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ABSTRACT

This paper reviews needs and current research on materials for high field superconducting magnets. The review is written from an American perspective, but recognizes that much of the most important current research is done interactively in parallel programs in the United States and Japan. High field superconducting magnets are used principally for magnetic fusion energy devices and accelerators for research in particle physics. The advance of both technologies has increased the field, size and performance requirements of such magnets. More stringent requirements are expected in the future. The needed materials include both structural materials for the magnet case and support structure and high field superconducting wire for the magnet winding. Current research on both classes of materials is briefly described and discussed.

I. INTRODUCTION

Many of the most challenging needs for cryogenic materials relate to the design and construction of high field superconducting magnets that operate at temperatures of 4K or below. Three distinct types of materials are used in these magnets: superconducting wire to carry the current that creates the field, high strength steels to position and support the superconductor, and electrical insulators to separate the wires from one another. There are important problems with each of these material types, but the present paper is biased toward metals and will focus on materials needs and current research in structural alloys and high field superconducting wire.
The materials needs for high field superconducting magnets are set by the performance targets of the high field magnets that will be used in future systems. High field superconducting magnets are central to a number of new technologies. Those that have excited major research and development efforts within the United States include magnetic fusion devices for energy production, magnetic accelerators for research in high energy physics, and 'whole body' nuclear magnetic resonance (NMR) for medical diagnosis. The relevant materials research is primarily intended to support magnetic fusion and high energy physics, while medical NMR will likely create a major market for superconducting wire, both current and projected systems require relatively low-field magnets that can be built with existing materials [1].

II. SUPERCONDUCTING MAGNET SYSTEMS AND MATERIALS NEEDS

The current systems and materials needs for magnetic fusion and high energy physics can be briefly outlined as follows:

A. Magnetic Fusion Energy

1. Fusion Energy Devices

The United States Department of Energy currently supports research on two basic concepts for magnetic fusion energy: toroidal confinement and mirror confinement [2]. The two concepts differ in how the cylindrical plasma is bounded along its length. Both use a high magnetic field to heat the plasma by compressing it into a cylinder, and must employ superconducting magnets to create that field if magnetic fusion energy is to be technically feasible or commercially viable.

In the toroidal concept the plasma is closed on itself to form a torus. The best-known machine design is the 'tokamak', which is the basis for the Tokamak Fusion Test Reactor (TFTR) [3] in Princeton, New Jersey, the major test facility within the United States (Figure 1). The TFTR employs a magnetic field that is approximately 10 tesla peak field at the magnet conductor (~5T at the plasma center). However, the magnets are copper electromagnets. It was decided to develop the superconducting magnet systems for toroidal machines separately from the test device itself. This decision led to the Large Coil Project at the Oak Ridge National Laboratory [4] in which six large 8T superconducting magnets are being assembled for magnet system tests (Figure 2). The six magnets were chosen to represent different design and materials concepts. Three of the six large coils were constructed in the United States, two in Europe, and one in Japan. Five utilize NbTi superconducting wire; the sixth, built by Westinghouse in the United States, utilizes Nb₃Sn.

A next-generation tokamak machine is now in design [5]. The device is called the TFCX (Tokamak Fusion Core Experiment). The physics objective of the TFCX is to achieve plasma ignition and demonstrate the technical feasibility of magnetic fusion as a power source. While the magnetic field of the TFCX is not firmly set, the toroidal field coils that confine the plasma will have a peak field near 10 tesla at the conductor. Both normal and superconducting options are under investiga-
tion. The poloidal field coils that shape the field will operate in the 7-8 tesla range, and will be superconducting.

A companion toroidal fusion experiment is the the Alcator DCT at the Massachusetts Institute of Technology [6]. This device will employ 10T toroidal magnets, generating a 7T field at the plasma center, and 7T poloidal magnets. Both magnet sets will be superconducting. The initial design envisages Nb$_3$Sn toroidal coils and NbTi poloidal coils.

In the mirror fusion concept the plasma is a linear cylinder that is sealed at its ends by an arrangement of high field magnets. The major facility in the United States is the Mirror Fusion Test Facility (MFTF-B) [7], located at the Lawrence Livermore National Laboratory. Figure 3 is an artist's drawing of the MFTF-B, which is now under construction at LLNL. The plasma is contained in a central tube of solenoid magnets. The solenoid set is completed at either end by a combination of magnets, including two small, high-field axicell 'choke coils', two transition magnets to shape the plasma, and a large 'ying-yang' mirror coil set. Current plans call for a peak field of 2-3 tesla in the solenoid, 12-13 tesla in the small axicells, 5-6 tesla in the transition coils, and 7-8 tesla in the ying-yang coil set. All magnets will be superconducting. Two of the high-field axicells will be hybrid magnets with NbTi outer solenoids and Nb$_3$Sn inserts. The current design employs NbTi conductors in all other magnets.

The first of the large ying-yang coil sets for the MFTF-B was constructed as part of an earlier project. The magnet case for one of the magnets in the pair is shown in Figure 4 to illustrate its size. The second ying-yang coil set and the axicells have been wound. As of this writing (May, 1984) the winding of the axicells and transition cells is underway.

2. Materials Needs

The materials needs of the magnetic fusion energy program include both structural alloys for the magnet case and high-field superconducting wire for the magnet winding.

a. Structural Alloys

The increasingly stringent specifications for magnet structural alloys result from the increasing size and field of the magnets, both of which raise the mechanical load on the magnet structure. While the systems that are now under construction have been designed with existing alloys (Table I shows the structural alloys used in the MFTF-B), future designs will almost certainly require alloys that offer superior combinations of strength and toughness at 4K.

The precise requirements depend on the structural support scheme as well as on the detailed magnet design. Two alternate methods of support are now in use. The most common is an exoskeletal configuration in which the superconductor receives its primary support from the external magnet case. This design method was used for the MFTF and for several of the coils in the large coil project. The alternate method is to place the superconductor within a high-strength conduit that provides
internal support. The exoskeleton can then be a relatively low-strength construction of aluminum or stainless steel. This method allows the superconductor to be a loose winding that is 'force-cooled' by flowing helium (Figure 5). The 'force-cooled conductor' concept was used in the Westinghouse Nb₃Sn coil for the LCP, and is the current method of choice for the Alcator DCT.

Conventional austenitic stainless steels are used in the magnet cases of the existing magnets that depend on exoskeletal support. As shown in Figure 6, these alloys offer a combination of strength and toughness at 4K that easily meets the design requirements for the MFTF [8]. They can be welded with available techniques and filler metals so that they preserve the needed strength and toughness [9]. It is likely, however, that future fusion magnets will require superior materials. Figure 6 also shows the design specifications developed by the Japan Atomic Energy Research Institute (JAERI) for a future tokomak machine [10]. These specifications call for a strength-toughness combination that is significantly beyond the capability of the existing stainless steels. While no specifically comparable studies have been done in the United States, preliminary design studies suggest that the JAERI specifications are reasonable design targets. It follows that new structural alloys are needed for the magnet case.

