

50 years of the Condensed Matter Physics Research Council of the Russian Academy of Sciences

(Scientific session of the Physical Sciences Division of the Russian Academy of Sciences, 9 November 2011)

DOI: 10.3367/UFNe.0182.201205g.0559

The scientific session of the Physical Sciences Division of the Russian Academy of Sciences (RAS) commemorating the 50th anniversary of the RAS Scientific Council on Physics of Condensed Media took place on 9 November 2011 at the conference hall of the P N Lebedev Physical Institute of the RAS.

The following reports were put on the session agenda posted on the website www.gpad.ac.ru of the RAS Physical Sciences Division:

(1) **Kveder V V** (Institute of Solid State Physics, RAS, Chernogolovka, Moscow region) “50th anniversary of the RAS Scientific Council on Physics of Condensed Media”;

(2) **Vul’ A Ya** (A F Ioffe Physical-Technical Institute, RAS, St. Petersburg) “Carbon nanostructures: from fullerenes to graphene. Achievements and unsolved problems”;

(3) **Glezer A M** (Bardin Central Research Institute of Ferrous Metallurgy, Moscow) “Development of new-generation multifunctional structural materials: guiding principles”;

(4) **Pyatakov A P** (M V Lomonosov Moscow State University, Moscow; A M Prokhorov General Physics Institute, RAS, Moscow) “Magnetoelectric phenomena and micromagnetism”.

The paper that follows is based on the oral report 3. The contents of report 4 are presented in expanded form in the review by A P Pyatakov and A K Zvezdin, “Magnetoelectric and multiferroic materials”, which will be published in one of the later issues of *Physics–Uspekhi*.

PACS numbers: 61.43. –j, 61.46. +w, 62.50. –p
DOI: 10.3367/UFNe.0182.201205h.0559

Creation principles of new-generation multifunctional structural materials

A M Glezer

1. Introduction

Modern technological advances impose ever growing demands on structural and functional metallic materials. The requirement of multifunctionality, i.e., of simultaneously combining high-level mechanical, physical, and chemical properties that are often hard to combine in one and the same material, is a qualitatively new demand [1].

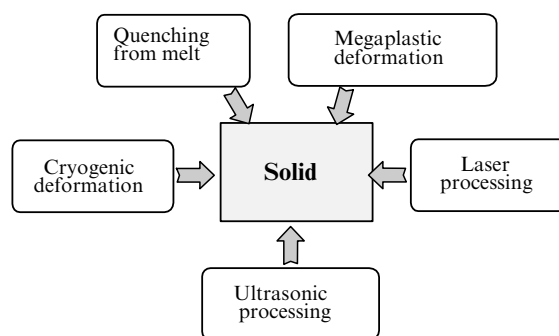


Figure 1. Methods of extreme actions discussed in the paper.

Standard materials science principles of structure and property formation, based on conventional thermal and deformation factors acting on a solid object, already fail to meet the new requirements, even though sometimes the desired results can be obtained by building expensive multi-component systems [2].

There is no doubt that the key to creating new unconventional properties of materials is to produce in them previously unknown, principally new structural states; exposing solids to extreme action may offer a method of producing such structures [3]. The famous British physicist R W Cahn expressed his conviction that materials pushed into extreme states constitute the latest achievements of materials science, and that this field will be actively pursued in the future [4]. Cahn defined states as extremal if they result from the extreme conditions of processing the material and, as a consequence, from the unique nature of the microstructure that forms during this processing [4]. Some types of such processing are demonstrated in Fig. 1.

2. Melt quenching

Following R W Cahn, we consider rapid quenching from melt (RQM) as the most efficient method of extremal processing,

A M Glezer State Scientific Center of the Russian Federation ‘Bardin Central Research Institute of Ferrous Metallurgy’, Kurdyumov Institute for Physical Metallurgy, Moscow, Russian Federation; National Research and Technology University ‘Moscow Institute of Steel and Alloys’, Moscow, Russian Federation
E-mail: a.glezer@mail.ru

which offers entirely new opportunities for creating advanced materials with unique combinations of properties. Several versions of RQM are currently available, and their different modifications make it possible to produce amorphous and microcrystalline alloys in the form of ribbons up to 100 μm thick and up to 300 mm wide [5]. And yet, the most common is the spinning technique [6] in which the molten metal is poured under pressure on a rapidly revolving cooling disk. Melt spinning produces ribbons from 20 to 100 μm thick, whose structure is determined by alloy composition and cooling rate. RQM on a rotating cooling disk is a very short-time process which depends on many physical and technological parameters. By changing or adjusting these parameters, one has the opportunity to achieve optimum conditions for RQM and to influence the structural state of quenched materials of a given chemical composition in the desired way.

The nature of the structure formed by rapid quenching changes significantly with increasing cooling rate. First, the polycrystalline structure gets considerably finer, then the solubility of components in the solid solution increases, depending on the alloy composition, and metastable crystalline phases may form. Finally, if the cooling rate is very high, crystallization gets completely suppressed due to lack of time and the shear viscosity of the system increases smoothly and continuously in the course of cooling. In its final form, the atomic structure inherent in the liquid state departs from the state of thermodynamic equilibrium and becomes homogeneously frozen at glass transition temperature T_g [7]. The resulting amorphous state is one of the key factors in conducting quenching experiments.

