set of interrelated optimization problems.

Выводы

The electric steelmaking process was analyzed, and a method for its mathematical modeling and optimization was developed. The proposed approach is distinguished

by a systematic treatment of a set of interrelated optimization problems, enabling the determination of acceptable process conditions within the metallurgical system using modeling and optimization methods. In addition, technological operating modes for implementing the electric melting technology are proposed.

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E. M. Zapol'skaya – processing of the results, design of the article.

И. А. Рыбенко – проведение термодинамических расчетов, обработка результатов.

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INDEX OF ARTICLES

"IZVESTIYA. FERROUS METALLURGY" FOR 2025, VOL. 68

HISTORY OF METALLURGY

- Tresvyatskii L.A., Protopopov E.V., Umanskii A.A., Khudoleev A.N. Kuznetsk armor: The first frontiers of victory 2

METALLURGICAL TECHNOLOGIES

- Baidin V.V., Umanskii A.A. Modeling and development of technological modes for production of grinding balls of increased hardness and impact resistance 3
- Bessonova O.V., Polevoi E.V., Oskolkova T.N., Komarova T.A. Comparative analysis of the structure and properties of hypereutectoid railway rails 6
- Bogdanova N.A., Zhilin S.G., Predein V.V. Strength characteristics of investment patterns obtained by compaction of waxy material powders in the field of centrifugal forces 6
- Goloviznin S.M., Bryukhanov I.Yu. Energy consumption analysis of multistage steel wire drawing 3
- Gromov V.E., Kononov S.V., Polevoi E.V. Anthology of rails produced by JSC EVRAZ United West Siberian Metallurgical Plant in the 21st century 1
- Karfidov A.O., Chichenev N.A., Vasil'ev M.V., Chicheneva O.N. Production of arc sheet elements by stepped bending method 3
- Kharchenko A.S., Sysoev V.I., Sibagatullin S.K., Dzyuba A.V., Savinov A.S., Kharchenko E.O. Modeling the distribution of components emitted from oiled scale between water, gas, and dust media in blast furnace dedusting plant 4
- Mozzhegorov M.N., Mashentseva M.S. Development of profile pipes production technology, providing higher accuracy of geometric parameters compared to foreign manufactures 1
- Nevskii S.A., Bashchenko L.P., Sarychev V.D., Granovskii A.Yu., Shamsutdinova D.V. Influence of inclined electric field on decay of a liquid jet during heat treatment and surfacing 1
- Pospelov I.D. Stiffness modulus of stands in finishing group of continuous wide-strip hot rolling mill 1
- Shakirov M.K., Protopopov E.V., Turchaninov E.B., Beletnitskii E.A., Fat'yanov S.S. Complex energy-dynamic model of BOF melting 3
- Shkirmontov A.P. Modeling the operation of ferroalloy furnace with increased power and electrode diameters 4
- Tepliyakova L.A., Kunitsyna T.S., Pechkovskii V.A., Kashin A.D. Tensile fracture of tempered martensitic steel 1
- Yunusov A.M., Polevoi E.V., Yunin G.N., Oskolkova T.N. Experience in producing rails from bainitic steels 30KhG2S2AFM and 30KhG2SAFN 2

ECOLOGY AND RATIONAL USE OF NATURAL RESOURCES

- Ibraev I.K., Ibraeva O.T. Environmental features of oxidative refining of phosphorous hot metal in heavy-duty converters 3

- Shlyarov V.V., Shlyarova Yu.A., Bashchenko L.P., Zagulyaev D.V. Waste reduction and implementation of environmentally safe and efficient production processes using high-silicon alloys of the Al – Si system 3
- Ulyeva G.A., Tuyskhan K., Matsugina E.M., Volokitina I.E., Akhmetova G.E. Actual progress of production waste disposal problems 2
- Vasina M.V., Bashchenko L.P. Promising designs of gas analyzers for metallurgy 4