The Nb₃Sn magnets that employ force-cooled conductors present unusual materials needs because of the way the conductor is manufactured [11]. The sheath material is wrapped around the wire and seam welded to make the conduit before the wire is heat-treated to form the Nb₃Sn phase. The conduit material must, therefore, have good strength and toughness at 4K after having been welded and exposed to temperatures in the range 700-750°C for times that may be as long as 120 hours. These requirements suggest the use of Fe- or Ni-based superalloys that are normally aged for long times in this temperature range. An Fe-based superalloy designated JBK-75, which is a modified version of conventional A286, was selected for the conduit case in the Westinghouse LCP coil. While this alloy appears successful for the LCP conductor, which was aged at 700°C for 30 hours, it is known to precipitate undesirable intermetallic phases if it is aged at higher temperatures or for much longer times, and is hence questionable for use in advanced systems such as the Alcator DCT [12]. Also, its coefficient of thermal expansion is undesirably high and strains the superconductor on cooling to 4K. It follows that new structural alloys are needed for force-cooled conductor conduits.

b. Superconductors

The challenging superconductor materials needs for magnetic fusion energy are due to the high fields required in many of the magnets, the high structural loads imposed, and the radiation environment in which the superconductor must operate.

The high fields that are needed for plasma confinement, 10-13T for the toroidal field coils in a tokomak, and 18-24T for the choke coils in a mirror machine, now appear to require the use of alloyed Nb₃Sn and establish a need for new superconductors with higher field capability to minimize or remove the need for a normal copper insert at fields above
Moreover, the high structural loads can impose a strain on the superconductor that decreases its critical current. Design or materials modifications that minimize strain or maximize strain tolerance are desirable.

Finally, \( \text{Nb}_3\text{Sn} \) superconductors are susceptible to radiation damage [13]. High-energy particles disorder the A15 phase and degrade the critical current. Since it is impractical from a cost perspective to shield the superconductor completely, some radiation must be accepted. The importance of this problem is suggested by the results of conceptual design studies on possible upgrades of the MFTF-B axicell that peg the life of the coil to its cumulative radiation dose [14]. New or modified materials are needed that offer improved radiation resistance.

In assessing the magnet materials needs of fusion systems it must be remembered that these devices are ultimately intended to produce electric power in competition with fossil and fission devices. The materials contribution to system cost and reliability is an increasingly important issue. The materials used in fusion reactors must be inexpensive as well as functional and reliable.

B. High Energy Accelerators

1. Accelerator Systems

High-field magnets are also used to accelerate particles to high energies for experiments in fundamental physics. As the physics devices have grown in size and energy it has been necessary to use superconducting magnets to minimize power consumption. Particle accelerators require two basic types of magnets [15]: dipole magnets to deflect or bend the collimated beam of high energy particles and quadrupole magnets to focus the beam. An example accelerator dipole magnet is shown in Figure 7. These magnets are designed as long, rather thin cylinders that have a small bore path for the particle beam, and are, therefore, small compared to the solenoids or toroidal field coils in a fusion machine. On the other hand, a very large number of magnets are required to fill the ring of a modern accelerator. The most ambitious project completed to date in the United States was the Energy Doubler/Saver (ED/S) Project at the Fermilab accelerator in Batavia, Illinois, which required the construction of over 800 NbTi dipole magnets generating fields of \( \sim 4 \) tesla [16].

The major accelerator project in the United States is the Superconducting Super Collider (SSC) [17], a plan for a colliding beam accelerator that will produce particles with energies near 20 TeV (versus \( \sim 1 \) TeV for the Fermilab ED/S). Such an accelerating ring will necessarily have a large diameter compared to that of the current ring at Fermilab. Its size is a strong function of the magnetic field of the bending magnets employed. A preliminary design study projects a 20 km ring diameter for 8T bending magnets. The diameter expands to 24 km if 6.5T magnets are used, and grows to 52 km for 3T bending magnets. For comparison, the present Fermilab ring is approximately 2 km in diameter. The use of high field magnets will, therefore, decrease both the number of magnets required and the associated construction costs. The problem is that the cost of the magnet itself increases with the field, and tends to jump
discontinuously as the field moves from 3T (superferritic magnets) to ~6.5T (near the realistic peak field for NbTi dipoles) to 8T or greater (requiring Nb₃Sn). For this reason there are active design studies within the United States exploring the superferritic, NbTi and Nb₃Sn materials options to identify the most economic choice.

2. Materials Needs

The key materials issues in accelerator magnets center on the superconductor. The magnet structure must be designed with care but, given the relatively small size of the magnets, has not required research toward new structural alloys. The superconductor materials problems arise from the small bore of the magnets and the large number of magnets that are required for the system. The large number of magnets means that the cost per magnet must be kept low and the design must be sufficiently simple that they can be made reliably in large quantities. The small bore has the consequence that relatively high overall current densities are required to achieve a high magnetic field, and must be reached in wire that is tightly wound. Given these needs the relevant materials research is focused on increasing overall critical current density, consistent with reproducibility and low cost, in a wire that is either tolerant to winding strains (as NbTi) or can be used with process or design modifications to eliminate winding strain (the 'react-and-wind' [18] and 'dog-bone' [19] approaches to Nb₃Sn).

The superconductor of choice for future accelerators depends largely on how well these materials and manufacturing problems are solved.

III. Structural Alloys for High Field Superconducting Magnets

Research toward new alloys for high field superconducting magnets can usefully be divided into two categories on the basis of its principal goal: alloys for the external magnet case and alloys for the internal superconductor conduit. The alloys that are under development for the magnet case are high strength structural steels, including nitrided austenitic Fe-Ni-Cr and Fe-Mn-Cr alloys and ferritic Fe-Ni cryogenic steels. The alloys that are being considered for the force-cooled superconductor conduit are Fe-based and Ni-based superalloys whose heat treatments are compatible with those of the superconducting wire. Both sets of alloys will probably find other applications as well. For example, the Fe-based superalloy A286 is used for the magnet support structure forgings for MFTF-B, and conventional stainless steels have been used for the force-cooled conductor conduits of NbTi magnets that do not require heat treatment after fabrication.

Structural alloy research utilizes the known relations between the microstructure of an alloy and its mechanical properties to select or design suitable materials.

