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Authors
Scanlan, Ronald M.
Malozemoff, Alexis P.
Larbalestier, David C.

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Superconducting Materials for Large Scale Applications

R. M. Scanlan, A. P. Malozemoff and D. C. Larbalestier

Abstract

Since the 1960's, Nb-Ti (superconducting transition temperature T_c = 9K) and Nb_3Sn (T_c = 18K) have been the materials of choice for virtually all superconducting magnets. However, the prospects for the future changed dramatically in 1987 with the discovery of layered cuprate superconductors with T_c values that now extend up to about 135 K. Fabrication of useful conductors out of the cuprates has been difficult, but a first generation of silver-sheathed composite conductors based on $(Bi,Pb)_2Sr_2Ca_2Cu_3O_{10}$ (T_c ~ 110K) has already been commercialized. Recent progress on a second generation of biaxially aligned coated conductors using the less anisotropic $YBa_2Cu_3O_7$ structure has been rapid, suggesting that it too might enter service in the near future. The discovery of superconductivity in MgB_2 below 39 K in 2001 has brought yet another candidate material to the large-scale applications mix. Two distinct markets for superconductor wires exist – the more classical low-temperature magnet applications such as particle accelerators, NMR and MRI magnets, and plasma-containment magnets for fusion power, and the newer and potentially much larger market for electric power equipment, such as motors, generators, synchronous condensers, power transmission cables, transformers and fault current limiters for the electric utility grid. We review key properties and recent progress in these materials and assess their prospects for further development and application.

Key words: Superconductors, superconducting wires, Nb-Ti, Nb_3Sn, MgB_2, BSCCO-2212, $Bi_2Sr_2CaCu_2O_{8-x}$ BSCCO-2223, $(Bi,Pb)_2Sr_2Ca_2Cu_3O_{10-x}$, YBCO, $YBa_2Cu_3O_{7-\delta}$, coated conductors.

I. Introduction

Superconducting conductors for electric power and magnet applications are poised for major change. For about 40 years virtually all magnets have been made out of Nb-Ti or Nb_3Sn, the cubic low temperature superconductors (LTS), which possess transition temperatures T_c of 9K and 18K, respectively [1]. In spite of the much higher T_c values of high-temperature superconductor (HTS) materials, ~90K (BSCCO-2212 and YBCO-123) and ~110K (BSCCO-2223), and strong progress in developing HTS wire, these materials have not yet made a serious dent in the commercial dominance of Nb-Ti and Nb_3Sn in the broader magnet market [1]. That market has for years consisted of magnetic resonance imaging (MRI) and nuclear magnetic resonance (NMR), and magnets for high-energy physics accelerators and plasma fusion devices, together with smaller niches for research magnets [2]. These applications were developed with LTS conductors and are well served by their low cost, coupled with their ability to be fabricated as strong, round wires with distinct filament structures, high current density and a high superconductor fill factor.

Nb-Ti conductors were scientifically understood and commercially optimized during the 1980's. Nb-Ti is in fact the only material whose fabrication process has been effectively optimized based on a detailed scientific understanding. This understanding is still being developed for the high field use of Nb_3Sn, especially for NMR and particle accelerator applications, with much recent progress.

Since the late 1980's, conductors made from the silver-sheathed, high-temperature superconductors (HTS) $Bi_2Sr_2CaCu_2O_8$ (BSCCO-2212) and $(Bi,Pb)_2Sr_2Ca_2Cu_3O_{10}$ (BSCCO-2223) have been in ongoing development [1]. Although the understanding of their materials science is still limited and process optimization is for now rather empirical, they have progressed to the point of being commercially available, with acceptable mechanical properties and practical current densities in the 30-77K range [3]. These are known as first generation HTS wires or conductors, and they are principally directed at electric power applications, a major new market for superconductivity [4]. Conductors made of BSCCO-2223 are being applied in significant prototypes of electric power equipment, including an 8 MVAR synchronous condenser [5], a 5000 hp industrial motor [5], a 5 MW, high-torque, ship propulsion motor [5], a 5 MVA transformer [6], and a variety of power transmission and distribution cables [7]. Very recently, a magnet of BSCCO-2212 wire has generated 5 T in a background field of 20 T, making it the highest field superconducting magnet [8], a result which clearly shows the potential of HTS materials to overtake LTS materials in magnet applications.