In accordance with the kinetic approach to the process of amorphization, any metallic alloy containing at least several percent of the second component can be amorphized on achieving a certain high cooling rate [8]. Therefore, the tendency to amorphization in an alloy of a given chemical composition is characterized by the critical cooling rate V_c ; in the case of applying the spinning technique, this corresponds to the critical thickness t_c of the ribbon being quenched [6]. The value of t_c for the maximum achievable cooling rate at melt spinning, 10^6 K s^{-1} , is typically below 70–80 μm for most systems undergoing amorphization [9]. It should be mentioned that a number of easily amorphized multicomponent metallic systems (mostly based on Pd–Cu, Ti–Zr, Zr–Cu, and Mg–Cu) have been discovered in recent years, for which V_c was found to be so low that such systems can be manufactured in the form of solid bars and ingots [10]. In reality, the overwhelming majority of amorphous and nanocrystalline alloys of practical importance are producible only in the form of ribbons several dozen micrometers thick.

The transition from the liquid (amorphous) state to the crystalline state may be treated as a disorder–order type transition [11]. In principle, such a transition can be implemented either in the process of cooling from the melt at a certain near-critical rate or by exposing a solid-phase amorphous state—obtained in turn by using RQM—to temperatures or deformations. Crystallization then occurs under conditions of a constant heat conveyance (at a constant or continuously increasing temperature); note that crystallization itself releases additional thermal energy. As a result of this, a structure consisting of two clearly distinguishable structural components is formed in most cases at a certain stage of thermal or deformation treatment of a system: an amorphous one, and a nanocrystalline one [12]. The characteristics of such a structure depend to a certain extent on the

cooling rate of RQM, on the subsequent heating, and on the annealing temperature and the quenching atmosphere, or on the parameters of deformational action.

A very different type of morphological structure can be created during the early stages of crystallization under the conditions of abrupt cooling of the melt, when efficient heat removal from the crystallizing system is implemented. Such combined amorphous–crystalline systems have been studied very little, but the mechanical properties produced in this way can be described as unique (Vickers hardness $HV = 15 \text{ GPa}$, and strength $\sigma_p = 4.8 \text{ GPa}$ [13]).

In principle, depending on the value of V_c and heat-removal parameters, RQM leads to implementation of five different scenarios and, correspondingly, five states that differ in structure and properties:

(1) the liquid phase is totally crystallized in the RQM process, and we deal with a single-phase or multiphase nano- or submicrocrystalline structure;

(2) RQM creates an amorphous state which undergoes partial or full crystallization in the subsequent cooling to below T_g . Another scenario is also possible: crystallization proceeds directly from the melt and other areas of the melt simultaneously switch to the amorphous state. This results in the formation of a combined amorphous–nanocrystalline structure;

(3) RQM forms a metastable amorphous state which is stable over a wide temperature range and is characterized by the absence of long-range (crystalline) order in the arrangement of atoms and, at the same time, by the presence of significant compositional and topological short-range order in the first coordination spheres. Amorphous alloys have high plasticity and high strength closely approaching the theoretical limit. Such a combination of mechanical properties turns amorphous alloys into unique subjects of both purely scientific and application-oriented interest [14];

(4) subsequent thermal treatment under appropriate conditions results in partial or complete crystallization of the RQM-produced amorphous state;

(5) subsequent exposure to very intense deformation results in partial or complete nanocrystallization of the RQM-produced amorphous state.

As evidenced by the foregoing, a single RQM session is sufficient for the formation of crystalline and amorphous structures of the first, second, and third types, while the formation of structures of the fourth and fifth types requires two stages (RQM + thermal or deformational treatment). The above-described structural classification of materials was proposed in Ref. [15]; it allows approaching the description of structural peculiarities found in RQM materials in a more physically reasonable manner and, as a result, getting closer to the goal of obtaining a desirable combination of the required physical and mechanical properties. It is important to emphasize that the nanocrystalline state obtained using RQM always arises in accordance with the ‘bottom-up’ principle (nanocrystals grow from the melt or from an amorphous matrix), which makes quenching from melt a close relative of the most progressive modern nanotechnologies established on the same principle [16].

The above analysis is a conclusive demonstration of the breadth and diversity of the range of structural states of nanomaterials that RQM helped to create (amorphous, amorphous–nanocrystalline, and nanostructured states). The physical and mechanical properties inherent in these structures are correspondingly diverse, and even unique.

We were able to resolve the following important issues in the process of developing promising amorphous alloys (AAs) obtained by RQM techniques. We have

- formulated the concept of regions of excess free volume, constituting characteristic defects responsible for the processes of plastic deformation and fracture of AAs [17];
- identified the nature of temper brittleness in AAs and suggested methods of purposefully affecting the temperature range of its manifestation [18];
- proposed a method for estimating the temperature- and time-stability of the mechanical properties of AAs [19];
- developed a method of cryogenic processing of AAs, resulting in improved magnetic and mechanical properties [20];
- discovered the plastification effect in the transition from the amorphous to the nanocrystalline state [21];
- developed carbide-hardened high-strength AAs [22];
- found a structural model of plastic deformation of AAs containing nanocrystals [23].