A. The Microstructure-Property Relations

The essential properties of structural alloys in the magnet case are strength and toughness; the alloys must not fail in service. Other properties are also important to optimize the performance of the magnet
1. The Strength-Toughness Relation

The structure-property relations governing the interplay between strength and toughness are shown in Figure 8. At moderate temperature all of the steels of interest fracture in a ductile mode and have reasonably high fracture toughness, as indicated by the toughness-temperature plot at upper left. However, the fracture toughness decreases as the alloy is made stronger, yielding the strength-toughness characteristic shown at upper right. As the temperature is lowered the yield strength increases, as illustrated in the plot at lower left. The rate of increase depends on the crystal structure and on the density of microstructural features such as interstitial solute atoms that represent thermally-activated barriers to dislocation glide. If the yield strength becomes too high relative to the critical stress for brittle fracture then the alloy becomes brittle, often over a fairly narrow temperature range that defines the ductile-brittle transition temperature. The dominant mode of brittle fracture below the transition is the easier of transgranular cleavage and intergranular separation.

To create an alloy that is both strong and tough at 4K one must first ensure that the alloy fractures in a ductile mode at 4K and then establish the toughness desired at the selected strength level. The ductile-brittle transition can be suppressed by lowering the strength of the alloy or moderating the increase in strength at low temperature, as naturally happens in the most promising austenitic steels and superalloys, or by increasing the alloy’s resistance to brittle fracture, as must be done in ferritic steels and in some austenitic steels. The proper metallurgical approach to improve resistance to brittle fracture depends on the fracture mode. If the brittle mode is intergranular the alloy should be modified to eliminate intergranular surfactants that cause embrittlement, as sulfur and phosphorus do in many steels, or changed to incorporate intergranular surfactants that promote cohesion, as boron does in ferritic Fe-Mn steels and Ni₃Al intermetallics. If the brittle mode is transgranular the alloy can usually be made more resistant to brittle fracture by refining its grain size.

Once the alloy is made tough at 4K one is left with the problem of maximizing its toughness in the ductile mode. Since the toughness is a monotonically decreasing function of the yield strength, as illustrated in the graph at upper right in Figure 8, the toughness at a given strength can only be increased by raising the strength-toughness characteristic as a whole. To do this one must address the mechanism of ductile fracture. Ductile fracture occurs through the nucleation of voids in the heavily deformed region just ahead of the crack tip. These grow and coalesce with the crack tip to accomplish crack growth. The dominant nucleation sites for ductile voids are inclusion particles within the alloy. The voids may form by particle cracking or particle-matrix decohesion, either of which is promoted if the slip pattern is planar, causing dislocation pile-ups and stress concentrations at the particle-matrix interface. It follows that ductile fracture can be made more difficult by decreasing the inclusion count or by promoting a distributed, homogeneous deformation. The inclusion density can be lowered by purifying the alloy. The homogeneity of plastic deformation depends on
the often complex interplay between the inherent deformation characteristics of the alloy, the initial defect density, and the nature and distribution of precipitate particles; each alloy must be treated as an individual case.

2. Other Important Properties

Other important properties of the magnet case alloy include its radiation resistance, its coefficient of thermal expansion, and its cost. The inherent magnetic properties of the structure are not usually a critical concern.

Two types of radiation resistance are relevant. The first concerns the mechanical degradation under long-time bombardment [20]. Radiation introduces defects that can lead to swelling, particularly in Ni-bearing FCC alloys such as austenitic steels, or to embrittlement, particularly in BCC alloys such as ferritic steels. Radiation-induced swelling is normally controlled by reducing the species, such as Ni, that produce He in the lattice under bombardment. Radiation-induced embrittlement is influenced by those metallurgical modifications, such as grain refinement, that normally improve ductile-brittle transition behavior.

The second class of radiation-related problems is the residual radiation in the alloy after it has completed its useful life [21]. If the alloy contains long-lived isotopes it presents a radioactive waste disposal problem that may be severe because of the large tonnage of steel involved. The metallurgical approach to this problem is to either minimize the radiation incident on the structure or delete the alloy species, such as Ni, Mo, and Nb, that produce long-lived isotopes.

The coefficient of thermal expansion is particularly important when the structure is closely interconnected with the superconductor, as it is in the structure of a tightly-wound dipole or the conduit of a force-cooled conductor. In that case the thermal contraction of the structural alloy can introduce strains into the superconductor that reduce its critical current. The problem is to match the relatively low coefficient of thermal expansion of the superconductor. The coefficient of thermal expansion depends on the composition and crystal structure. It is reasonably low in ferritic steels, and can be made very low in FCC alloys that contain high concentrations of Ni and Co. Alloys in the latter class are, however, expensive.

The metallurgical principles that underlie research toward superior magnet case alloys are illustrated by the alloy types that are under development now.

B. Structural Alloys for the Magnet Case

The alloys in the magnet case of a fusion reactor must combine strength, toughness and fatigue resistance in welded structures at 4K. Response to radiation is of interest, and may become critical in future machines that will operate at end-of-life neutron fluences beyond \( 10^{19} \) neutrons per cm\(^2\) (E\(\geq\)0.1 MeV). Alloy cost is an important consideration. American magnet designers do not insist that the alloys be stainless. Three classes of alloys may meet these specifications: modified Fe-Ni-Cr
(i.e., class 300) austenitic steels, modified Fe-Mn-Cr (i.e., class 200) austenitic steels, and modified Fe-Ni or Fe-Mn ferritic cryogenic steels. Each class is the subject of active research and development.

1. **Fe-Ni-Cr Austenitic Steels**

The strength-toughness characteristic of the conventional Fe-Ni-Cr stainless steels is presented in Figure 6 [22]. These alloys have adequate strength and toughness to satisfy the specifications for current systems, but do not have the properties that may be needed for future machines, for example, the JAERI specifications [XX] given in the figure.

The simplest way to improve the strength-toughness properties of these steels is to utilize high-purity melting practice to minimize the inclusion content. Researchers at Nippon Steel have used this approach to achieve an excellent combination of strength and toughness [23]. The initial mechanical property data for these alloys are attractive, but American researchers are concerned by their high potential cost. It is also unclear whether these high-purity alloys can be welded so as to maintain comparable purity, and comparable properties, in the weld metal and heat affected zone.

An alternate approach is to add manganese to the Fe-Ni-Cr base alloy, which increases the solubility of interstitial nitrogen and permits the alloys to be solution-hardened to high strength. The high strength alloy Nitronic 40 (21Cr-6Ni-9Mn-XN) is an example. These alloys reach very high strength levels at 4K, but also tend to embrittle so that their fracture toughness is relatively low [24]. The current versions of the Nitronic alloys do not meet the JAERI specifications. Difficulties have also been encountered in the welding of these alloys, due to the retention of brittle δ-ferrite or the formation of intermetallics in the weld metal [25].

The most direct solution to the problems of the nitronic alloys would appear to be a significant increase in the manganese content, but if this is done the alloys become members of the class of Fe-Mn based alloys discussed below.

2. **Fe-Mn-Cr Austenitic Steels**

The second important class of austenitic steels includes the steels that are based primarily on the Fe-Mn-Cr ternary. These steels have been under intensive development for a variety of applications in recent years, principally in Japan. They offer a good combination of low temperature mechanical properties at relatively low cost, and have been tailored for both 77K and 4K applications.

