PHILOSOPHICAL TRANSACTIONS OF THE ROYAL SOCIETY A
MATHEMATICAL, PHYSICAL & ENGINEERING SCIENCES

Discussion

G. C. Smith, T. Gladman, J. H. Van Der Veen, W. B. Morrison, R. G. Baker, S. F. Pugh, R. Dolby, F. M. Burdekin, D. B. Clay, A. P. Miodownik, W. L. Mercer, C. Edeleanu, G. G. Saunders, W. J. Murphy, K. Sachs, M. Cohen, J. E. Roberts, L. M. Wyatt, J. Savas, J. Barford, W. G. Ferguson, T. N. Baker, R. McCallum and J. A. Chapman

Phil. Trans. R. Soc. Lond. A 1976 **282**, 346-357
doi: 10.1098/rsta.1976.0055

3.5 A forward look at some steel developments based on physical metallurgy

By M. Cohen and W. S. Owen
Department of Materials Science and Engineering, Cambridge, Massachusetts, U.S.A.

It is projected that research on the physical metallurgy of steel during the next decade will be directed towards translating the impressive gains in our understanding of materials science into materials engineering. This will require the physical metallurgist to have an awareness of the economic and social environments in which the steel industry operates, the nature and special problems of technology transfer in the industry and the interdependencies of physical, chemical and process metallurgy. These points are illustrated by brief discussions of some areas of materials science of direct interest in the physical metallurgy of steel, some expected developments in low-alloy and in high-strength steels, and by a short review of a few of the more important problems awaiting practical solutions.

THE RESEARCH ENVIRONMENT

It has been pointed out (Old 1972) that the common practice of reporting a company's research and development (R and D) expenditures as a ratio of net sales can be misleading. Instead two ratios are suggested: one, R and D expense divided by cash flow, and the other, technology-based capital investment less tax credit divided by cash flow. Typically, on this basis, a U.S. steel company might have an R and D investment ratio of 0.03 and a technology capital investment ratio greater than 0.60. Thus, about 65% of the total cash flow is invested in technology. This is another way of saying that the steel industry is both technology and capital intensive. For a company's R and D programme to be cost-effective, it must develop technology in which it will prove profitable to invest at some not too distant time. The R and D investment ratio is so much smaller than the capital investment ratio that it is inevitable that much of the research done in a company's R and D programme will not contribute directly to a new or modified technology which will attract capital investment. In this sense, for many years research was considered to be relatively cheap and few serious attempts to assess its cost effectiveness or to direct it into profitable channels were made. In another sense, research of this kind was considered by many steel companies to be a luxury so that cuts in research budgets occurred early in periods when economies in operating costs were considered necessary.

In recent years there has been a substantial change in outlook. Basic research is no longer relatively cheap and it is realized that in many cases it is more profitable to buy, at home or abroad, the technology in which large investments are to be made. The communication of technical information within the steel industry, nationally and internationally, is remarkably unrestricted (Hiestand 1974). In the United States this situation dates back to 1922 when, in settlement of a major antitrust prosecution, the U.S. Steel Corporation agreed to make technological information available to its competition to help them expand. The selection of a technology in which a company decides to invest is influenced greatly by the local conditions: its geographical location in relation to ore and fuel sources, the nature of the products required by its market, the technology in which capital is already invested, the local environmental constraints, etc. Thus there is a need, which is being recognized increasingly by the U.S. steel industry, to direct its research towards those problems of major importance to the technology investment programme of the individual company.

Of course, some research on steel is done by industries which are large users of steel. The results of such work are used either to influence the technological development programmes of the steel companies wishing to supply the product, or to change or improve their own fabrication or heat-treatment processes. Usually, the technology capital investment ratio is not so large in the latter case as it is in the primary steel industries because the equipment is on a smaller scale. Consequently, the rate of technology transfer out of the laboratory onto the production line is more rapid. Virtually all of the steelmaking technology currently in use throughout the world is based on physical-chemical metallurgy that is at least a quarter of a century old. In contrast, heat-treatment processes in automobile plants, for example, are changing rapidly. In some cases the technology transfer interval may be as short as three to five years.