The following problems were successfully solved in the course of developing promising RQM-produced nanocrystalline alloys (NCAs). We have

- shown that, far from slowing down the course of phase transitions in solids, RQM can in fact accelerate them [24];
- established that athermal and thermoelastic martensitic transformations are suppressed for a critical size of nanocrystals [25];
- optimized the conditions for obtaining Finemet type alloys with unique magnetic and mechanical properties [26];
- suggested a new mechanism of plastic flow in NCAs: low-temperature grain-boundary microslip [27];
- shown that the anomaly in the Hall–Petch relationship in NCAs stems from the change in the mechanism of plastic deformation: from dislocation-assisted flow to the grain-boundary microslip [28];
- suggested a new classification of nanocrystals based on their mechanical behavior [29], and
- established structural mechanisms of nanocrystal destruction [30].

Figure 2 gives a schematic demonstration of the most impressive examples of obtaining a new generation of multifunctional alloys using RQM techniques. Furthermore, another large independent group of amorphous alloys comprises those which are successfully used as reinforcing elements in ceramic-, plastic-, and concrete-based composites. Using high-strength amorphous alloys as materials for

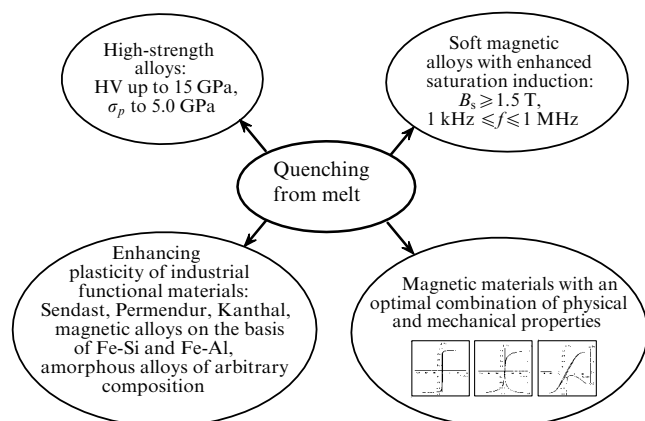


Figure 2. Creation of new multifunctional materials by RQM techniques.

abrasive and cutting tools also appears very promising. There can be no doubt that the fields of application of amorphous and nanocrystalline alloys produced by RQM techniques and possessing valuable mechanical, magnetic, electrical, and corrosion properties will be constantly widening.

3. Megaplastic deformation

The effect of very large plastic deformations on a metallic solid should definitely be included in the class of extreme phenomena. The interest in this method of action on the materials has grown considerably over recent years, as it promises the possibility of significantly improving the physical and mechanical properties of metallic materials [31]. This is caused to a great extent by the formation of different types of nanostructured states, and in particular by nanocrystallization in the course of processing amorphous alloys quenched from melt.

Plastic deformation in which the true strain ϵ is greater than 1 and can reach magnitudes above 7–8 is usually referred to in the literature in Russian as *high-intensity plastic deformation*. In fact, this term does not appear to be very satisfactory. Indeed, ‘highly intense natural process’ typically stands for a process unfolding at a high rate [32]. Estimates show that the rate of strain growth at very high deformations is typically not very high (in the range of $10^{-1} - 10^1 \text{ s}^{-1}$), i.e., it falls into the transition region between static and dynamic strain rates and corresponds to rates typical of conventional rolling. Such deformations cannot, therefore, be regarded as intense. The term used in publications in English, *severe plastic deformation* [33] is more appropriate, and can be translated into Russian as ‘strict’, ‘hard’, ‘deep’, but preferably ‘strong’ plastic deformation [34].

A different, more physically rigorous, Russian term was suggested in paper [35]. It can be traced back to the general philosophy concepts of the matter surrounding us. It is well known [36] that natural sciences separate the material world into three scale levels: the microcosm in which phenomena unfold on the scale of individual atoms and molecules; the macrocosm of human perception of the world: meters, kilograms, second, and the megaworld with its astronomical scale. There is a close analogy between the above-described scales of matter organization and the levels of plastic deformation. Indeed, we know well the process of microplastic deformation which is observed at stresses not exceeding the macroscopic yield stress, and the macroplastic deformation process which is realized at stresses exceeding the yield stress [37]. It is then logical, in the spirit of the above analogy, to refer to very large plastic deformation as megaplastic deformation (MPD), which conforms to the logic of evolution of any material phenomenon.

Among the methods of producing gigantic strains, the most widely used are torsion under pressure in a Bridgman high-pressure chamber (TPBC), equichannel angular pressing (ECAP), torsional extrusion, and accumulated rolling [38]. Plastic strains produced by these techniques are so huge that conventional magnitudes of relative strains become meaningless, and one has to turn to true logarithmic strain ϵ .

Summarizing a large number of experimental studies into the structure of materials subjected to megaplastic deformation, we can state that observations reveal a complex combination of defect structures containing low-angle and high-angle grain boundaries whose relative percentages vary, and also defect structures inside the grains of various degrees of perfection. The three-dimensional statistical assessment of

the resulting structure of grains, popular in the literature, yields at best a ratio of the high-angle to low-angle boundaries in the material, and very little information on the nature of the physical processes which actually take place under severe plastic deformation. On top of it, this information is unfortunately rather contradictory and ambiguous, since different experiments, dealing with the same materials subjected to similar deformation conditions, typically observe different structural states. Note that the true nanostructured state ($d < 0.1 \mu\text{m}$) is not formed in every one of the cases. In steels and alloys, MPD is not infrequently accompanied by phase transitions: precipitation and dissolution of the phases, martensitic transformation, and amorphization [39]. As a rule, amorphization occurs at relatively low temperatures in intermetallics or multicomponent systems. In our studies, we proved it possible to increase strength after MPD at room temperature by 25% on retention of a sufficient plasticity by using ECAP for high-strength martensite-aging steels just via significant fragmentation of the products of the reverse $\alpha \rightarrow \gamma$ transformation (the size of γ particles being about 20–30 nm) [40].