The Fe-Mn system differs from the Fe-Ni system mainly through its propensity toward mechanical twinning and transformation to the hexagonal ε-martensite phase [26]. The two phenomena are related. A twin or fault on the basal plane in the FCC structure introduces an element of hexagonal phase; ε-martensite can be created by periodic twinning or faulting. The hexagonal ε-martensite is found in quenched structures at Mn contents that are intermediate between the α' martensite and stable γ
austenite regions, i.e., between ~10 and ~30 weight percent Mn in the Fe-Mn binary (Figure 9). The ε-martensite also forms through a strain-induced transformation of the residual γ in alloys that fall in this composition range (Figure 9). The ε phase contributes to strength, but appears to be detrimental to the toughness [27]. There is a deep trough in the cryogenic toughness of the alloy at about 25 weight percent Mn (Figure 10) that is more pronounced than can be explained from the strength increase alone.

An appropriate addition of chromium, carbon or nitrogen decreases the tendency toward ε-phase formation and tends to improve the strength-toughness relation. Both carbon and nitrogen are also potent interstitial strengthening species in Fe-Mn (Figure 11). Nitrogen is the preferred additive; carbon forms precipitates during heat treatment or welding that degrade the toughness (Figure 12) [28]. Since Mn increases the solubility of N in austenite it is possible to use the nitrogen addition to increase low temperature strength to a very high level. Chromium stabilizes the austenite phase in high Mn alloys, and also imparts corrosion resistance if it is present in an amount above about 12 weight percent. Fe-Mn-Cr alloys can hence be made stainless. But since brittle intermetallic phases can form at higher Cr contents, particularly in weldments and castings, a small amount of Ni is often added to the stainless grades to stabilize the γ phase [29].

An Fe-Mn-based alloy that is stabilized in the γ phase still tends to deform largely by microtwinning [30]. The result is a high work hardening coefficient and a very complex substructure in the deformed alloy. This appears to be the reason that the Fe-Mn alloys are relatively tolerant of inclusion particles and have good strength-toughness characteristics even when they are rather impure. Similarly, the Fe-Mn-based austenites respond very well to mechanical work, adding strength with a relatively small loss in toughness [31].

Two classes of Fe-Mn-Cr alloys have been developed in recent years in Japan for cryogenic structural use. They differ in corrosion resistance, and hence in Cr and Ni content. The non-stainless grades include the 25Mn-5Cr alloy that was developed some years ago at Nippon steel for use at 77K [32] and the 32Mn-7Cr alloy that was subsequently proposed by Japan Steel Works [33]. Stainless cryogenic grades have been proposed by Kobe Steel [34] and by Kawasaki Steel [35].

The exceptional cryogenic properties that can be achieved in the stainless Fe-Mn-Cr alloys are indicated by the data shown in Figure 13, which are taken from recent work on a Kobe Steel alloy (18Mn-16Cr-5Ni-0.22N) at the Lawrence Berkeley Laboratory [31]. Almost irrespective of its processing the alloy has a strength-toughness combination at 4K that is superior to that of the conventional stainless steels. It preserves its comparative advantage in toughness after the addition of mechanical work through either hot- or cold-rolling. If the alloy is tested as-rolled at 1200°C it essentially meets the JAERI specifications for 4K strength and toughness. Cold rolling produces an even better strength-toughness combination, but it is difficult to see how cold rolling can be applied to plates of the thickness needed for the magnet case.

This class of alloys also has favorable fatigue resistance [36].
Figure 14 compares the fatigue crack growth rate at cryogenic temperature to that of the Fe-Cr-Ni stainless steels 304 and 304LN. The crack growth rate of the hot-rolled alloy is well below that of 304LN, the alloy used in the construction of the MFTF magnet case. It is not as low as that of the metastable austenitic alloy 304, but has substantially higher strength.

The properties of this class of stainless Fe-Mn-Cr-N steels were confirmed in further work on a 22Mn modification done jointly by Kobe Steel and JAERI [34]. It seems clear that the alloy can be made in commercial heats that meet the JAERI specification. The weldability of these alloys is, however, unclear. The extremely favorable strength-toughness combination is due to a significant degree to the residual mechanical work in the microstructure. Welding procedures must be developed to preserve or duplicate this microstructure so that welded plates will have comparable properties.

Given the attractive properties of the Fe-Mn alloys tested to date, further research can profitably be done on several alloy variants. The most attractive modification for the U.S. fusion program is the elimination of Ni. The Ni alloy addition adds to alloy cost, and has the additional disadvantage that Ni produces long-lived reaction products in a radiation environment (Fe-Mn-Cr alloys are exceptionally free of long-lived isotopes). While it may be necessary to sacrifice the corrosion resistance of the alloy to eliminate Ni and maintain strength and toughness, corrosion resistance is not considered an essential requirement by U.S. magnet designers.

3. Ferritic Cryogenic Steels

The focus on austenitic steels for the superconducting magnet case has two origins: their favorable low-temperature ductility in the as-cooled condition and their low magnetic susceptibility. The latter requirement has largely disappeared as studies have shown that the presence of saturated ferromagnetic materials will not adversely affect performance. In fact, the modified A286 superalloy used for the conductor conduit in the Westinghouse large coil is ferromagnetic at 4K, and ferritic alloys are serious candidates for the reactor first wall that sits within the magnetic field. The low-temperature brittleness of the ferritic alloys remains a problem. Even the best of the ferritic alloys that are commercially available have ductile-brittle transition temperatures above 4K.

On the other hand, the ferritic alloys offer specific advantages for use in the magnet case. They are relatively inexpensive, familiar alloys that naturally reach strength levels near 200 ksi (1300 MPa) at 4K. Alloy design efforts have been undertaken in both the United States and Japan to create ferritic steels that are suitable for the magnet case.

The problem with the ferritic steels is the ductile-brittle transition, and the metallurgical challenge is to devise alloy treatments that suppress the transition temperature \( T_B \) to below 4K. The starting alloys are the Fe-Ni ferritic steels that have been successfully used at temperatures of 77K or below, Fe-(9-12)Ni.
The source of the ductile-brittle transition in 9-12Ni steels is relatively straightforward, and can be understood by reference to the 'Yoffee diagram' in the lower left of Figure 8. These alloys naturally quench into a lath martensitic structure whose basic element is a 'packet' of parallel laths (Figure 15) [38]. While the substructure of a packet appears refined in an optical or bright-field transmission electron micrograph, crystallographic studies show that the laths share a common crystallography (Figure 16). They hence have a common [100] cleavage plane, and cleave as a unit under high stress. When the alloy is cooled its strength increases dramatically until the cleavage stress for the martensite packet is exceeded in the highly stressed region ahead of the crack tip. The alloy then fractures in a brittle manner, predominantly through the successive cleavage of martensite packets along [100] planes as illustrated in the profile transmission electron fractograph shown in Figure 17 [39].