In materials research generally, the years between 1946 and 1970 were a period in which the potency of the structure-property approach of the new materials science was realized and exploited. The expansion of our knowledge of structure at all levels, our techniques for studying structures, and our understanding of the often subtle interplay between properties and structure were explosive. It also coincided with the period in which basic research, whether supported financially by industry or federal agencies, was relatively undirected. Physical metallurgists working on steel were swept up by the new science. Research in physical metallurgy done during this period was enormously productive from a scientific viewpoint. However, as the subject developed, reaching increasingly towards materials science, the couplings with chemical and process metallurgy and with engineering design were weakened. By the start of the 1970s it became clear that this trend was reversing. The economic and technological considerations in the steel industry which forced this change have been referred to earlier. There was an additional reason. The rate of development of materials science was so rapid in the postwar years that useful, although not always completely satisfying, solutions to most, but not all, the scientific problems relevant to steel technology were found by 1970. Many fascinating and difficult problems remain for the next generation of materials scientists but it cannot be claimed, at least as far as the steel industry is concerned, that solutions to those problems are quite so crucial to the development of steel technology as was the case a few decades ago. The need now and in the next decade is to develop the engineering of materials and, in so doing, to utilize the many important insights into structure-property relationships which have been acquired during the extraordinary growth of the science of materials.

At present, most research on the physical metallurgy of steel, and especially that carried out by steel-producing companies, has limited and circumscribed objectives. It is concerned with improving the yield strength or the fracture toughness of hot-worked products, the formability of cold-rolled sheet, the corrosion-fatigue resistance of stainless steel, the high-temperature creep resistance of highly alloyed steel forgings, etc. The development sector of the R and D effort is concerned with the utilization of the findings of the research and the development of a new or improved product for which, it is believed, there is a market or for which a market can be created. In this total process, two-way interactions between the materials engineers and the potential users are of crucial importance. It seems probable that this interaction will increase still further in the future. Users will have a greater influence on research planning, even in those sectors involving a large tonnage of steel in situations where there is no alternative material. This is likely to be the future situation in the construction industry, including the construction of offshore oil rigs, power plants, ships, pipelines and similar structures. Of course, other research and development carried out by steel companies is aimed at improving the cost-effectiveness of process operations within the company and, consequently, usually does not require extensive interaction with user industries. To an increasing extent, effective work on problems in this area requires the cooperative efforts of physical, chemical and process metallurgists.

In the industries producing materials, the future research directions likely to be most rewarding are often difficult to identify. The development of high-strength, low-alloy (h.s.l.a.) steels provides an example. Over the last decade or more, much imaginative and painstaking research has led to remarkable improvements in the properties of these steels (Korchynsky & Stuart 1970). Hot-rolled ferritic steels with a yield strength of 550 N/mm^2 (80000 lbf/in^2) coupled with good ductility are now available and it is probable that similar steels with strengths in excess of 700 N/mm^2 (100000 lbf/in^2) will be developed commercially in the near future. By reducing the carbon and sulphur levels and applying the results of modern research on sulphide-particle-shape control, it is possible to design these steels to have good weldability, formability and directionality of properties combined with acceptable corrosion and fatigue resistance. To take advantage of the higher available strength and to reduce the cost, it is necessary for the designer of engineering structures to achieve the required function with substantially less steel than would be needed had older conventional steels been used. This inevitably leads to difficulties with the rigidity of the structure which, in some cases, necessitates radical changes in the design. It might be argued that in these circumstances, future research on h.s.l.a. steels should be directed towards improving the rigidity of the steel itself (Kiessling 1974). In many designs, a 10% increase in the shear modulus would greatly ease the designers difficulties. However, because the elastic moduli are not structure-sensitive properties, one may question whether even this modest goal can be achieved.

A careful analysis of the total system may reveal that the h.s.l.a. steels already available cover a range of strength greater than can be used economically in large structures in which rigidity is important. If so, it would be more profitable to concentrate future research on achieving the desired combination of properties by a different and cheaper route. This might involve the utilization of a different 'structure', but is more likely to require consideration of the total processing of the steel. Microalloyed h.s.l.a. steels with the shape of the sulphide particles controlled by the addition of rare earths (cerium) or, in the case of niobium-treated steel, the addition of zirconium, must be fully killed. For many plants with a large financial investment in ingot technology, these processes are prohibitively expensive. If comparable strengths, impact shelf-energies (resistance to ductile fracture) and directional properties could be achieved in semi-killed or rimming steels without significant increases in the cost of the processing, the benefit to the construction industry could be much greater than further improvements in the properties of killed steels. Such research is likely to be concerned at least as much with the chemistry of the steelmaking process and the processing history as with the structure-property relationships of conventional physical metallurgy. The metallurgists engaged in such work would be called upon to apply a much wider range of knowledge and skill than has been the case in the recent past. Recognizing this trend, at M.I.T. we have recombined the graduate programmes in chemical, process and physical metallurgy after more than twenty years of separation.