Simultaneously, the HTS material $RBa_2Cu_3O_7$, where R is a rare earth atom or yttrium, has been developed in a coated conductor format [9,10]; we will use YBCO to denote this material family, since $YBa_2Cu_3O_7$ is the most commonly used member. This technology is rapidly moving from R+D to scale-up and has already enabled a first short power cable prototype carrying a commercial-level current [11]. Coated conductors are often called second-generation HTS wires or conductors. In addition, some developmental conductors are starting to be fabricated from the recently discovered superconducting material MgB_2 [12].

The situation in 2003 is distinctly different than that of 5 years ago. For the first time, one can say that the Nb-based LTS conductors are approaching their limits, and that BSCCO and YBCO materials, and perhaps MgB_2, are ready to surpass LTS capabilities in higher field and higher temperature domains, thus opening up new markets.

The largest potential market for HTS conductors lies in the electric power arena and involves power transmission cables, high-power industrial and ship propulsion motors, utility generators, synchronous condensers, fault-current limiters and transformers. Study after study, such as Vice President Cheney's Task Force on National Energy Policy [13], the Secretary of Energy's National Transmission Grid Study [14], and Grid 2030 [15], the vision of the future transmission system from the DOE Office of Electric Transmission and Distribution, have highlighted the critical reliability and capacity issues in the US power grid, and specifically noted the beneficial potential impact of superconducting technology in addressing these issues. The northeast blackout of Aug. 14, 2003 has elevated calls for action to a crisis level. Momentum from these events is propelling HTS conductors into commercialization.

In this article we review key properties and technical progress in superconducting materials for large-scale applications during the last few years, and assess their prospects for further development and application. We limit discussion to materials in the form of flexible, long-length wires rather than bulk pellets and rods, which have their own array of issues and applications.

II. Conductor Requirements for Power and Magnet Technology

Superconducting conductors for large-scale applications are round or tape-shaped wires in which one or more superconducting filaments are embedded in a matrix consisting at least partially of a normal metal, such as Cu or Ag, which provides protection against magnetic flux jumps and thermal quenching [1,16] (Figure 1). Such wires must have sufficient strength to withstand conductor fabrication, coil winding cabling, the thermal stresses of cool-down, and operational electromagnetic stresses. They must be capable of carrying operating currents, dc or ac, of hundreds of amperes, and often of being cabled to carry thousands of amperes. In particular, the minimum critical tensile stress before loss of critical current density should be in the range of 100 MPa or higher, and minimum tensile, compressive and bending strains before degradation must be several tenths of a percent. Engineering (total cross-section) current density J_e must attain 10^4-10^5 A/cm^2 in magnetic fields which extend from ~0.1 to 25 T, depending on the application. The superconductor current density J_c is J_e/f, where f is the fill fraction of superconducting material in the total conductor cross-section. Operating temperatures tend to be as low as possible for ultra high-field (~25 T) dc applications, for example 2K in superfluid helium, but are as high as possible for ac power applications, since ac losses generated at low temperatures exact a stiff cooling penalty of many times the Carnot efficiency in practical cryogenic environments (see paper X in these proceedings)..

Estimates of the maximum acceptable price for different applications range from ~\$1 to 100 per kiloamp-meter, where kiloamp (kA) refers to the operating current level. In particular, Nb-Ti wire typically sells for \$1/kAm but is limited to operation in the helium temperature range. High temperature superconductors operating at 20-77 K are expected to be economical for some applications in the \$10-100/kAm range. In power equipment, copper wires typically operate in the range of \$15-25/kAm; this sets a benchmark for superconducting wire. As a whole, LTS conductors are used for applications where few or no conventional alternatives exist, while HTS materials must compete against copper in electric power technology, where cost pressures are almost always significant.