The most direct way to decrease $T_B$ in a ferritic steel is to refine the effective grain size, which is the martensite packet size in the best Fe-Ni cryogenic steels. There are two metallurgical techniques that may be used to accomplish this. The conventional approach is to give the alloy an intercritical temper so as to introduce a controlled distribution of precipitated austenite phase along the martensite lath boundaries (Figure 18) [38-40]. This austenite breaks up the crystallographic alignment of the martensite packets (in a somewhat subtle way), increases the resistance to cooperative cleavage fracture, and decreases the ductile-brittle transition temperature.

The second approach is to heat treat the alloy to reduce the packet size. This can be accomplished by thermal cycling treatments that either recrystallize the prior austenite grains, decreasing the maximum packet size (Figure 19), or directly destroy the crystallographic alignment of adjacent laths to achieve an extremely fine effective grain size (Figure 20) [41]. The thermal cycling treatments used for bulk alloys must employ relatively slow heating and cooling rates. The most favorable treatments are those that alternate a thermal cycle into the austenite stability field ($\alpha + \gamma$ reversion treatment) with a thermal cycle into the upper portion of the $\alpha + \gamma$ stability field (an intercritical anneal). It was found some years ago [42] that an four-step cycle of this type, designated the '2B' treatment, could produce an exceptional combination of strength and toughness near 4K in an Fe-12Ni-0.25Ti alloy. Two more recent versions of the thermal cycling treatment are diagrammed in Figure 21 [43].

The 4K strength-toughness combinations obtained in Fe-12Ni-0.25Ti that was processed through the thermal cycle schedules described above are plotted in Figure 22, and compared to the strength-toughness characteristic for the conventional stainless steels and to the JAERI goal. Ferritic alloys can be reproducibly made that have 4K strength-toughness combinations above those of the conventional stainless steels. While these alloys do not yet have the toughness required by the JAERI goal, detailed fractography shows that the fracture mode in the best alloys is still not completely ductile. It follows that further improvements in processing should yield a substantial addition to the toughness. Research on these alloys is continuing.
A surprising and important property of the ferritic alloys is their weldability for 4K service. Fundamental research has shown that the best heat treatment for the Fe-Ni alloys is a rapid thermal cycle into the γ field. As was illustrated in Figure 20, a rapid reversion cycle directly decomposes the lath alignment within a martensite packet to create an extremely fine effective grain size [41]. The rapid heating and cooling rates employed are impractical for the processing of bulk alloys, but are naturally imparted if the alloy is welded in a fine wire, multipass process in which each deposited bead is repeatedly heated by subsequent passes. An appropriate gas-tungsten-arc (GTA) welding technique was developed some years ago in a joint project between Kobe Steel and Nippon Kokan to weld 9Ni steel for use at 77K and above [44]. A suitable modification of this process has been shown to yield weldments that have exceptional toughness at 4K [45].

The viability of the rapid thermal cycle, or multipass welding scheme is perhaps best illustrated by the compact-tension fracture toughness sample shown in Figure 23 [46]. This sample was cut from a plate that was built up entirely of Kobe 11Ni weld metal using a multipass GTA welding process. The plate is hence entirely weld metal, and yet had a fracture toughness above 110 ksi·in$^{1/2}$ at 4K. Since there were flaws in the weldment that appear as fairly large brittle regions in the fracture surface, an improved welding technique will almost certainly produce a substantially higher toughness.

It follows from these results that ferritic cryogenic steels can be made to have an attractive combination of strength and toughness at 4K. These alloys should also be useful for high field magnet structures.

C. Structural Alloys for Force-Cooled Conductors

A force-cooled conductor is shown in Figure 24. It consists of a relatively loose bundle of superconducting wire inside a metal conduit. Helium flows through the interstices of the wire bundle to cool the superconductor. This geometry offers advantages in magnet construction that have made it the configuration of choice for several large systems, including the Westinghouse large coil and the planned Alcator DCT.

The conduit for the force-cooled conductor is typically made by wrapping a sheet of metal around the wire bundle, seam-welding it to form a tube, and squaring the tube so that the conductor can be stacked in a compact winding [47]. If the superconductor is NbTi the conductor sheath can be made of any suitable cryogenic steel. Conventional stainless steels have been used. But if the conductor is Nb$_3$Sn that must be reacted after assembly the conductor sheath must be made of an alloy that can tolerate the long-time, high-temperature reaction treatment and still preserve a good strength-toughness combination at 4K.

The conventional stainless steels are generally unsuitable for Nb$_3$Sn conductor conduits because they become sensitized or form intermetallic phases in the weldment after prolonged heating at high temperature. The materials of choice have been superalloys [48], since these are normally processed by aging for long times at temperatures in the range used for the reaction heat treatment of Nb$_3$Sn, 700–750°C.
The Westinghouse large coil used a bronze-processed Nb$_3$Sn conductor that was reacted at 700°C for 30 hrs. The alloy chosen for the conduit case was JBK-75, a modified version of A286 [49]. Its nominal composition is Fe-29Ni-15Cr-2Ti-1.25Mo-0.25Al-0.25V-0.016C [XX]. It hardens during heat treatment through the precipitation of the ordered intermetallic phase γ' Ni$_3$(Ti,Al,Mo), an ordered precipitate that coarsens slowly. Like most precipitation-hardened alloys JBK-75 has a yield strength that depends only slightly on test temperature, but exceeds 150 ksi (1000 MPa) at 4K after a slight cold work followed by heat treatment for 30 hrs. at 700°C. Since the alloy ductility and thin sheet toughness remain high after autogenous GTA welding, it meets the obvious mechanical requirements for use in the conductor sheath.

However, JBK-75 is less suitable for the conductor sheath in the Alcator DCT [50]. This device is planned to operate at a higher field (10T versus 8T for the Westinghouse coil), which places more stringent requirements on the performance of the Nb$_3$Sn conductor. To improve the critical current it is desirable to use a double-aging treatment, such as 700°C/2 days plus 730°C/2 days. The combination of higher aging temperature and longer aging time initiates the reconfiguration of the intermediate γ' precipitate into the equilibrium η phase along the grain boundaries, with a concomitant loss of strength and toughness. Moreover, the JBK-75 undergoes a much greater thermal contraction on cooling to 4K than the superconductor does. The excess contraction compresses the superconductor, causing a significant decrease in its critical current at 10T. These problems suggest the need for a more compatible sheath material.