The most coherent concept of severe plastic deformations was suggested by V V Rybin [41]. Using the concepts of the dominant role of disclination modes when implementing large plastic deformations and related fragmentation processes, Rybin was able to correctly describe the phenomena occurring at significant strains close to $e = 1$. In accordance with the disclination concept, the size of the fragments — the basic structural elements — decreases continuously as strain increases and reaches a constant minimum value of $0.2 \mu\text{m}$. This, in fact, means that the transition to the range of the nanostructured state and the creation of fragments (grains) smaller than 100 nm are hardly possible in the disclination mode. S A Firstov et al. [42] pointed to the fact that the transition to the highly strained state is accompanied by abrupt changes in the structure of the material at a certain critical value e_c . In the case of technical-grade pure iron, one has $e_c \approx 1$. Also, the mechanical behavior of materials changes: strain hardening at high strains obeys a linear rather than parabolic law. It was shown in a number of studies (see, e.g., paper [43]) that a very large number of excessive point defects (mainly vacancies) are generated in the structure as MPD increases; these defects are capable of stimulating the occurrence of diffusive phase transitions in the course of deformation building up.

In Ref. [35] we examined the energy aspects of the behavior of solids under loading. When a solid body of finite size is subjected to a mechanical load, certain strain energy is ‘pumped’ into it (Fig. 3). Plastic deformation is an obvious ‘dissipation channel’ for this energy. Once the potential of this

channel is exhausted, another channel may show its worth — mechanical destruction. If, however, elastic energy is sufficiently high, plastic deformation can, in principle, initiate additional ‘dissipation channels’: dynamic recrystallization, phase transitions, and the release of thermal energy (see Fig. 3). In the case of MPD, when the stress component of uniform compression is large, the formation and growth of cleaving cracks are partly or completely suppressed and, hence, crystal destruction is significantly inhibited. In other words, application of ECAP, TPBC, or similar schemes of loading drives a solid body to deforming without destruction. If Rybin’s concepts [41] are correct, plastic deformation is efficient up to a certain limit corresponding to the formation of a critical defect structure, after which other physical processes (dynamic recrystallization, phase transformations, and/or heat release) are expected to act as the main dissipation channels. In Ref. [44] we proposed three possible scenarios of further development. If dislocation (disclination) rearrangement processes are facilitated in a material (for example, in pure metals), plastic deformation is followed by low-temperature dynamic recrystallization. The local regions of the structure are ‘cleansed’ of defects, and plastic flow resumes in the new recrystallized grains due to dislocation and disclination modes. In this case, dynamic recrystallization acts as a powerful additional channel of elastic energy dissipation. If the mobility of the plastic deformation carriers is relatively low (as, for example, in solid solutions or intermetallics), phase transitions step forward as a powerful additional channel for dissipation of elastic energy. Most often, this is a ‘crystal \rightarrow amorphous state’ transition. As a result, the plastic flow is localized in the amorphous matrix in the absence of effects of strain hardening and of the accumulation of large internal stresses. It appears that an intermediate case also occurs, in which disclination rearrangements may serve as an additional channel of dissipation, producing stabilization of the fragmented structure observed in some experiments as MPD unfolds.

Obviously, the transition from one scenario of structural rearrangements to another is also affected by the parameter T_{MPD}/T_m , where T_{MPD} is the MPD temperature (taking into account the possible effect of heat release), and T_m is the melting point. A detailed analysis of the structural processes proceeding in the course of severe plastic deformation allowed us to formulate three fundamental principles of the MPD implementation: the principle of hydrostatic functionality of the strained state under MPD; the principle of nondislocational modes of elastic energy dissipation under MPD, and the principle of cyclic behavior of structural and phase transitions under MPD. The significance of the first two is clear from the above. As for the principle of cyclic behavior, it is of a synergistic nature under the conditions of an open system in which, in the course of MPD, strain energy is constantly ‘pumped’ in [45]. If we examine a specific microscopic volume of the specimen undergoing deformation, the plastic deformation process that follows dynamic recrystallization or amorphization in the newly formed recrystallized grains or in the amorphous phase domain restarts, so to speak, from scratch. Then, deforming stress again produces accumulation of defects in this microscopic volume, and the process is repeated. The cyclic behavior of structure formation under MPD in the conditions of dynamic recrystallization was first explained by V A Likhachev et al. [46] for pure metals and then repeatedly confirmed by other experimenters. The cyclic behavior under MPD in the case of

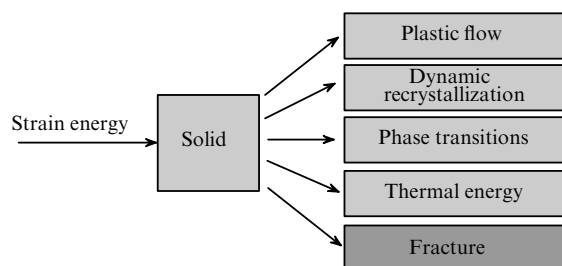


Figure 3. Energy principles of mechanical action on solids subjected to MPDs.

‘crystal \leftrightarrow amorphous state’ phase transitions in intermetallics has recently been reported in paper [47].