MATERIAL SCIENCE

- Aleutdinova M.I., Fadin V.V. Contact characteristics of C235 steel in dry sliding against C45 steel under high-density alternating current at different transformation coefficients of supply transformer 6
- Aleutdinova M.I., Fadin V.V. Features of steel contact surface wear in sliding under low contact pressure 3
- Belomytsev M.Yu., Mikhailov M.A., Kozlov D.A., Mikhailov A.M., Karavatskii I.I. Influence of heat treatment modes on the properties of 56DGNKh (Cu20Ni20Mn2Cr) alloy 1
- Goikhenberg Yu.N., Kulakov B.A., Polukhin D.S. Non-metallic inclusions in micropores of massive steel castings 3
- Gromov V.E., Minenko S.S., Chapaikin A.S., Semin A.P., Shlyarova Yu.A. Prospects for creation of high-speed high-entropy steels 4
- Ishmametov D.A., Pomel'nikova A.S., Petelin A.L. On the influence of rare-earth oxide additives on kinetics of borated layer formation and boron diffusion along grain boundaries during steel boriding 2
- Klevtsova N.A., Valiev R.Z., Klevtsov G.V., Fesenyuk M.V., Pigaleva I.N., Balashov V.M. Martensitic transformations in metastable austenitic steel with coarse-grained and ultrafine-grained structure during torsion 3
- Klimov S.A., Noskov F.M., Tokmin A.M., Masanskii O.A. Production of carbide steels based on high-speed steel by induction surfacing 6
- Koltygin A.V., Bazhenov V.E., Belova A.A., Sannikov A.V., Lyskovich A.A., Belov V.D., Shchedrin E.Yu. Influence of heat treatment on structure, phase composition, hardness and electrical conductivity of VZHL14N-VI nickel superalloy 1
- Orlova D.V., Shlyakhova G.V., Nadezhkin M.V. Peculiarities of deformation localization in additive material with structural-phase heterogeneity 3
- Permyakova I.E., Ivanov A.A., Lukina I.N., Kostina M.V., Dyuzheva-Maltseva E.V. Surface engineering of Fe_{53.3}Ni_{26.5}B_{20.2} amorphous alloy by excimer laser 6
- Popova M.V., Polunin A.M. Structure and properties of special-purpose alloys after annealing 6
- Popova N.A., Gromov V.E., Yur'ev A.B., Nikonenko E.L., Porfir'ev M.A. Hardening of surface layers of long rail heads during long-term operation 6
- Vlasov I.V., Gordienko A.I., Semenchuk V.M. Effect of thermal cycles on formation of pearlitic heat-resistant steel structure under wire arc additive manufacturing 1

Yaresko S.I., Balakirov S.N., Oskolkova T.N. Limiting energy characteristics during laser pulse treatment of tungsten-cobalt hard alloys	4
Zorya I.V., Poletaev G.M., Bebikhov Yu.V., Semenov A.S. Initiation of melting at tilt grain boundaries in austenite depending on the misorientation angle	2

INNOVATIONS IN METALLURGICAL INDUSTRIAL AND LABORATORY EQUIPMENT, TECHNOLOGIES AND MATERIALS

Evstigneev A.I., Chernyshova D.V., Odínokov V.I., Dmitriev E.A., Evstigneeva A.A., Koloshenko Yu.B., Potyannikhin D.A. Modeling and optimization of the effect of temperature seams on stress-strain state of spherical metal casting molds	4
Evstigneev A.I., Chernyshova D.V., Odínokov V.I., Dmitriev E.A., Evstigneeva A.A. Optimization modeling of crack resistance of ceramic shell mold during cooling of steel casting in it	3
Evstigneev A.I., Odínokov V.I., Chernyshova D.V., Evstigneeva A.A., Dmitriev E.A. Effect of annular seams on stress-strain state in cylindrical ceramic shell mold during solidification of a steel casting in it	6
Kulikov B.P., Storozhev Yu.I., Potapenko A.S. Metallization of ore-coal briquettes in an annular furnace heated by generator gas	4
Lekhov O.S., Bilalov D.Kh. Calculation of temperature and thermoelastic stresses in strikers during production of hollow steel billets in a unit of combined casting and deformation. Part 1	4
Lekhov O.S., Bilalov D.Kh. Calculation of temperature and thermoelastic stresses in strikers during production of hollow steel billets in a unit of combined casting and deformation. Part 2	6
Sivkov A.S., Chibizova S.I., Apasova A.D. Improving the design and thermal operation of a rotary kiln for production of metallurgical lime from chalk	4
Zhang C., Kozlov V.N., Chinakhov D.A., Klimenov V.A., Chernukhin R.V. Analysis of processing 40Kh13 martensitic stainless steel billet obtained by wire electron beam additive manufacturing	6