The alloys that are currently under study for the Alcator sheath are low-expansion Fe-Ni-Co superalloys [51]. An example is Incoloy 903: Fe-39Ni-14Co-3Nb-1.4Ti-0.9Al. Incoloy 903 has a coefficient of thermal contraction that is almost identical to that of Nb$_3$Sn between room temperature and 4K. It hardens through the precipitation of the ordered intermetallic Ni$_3$(Nb,Ti,Al). The niobium addition improves the aging response and also slows the overaging kinetics so that the alloy is more tolerant of long-time, high-temperature heat treatments.

The change in the sheath alloy for the Alcator magnet is one example of the probable trend in alloys for the force-cooled conductor conduit. These will be chosen to provide both the needed cryogenic structural properties and the appropriate match to the functional properties of the superconductor. The alloys will be tailored to the design, and may differ from one magnet to another.

IV. High Field Superconducting Wire

The function of the wire in a superconducting magnet is to carry the circulating current that creates the magnetic field. Higher current in the wire makes it possible to generate a higher field or to produce a given field with fewer turns in the winding. The maximum current that can be carried by a superconducting wire is, hence, one of its most important properties. But a superconducting material has a critical current density ($J_c$) above which superconductivity is lost. The criti-
cal current density decreases as either the temperature or the magnetic field is increased. $J_c$ also decreases when the wire is strained. The critical current of the wire ($I_c$) is the product of the $J_c$ of the superconducting phase and cross-sectional area of the superconductor in the wire. The areal fraction of the superconductor in the wire is limited either by manufacturing constraints (as in bronze-process Nb$_3$Sn) or by stability considerations, which require that there be an alternate conducting path, usually provided by copper, to prevent the superconductor from overheating and quenching if thermal or current transients cause a local excursion beyond the capability of the superconducting phase.

These considerations suggest that the best superconducting wire will contain the maximum areal fraction of that particular superconducting phase that has the highest critical current at the field at which the wire is to be used. But other factors influence the actual choice of the superconductor. The practical materials with the best critical current characteristics are the A15 phases, such as Nb$_3$Sn, Nb$_3$Al, Nb$_3$Ge and V$_3$Ga. But these are expensive, difficult to manufacture, and troublesome to handle and wind. For this reason the alloy superconductor NbTi is widely used for high field magnets.

However, the relatively low critical current of NbTi at 4K limits its use to magnets that have modest peak fields, ~6.5T in accelerator dipoles, ~8T in solenoids. Its capability can be improved somewhat by alloying with heavy elements such as Ta and Hf, and can be raised dramatically by superfluid cooling to 1.6K. But current engineering analyses suggest that it is advantageous to change the superconductor to Nb$_3$Sn at fields above those cited.

Nb$_3$Sn superconductors have undergone rapid development in recent years to improve both high- and low-field properties, maximize reliability and minimize cost. This development is continuing. But even with recent improvements Nb$_3$Sn is less useful for magnets that have fields above about 16T. Very high field magnets will require superconducting phases with higher critical fields. The only one of these for which a practical manufacturing process exists is V$_3$Ga. But this phase is inherently expensive because of the high cost of gallium. There are hence active research programs on the manufacturing of superconducting wire from promising alternative A15 phases such as Nb$_3$Al.

The metallurgy of NbTi is well documented elsewhere [52]. The present discussion will concentrate on the development of Nb$_3$Sn and on research toward new materials for use at very high fields.

A. Nb$_3$Sn Superconducting Wire

The Nb$_3$Sn wires of greatest current interest are multifilamentary wires in which thin filaments of niobium are embedded in copper or bronze and then reacted with tin to form filaments of the A15 phase. To maximize critical current these should contain a high areal fraction of an inherently good Nb$_3$Sn conductor. The attainable areal fraction depends on the manufacturing procedure.

1. Manufacturing Techniques
The Nb₃Sn conductors that are now being studied in the United States include several different Nb configurations and Sn sources. The four most promising are:

1. **Bronze-process wire** [53]. In this, the most common configuration, Nb filaments are inserted into a Cu-Sn bronze matrix and drawn into a fine, multifilamentary wire. The composite is heated in the range 650-800°C to form Nb₃Sn at the surface of the filaments by extracting Sn from the bronze. Since the maximum solubility of Sn in the bronze is ~8 wt. percent, these wires have a limited areal fraction of Nb₃Sn in the wire cross section. However, beneficial dopant species such as Ti, Ta or Mg can be added either to the bronze or to the Nb to improve the superconducting properties of the A15 phase. An example wire that was manufactured by Oxford Airco is shown in Figure 25.

2. **In-situ wire** [54]. In the in-situ process a Cu-Nb ingot is directionally-solidified into a bar. Since Cu and Nb are almost insoluble in one another in the solid state Nb precipitates during casting. Directional solidification causes the Nb phase to form in dendritic filaments along the axis of the bar. The bar is then drawn into a fine wire that contains very fine Nb filaments. The wire can be coated or co-drawn with tin, and reacted to form Nb₃Sn. The most active current research on the in-situ process is at the Ames Laboratory of the Department of Energy.

3. **Internal tin multifilamentary wire** [55]. The areal fraction of Nb₃Sn in the cross section of a multifilamentary wire can be increased by providing an internal source of tin to supplement that available from the bronze. A practical method for doing this is to draw a multifilamentary wire of Nb in copper or bronze, and then co-draw the wire with filaments of pure Sn. An example that was manufactured by Intermetrics General (IGC) is shown in Figure 26. The wire is usually given a rather complex heat treatment in which the wire is first heated to bind the Sn into a Cu-Sn intermetallic, and then reacted at higher temperature to form the A15 phase.

4. **Modified jelly-roll wire** [56]. Multifilamentary wire is relatively difficult to manufacture because of the need to draw the Nb-bronze composite into fine filaments. An alternate approach, under development by Teledyne Wah Chang, is to slice and stretch a thin sheet of niobium to form a continuous mesh. The mesh is then laminated with bronze or copper and continuously wrapped around a tin wire. A number of these 'jelly-rolls' are compacted and co-drawn into a wire that contains a high density of fine Nb filaments, and reacted to form the superconductor. The modified jelly-roll process is attractive for use in steady field magnets, but is disadvantageous for pulsed or variable field magnets because the interconnectivity of its filaments lead to relatively high AC losses.

In addition to methods above, which have been used to make commercial lengths of wire, a number of other approaches have been explored in the laboratory. These include thin-film deposition [57] and powder methods [58,59]. If a thin film of Nb₃Sn is deposited under carefully controlled conditions the film can be made very nearly stoichiometric
and has exceptional inherent superconducting properties. Thin-film deposition is useful for fundamental work, but does not seem practical for the manufacture of wire in useful lengths. The most thoroughly researched of the powder processes is the 'liquid infiltration' method [59] in which a porous powder of Nb is passed through a bath of liquid tin and given a short high-temperature treatment to form the A15 phase at the powder interface. The infiltration method yields a wire with a high critical current and good strain resistance, but is prodigal in its use of niobium, the expensive constituent of Nb$_3$Sn. The problem of scaling the infiltration technique to large billets is also unsolved.