Calculations show that, for low Peierls barriers and in the presence of an efficient dissipation channel, the plastic flow curve is cyclic and displays a certain ‘wavelength’ $\Delta\varepsilon$ [44]. The fact that we almost never record such flow curves in experiments is not inconsistent with these theoretical predictions. The process of plastic flow is known to proceed extremely nonuniformly at every stage, and different domains of the deformed crystal are at various stages of their evolution.

It is important to emphasize that MPD as such is not a guarantee that a nanocrystalline state will be formed, with crystallites of less than 100 nm in size separated by high-angle or interphase boundaries. For example, this is practically impossible in pure metals with high dislocation mobility. Important factors in the formation of nanostructures under MPD are the occurrence of martensite and diffusive type phase transformations, twinning, and the transition to an amorphous state. We can therefore generate various types of nanostructures when stimulating phase transformations or twinning by varying the temperature and chemical composition of materials.

A hypothesis that MPD leads to the formation of ‘unconventional’ nonequilibrium grain boundaries [48] is very often mentioned now in the literature. A number of authors maintain that these boundaries are responsible for anomalous phenomena such as slipping, diffusion, and interactions with various lattice defects and, therefore, may be responsible for high magnitudes of strength and ductility. Our energy model of MPD makes it unnecessary to involve notions of unconventional strongly nonequilibrium grain boundaries. Furthermore, the boundaries formed during dynamic recrystallization can, in principle, be attributed to two types of boundaries: ordinary high-angle boundaries arising during dynamic recrystallization, and fragmented high-angle boundaries created in the course of deformation. It is probable that both types of boundaries possess the same properties as any other grain boundaries. We can assume that the concentration of defects in grain boundaries of the first type and in the contiguous microscopic volumes of crystals is lower, and these domains are responsible for the tendency of a material to plastic flow, while boundaries of the second type, having deformation origin, and microscopic volumes of crystals contiguous to them contain more defects and are responsible for hardening in accordance with the Hall–Petch relationship.

Figure 4 plots the volume fraction of both types of grain boundaries in pure iron, evaluated by us using electron microscopy after various stages of TPBC at room temperature. We see that as the number n of revolutions of the movable anvil increases, the fraction α_f of fragmented high-angle grain boundaries decreases and reaches saturation ($\alpha_{f0} = 0.6$) at $n = 3$; the fraction α_r of recrystallized grain boundaries correspondingly increases and also reaches saturation ($\alpha_{r0} = 0.4$) at $n = 3$. The ratio $K = \alpha_f/\alpha_r$ remains virtually unchanged in subsequent stages. In accordance with our assumption, it is the value of K that dictates the relative roles of strength and ductility in materials subjected to MPD. This conclusion is in some contradiction with the currently fashionable point of view that the efficiency of MPD is determined by the overall fraction of high-angle grain boundaries created under MPD, regardless of the nature of the grain boundaries [33].

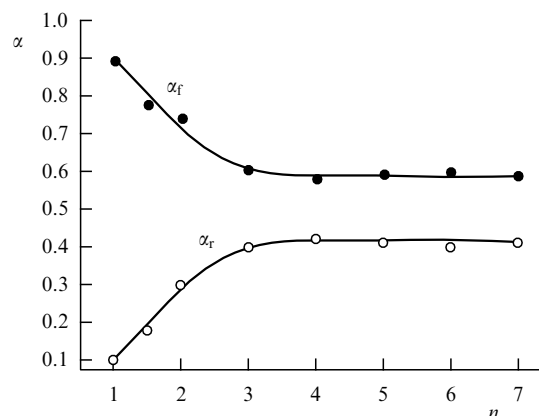


Figure 4. The volume fraction of high-angle grain boundaries formed in the course of fragmentation (α_f) and under dynamic recrystallization (α_r) as a function of the increasing number n of full revolutions of the movable anvil in the Bridgman chamber during MPD of α -Fe at room temperature.

4. Cryogenic deformation

It is a known fact [25] that martensitic transformation is suppressed in nanocrystalline structures. The higher the strain created by transformation, the larger the critical size that inhibits the growth of martensite crystals [49]. This factor makes it impossible to obtain nano-sized crystals of martensite, which are obviously capable of greatly enhancing the strength and improving a number of other functional properties of metallic materials, under normal conditions of processing. However, the low-temperature deformation, which has a number of significant features, both for generating dislocation structures and for activating martensitic transformations, has been known for quite some time now [50]. In view of this, we conducted experiments on low-temperature (77 K) deformation of Fe–Cr–Ni alloys having a single-phase face-centered cubic (fcc) lattice of γ phase at room temperature, which remained stable (it did not transform into a body-centered cubic (bcc) lattice of α -martensite either under cooling to 77 K or under plastic deformation at room temperature). We found that under certain experimental conditions simultaneous cooling to cryogenic temperatures and plastic deformation (by rolling) are capable of stimulating martensite $\gamma \rightarrow \alpha$ transformation, and, if the amount of α -martensite is high (volume fraction above 0.6), it can very significantly (by a factor of almost three) increase the strength and hardness of chromium–nickel alloys. Low amounts of Al dopant contributed to dispersion hardening and led to an additional increase in strength by another 15% after low-temperature annealing [51].