III. Basic Physical Properties of Superconducting Wires

We give only a brief summary of the basic physical concepts underlying the performance of superconducting materials in conductors; these are described more extensively elsewhere [1,9,16,17]. These materials exhibit superconducting properties in a region below the interdependent values of critical temperature T_c, upper critical magnetic field H_{c2} and critical current density J_c. The maximum critical temperature for metallic low temperature superconductors (LTS) is 23 K; it is 40 K for MgB_2 [12], and about 135 K for high temperature superconductors (HTS) based on cuprate (copper oxide) compositions [18].

Figure 2 presents the magnetic-field/temperature (H-T) phase diagram for the three presently commercial conductor materials, Nb-Ti, Nb_3Sn, and $(Bi,Pb)_2Sr_2Ca_2Cu_3O_{10}$ (BSCCO-2223) and three developing conductor materials $YBa_2Cu_3O_7$ (YBCO), BSCCO-2212, and MgB_2. Their very different phase diagrams result from their distinctly different crystal structures and physical parameters. All six are type II superconductors for which bulk superconductivity exists up to an upper critical field $H_{c2}(T)$, that can be highly anisotropic and exceeds 100 T for BSCCO-2223 and YBCO. However, applications of HTS materials are limited by a lower characteristic field, the so-called "irreversibility field" $H^*(T)$, at which the bulk J_c vanishes [19].

Under equilibrium conditions magnetic flux penetrates the bulk of a Type II superconductor above a lower critical field H_{c1}, which is <100 mT for the materials under consideration [19]. This magnetic flux exists as an array of flux-quantized line vortices or fluxons, which can be spatially ordered or disordered, static or liquid-like. Each vortex is a tube of radius of the London penetration depth $\lambda(T)$, in which superconducting screening currents circulate around a small non-superconducting core of radius $\xi(T)$, where $\xi(T)$ is the superconducting coherence length. The flux carried by the screening currents in each vortex equals the flux quantum $\phi_0 = 2\times10^{-15}$ Wb. Bulk superconductivity disappears when the normal cores overlap at the upper critical field, $H_{c2}(T) = \phi_0/2\pi\mu_0\xi(T)^2$.

Type II superconductors can carry bulk current density only if there is a macroscopic fluxon density gradient defined by the Maxwell equation $\nabla\times\mathbf{B} = \mu_0\mathbf{J}$. This gradient must be sustained by the pinning of vortices at microstructural defects, and much process development is oriented to optimizing this "flux pinning" to maximize the current carrying capacity. At absolute zero, superconducting current can flow without any loss (i. e. at zero voltage) up to a critical current J_c, corresponding to the maximum pinning strength. However, at finite temperature, thermal activation can cause "flux creep" of vortices down the macroscopic fluxon density gradient (actually a small amount of creep can occur even at absolute zero because of quantum tunneling of vortices, observed in HTS materials) [19-21]. In most materials with random microstructural defects, the vortex structure is disordered or glassy in nature, and vortex glass theory predicts that flux creep gives rise to a $V\sim I^n$ voltage-current relationship, where n is called the "index value" [20-21]. Non-uniformities in material properties on nm or greater length scales can also give rise to this power-law relationship. The index value must be well above 30 to allow the superconducting wire to operate in a quasi-persistent mode, with minimal current decay over time. This condition can be achieved in Nb-Ti and Nb_3Sn but only rarely in HTS wires where typical index values range from 15-30. Because the voltage-current curve is continuous, there is no absolute measure of J_c. However, voltage increases rapidly enough with current to enable a practical definition of J_c at a characteristic electric field, typically chosen to be either 0.1 or 1 μV/cm. For magnet applications, even lower electric fields may be appropriate.

Above J_c, or at magnetic fields above a characteristic "irreversibility field" H*, which is a function of temperature and which is much lower than H_{c2} for HTS materials, major changes occur in the vortex structure [19]. These changes have been described by a variety of theories, for example the vortex glass melting theory, in which vortices above H* are in an unpinned, liquid-like state. Under these conditions, current will drive a dissipative, flux-flow state, with a resistivity which, to first approximation, is given by the fraction of the material occupied by the normal cores of the vortices. This simple concept leads directly to the Bardeen-Stephen formula for flux flow resistivity $\rho_{FF} = \rho_N B/\mu_0 H_{c2}$, where ρ_N is the normal state resistivity [17].