2. Composition and Microstructure Control

The developments cited above have the result that Nb$_3$Sn wires can be made with useful areal fractions of Nb$_3$Sn. The second factor that controls the critical current is the quality of the Nb$_3$Sn itself. Its critical current density at given field, $J_c(H)$, depends on its composition and its microstructure. The most important compositional parameter of the Nb$_3$Sn phase is its stoichiometry [60], though there is evidence to suggest that its inherent critical current can be improved by doping with suitable impurities such as Ti, Ta, and possibly Mg [61-63]. The most important microstructural factor is the density of efficient internal flux-pinning sites. Since the most effective flux-pinning sites are grain boundaries [64], the A15 grain size should be as small as possible.

A third important parameter is the internal strain in the Nb$_3$Sn [65]. In a wire that contains linear filaments the A15 phase is compressed on cooling from the reaction temperature to 4K since its coefficient of thermal expansion is lower than that of the bronze matrix. The net compression increases with the bronze/niobium ratio in the composite wire, and degrades $J_c$. Since some of this compression is recovered when the wire is strained by the Lorentz force generated during magnet operation, the optimum value of the internal strain depends on the design and operating characteristics of the magnet.

a. Composition and microstructure in bronze-process wire

Recent research has shown that the composition and microstructure of the Nb$_3$Sn in a bronze-processed wire are intimately connected [66]. The interconnection is illustrated in Figure 27.

The microstructure of the wire, as determined by high-resolution transmission electron microscopy, shows that the Nb$_3$Sn layer is a composite of three concentric shells [66]. The inner shell is made up of columnar grains that radiate from the Nb interface, as shown schematically in Figure 27 and illustrated in the transmission electron micrograph in Figure 28. These grains are the apparent result of the Nb–Sn reaction at the Nb interface. The columnar grains are dislocated, as shown in the figure. The dislocations probably derive from the strain in the A15 as it grows in a cylindrical geometry. The dislocations coalesce into cell boundaries that decompose the columnar grains into a network of very fine cells or subgrains that form the intermediate shell in the Nb$_3$Sn layer. These fine grains eventually grow to form a coarse-grained shell around the outside of the filament.
Since $J_c(H)$ is a decreasing function of the grain size, the most important element of the A15 layer is the fine-grained intermediate shell. The grain size within this shell increases with the reaction temperature for a filament of given diameter, as shown in Figure 29. However, for the particular wire studied (Fig. 25) the areal fraction of fine-grained material within the shell passes through a maximum as the aging temperature is raised. The areal fraction of fine-grained material at nearly complete reaction is peaked at about the 700°C reaction temperature.

The composition within the A15 layer is affected by the reaction temperature and coupled to the microstructure. As shown schematically in Figure 27 and illustrated by the STEM-EDAX traces shown in Figure 29, the tin content in the layer drops monotonically from an apparently superstoichiometric value at the bronze interface to a substoichiometric composition at the niobium interface, but has a relatively flat profile in the center of the layer. This flattened profile corresponds in dimension to the fine-grained intermediate layer, and falls at a nearly stoichiometric composition.

The plateau in the composition profile is presumably due to the intergranular mechanism of diffusion of Sn through the reacted layer. When the diffusion path is through the grain boundaries, the effective diffusivity increases as the grain size is made smaller. The fine-grained shell hence has a comparatively high diffusivity, and functions as a diffusive layer between shells that function as relative diffusion barriers, homogenizing the tin content. The degree of homogenization depends on the reaction temperature. At high temperature the total gradient across the reacted layer is smaller, and the composition plateau is flatter and more nearly stoichiometric.

The critical current in the Nb₃Sn depends on the grain size in the fine-grained layer, the areal fraction of the fine-grained layer, and the composition gradient. The former factor is optimized by aging at low temperature, the second is optimal at intermediate temperature, and the third has its best value for relatively high temperature aging. The three factors affect the critical current differently at different fields. Fine grain size and a large fraction of fine-grained material are most important at lower fields while the composition is most important at high fields. Consequently the wire has its best properties at fields near 10T when it is aged near 700°C, but has better properties at higher fields when it is reacted at higher temperature.

Similar microstructural considerations determine the optimal aging time. As the wire is aged the thickness of the Nb₃Sn layer increases until the niobium is totally consumed. On the other hand the grain size within the fine-grained layer increases with aging time. The balance of these two effects usually has the consequence that the overall critical current is highest when the wire is treated to slightly less than complete reaction, leaving a small residual Nb core.
b. Double-aging treatments

Given the balance of metallurgical effects that determine the critical current of the Nb$_3$Sn reacted layer it is possible to design heat treatments that improve $J_c(H)$ [66,67]. The simplest of these are double-aging treatments. Low-temperature aging establishes a fine grain size in the intermediate shell and a high defect density in the inner columnar shell, but results in a relatively poor composition profile (Fig. 29). High-temperature treatments improve composition, but at a cost in the grain size. However, high temperature also accelerates the diffusion-controlled processes, including the chemical distribution through the layer and the polygonization of the dislocation structure in the columnar shell. It follows that a double-aging treatment in which a low-temperature aging is followed by aging at higher temperature should thicken the fine-grained shell and improve its composition. If the second aging temperature is not too high these beneficial effects should be achieved without significant grain growth.

Figures 30-32 show the microstructural consequences of two sets of double-aging treatments of the Airco wire that was pictured in Fig. 25. An incomplete treatment at 700°C followed by a second aging at 730°C increases the areal fraction of the fine-grained shell (Fig. 30) and improves the composition profile (Fig. 31). The result is a substantial increase in the critical current at all fields (Fig. 32).

The benefit is lost if the reaction temperatures are too widely separated or if the second aging treatment is too high. Aging at 800°C following an initial age at 650°C causes exaggerated grain growth, decreasing the fine-grained fraction, leaving a poor composition profile, and decreasing the critical current relative to that in the best single-aged condition.

c. Chemical modification to improve microstructure

While double-aging treatments are useful they do not affect the state of the coarse-grained shell that surrounds the superconducting filament. This shell may be relatively small in thickness, but still accounts for a significant fraction of the total A15 because it encases the outer radius of the filament. To mitigate or eliminate the coarse-grained layer it is useful to employ chemical species that retard grain growth.

The most useful of the growth-retarding species investigated to date is magnesium [68,XX]. If Mg is added to the bronze it segregates strongly to the Nb$_3$Sn layer during the reaction (Fig. 33), and has proven strikingly effective in suppressing grain coarsening in bronze-processed multifilamentary wires made in the laboratory (Fig. 34). The composition distribution within the wire is also improved (Fig. 35). The result is a pronounced increase in critical current relative to that of Mg-free wire (Fig. 36). The critical current increases with Mg content up to at least 0.62 wt. percent; there are experimental indications that the optimum Mg content is near this level.