PHYSICO-CHEMICAL BASICS OF METALLURGICAL PROCESSES

Bol'shov L.A., Korneichuk S.K., Bol'shova E.L. Thermodynamic second-order interaction coefficients of nitrogen with nickel and chromium in liquid steel	2
Dmitriev A.N., Burova Yu.E., Vit'kina G.Yu., Barbin N.M., Terent'ev D.I. Thermodynamic modelling of reduction of iron ore materials by hydrogen-containing gases	2
Kazakevich G.A., Popov A.Yu. Construction of liquidus surface of Fe – B – Mn – C – Cr five-component diagram	1
Sheshukov O.Yu., Mikheenkova M.A., Egiazar'yan D.K., Mikheenkova A.M., Kleonovskii M.V., Matyukhin O.V. Peculiarities of the phase composition formation of steel-making slags and evaluation of the possibility of obtaining mineral binders with low CO ₂ generation on their basis	2
Sheshukov O.Yu., Nevidimov V.N., Nekrasov I.V., Metelkin A.A., Tsepelev V.S. Development of research on the physico-chemical properties of oxide and metal melts	1

Smirnov K.I., Gamov P.A., Roshchin V.E., Samolin V.S. Thermodynamic analysis of conditions for iron and titanium separation in ilmenite concentrate by selective reduction of elements	2
Vusikhis A.S., Mikheenkova M.A., Leont'ev L.I., Agafonov S.N. Effect of boron oxide additives on viscosity and melting point of the CaO – SiO ₂ – Al ₂ O ₃ – MgO system	3

INFORMATION TECHNOLOGIES AND AUTOMATIC CONTROL IN FERROUS METALLURGY

Dmitriev A.N., Vit'kin D.A., Zolotykh M.O., Vit'kina G.Yu. Mathematical model of blast furnace hearth condition based on data from thermocouples in refrigerator belts	3
Dmitriev A.N., Vit'kin D.A., Zolotykh M.O., Vit'kina G.Yu. Mathematical model of the blast furnace main trough lining condition	4
Rybenko I.A., Protodopov E.V., Zapol'skaya E.M. General principles of modeling and optimization of electric steel-making process	6
Shalaevskii D.L. Methodology for calculating the flatness of cold-rolled steel strips	4
Sokolovskaya E.A., Bosov E.V., Kudrya A.V., Kodirov D.F., Alekseev V.I. Possible consequences caused by reducing the number of samples during acceptance tests of a unit of metal products to assess its quality	3
Spirin N.A., Gurin I.A., Lavrov V.V. Information modeling system for assessing instability of blast furnace functioning	4
Zimin A.V., Kornet M.E., Burkova I.V., Zimin V.V. Using digital simulation models to optimize the operation of complex production facilities	2

ECONOMIC EFFICIENCY OF METALLURGICAL PRODUCTION

Chernikova O.P., Afanas'eva O.V., Afanas'ev E.G. Directions of decarbonization of Russian ferrous metallurgy	1
Fastykovskii A.R., Musatova A.I., Martyushev N.V. Development of models for functioning of drawing equipment for multi-mill servicing	1

IN THE ORDER OF DISCUSSION

Murashov V.A., Stroganov K.V., Bastynets A.K., Lvov D.D. Development of a continuous extra-furnace steel processing unit	1
Roshchin V.E., Roshchin A.V. Electrochemistry of reduction processes and prospects for the development of reduction technologies	4

BASED ON THE MATERIALS OF THE 16TH INTERNATIONAL SCIENTIFIC CONFERENCE “PHYSICO-CHEMICAL FOUNDATIONS OF METALLURGICAL PROCESSES” NAMED AFTER ACADEMICIAN A.M. SAMARIN

Agafonov S.N., Udоеva L.Yu., Vusikhis A.S., Leont'ev L.I. Assessment of the technological possibility of joint processing of ilmenite and perovskite concentrates	5
---	---