From these considerations we conclude that even in relation to those industries which are large-scale users of steel, future directions will be dictated by the conclusions drawn from analyses of total engineering systems; analyses which will incorporate economic and environmental factors as well as considerations of hardware design and materials technology.

This conclusion is more easily recognized when the materials needs of product-oriented industries are considered: industries which use special purpose steels in combination with a variety of other metallic and non-metallic materials. The automobile industry is a good example. No longer does the automobile designer expect to find an optimum combination of properties in a single material. The function of each component in the total system is achieved by a combination, often a composite, of materials. The cost of the component is a sensitive function of the way the component is made, as well as of the cost of the basic materials. In these circumstances, steel, alone or in a composite, must compete with the whole range of engineering materials. Frequently, there is no unique solution to a particular design problem. The choice of materials is then dictated by considerations of quality control, reliability and, of course, primarily, cost. The solution adopted is very likely to vary from company to company or even from one location to another within the same company.

It is now becoming evident that three of the factors which have profound influences on the materials systems approach to functional design are the national materials policy, the content of the materials technology bank, and the rate at which stored technology can be translated into practical applications. (See, for example, the reports of the Committee on the Survey of Materials Science and Engineering, National Academy of Sciences 1974, and of the National Commission on Materials Policy 1973.) To an increasing extent, national policies are deciding the relative cost and availability of steel and the many materials with which it competes. Obvious examples include the impact that recent environmental protection regulations have had on blast furnace, open-hearth and other steelmaking processes, and the dramatic shifts in the relative costs of materials produced by the unexpected and urgent need to conserve expensive energy. The increase in the cost of polymeric materials resulting from the much increased cost of petrochemicals is likely to have an important effect on the relative attractiveness of metals, including steel, and polymers in many materials systems.

Some effects of national materials policies are less obvious. Changes in the United States policy on stockpiling of materials have resulted in significant changes in the availability of some materials. To a larger extent than is usually appreciated, national policy dictates the general problem areas to which funds made available by the federal government are applied. In materials, there has been a major redeployment of research effort to attack numerous materials problems considered to be important in relation to the development of new or improved sources of energy. One surprising consequence of this has been the re-establishment of a high priority for research on high-strength and high-alloy special steels for use in power-generation machinery.

In the past, the bank of available knowledge about achievable materials properties and viable materials process technology has been filled in an unorganized fashion. Many people who worked on the physical metallurgy which became the basis for the development of ausformed steels have been disappointed to discover that these steels have found only limited application in practice. Ausformed steels were the result of research and development motivated primarily by the desire to exploit as fully as possible the then current research trends. When considered in a total materials-system design, ausformed steels were seldom selected because of the poor stress-corrosion resistance, difficulties with processing and forming, and limited range of shapes and thicknesses in which good mechanical properties can be achieved. The knowledge and expertise resulting from the substantial research and development effort devoted to ausformed steels a few years ago has become a part of the existing materials technology bank. Whether or not at some future time ausformed steels will provide the solution to as yet unidentified materials design problems is difficult to predict.

It is becoming increasingly clear that materials research in the United States should and will be more strongly focused on specified goals than has been the case in the past. Producer industries will be influenced by the results of their materials-systems analyses in deciding what materials technologies need to be added to the technology bank. The expenditure of federal funds will be much influenced by the developing national materials policy. A possible future development connected with materials policy which might be of special importance and interest to the steel industry concerns manganese. The United States steel industry is almost completely dependent on foreign supplies of ferromanganese and for this reason there has been a great deal of interest in the possibility of extracting manganese from the nodules of metallic oxides in great abundance in some areas of the deep oceans. Substantial resources have already been expended on developing the technologies necessary to mine these deposits and to extract from them cobalt, nickel and copper (Anon 1972, 1973, 1974; Rothstein & Kaufman 1973). To extract the manganese from the oxide waste is probably relatively simple. One econometric study has concluded that if the extraction of manganese from this source were put into large-scale production, by 1980 manganese could become the next cheapest metal after iron. Of course, to warrant large-scale production, adequate markets must be available. It is clear that at present there is available insufficient understanding of the technological consequences of substituting manganese or manganese oxide for ferromanganese in the steelmaking process. Nor is enough known about the physical metallurgy of iron-manganese and manganese-base alloys to indicate whether or not useful and competitive properties can be developed in such alloys, thus enlarging the market for manganese products. If the federal government decides, as a part of the national materials policy, to encourage the production of manganese from oceanic deposits, the use of research resources to explore the chemical and physical metallurgy of alloys containing manganese follows logically. Such a decision would have an important influence on the future trends of research on steel.