Long conductors for bulk applications are always based on polycrystalline materials, and the influence of grain boundaries is thus decisive. In metallic LTS and MgB_2, current flow is uniform through the polycrystalline superconducting matrix unless gross obstructions such as cracks occur. In fact, grain boundaries are often beneficial flux pinning centers, so that J_c typically increases as the grain size decreases [22].

However, HTS materials have low carrier density and short coherence lengths, factors that predispose grain boundaries to act as obstacles to current flow [23]. The most unambiguous evidence for this comes from detailed studies of YBCO films grown epitaxially on bicrystals of $SrTiO_3$ or Y_2O_3–stabilized ZrO_2 (YSZ) substrates. The critical current density J_b across the grain boundary drops exponentially below that within the grains, $J_b = J_c \exp(-\theta/\theta_c)$, as a function of the misorientation angle θ between the neighboring crystallites [24-26], where $\theta_c \approx 2\text{-}5^o$, depending on the value of the intragrain J_c (Figure 3).

At angles of about 10° and below, the structure of the grain boundary breaks up into identifiable edge dislocations spaced by $b/2\sin(\theta/2)$, where $b \approx 0.4$ nm is the Burgers vector. The cores of these dislocations in HTS materials are insulating. Above a misorientation angle θ of 5-7^o, the spacing between the cores is less than the coherence length ξ, and the grain boundary starts to behave like a Josephson weak link. At lower misorientation angles, the dislocation cores are separated by only modestly strained material, which enables strong coupling and current flow between the grains at the bulk J_c level. The possibility of fabricating highly textured substrates coated with an epitaxial HTS layer, in which θ is small enough to permit strong coupling and hence significant supercurrent flow, is the basis for the coated conductor class of HTS wire [9].

The mechanism of current flow across grain boundaries in BSCCO-based conductors may be different [27,28]. Although these materials are strongly textured around a common c-axis in wires, the in-plane orientations of their grains' ab-planes are random, causing direct lateral current flow across high-angle grain boundaries to be exponentially weak. The observed current flow appears instead to involve redistribution of the current across the large-area twist boundaries separating c-axis-aligned grains, allowing current to bypass high-angle grain boundaries and other obstacles.

The measurement of critical current for LTS and HTS conductors has evolved in distinctly different ways. LTS wires are always tested in magnetic fields appropriate to the application, e.g. 5-7T for Nb-Ti and 10-15T for Nb_3Sn at an electric field criterion that approximates magnet performance, often 0.1 μV/cm [29]. However the standard measurement for HTS materials, because of its simplicity, is in self field at 77K at 1 μV/cm. The drawback of this test is that J_c drops rapidly with even weak applied or self magnetic fields, making results dependent on the conductor geometries. In the case of BSCCO-2223 conductors, proper comparison to avoid these self-field effects requires a field of at least 0.1T at 77K [30]. As applications develop, testing appropriate to the operating conditions will be established.

Under conditions of time-varying current and/or magnetic field, superconductors are subject to ac losses. There are three types of losses in addition to the usual eddy current losses of the normal metal matrix: (1) hysteretic, due to the motion of vortices through the superconductor, (2) interfilament "coupling" losses, due to induced currents flowing between filaments across the normal metal matrix within a multifilamentary strand, and (3) interstrand "coupling" losses, due to eddy currents flowing between the stands in a multi-strand conductor (see ref [16]). These losses can be mitigated, but not completely eliminated, as follows: In low magnetic field environments in which the superconductor is only partially penetrated by the ac magnetic flux, the ac loss varies inversely with J_c and therefore can be mitigated by increasing J_c. In high field environments in which the superconductor is fully penetrated by ac magnetic flux, hysteretic loss, normalized to the current capacity of the wire, is proportional to d, the superconductor filament diameter (or width perpendicular to the applied field), provided the filaments are decoupled. Thus, practical superconductors for ac applications are subdivided into fine filaments imbedded in a normal metal matrix so that d is reduced to a small fraction of the wire diameter, Figure 1. Round configurations are favorable compared to tape-shaped wires, unless ac fields can be oriented along the wide face of the tapes to minimize d; round wires facilitate cabling with transposition and reduce the maximum transverse dimension as compared to a tape.