Structural analysis evidenced that such a strong reinforcing effect of cryogenic deformation stems from the formation of martensite nanocrystals 10–50 nm in size (Fig. 5) and of highly fragmented austenite crystals no larger than 100 nm which survived low-temperature tempering. There is no doubt that the combined effect of plastic deformation and temperature—each of these factors being rather inefficient—may serve as an efficient technique for creating a unique structure and unique properties of materials, which can also be successfully implemented in other metallic systems.

Unfortunately, the insufficient size of this article does not allow us to describe the efficiency of the other two extreme



Figure 5. Nano-sized crystals of α -martensite produced under cryogenic ($T = 77$ K) deformation in Fe–Ni–Cr–Al alloy; dark-field electron microscopy image in α -phase reflection.

forms of processing mentioned in Fig. 1: laser and ultrasonic treatments.

5. The principle of combined extreme actions

The processes of structure formation discussed in Sections 2–4 illustrated the diversity of nanostructured states that can be produced when materials in the condensed state are subjected to extreme actions. RQM, MPD, and laser irradiation techniques help produce unusual structures which give rise to materials with unique physical and mechanical properties. Under certain conditions, cryogenic deformation creates martensitic nanostructures of considerably higher strength and ductility and satisfactory corrosion resistance. Ultrasonic treatment produces noticeable relaxation of internal stresses and appreciable hardening of surface layers.

Our new idea was to combine the above extreme effects and ‘align’ them in a joint technological chain in the hope that this would result in qualitative changes in the nature of the final structure and, hence, in potentially unique properties of metallic materials subjected to integrated extreme treatments (Fig. 6). We have already produced some links of this chain (RQM leading to amorphous state + MPD) [52]. This did give us unique magnetic and mechanical properties of metal–metalloid type alloys; earlier noncombined approaches had failed to produce them [53] (Fig. 7). It is possible to align other, not yet scrutinized, technological chains, such as MPD + exposure to ultrasound, or RQM + cryogenic deformation. The chain of maximum interest is, of course, the four-link chain: RQM producing an amorphous, amorphous–nanocrystalline, or nanocrystalline structure + MPD + cryogenic deformation (or MPD at cryogenic temperatures) + ultrasonic treatment. However, integrated processing at this level of complexity requires

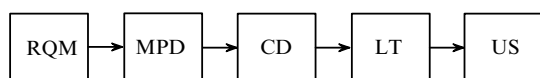


Figure 6. The principle of integrated extreme actions: RQM — quenching from melt, MPD — megaplastic deformation, CD — cryogenic deformation, LT — laser treatment, US — ultrasonic treatment.

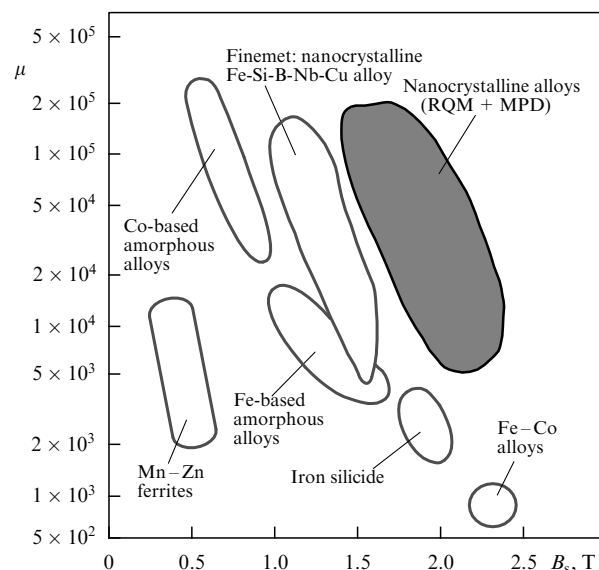


Figure 7. New multifunctional soft magnetic alloys created by integrated extreme-conditions treatments RQM + MPD.

‘synchronization’ of the parameters of all actions impacting the material if one wishes to achieve a qualitatively new structure and structure-sensitive properties.

Such synchronization should definitely be pursued by further research, via successive analysis of various scenarios of extreme actions and subsequent optimization of the conditions of their implementation. This approach to materials research is totally original and remains insufficiently represented in the literature.

6. The principle of grain boundary engineering

Grain boundaries constitute the most important element of the structure of materials. Grain boundaries offer a lever for controlling the physical and mechanical properties of materials. For example, the formation of grain boundary segregations of ‘harmful’ impurities can embrittle a material and significantly reduce its strength. On the contrary, deliberate controlled creation of segregations of ‘useful’ impurities at the grain boundaries, which enhance their cohesive strength, may drastically improve the mechanical properties of materials. Influencing the structure and phase composition of grain boundaries in a desired way, we learn to control the properties of materials. The principle of grain boundary engineering can be successfully applied, for instance, to alloys obtained using RQM or MPD. A research project that we have completed recently [54] is a good illustration of the promise of this approach. It was shown that incorporation of boride phases Ti_2B and TiB_2 about 5 nm in size into grain boundaries of nanoparticles in Ti–Ni–Cu-based nanocrystalline alloys suppresses, in accordance with the principle of grain boundary engineering, the low-temperature process of grain-boundary slipping and shifts the region of the Hall–Petch relationship anomaly to lower values of the average size of nanocrystals. This led to a 20% increase in the maximum normalized value of microhardness, which grew to the 95% level of the theoretical limit of this characteristic (Fig. 8).