Aleksandrov A.A., Anuchkin S.N., Kanevskiy A.G., Remyantseva S.B., Grigorovich K.V. Interaction of Al_2O_3 -based ceramics on with slag melt 45 % CaO – 40 % Al_2O_3 – 10 % SiO_2 – 5 % MgO	5	Sabirov E.R., Pokolenko A.Yu., Bizhanov A.M., Bersenev I.S., Spirin N.A. Carburization of pellets to a carbon content of more than 4.5 % during metallization in shaft furnaces	5
Babenko A.A., Shartdinov R.R., Lobanov D.A., Smetanikov A.N., Upolovnikova A.G., Gulyakov V.S. Features of application of boron-containing slags in stainless steel smelting	5	Safarova D.E., Bazaleeva K.O., Ponkratova Yu.Yu., Alekseev A.V. Dependence of structure and properties of VT23 alloy on laser deposition parameters	5
Drozin A.D., Dudorov M.V. Criteria for the existence of local equilibria in supercooled melts	6	Shvetsov D.P., Chikova O.A., Tsepelev V.S., Sinitsin N.I., V'yukhin V.V. Physico-chemical analysis of 12Kh18N9TL steel melts for quality control of cast products	5
Grigorovich K.V. Results of the 16 th International Scientific Conference “Physicochemical Foundations of Metallurgical Processes” named after Academician A.M. Samarin	5	Sivtsov A.V., Sheshukov O.Yu., Egiazaryan D.K., Tsymbalist M.M., Orlov P.P. Influence of amphoteric oxides content on valve effect characteristics of electric arc	5
Grigor'ev E.V., Kapelyushin Yu.E., Bil'genov A., Stepanov D.V., Khalikulov A.A. Removal of chlorine from electric arc furnace dust by oxidative roasting	5	Vusikhis A.S., Tyushnyakov S.N., Pikulin K.V., Agafonov S.N., Leont'ev L.I. Boron distribution between metal and slag during melting of metalized siderite concentrate in electric furnace	5
Glukhov I.V., Voronov G.V., Sheshukov O.Yu., Kalganov M.V. Comparative analysis of kinetic and diffusion modes of natural gas combustion in EAF burners	5	To the 50th Anniverssary of O.V. Zayakin	6
Gudov A.G., Burmasov S.P., Smirnov L.A. Formation of iron-carbon melts and technological aspects of carburization ...	5	To the 90th Anniversary of Nikolai Filippovich Yakushevich	3
Kel' I.N., Zhuchkov V.I., Zayakin O.V. Correlation between the structure and properties of ferroalloys	5	To the Anniversary of Aleksei Borisovich Yur'ev	2
Olyanina N.V., Beltyukov A.L., Ladyanov V.I. Viscosity of liquid alloys of cobalt with silicon and boron: Experiment and calculation	5	Yur'ev A.B., Temlyantsev M.V. To the 95 th Anniversary of SibSIU – the first university in Kuzbass!	3

УКАЗАТЕЛЬ СТАТЕЙ, ПОМЕЩЕННЫХ В ЖУРНАЛЕ «ИЗВЕСТИЯ ВЫСШИХ УЧЕБНЫХ ЗАВЕДЕНИЙ. ЧЕРНАЯ МЕТАЛЛУРГИЯ» ЗА 2025 ГОД, ТОМ 68

ИСТОРИЯ ОТРАСЛИ

- Тресвятский Л.А., Протопопов Е.В., Уманский А.А., Худолеев А.Н. Кузнечная броня: первые рубежи победы 2