Typical multifilamentary superconductors may contain from several tens (MRI magnet) to several thousands of filaments (NMR, fusion, or accelerator magnets), with diameters from ~50 μm to 0.1 micron, depending upon the application. For Nb-Ti, the effective filament size, d_{eff}, is usually the physical filament diameter. However, for Nb_3Sn and all multifilamentary HTS conductors, the effective filament size can be much larger, since the filaments may coalesce during the reaction step needed to produce Nb_3Sn, while for BSCCO-2212 or BSCCO-2223, superconducting grains grow through the Ag matrix to form a superconducting path between adjacent filaments.

Because the matrix material is typically a good normal metal such as Cu in LTS or Ag in HTS, the interfilamentary resistance is typically low enough under magnet-charging rates to allow coupling currents to flow between filaments and thus shield the interior of the wire from flux changes. Twisting the conductor or transposing the filaments can reduce these coupling currents, so that the magnitude of the induced emf driving the currents is lessened and periodically reverses along the wire length. Typical twist pitches are 10 times the wire diameter, with a ratio of 5 being the minimum possible. For large magnets requiring high currents, the conductor must be built up by cabling many strands together, preferably in a fully transposed geometry. The interstrand resistance then controls the magnitude of strand-to-strand coupling currents. This source of loss can be reduced by increasing the interstrand resistance, for example by adding a high resistance alloy layer such as Cu-Ni or a Cr-plate in LTS conductors, or by applying an insulating coating to the strands.

One final issue of importance for conductor properties concerns whether current is percolative or flows uniformly through the whole cross-section. For a uniform conductor, division of the measured critical current by the cross-sectional area of superconductor, A_{sc}, then gives the intrinsic critical current density J_c determined by the flux pinning strength of the material. For Nb-Ti and Nb_3Sn certain extrinsic defects to be described later affect the value of A_{sc} but introduce an error in calculating J_c of no more than 5-10%. For HTS materials, this error may be much larger, because of percolative current flow arising from a combination of intrinsic obstacles like grain boundaries and extrinsic defects such as large insulating second phase particles or cracks [9]. One of the central tasks of conductor development has been to understand such obstacles to current flow and to eliminate them as best one can, finally asking how close J_c can be brought to the ultimate possible current density, which is the depairing current density of the vortex currents, J_d [17]. In Nb-Ti, which has exceptionally strong pinning, J_c is of order 0.1 J_d [31], a value now being approached by YBCO coated conductors under certain circumstances. However, it is always worth questioning whether the active cross-section carrying current differs from the whole cross-section, especially with HTS conductors and perhaps also with MgB_2 [9,32, 33].

IV. Niobium-Titanium alloy

Nb-Ti alloy superconductors have been the "workhorse" materials of the superconductor industry for the past 40 years [1]. They were discovered in the 1960's to have a high upper critical field (~11 T at 4.2 K and 14 T at 2 K), to co-draw well with Cu and to have good ductility. Early magnets performed poorly due to premature quenching, which was not understood at the time. However, the importance of subdividing the superconductor in order to provide intrinsic stability against flux jumps was recognized in the late 1960's [16]. This discovery, together with the discovery by the Rutherford group [34] that twisting the wire would reduce filament coupling, led to wires with greatly improved properties. Several major milestones were achieved in the 1980's using Nb-Ti, including the first superconducting accelerator, the Tevatron, in 1983, and the first large-scale superconducting commercial application, Magnetic Resonance Imaging (MRI), in the early 1980's.