7. Conclusion

We have analyzed the basic laws guiding the formation of the structure and of the unique physical and mechanical proper-

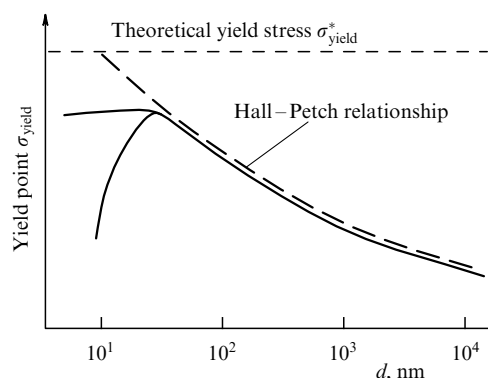


Figure 8. Application of the principle of grain boundary engineering to creating a superhigh-strength state approaching the theoretical limit for solids.

ties of metallic materials under extreme actions (quenching from melt, megaplastic and cryogenic deformation). We suggested the principle of combined extreme actions and applied it to developing multifunctional materials combining high strength and the desired level of physical properties. The method of grain boundary engineering allowed us to create model alloys of real strength approaching the theoretical limit. This paper discussed only two basic consequences of the analysis, given above, of extreme actions on the structure and properties of metallic materials, but readers impressed by the results will undoubtedly find for themselves a number of other useful ideas for new and efficient ways of designing promising materials for the near future. At any rate, we very much hope they will.

Acknowledgments

The author dedicates this article to the memory of the outstanding Russian physicist *Yury Andreevich Ossipyan*. Russia's scientific community has recently celebrated his 80th birthday. The author expresses his deep appreciation to his students and the colleagues who actively participated in our joint experiments.

This work was supported by the Russian Foundation for Basic Research (grants 09-02-00831, 09-02-90439, and 11-02-12063) and the Federal Target Program of the RF Ministry of Education and Science 'Scientific and Scientific-Pedagogical Personnel of Innovative Russia' (contract P2291).

References

- Osip'yan Yu A, Mogutnov B M, Glezer A M *Izv. Ross. Akad. Nauk Ser. Fiz.* **66** 1230 (2002)
- Firstov S A et al., in *Sovremennye Problemy Fizicheskogo Materialovedeniya* (Current Problems of Physical Materials Science) (Trudy IPM NANU, Issue 18) (Kiev: IPM, 2009) p. 140
- Fortov V E *Ekstremal'nye Sostoyaniya Veshchestva* (Extreme States of Matter) (Moscow: Fizmatlit, 2009); *Extreme States of Matter on Earth and in the Universe* (Berlin: Springer, 2011)
- Cahn R W *The Coming of Materials Science* (Amsterdam: Pergamon Press, 2001) [Translated into Russian (N. Novgorod: Izd. Nizhegorodskogo Gos. Univ., 2011)]
- Luborsky F E (Ed.) *Amorphous Metallic Alloys* (London: Butterworths-Heinemann, 1983) [Translated into Russian (Moscow: Metallurgiya, 1987)]
- Filonov M R, Anikin Yu A, Levin Yu B *Teoreticheskie Osnovy Proizvodstva Amorfnykh i Nanokristallicheskikh Splavov Metodom Sverkhbystroi Zakalki* (The Theoretical Basis of the Production of Amorphous and Nanocrystalline Alloys by Ultrarapid Quenching) (Moscow: MISiS, 2006)
- Zolotukhin I V, Kalinin Yu E *Usp. Fiz. Nauk* **160** (9) 75 (1990) [*Sov. Phys. Usp.* **33** 720 (1990)]
- Sudzuki K, Fudzimori H, Hashimoto K (Eds) *Amorphous Metals* (London: Butterworths, 1983) [Translated into Russian (Moscow: Metallurgiya, 1987)]
- Cantor B (Ed.) *Rapidly Quenched Metals III* (London: Metals Society, 1978) [Translated into Russian (Moscow: Metallurgiya, 1983)]
- Inoue A, in *Amorphous and Nanocrystalline Materials: Preparation, Properties, and Applications* (Eds A Inoue, K Hashimoto) (Berlin: Springer, 2001) p. 1
- Frenkel' Ya I *Vvedenie v Teoriyu Metallov* (Introduction to the Theory of Metals) (Leningrad: Nauka, 1972)
- Glezer A M *Russ. Khim. Zh.* **XLVI** (5) 57 (2002)
- Glezer A M et al. *Fiz. Met. Metalloved.* **64** 1106 (1987)
- Glezer A M et al. *Mekhanicheskoe Povedenie Amorfnykh Splavov* (Mechanical Behavior of Amorphous Alloys) (Novokuznetsk: Izd. SibGIU, 2006)
- Glezer A M *Materialovedenie* (3) 10 (1999)
- Tret'yakov Yu D *Vestn. Ross. Akad. Nauk* **77** 3 (2007) [*Herald Russ. Acad. Sci.* **77** 7 (1990)]
- Betekhtin V I, Kadomtsev A G, Kipyatkova A Yu, Glezer A M *Fiz. Tverd. Tela* **40** 85 (1998) [*Phys. Solid State* **40** 74 (1998)]
- Glezer A M *J. Physique IV* **8** Pr8-175 (1998)
- Glezer A M et al. *Fiz. Met. Metalloved.* **80** (2) 142 (1995)
- Zaichenko S G et al., RF Patent No. 2154869 (1999)
- Glezer A M, Permyakova I E, Manaenkov S E *Dokl. Ross. Akad. Nauk* **418** 181 (2008) [*Dokl. Phys.* **53** 8 (2008)]
- Glezer A M et al., RF Patent No. 2001457 (1989)
- Glezer A M et al., in *IV Mezhdunar. Konf. "Deformatsiya i Razrushenie Materialov i Nanomaterialov"*, Moskva, 25–28 Oktyabrya 2011 g. (IV Intern. Conf. "Deformation and Mechanical Failure in Materials and Nanomaterials", Moscow, 25–28 October, 2011) *Sbornik Materialov* (Collected Papers) (Moscow: IMET RAN, 2011) p. 303
- Glezer A M et al. *Dokl. Ross. Akad. Nauk* **407** 478 (2006) [*Dokl. Phys.* **51** 197 (2006)]
- Glezer A M et al. *J. Nanopart. Res.* **5** 551 (2003)
- Shurygina N A et al. *Izv. Ross. Akad. Nauk Ser. Fiz.* **76** 52 (2012) [*Bull. Russ. Acad. Sci. Phys.* **76** 44 (2012)]
- Glezer A, Pozdnyakov V *Nanostruct. Mater.* **6** 767 (1995)
- Pozdnyakov V A, Glezer A M *Pis'ma Zh. Tekh. Fiz.* **21** (1) 31 (1995) [*Tech. Phys. Lett.* **21** 13 (1995)]
- Glezer A M *Deformatsiya Razrushenie Materialov* (2) 1 (2010)
- Pozdnyakov V A, Glezer A M *Fiz. Tverd. Tela* **47** 793 (2005) [*Phys. Solid State* **47** 817 (2005)]
- Andrievski R A, Glezer A M *Usp. Fiz. Nauk* **179** 337 (2009) [*Phys. Usp.* **52** 315 (2009)]
- Umanskii Ya S, Finkel'shtein B N, Blanter M E *Fizicheskie Osnovy Metallovedeniya* (Physical Principles of Metals Science) (Moscow: Metallurgizdat, 1949)
- Valiev R Z et al. *JOM* **58** (4) 33 (2006)
- Amosova N N et al. *Bol'shoi Anglo-Russkii Slovar'* (Great English-Russian Dictionary) Vol. 2 (Moscow: Russkii Yazyk, 1988) p. 427
- Glezer A M *Izv. Ross. Akad. Nauk Ser. Fiz.* **71** 1764 (2007) [*Bull. Russ. Acad. Sci. Phys.* **71** 1722 (2007)]
- Golovin Yu I *Universal'nye Printsipy Estestvoznaniya* (Universal Principles of Natural Sciences) (Tambov: Izd. Tambov. Gos. Univ., 2002)
- Shtremel' M A *Prochnost' Splavov* (Strength of Alloys) Vol. 2 (Moscow: MISiS, 1997)
- Valiev R Z *Russ. Nanotekhnol.* **1** (1–2) 208 (2006)
- Glezer A M, Gromov V E *Nanomaterialy, Sozdannye Putem Ekstremal'nykh Vozdeistvii* (Nanomaterials Created by Applying Extreme Action) (Novokuznetsk: Inter-Kuzbass, 2010)
- Glezer A M et al. *Izv. Ross. Akad. Nauk Ser. Fiz.* **69** 1350 (2005)
- Rybin V V *Bol'shie Plasticheskie Deformatsii i Razrushenie Metallov* (Large Plastic Deformations and Destruction of Metals) (Moscow: Metallurgiya, 1986)
- Firstov S A et al. *Izv. Vyssh. Uchebn. Zaved. Fiz.* (3) 41 (2002) [*Russ. Phys. J.* **45** 251 (2002)]