МЕТАЛЛУРГИЧЕСКИЕ ТЕХНОЛОГИИ

- Байдин В.В., Уманский А.А. Моделирование и разработка технологических режимов производства мелющих шаров повышенной твердости и ударной стойкости 3
- Бессонова О.В., Полевой Е.В., Осколкова Т.Н., Комарова Т.А. Сравнительный анализ структуры и свойств железнодорожных рельсов заэвтектоидного состава 6
- Богданова Н.А., Жилин С.Г., Предеин В.В. Прочностные характеристики выплавляемых моделей, получаемых уплотнением порошков воскообразных материалов в поле действия центробежных сил 6
- Головизнин С.М., Брюханов И.Ю. Анализ энергозатрат многократного волочения стальной проволоки 3
- Громов В.Е., Коновалов С.В., Полевой Е.В. Антология рельсов производства «ЕВРАЗ Объединенный Западно-Сибирский металлургический комбинат» XXI века 1
- Карфидов А.О., Чиченев Н.А., Васильев М.В., Чиченева О.Н. Изготовление дуговых листовых элементов методом ступенчатой гибки 3
- Мозжегоров М.Н., Машенцева М.С. Разработка технологии производства профильных труб, обеспечивающей более высокую точность геометрических параметров по сравнению с зарубежными производителями 1
- Невский С.А., Башенко Л.П., Сарычев В.Д., Грановский А.Ю., Шамсугдинова Д.В. Влияние наклонного электрического поля на распад струи жидкости в процессах термообработки и наплавки 1
- Поспелов И.Д. Исследование модуля жесткости клетей чистовой группы непрерывного широкополосного стана горячей прокатки 1
- Теплякова Л.А., Куницына Т.С., Печковский В.А., Кашин А.Д. Разрушение отпущенной мартенситной стали при растяжении 1
- Харченко А.С., Сысоев В.И., Сибгатуллин С.К., Дзюба А.В., Савинов А.С., Харченко Е.О. Моделирование распределения компонентов, выделившихся из замасленной окалины, между водо-газо-пылевыми средами системы газоочистки доменной печи 4
- Шакиров М.К., Протопопов Е.В., Турчанинов Е.Б., Белецкий Е.А., Фатьянов С.С. Комплексная энерготермическая модель конвертерной плавки 3
- Шкирмонтов А.П. Моделирование работы ферросплавной печи при увеличении мощности и диаметра электродов 4
- Юнусов А.М., Полевой Е.В., Юнин Г.Н., Осколкова Т.Н. Опыт производства рельсов из бейнитной стали марок 30ХГ2С2АФМ и 30ХГ2САФН 2

ЭКОЛОГИЯ И РАЦИОНАЛЬНОЕ ПРИРОДОПОЛЬЗОВАНИЕ

- Васина М.В., Башенко Л.П. Перспективные конструкции газоанализаторов для металлургии 4
- Ибраев И.К., Ибраева О.Т. Экологические особенности окислительного рафинирования фосфористых чугунов в большегрузных конвертерах 3
- Ульева Г.А., Туысхан К., Мацугина Е.М., Волокитина И.Е., Ахметова Г.Е. Современное состояние проблемы утилизации отходов производства 2
- Шляров В.В., Шлярова Ю.А., Башенко Л.П., Загуляев Д.В. Уменьшение количества отходов и внедрение экологически безопасных и эффективных производственных процессов при использовании высококремнистых сплавов системы Al – Si 3

МАТЕРИАЛОВЕДЕНИЕ

- Алеутдинова М.И., Фадин В.В. Особенности изнашивания контактной поверхности стали при скольжении под низким контактным давлением 3
- Алеутдинова М.И., Фадин В.В. Характеристики контакта стали Ст3 при сухом скольжении по стали 45 под переменным электрическим током высокой плотности при разных коэффициентах трансформации питающего трансформатора 6
- Беломытцев М.Ю., Михайлов М.А., Козлов Д.А., Михайлов А.М., Каравацкий И.И. Исследование влияния режимов термической обработки на свойства сплава 56ДГНХ 1
- Власов И.В., Гордиенко А.И., Семенчук В.М. Влияние термических циклов на формирование структуры жаропрочной стали перлитного класса в условиях проволочного электродугового аддитивного производства 1
- Гойхенберг Ю.Н., Кулаков Б.А., Полухин Д.С. Неметаллические включения в микропорах массивных стальных отливок 3
- Громов В.Е., Миненко С.С., Чапайкин А.С., Семин А.П., Шлярова Ю.А. Перспективы создания быстрорежущих высокоэнтропийных сталей 4
- Зоря И.В., Полетаев Г.М., Бебихов Ю.В., Семенов А.С. Инициация плавления на границах зерен наклона в аустените в зависимости от угла разориентации 2
- Ишмаматов Д.А., Помельникова А.С., Петелин А.Л. К вопросу о влиянии добавок оксидов редкоземельных элементов на кинетику образования борированных слоев и диффузию бора по границам зерен при борировании сталей 2
- Клевцова Н.А., Валиев Р.З., Клевцов Г.В., Фесенюк М.В., Пигалева И.Н., Балашов В.М. Мартенситные превращения в метастабильной аустенитной стали с круп-