The properties of Nb-Ti alloy superconductors improved slowly throughout the 1960's and early 1970's, so that by the time conductor was ordered for the Tevatron, a critical current specification of 1800 A/mm^2 (5 T, 4.2K) could be achieved. Although a recipe for improving J_c had been developed along empirical lines by industrial manufacturers, the technology lacked a good fundamental understanding. In the early 1980's, a collaboration of university, national laboratory and industrial partners, led by the University of Wisconsin Applied Superconductivity Center, was formed to improve the understanding of flux pinning in Nb-Ti and to incorporate this improved understanding into the manufacturing technology for Nb-Ti [35]. The driving force at the time was the high performance requirements for the Superconducting Super Collider, and much of the work was sponsored by the Office of High Energy Physics of the US Department of Energy (DOE). The steps taken to improve Nb-Ti performance can be divided into two categories--extrinsic and intrinsic, and an apt description of the process of systematic improvement, coined at one of the low temperature superconductor workshops, is "peeling the onion".

First, several extrinsic limitations were discovered and corrected. These include the formation of intermetallics at the Cu/Nb-Ti interface during the precipitation heat treatment steps [36]. This was corrected by the use of a Nb diffusion barrier wrapped around the Nb-Ti filaments during billet construction [37]. Another problem is the tendency of the filaments to undergo distortion during fabrication, called "sausaging". Sausaging is caused by the difference in mechanical properties between the Nb-Ti and Cu matrix, and can be prevented by placing the filaments in a dense stack in the Cu matrix, rather than having them widely spaced [38]. Adjusting the placement and temperature of the multiple heat treatments required to achieve a high Jc can also help reduce the differences in mechanical properties between the Nb-Ti filaments and the Cu matrix.

The first intrinsic limitation that was found and corrected was the inhomogeneity of the Nb-Ti alloy, caused by a distribution of Ti-rich regions [39]. These responded very differently to the precipitation heat treatment and thus prevented establishing an optimum heat treatment to yield a uniform and predictable distribution of alpha-Ti precipitates necessary for effective flux pinning [40]. As a result of the improved understanding of Nb-Ti alloys, the current density was increased by about 100 %, to >3700 A/mm^2 at 5 T, 4.2 K [41].

Multifilamentary Nb-Ti fabricated by conventional technology is a mature field, and the rapid improvements seen from 1983 to 1990 have slowed. However, a new approach to the fabrication of Nb-Ti using "artificial" or engineered pinning centers (APC's) was suggested in 1985 and continues to enable progress [42,43]. The basic idea is to incorporate a normal metal into the Nb-Ti at the beginning of the fabrication process and then to co-reduce both Nb-Ti and the pinning center material by extrusion and drawing until the normal metal component has the correct size and spacing for optimum vortex pinning. The original approach was to drill holes in a Nb-Ti ingot and load metal rods that are compatible with the Nb-Ti (e.g., Nb). However, a more practical method from the manufacturing standpoint is to assemble a jellyroll of alternating Nb-Ti and Nb sheets. This jellyroll is loaded into a Cu can, compacted, and extruded to produce a "clad monofilament" element, which is cut into short lengths, loaded into the final Cu can, and extruded. The resulting multifilamentary composite is then drawn to final wire size. Using this approach, long lengths of Nb-Ti superconductor have been produced with J_c values exceeding those obtained for conventionally processed Nb-Ti, in the field range from 2-7 T. Initially, the high field (6-8 T) performance was not as good for the APC wires; however, recent results with ferromagnetic pinning centers in place of the Nb pinning centers have shown excellent properties at higher fields as well [43,44]. The best J_c values for both conventional and APC Nb-Ti are shown in Figure 4. In this graph we also show some recent optimization using very long heat treatment times by the Khar'kov group that have driven the Nb-Ti critical current density to over 4000 A/mm^2 at 5T, 4.2K [45].