From this brief review of the context in which materials research is expected to develop, it is obvious that any attempt to predict future trends by extrapolation from current interests and activities is very likely to be misleading. However, in the present circumstances, we see no alternative course of action available to us.

The application to engineering

Since the Second World War the physical metallurgy of steel, as of other metals, has been dominated by the study of structure-property relationships. The advances in our understanding of the science of many basic phenomena have been spectacular. Significant insights have been obtained into work-hardening, age-hardening, solid-solution hardening, plastic flow and fracture, to mention only a few phenomena. To apply this basic physical metallurgy to practical problems often requires a connexion to engineering which is difficult to make.

Much careful work on ferrite-pearlite annealed or normalized steels has led to the establishment of reliable quantitative relationships between the yield strength and such structural parameters as grain size, solid-solution hardening, volume fraction of pearlite, and pearlite spacing (Baird & Preston 1972). The quantitative effect of precipitation hardening by carbide particles has not been established by direct methods, but, qualitatively, the influence on strength and ductility is well understood. This and parallel work on the control of the size, shape and distribution of sulphide particles and other inclusions provided the scientific basis on which modern h.s.l.a. steels were developed. Similarly, the relative potency of the various structural strengthening mechanisms in quenched and tempered steels have been determined (Cohen 1963; Owen 1970), although the integrated effect produced by two or more of these mechanisms is not yet understood quantitatively.

The prediction of the fracture toughness of a steel from microstructural parameters is much more uncertain. The problem is compounded if the steel is to operate in a corrosive environment. For a steel, tested under non-corrosive conditions, with a yield strength greater than about 1100 N/mm^2 (160 klbf/in^2) or sufficiently thick to ensure that the specimen fractures under plane-strain conditions at a lower strength level, the fracture toughness parameter, K_{Ic}, as determined by direct application of linear fracture mechanics, is a satisfactory measure of the fundamental resistance of the steel to the propagation of a crack. This measure of fracture toughness is widely used in successful designs based upon maximum-permissible flaw concepts (Pellini 1968). However, even in these circumstances, knowledge of quantitative relationships between the fracture toughness and microstructural parameters is lacking. Attempts to obtain some general qualitative understanding of possible relationships through intermediary simple mechanical parameters such as the strain-rate dependence of the flow stress or the strain-hardening index have met with only modest success (Hahn & Rosenfield 1966). About all that is known with any confidence is that structural changes which increase the strain-hardening index are likely to improve the K_{Ic} value. This information is of limited value as a guide to the development of ductile steels because the number of different strain-hardening mechanisms, and consequently the number of microstructures in which they operate, is small.

When a steel has a yield stress of less than 1100 N/mm^2 (160 klbf/in^2) and is thinner than the critical thickness, no satisfactory measure of the fracture toughness is known, and consequently, there is no understanding of the influences of changes in the structure. This pair of coupled problems was discussed extensively at the recent Kyoto Conference. Liebowitz (1974) was of the opinion that advances in our understanding of the mechanics of fracture offer some hope that the former problem may be solved by developments of the J-integral concept or in some other way, but there appears to be no reason to believe that solutions to the structure-property problems are within sight. Some structural-hardening mechanisms which, until now, have received little attention will be mentioned later. There is, as yet, no indication of how useful they might be. Stress-corrosion failure cannot be prevented by the control of structural parameters. It is probable that, at least in the near future, the use of high-strength steels in even mildly corrosive atmospheres will be limited by the development of effective protective surface coatings (probably polymers), but the reliability of these solutions to the problem is in doubt at the present time.

The concepts of linear fracture mechanics have been extended, with as yet limited success, to explain fatigue behaviour of steels. The problem is perhaps more complex than that of static brittle behaviour but, undoubtedly, solutions to both phenomena will be found within the same theoretical framework.