нозернистой и ультрамелкозернистой структурой при кручении	3
Климов С.А., Носков Ф.М., Токмин А.М., Масанский О.А. Получение карбидосталей на основе быстрорежущей стали методом индукционной наплавки	6
Колтыгин А.В., Баженов В.Е., Белова А.А., Санников А.В., Лыскович А.А., Белов В.Д., Щедрин Е.Ю. Влияние термической обработки на фазовый состав, структуру, твердость и электропроводность никелевого жаропрочного сплава ВЖЛ14Н-ВИ	1
Орлова Д.В., Шляхова Г.В., Надежкин М.В. Особенности локализации деформации в аддитивном материале со структурно-фазовой неоднородностью	3
Пермякова И.Е., Иванов А.А., Лукина И.Н., Костина М.В., Дюжева-Мальцева Е.В. Инженерия поверхности аморфного сплава $Fe_{53,3}Ni_{26,5}B_{20,2}$ эксимерным лазером	6
Попова М.В., Полунин А.М. Структура и свойства сплавов специального назначения после отжига	6
Попова Н.А., Громов В.Е., Никоненко Е.Л., Порфирьев М.А., Юрьев А.А. Упрочнение поверхностных слоев головки длинномерных рельсов при длительной эксплуатации	6
Ярьско С.И., Балакиров С.Н., Осколкова Т.Н. Предельные энергетические характеристики при лазерной импульсной обработке вольфрамокобальтовых твердых сплавов	4

ИННОВАЦИИ В МЕТАЛЛУРГИЧЕСКОМ ПРОМЫШЛЕННОМ И ЛАБОРАТОРНОМ ОБОРУДОВАНИИ, ТЕХНОЛОГИЯХ И МАТЕРИАЛАХ

Евстигнеев А.И., Одинокое В.И., Чернышова Д.В., Евстигнеева А.А., Дмитриев Э.А. Влияние кольцевых швов на напряженно-деформированное состояние в цилиндрической керамической оболочковой форме при затвердевании в ней стальной отливки	6
Евстигнеев А.И., Чернышова Д.В., Одинокое В.И., Дмитриев Э.А., Евстигнеева А.А. Оптимизационное моделирование трещиностойкости керамической оболочковой формы при охлаждении в ней стальной отливки	3
Евстигнеев А.И., Чернышова Д.В., Одинокое В.И., Дмитриев Э.А., Евстигнеева А.А., Колошенко Ю.Б., Потанин Д.А. Моделирование и оптимизация влияния температурных швов на напряженно-деформированное состояние сферических металлических литейных форм	4
Куликов Б.П., Сторожев Ю.И., Потапенко А.С. Металлизация рудноугольных брикетов в кольцевой печи, отапливаемой генераторным газом	4
Лехов О.С., Билалов Д.Х. Расчет температуры и термоупругих напряжений в бойках при получении стальных полых заготовок на установке совмещенного литья и деформации. Часть 1	4
Лехов О.С., Билалов Д.Х. Расчет температуры и термоупругих напряжений в бойках при получении стальных полых заготовок на установке совмещенного литья и деформации. Часть 2	6
Сивков А.С., Чибизова С.И., Апасова А.Д. Направления совершенствования конструкции и тепловой работы	

вращающейся печи для реализации технологии производства металлургической извести из мела	4
Чжан Ц., Козлов В.Н., Чинахов Д.А., Клименов В.А., Чернухин Р.В. Анализ процесса обработки заготовки из мартенситной нержавеющей стали 40X13, полученной методом проволочного электронно-лучевого аддитивного производства	6