43. Gapontsev V L, Kondrat'ev V V *Dokl. Ross. Akad. Nauk* **385** 608 (2002) [*Dokl. Phys.* **47** 576 (2002)]
44. Glezer A M, Pozdnyakov V A *Deformatsiya Razrushenie Materialov* (4) 9 (2005)
45. Glezer A M, Metlov L S *Fiz. Tverd. Tela* **52** 1090 (2010) [*Phys. Solid State* **52** 1162 (2010)]
46. Likhachev V A et al. *Fiz. Met. Metalloved.* **45** (1) 163 (1978)
47. Glezer A M, Sundeev R V, Shalimova A V *Dokl. Ross. Akad. Nauk* **440** 39 (2011) [*Dokl. Phys.* **56** 476 (2011)]
48. Valiev R Z, Aleksandrov I V *Nanostrukturnye Materialy, Poluchennye Intensivnoi Plasticheskoi Deformatsiei* (Nanostructural Materials Produced Using Severe Plastic Deformation) (Moscow: Logos, 2000)
49. Blinova E N et al. *Materialoved.* (12) 3 (2007)
50. Startsev V I, Il'ichev V Ya, Pustovalov V V *Plastichnost' i Prochnost' Metallov i Spлавov pri Nizkikh Temperaturakh* (Plasticity and Strength of Metals and Alloys at Low Temperatures) (Moscow: Metallurgiya, 1975)
51. Glezer A M et al. *Deformatsiya Razrushenie Materialov* (10) 23 (2011)
52. Glezer A M et al. *Izv. Ross. Akad. Nauk Ser. Fiz.* **73** 1310 (2009) [*Bull. Russ. Acad. Sci.* **73** 1240 (2009)]
53. Glezer A M et al., Patent RF No. 2391414 (2008)
54. Glezer A M et al. *Deformatsiya Razrushenie Materialov* (11) 1 (2011)