High-temperature deformation (creep) is reasonably well understood. Recently, there have been substantial advances in our understanding of thermally activated deformation processes at high temperatures. Future development of high-temperature creep resistant materials is unlikely to be retarded by any lack of understanding of the basic phenomena.

Some developments possible in the future

Low- and medium-strength steels

The use of carbides precipitated in the parent austenite during controlled hot rolling to retard recrystallization of the austenite has been a key factor in the development of fine-grained h.s.l.a. ferritic steels. To date, this process has been developed empirically. It is possible that if the kinetics of nucleation and growth of the various carbides and nitrides were understood in terms of the temperature and rate of deformation and the chemistry of the austenite, ferrite grain sizes substantially smaller than those currently attainable could be obtained in commercial practice. The physical metallurgy of the problem is complex and difficult to study experimentally. However, the results obtained in a number of recent studies using either hot-torsion testing or rapidly quenched specimens coupled with systematic optical and electron microscopy indicated that experimental studies are possible and rewarding (Hansen 1975). The lower limit to the ferrite grain size seems to be a few micrometres. The Hall–Petch relation between strength and grain size holds even at this extreme of the range (Baird & Preston 1972), predicting that, in principle, steels with a yield strength approaching the theoretical strength could be obtained by this means. Grain refinement has an important advantage over other strengthening mechanisms; it enhances the ductility. Still finer ferrite spacing can be obtained by large cold deformation. In this range, however, the Hall–Petch relation is replaced by a linear relation between the strength and the reciprocal of the ferrite dimension (Langford & Cohen 1969). It is possible that similar cold-worked structures could be obtained with much smaller deformation by control of the precipitation of carbides or nitrides during the total thermomechanical treatment of suitably designed steels. Dislocation-interstitial-precipitate interactions during cold and warm working of steel in the ferritic condition is also a neglected topic. It has been known for many years that large increases in the strength of quenched steels can be obtained by dynamical strain-ageing. Unfortunately, this form of hardening is usually accomplished by a deterioration in the fracture resistance (blue-brittleness) and unstable plastic flow. In the last few years it has been recognized that dynamical strain-ageing of steels containing carbide or nitride particles is a phenomenon involving only freshly generated moving dislocations. It requires a flux of interstitial atoms from redissolving precipitate particles to the glissile dislocation (Kalish 1966; Roberts & Owen 1967). It is possible to induce dynamical strain-ageing in such a way that the steel is not excessively brittle and that unstable plastic flow occurs only at temperatures which are higher, and strain-rates which are faster, than those of practical interest. The importance of resolution of carbides and nitrides is clear, but it is not known whether or not significant differences in the dynamical strain-ageing behaviour can be obtained by using different precipitate compositions. In particular, the behaviour of the various precipitates possible in microalloyed steels have not been explored. It is possible that the exploitation of dynamical strain-ageing to develop a structure which contains an optimum dispersion of microalloy carbides or nitrides could lead to the development of steels with very attractive strength-ductility properties at temperatures and strain-rates below the lower limit of the blue-brittleness range.

Most of the h.s.l.a. steels developed to date have a ferritic substructure typical of hot-rolled steel. This dislocation substructure provides some, but probably a small, contribution to the strengthening, but it is an important source of heterogeneous nucleation sites for the precipitation processes. Recently, there has been much interest in utilizing the denser dislocation substructure developed in steels with an 'acicular-ferrite' structure obtained by suitable control of the chemistry and the cooling rates of microalloyed steels. This structure is clearly related to the bainitic structure which can be developed in low-carbon alloy steels and to the 'lathe' structure found in low-carbon and low-alloy martensites. In spite of much attention, this family of structures is not well understood. It is generally assumed that a transformation involving a lattice-shape-change (martensite transformation) is an essential step in the development, but it is unclear whether or not this shape change is the same in all cases. The relationship of the dislocation substructure to the morphology of the transformation product is also uncertain (Owen 1970). The substructure of acicular-ferrite is closely related to that produced in slowly cooled high nickel (14–18% nickel) steels of the maraging type. In low-alloy or microalloy steels accelerated cooling is required, but the chemistry can be adjusted to ensure that the cooling rates which are necessary are within the range attainable economically. The development of a high dislocation-density substructure by a transformation requiring only practical accelerated-cooling instead of cold working involving significant deformations is attractive commercially. In the future, undoubtedly there will be much study of these transformations and of the properties which can be developed in steels with this class of substructure.