ФИЗИКО-ХИМИЧЕСКИЕ ОСНОВЫ МЕТАЛЛУРГИЧЕСКИХ ПРОЦЕССОВ

Большов Л.А., Корнейчук С.К., Большова Э.Л. Термодинамические параметры взаимодействия второго порядка азота с никелем и хромом в жидкой стали	2
Вусихис А.С., Михеенков М.А., Леонтьев Л.И., Агафонов С.Н. Влияние добавок оксида бора на вязкость и температуру плавления системы $CaO - SiO_2 - Al_2O_3 - MgO$	3
Дмитриев А.Н., Бурова Ю.Е., Витькина Г.Ю., Барбин Н.М., Терентьев Д.И. Термодинамическое моделирование процесса восстановления железорудных материалов водородсодержащими газами	2
Казакевич Г.А., Попов А.Ю. Построение поверхности ликвидус пятикомпонентной схемы диаграммы $Fe - B - Mn - C - Cr$	1
Смирнов К.И., Гамов П.А., Рошин В.Е., Самолин В.С. Термодинамический анализ условий разделения железа и титана в ильменитовом концентрате селективным восстановлением элементов	2
Шешуков О.Ю., Михеенков М.А., Егназарьян Д.К., Михеенков А.М., Клеоновский М.В., Матюхин О.В. Особенности формирования фазового состава сталеплавиных шлаков и оценка возможности получения на их основе минеральных вяжущих веществ с низкой генерацией CO_2	2
Шешуков О.Ю., Невидимов В.Н., Некрасов И.В., Метелкин А.А., Цепелев В.С. Развитие исследований физико-химических свойств оксидных и металлических расплавов	1

ИНФОРМАЦИОННЫЕ ТЕХНОЛОГИИ И АВТОМАТИЗАЦИЯ В ЧЕРНОЙ МЕТАЛЛУРГИИ

Дмитриев А.Н., Витькин Д.А., Золотых М.О., Витькина Г.Ю. Математическая модель состояния горна доменной печи на основе показаний термпар, находящихся в поясах холодильников	3
Дмитриев А.Н., Витькин Д.А., Золотых М.О., Витькина Г.Ю. Математическая модель состояния футеровки главного желоба доменной печи	4
Зимин А.В., Корнет М.Е., Буркова И.В., Зимин В.В. Использование цифровых имитационных моделей для оптимизация сложных производственных объектов	2
Рыбенко И.А., Протопопов Е.В., Запольская Е.М. Общие принципы моделирования и оптимизации электросталеплавиного процесса	6
Соколовская Э.А., Босов Е.В., Кудря А.В., Кодиров Д.Ф., Алексеев В.И. Оценка различий в информативности результатов прямо-сдаточных испытаний металлопродукции при статистическом анализе больших массивов данных производственного контроля в металлургии	3

Спирин Н.А., Гурин И.А., Лавров В.В. Информационно-моделирующая система оценки нестабильности функционирования доменной печи	4
Шалаевский Д.Л. Методика расчета плоскостности стальных холоднокатаных полос	4

ЭКОНОМИЧЕСКАЯ ЭФФЕКТИВНОСТЬ МЕТАЛЛУРГИЧЕСКОГО ПРОИЗВОДСТВА

Фастыковский А.Р., Мусатова А.И., Мартюшев Н.В. Разработка моделей функционирования волочильного оборудования при многостановом обслуживании	1
Черникова О.П., Афанасьева О.В., Афанасьев Е.Г. Направления декарбонизации Российской черной металлургии	1

В ПОРЯДКЕ ДИСКУССИИ

Мурашов В.А., Строгонов К.В., Бастынец А.К., Львов Д.Д. Разработка агрегата внепечной обработки стали непрерывного действия	1
Рошин В.Е., Рошин А.В. Электрохимия восстановительных процессов и перспективы развития восстановительных технологий	4

ПО МАТЕРИАЛАМ XVI МЕЖДУНАРОДНОЙ НАУЧНОЙ КОНФЕРЕНЦИИ «ФИЗИКО-ХИМИЧЕСКИЕ ОСНОВЫ МЕТАЛЛУРГИЧЕСКИХ ПРОЦЕССОВ» ИМЕНИ АКАДЕМИКА А.М. САМАРИНА