Since Mg refines the grain size within the reacted layer and increases the areal fraction of the fine-grained shell there is little
point in double-aging a Mg-doped wire. The experimental results [68] confirm that double-aging has little effect on the critical current of these wires.

d. Chemical modification to improve the A15

The critical current of Nb₃Sn wire can also be increased by adding chemical species that improve the inherent superconducting characteristics of the phase. Both titanium [XX] and tantalum [XX] are believed to increase the critical field of Nb₃Sn, and have been added to bronze-processed wire through either the niobium or the bronze. The experimental data also suggests that Mg may enhance the superconducting properties of Nb₃Sn as well as improving its microstructure [XX].

It is, however, difficult to discuss the inherent effect of these additives on the basis of current data. The detailed metallography necessary to separate chemical from microstructural effects has not yet been done.

e. Strain and radiation tolerance

Two aspects of the Nb₃Sn wire that affect its utility in fusion and accelerator magnets are its tolerance for mechanical strain [69] and radiation [70].

The strain tolerance of the wire determines the limits of winding and operating strain to which it can be subjected without seriously degrading its performance. The strain tolerance depends on the bronze to niobium ratio [71] and on the presence and density of voids, cracks and other defects in the wire [72], but beyond these obvious factors the metallurgical features that determine the strain tolerance are not well understood. In current designs strain tolerance is often incorporated in the magnet design, for example, by using wind-and-react techniques to make the winding and structural constraints to limit its strain during operation.

Radiation tolerance is particularly important in magnets for fusion energy. In current designs the superconductor will necessarily be exposed to some radiation, and the resulting degradation may be its life-limiting feature [XX]. The basic reason for the decrease in the critical current after radiation is straightforward [73]: radiation-induced defects disorder the superconductor. It is not yet known whether metallurgical modifications can be utilized to mitigate radiation damage in Nb₃Sn.

f. Metallurgical factors affecting internal-tin wire

As discussed above internal elements of pure tin can be incorporated into multifilamentary or jelly-roll wire to increase the total attainable fraction of Nb₃Sn in the cross section and hence improve the overall critical current of the wire. The internal tin introduces both problems and opportunities [74].

The problems are associated with the low melting point of the tin and the limited solubility of tin in copper. Since Sn expands signifi-
cantly on melting a direct heat treatment of a Sn-bearing wire creates a high internal pressures that may damage the wire (though recent experiments suggest that this may not be a major problem). The alternative is to use a very long, low-temperature treatment to convert the tin into a Cu-Sn intermetallic after wire drawing (the Cu-Sn intermetallics tend to be brittle) but before final reaction. The subsequent heat treatment must also be chosen with caution since Cu and Sn form a variety of intermetallic phases that have different specific volumes, diffusivities and mechanical properties.

The advantages offered by the internal tin process include the possibility of using a dense distribution of very fine filaments and reacting these under sustained conditions of high supersaturation, which accelerates grain growth and reduces the total reaction time. The former effect shrinks the initial grain size; the latter minimizes grain coarsening.

Internal-tin superconducting wires have been made in useful lengths and heat-treated to achieve very high overall critical currents [74,75]. The fundamental metallurgical research that is needed to maximize the properties of these wires is just getting underway.

B. Advanced Superconductors

New high-field superconducting wires are needed for magnets that will operate at fields above about 16T. Bronze-processed V₃Ga is currently available for these applications [76,77], but is inherently costly.

The attractive A15 compounds for very high field use are Nb₃Al, Nb₃Ge, and their alloys [78]. These compounds have exceptionally high critical fields, but are difficult to make into wire in practical lengths. The problem is two-fold. First, since the A15 phases are brittle they must be drawn into wire before reaction, like Nb₃Sn and V₃Ga. But no bronze-like process has yet been discovered that would provide an effective carrier for the Al or Ge. These elements may be incorporated into the wire in a pure form, but that approach confronts the second problem, which is that the stoichiometric A15 composition is not included in the equilibrium phase diagram. If either compound is made by direct equilibrium reaction it is off-stoichiometric and has relatively poor superconducting properties.

The most promising approach to overcome these problems is to use a composition-modified powder process in which Nb powders are mixed with Al or Ge and with a third element that promotes a stoichiometric phase. The rapid reaction of a fine powder can, under appropriate circumstances, yield a reaction product that is more nearly stoichiometric than the equilibrium phase diagram permits. Binary Nb-Al powders have been processed into laboratory wires of reasonable lengths that have been found to have very promising properties [79,80]. Appropriate ternary additions may exist to promote a stoichiometric product and permit the practical manufacture of conducting wires that exploit the superior properties of these phases. Other techniques, such as direct precipitation [81] and liquid infiltration [82], have also been used with moderate success.
Other type II superconducting compounds, such as NbN, are also of interest for both their high critical fields and their possibly enhanced resistance to mechanical strain [83] and radiation damage [84]. However, these phases also lack a practical processing scheme for manufacturing a wire. The current level of metallurgical research on the non-A15 superconducting phases is low. It is not possible to predict whether or when these phases will be utilized in practical wire.

V. Conclusion

The review that is contained in these pages covers only part of the variety of engineering initiatives and materials research activities that bear on the development of high field superconducting magnet technology in the United States. A number of devices were not described, and major materials research efforts on subjects such as superconductor insulation, NbTi technology, mechanical property testing, and alloy development for specific structural requirements in particular devices are not discussed. Nonetheless, the work that is described here should be sufficient to show that cryogenic materials research is an active and exciting area that addresses important technological needs in real devices and is systematically satisfying those needs through sophisticated materials research.
REFERENCES

5. C.D. Henning and E.N.C. Dalder, in Proceedings, Structural Mechanics in Reactor Technology, West Berlin, Germany, August 1979
19. Y. Tomota and J.W. Morris, Jr., ISIJ, Spring Meeting, April, 1984
34. Nippon Kokan, K.K., presented at 1980 AWS 61st Annual Meeting, Session 22
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Fig. 23. Schematic drawing of the microstructure and composition profile through the reacted layer in a bronze-process wire.
Fig. 24. Transmission electron micrograph showing elongated grains near the Nb core in a bronze-process wire. Groups of dislocations are indicated by arrows.
Fig. 25. The fine-grained central shell and part of the coarse-grained outer shell in a bronze-process wire.
Fig. 26. The coarse-grained outer layer at the bronze interface in a bronze-process wire.
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Fig. 30. The Sn composition profile through the reacted layer of wires given the double-aging treatments shown.
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Fig. 33. The Sn composition profile through the reacted layer of bronze process wires containing 0.6 and 0.25 Mg.
Fig. 34. The critical current at 15 T for bronze-process wires containing Mg in the bronze, as a function of the reaction temperature. The symbol 67XX means that the bronze contains 6.7Sn plus $0.\text{XX} \text{Mg}$. 
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