Solid-solution strengthening by carbon and nitrogen and precipitation hardening by carbide or nitride particles are the hardening mechanisms operative in nearly all ferritic steels. Jack (Driver, Hanley & Jack 1972) has demonstrated the existence of another mechanism in ternary alloys of a composition in which aggregations of, for example, manganese and nitrogen atoms can occur by a transformation which has some of the features of a spinodal decomposition. In some steels these aggregates, which can form in either austenite or ferrite, are similar to g.p. zones in aluminium alloys, but in other steels they appear to be very small coherent precipitates. Transformations of this kind increase the hardness. Steels with this hardening mechanism have not yet been designed or produced in sizes large enough to allow fracture-toughness testing. Thus, it is not yet known whether or not the mechanism can be the basis for the development of a new class of structural steels.

It has long been recognized that if steels containing very little or no carbon and nitrogen could be designed and produced economically, they would have a number of advantages over the traditional carbon steels. Welding problems would be reduced and loss of ductility due to the segregation of massive carbides at grain boundaries would be avoided. Work at Berkeley (McMahon & Thomas 1973) has shown that brittleness due to interphase carbides in Fe–Cr–C steels can be eliminated by a simple heat treatment. Steels used at temperatures at which creep is significant deteriorate with time because of the instability of carbide or nitride particles. At these temperatures, carbides and nitrides coarsen rapidly primarily because of the rapid diffusion rates of the interstitial atoms in ferrite. Thus, the idea of using a precipitate containing only elements which in the iron lattice dissolve substitutionally is attractive. The only carbon-free structural steel which utilizes precipitation strengthening and is in large-scale commercial use is maraging steel in which the effective precipitates are of the $Ni_3(Al, Ti)$ type. This steel, used for its room-temperature and low-temperature properties, is not attractive as a creep resistant material. Decker (1969) has summarized the results of the few known attempts to use intermetallic compounds in creep-resistant steels. He concluded that if Laves phases are present in significant concentration they have little effect on creep properties but they increase room-temperature brittleness. Similar results were obtained by Mihalisin, Bieber & Grant (1968) who investigated the creep and fracture properties of a steel containing sigma phase. These undesirable properties seem to be a result of the formation of grain-boundary films of either Laves or sigma phase. Jones, Parker & Zackay (1971) showed how a continuous network of Laves phase at the ferrite grain boundary in an Fe–Ta–Cr alloy can be dispersed and caused to spheroidize by a simple heat treatment. The properties of the heat treated alloy are not unusual, but the brittleness usually associated with Laves or sigma phases is removed (Zackay, Parker & Bhandarkar 1972). Clearly, much work along these lines remains to be done. It is possible that by proper design of alloy composition and thermomechanical treatments, high-strength alloys with good creep stability can be developed from carbon-free iron alloys containing only small concentrations of elements such as tantalum. Since stainless steels with chromium as an essential alloying element are particularly susceptible to the formation of sigma phase, it might well be possible to turn this difficulty to an advantage. Developments along these lines in the future can be expected to lead to many interesting and possibly useful steels.

High-strength steels

Although detailed understanding of the relationships between structure and fracture toughness is lacking, it has been evident for some years that in order to obtain a steel with high strength combined with fracture toughness a high work-hardening rate is required. Hence, a steel with a high dislocation density and a dispersion of fine precipitate particles is required. By thermomechanical treatment of steels of suitable composition excellent combinations of strength and toughness can be obtained. The practical limit of strength for a steel with a yield stress/K_{Ic} ratio of 1.0 $\mathrm{in}^{-\frac{1}{2}}$ appears to be about 1375 $\mathrm{N/mm^2}$ (200 $\mathrm{klbf/in^2}$). There are, of course, many drawbacks to this approach to high strength, not the least of which is the necessity to reduce the cross-section by more than 50% during the thermomechanical treatment.