Агафонов С.Н., Удоева Л.Ю., Вусихис А.С., Леонтьев Л.И. Оценка технологической возможности совместной переработки ильменитовых и перовскитовых концентратов	5
Александров А.А., Анучкин С.Н., Каневский А.Г., Румянцева С.Б., Григорович К.В. Взаимодействие керамики на основе Al_2O_3 со шлаковым расплавом 45 % CaO – 40 % Al_2O_3 – 10 % SiO_2 – 5 % MgO	5
Бабенко А.А., Шартдинов Р.Р., Лобанов Д.А., Сметанников А.Н., Уполовникова А.Г., Гуляков В.С. Особенности применения борсодержащих шлаков при выплавке нержавеющей стали	5
Вусихис А.С., Тюшняков С.Н., Пикулин К.В., Агафонов С.Н., Леонтьев Л.И. Распределение бора между	

металлом и шлаком в процессе плавки металлизированного сидеритового концентрата в электропечи	5
Глухов И.В., Воронов Г.В., Шешуков О.Ю., Калганов М.В. Сравнительный анализ кинетического и диффузионного режимов горения природного газа в горелках дуговой сталеплавильной печи	5
Григорович К.В. Итоги XVI Международной научной конференции «Физико-химические основы металлургических процессов» имени академика А.М. Самарина	5
Григорьев Е.В., Капелюшин Ю.Е., Бильгенов А., Степанов Д.В., Халикулов А.А. Удаление хлора из пыли дугового сталеплавильного производства окислительным обжигом	5
Гудов А.Г., Бурмасов С.П., Смирнов Л.А. Исследование формирования железоуглеродистых расплавов и технологические аспекты науглероживания	5
Дрозин А.Д., Дудоров М.В. Критерии существования локальных равновесий в переохлажденных расплавах	6
Кель И.Н., Жучков В.И., Заякин О.В. Корреляция структуры и характеристик ферросплавов	5
Олянина Н.В., Бельтюков А.Л., Ладьянов В.И. Вязкость жидких сплавов кобальта с кремнием и бором: эксперимент и расчет	5
Сабиров Э.Р., Поколенко А.Ю., Бижанов А.М., Берснев И.С., Спирин Н.А. Исследование процесса науглероживания окатышей до содержания углерода более 4,5 % при металлизации в шахтных печах	5
Сафарова Д.Э., Базалеева К.О., Понкратова Ю.Ю., Алексеев А.В. Зависимость структуры и свойств сплава ВТ23 от параметров лазерного выращивания	5
Сивцов А.В., Шешуков О.Ю., Егизарьян Д.К., Цымбалист М.М., Орлов П.П. Влияние содержания амфотерных оксидов в шлаке на характеристики вентильного эффекта электрической дуги	5
Швецов Д.П., Чикова О.А., Цепелев В.С., Сеницин Н.И., Вьюхин В.В. Физико-химические исследования расплавов стали 12X18H9ТЛ для управления качеством литых изделий	5
К 90-летию Николая Филипповича Якушевича	3
К юбилею Алексея Борисовича Юрьева!	2
Члену-корреспонденту РАН О.В. Заякину 50!	6
Юрьев А.Б., Темлянец М.В. Сибирскому государственному индустриальному университету 95!	3

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Comparative analysis of the structure and properties of hypereutectoid railway rails

Strength characteristics of investment patterns obtained by compaction of waxy material powders in the field of centrifugal forces

Hardening of surface layers of long rail heads during long-term operation

Production of carbide steels based on high-speed steel by induction surfacing

Contact characteristics of C235 steel in dry sliding against C45 steel under high-density alternating current at different transformation coefficients of supply transformer

Structure and properties of special-purpose alloys after annealing

Surface engineering of $\text{Fe}_{53.3}\text{Ni}_{26.5}\text{B}_{20.2}$ amorphous alloy by excimer laser

Criteria for the existence of local equilibria in supercooled melts

Effect of annular seams on stress-strain state in cylindrical ceramic shell mold during solidification of a steel casting in it

Calculation of temperature and thermoelastic stresses in strikers during production of hollow steel billets in a unit of combined casting and deformation. Part 2

Analysis of processing 40Kh13 martensitic stainless steel billet obtained by wire electron beam additive manufacturing

General principles of modeling and optimization of electric steelmaking process

To the 50th Anniverssary of O.V. Zayakin

Index of articles of "Izvestiya. Ferrous Metallurgy" for 2025, vol. 68

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