A major advance in the technology of high-strength steels is likely to occur only when ways are found of utilizing different and more effective work-hardening mechanisms. Available mechanisms are not numerous. Three interrelated mechanisms deserve consideration: hardening by the formation of strain-induced martensite, Hadfield hardening, and hardening by the formation of stacking faults. The first of these has been utilized in t.r.i.p. steels. These are designed to have a microstructure after thermomechanical treatments which is predominantly martensitic but which contains a significant volume-fraction of retained austenite of a composition such that with further deformation more martensite forms by strain-induced nucleation. Large heats of t.r.i.p. steels have been made and the excellent strength and fracture-resistant properties under normal conditions of temperature and strain rate have been confirmed. Unfortunately, the fracture properties are very sensitive to strain rate and temperature. At room temperature and high strain rates, the temperature in the small plastic zone ahead of a running crack rises above the M_{d} temperature and the formation of strain-induced martensite is inhibited. The steel then loses toughness. To produce t.r.i.p. steels with a wider tolerance for extremes of temperature and strain rate, it is necessary to design the steel composition such that the M_{s} temperature is near room temperature and the M_{d} temperature is as high as possible. The recent ideas of Olson (1974), utilizing the strain-induced martensite hardening mechanism, provide a rational scientific basis for the development of such steels. The Olson model of nucleation assumes that the first step in martensite nucleation is faulting on planes of closest packing. The stacking-fault energy is shown to consist of both volume-energy and surface-energy contributions. When the volume-energy contribution is negative, the fault energy decreases with increasing fault thickness. This condition leads to the spontaneous formation for a martensitic embryo. The model has been elaborated to explain both thermal and strain-induced nucleation and provides clear indications of the relative importance of the various factors which affect these two types of martensite formation. From these ideas, a first attempt to design a t.r.i.p. steel which is relatively insensitive to strain-rate effects has been made. The project is not sufficiently advanced to allow definite conclusions to be drawn at this time.

The strain-hardening that occurs in Hadfield austenitic steels is unusually large and a unique phenomenon. It is much larger than in steels with the same manganese content (12%) but lower carbon. Evidently, the low stacking-fault energy in the austenite and the unusually high concentration of carbon (1.2%) are together responsible for the effect. Hadfield steels are often quoted as an example of the toughness and strength which can be developed by the production of strain-induced martensite during deformation but, in fact, no new phase is formed martensitically during the deformation of Hadfield steel (White & Honeycombe 1962). The mechanism deserves serious study. High-strength, high-ductility steels produced by this mechanism will be very attractive economically if manganese becomes a cheap metal.

Finally, mention should be made of the possibility of utilizing the relatively high work-hardening rate of a low stacking-fault energy austenite. Information on this effect is sparse, but there are sufficient data to suggest that the effect might be useful. There is a suspicion that metals with a low stacking-fault energy are particularly susceptible to brittle fracture and that they have a low fatigue limit. Unfortunately, most of the tests which have been done to date have used material which was insufficiently pure to exclude the possibility of extraneous effects, such as grain-boundary carbide formation, invalidating the conclusions.

Priorities

The projected developments discussed here are extensions of topics currently of interest. They do not represent the most important research topics as predicted by experts looking at the total materials needs. In fact, in the COSMAT (1974) study of research priorities, physical metallurgy of ferrous metals and alloys received a low overall research priority rating, well below that of ceramics, glasses, polymers and composites. However, some phenomena which are important in steel metallurgy were recommended for urgent attention. These included corrosion, oxidation and other surface phenomena. Undoubtedly, these will become the focus of much sophisticated research effort. The techniques essential to rapid advances are available: l.e.e.d., h.e.e.d., Auger spectroscopy, and e.s.k.a.

We have emphasized that, in the next decade, attention will most probably be given to translating the lessons learned by the study of structure–property relationships into engineering practice. Much of the research will move in this direction also. The major areas of research are likely to be: reliability and reproducibility of properties; performance of materials as assessed over their lifetimes; and problems associated with inspecting and testing materials for specific uses. Design is likely to move more and more toward the adoption of critical flaw-size concepts. This means that much research effort will have to be devoted to studies of methods of non-destructive testing, particularly to the development of reliable methods for measuring spatial flaw-size distribution. Indeed, it appears that a substantial portion of the research thrust at steel companies will be directed to this end in the next five or ten years.

References (Cohen & Owen)

Anon 1972 *Ocean Industry*, July 1972, p. 58.

Anon 1973 *Kennecott Copper Company Annual Report*, 1973.

Anon 1974 *Chemical Engineering* 31, May 1974 p. 34.

Baird, J. D. & Preston, R. R. 1972 *Proc. Int. Symp. on Processing and Properties of Low Carbon Steels*, MD-AIME 1972.

Cohen, M. 1963 *J. Iron Steel Inst.* **201**, 833.

Committee on the Survey of Materials Science and Engineering (COSMAT) 1974 *Materials and man's needs.* Washington, D.C.: National Academy of Sciences.

Decker, R. F. 1969 Steel strengthening mechanisms. *Symposium Climax Molybdenum Company*, p. 147.

Driver, J. H., Hanley, J. R. & Jack, K. H. 1972 *Scan. J. Met.* **1**, 211.

Hahn, G. T. & Rosenfield, A. R. 1966 *Mechanics and metallurgy of brittle crack extension.* ASCE Mechanics Division, Washington DC, 1966.

Hansen, S. 1975 Structure–property relations in two high-strength, low-alloy steels. M.S. thesis M.I.T.

Hiestand, D. L. 1974 *High-level manpower and technological change in the steel industry* p. 43. N.Y.: Praeger.

Jack K. H. 1972 *Scan. J. Met.*, **1**, 195.

Jones R. H., Parker, E. R. & Zackay, V. F. 1971 *Proceedings Fifth International Materials Symposium.* Berkeley, CA. 1971, 829.

Kalish, D. 1966 The effect of strain tempering on strength and toughness. Sc.D. thesis, M.I.T.

Kiessling, R. 1974 Will steel appeal in A.D. 2000? *Metals & mater.*, February 1974.

Korchynsky, M. & Stuart H. 1970 *Symposium on low-alloy high-strength steels.* The Metallurgy Companies, Nuremberg, p. 17.

Langford, G. & Cohen, M. 1969 *Trans. Am. Soc. Metals* **62**, p. 623.

Liebowitz, H. 1974 *Proc.* 1974 *Symp. on the Mechanical Behavior of Materials.* Society of Materials Science, Japan.

McMahon, J. A. & Thomas, G. 1973 *Proc. Third Int. Conf. of the Strength of Metals and Alloys* **1** 20–25.

Mihalisin, J. R., Beiber, C. G. & Grant R. T. 1968 *Trans. A.I.M.E.*, **242**, 1968, 2399.

National Commission on Materials Policy 1973 *Materials needs and the environment.* Washington, D.C.: U.S. Government Printing Office.

Old, B. S. 1972 R and D investment: sharpening the focus. Management Notes, A. D. Little, Inc.

Olsen, G. 1974 A general mechanism of martensitic nucleation. Sc.D. thesis, M.I.T., 1974.

Owen, W. S. 1970 *Proc. Second Int. Conf. on the Strength of Metals and Alloys*, ASM, p. 198.

Pellini, W. S. 1968 *Advances in fracture toughness characterization procedures and in quantitative interpretations to fracture-safe design for structural steels* Naval Research Laboratory Report 6713, 3 April, p. 44.

Roberts, M. J. & Owen, W. S. 1967 The strength of martensitic iron–nickel alloys. *Trans. Am. Soc. Metals*, **60**, 687.

Rothstein, A. J. & Kaufman, R. 1973 *Proc. Offshore Tech. Conf.* Paper 1758, p. 323.

Schultz, C. E. 1971 *Proc. Offshore Tech. Conf.* Paper 1364, p. 839.

White, C. H. & Honeycombe, R. W. K. 1962 *J. Iron Steel Inst.*, June 1962, 457.

Zackay, V. F., Parker, E. R. & Bhandarkar, D. 1972 *John Dorn Memorial Symposium.* AIME, Cleveland, OH, 1972.

KEYNOTE SPEECH

By K. J. Irvine (*British Steel Corporation*)

At the I.S.I. Centenary Meeting six years ago I was asked to review the developments in Physical Metallurgy during this period. It is a point of some interest, that it was only during the last few years of this period that much attention had been given to structural steels. However when work started rapid progress was made in understanding and then in exploiting the physical metallurgy of these steels. The work has continued since that time and now it is possible to present a very complete picture as illustrated by the technical papers in this Session.

The relations between microstructure and physical properties, so intensively exploited since the mid-1950s, have led to significant developments in the field of high strength low alloy steels. Among the early work dealing with the quantitative effects of microstructure on properties was that of Gensamer which described the effects of interlamellar spacing on the strength of pearlite. In the early 1950s, Hall reported the quantitative relation between yield strength and grain size in mild steel, leading to the development of the widely accepted Hall–Petch relation. Also in the 1950s, considerable attention was given to aspects of precipitation hardening, and