The present rate of production for Nb-Ti superconductor is over 200 tons/year, with one project, the Large Hadron Collider of CERN, accounting for about half of the total. The LHC superconductor procurement will be complete in 2006. It is unlikely that the growth rate of MRI systems will be adequate to make up for this lost LHC volume, and the next large international fusion magnet project, International Tokomak

Experimental Reactor (ITER), will utilize mostly Nb_3Sn. However, much of the existing Nb-Ti production capacity and equipment will very likely be needed for the ITER Nb_3Sn procurements (see below). Thus, while the LTS industry is expected to undergo a transition, overall production volume is expected to remain steady.

V. Niobium-Tin

Nb_3Sn in a prototype wire form was the first material to show the possibility of high-field superconductivity in 1961; its H_{c2} reaches ~30T at 2K, substantially exceeding Nb-Ti and making it the prime candidate for higher magnetic field systems [1]. Progress was initially limited, primarily due to the fact that Nb_3Sn is a brittle intermetallic compound which could only be made as a tape, while Nb-Ti is a ductile alloy, enabling very flexible architectures. However, starting in the early 1970's, practical Nb_3Sn multifilamentary conductors were successfully developed with the bronze route and many laboratory and high field NMR magnets were and are still being made. As applications moved to fields exceeding the practical operating field of Nb-Ti (about 9 T at 4.2 K and 12 T at 1.8 K), there was strong interest in Nb_3Sn, driven also by plasma containment applications.

There are several processes for fabricating multifilamentary Nb_3Sn wires that permit processing ductile precursors down to final wire size, followed by reaction of the components to form the Nb_3Sn intermetallic compound. The so-called bronze process involves co-drawing Nb rods in a matrix of Cu-Sn alloy, which is made as Sn-rich as practical, while still maintaining good ductility (around 15 wt % Sn). However, much residual Cu is left behind after reaction, making A_{sc} a small (~0.2) fraction of the conductor package. To obtain a higher volume fraction of Nb_3Sn, and hence higher overall J_c values, the internal Sn process was introduced. This multifilamentary composite contains Nb filaments in a Cu matrix and the Sn is introduced as a separate component, usually as a core added to the Nb-Cu multifilamentary composite. These two processes, together with the newer powder-in-tube process, are illustrated in Figure 5.

Recently, the two principal technology drivers for Nb_3Sn have been magnets for high energy physics (HEP) accelerators and for magnetic fusion energy (MFE); a third, though less publicly discussed one, has been very high field NMR magnets used for the 800 and 900 MHz class spectrometers. In the early 1990's, both HEP and MFE initiated programs to develop superconductors and magnets for high field (>12 T) applications. The focus of the HEP effort was dipole and quadrupole magnets, while the fusion program focus was solenoidal and toroidal field coils for the international tokamak program (ITER) [46]. Both programs chose Nb_3Sn as the most promising conductor for high fields, and conductor development programs were initiated in industry. During the 1990's, the ITER program was the primary customer for Nb_3Sn, ordering more than 20 tons of wire from companies in Europe, Japan, Russia, and the U.S. This program helped establish a large-scale production capability for Nb_3Sn, and stimulated further development of bronze, internal Sn and powder-in-tube (PIT) multifilamentary wire fabrication techniques [47, 48].

As a result of the accelerator dipole design efforts at the U.S. high energy physics laboratories, it became clear that higher field, cost-effective magnets required higher current density and reduced manufacturing costs. Thus, in 1999, DOE initiated a new conductor development program for Nb_3Sn [49], with the goal of providing a cost-effective, high-performance superconductor for next-generation high-energy-physics colliders, as well as upgrades for existing colliders at Fermi National Accelerator Laboratory and CERN. The target conductor specifications are listed in Table I. The emphasis is on Nb_3Sn made by industrial partners with large-scale Nb_3Sn production experience. Improvements in the critical current density J_c in the non-copper part of the wire has been the highest priority, followed by reduction in the effective filament diameter d_{eff}, and then by lower cost.

The first-priority goal of the program, achieving an increase in the non-copper J_c(12T) from 2000 to 3000 A/mm^2, was reached in 2002 [50]. This goal was achieved without an increase in cost per kilogram, so the cost-performance ratio has decreased significantly (Figure 6). This new wire not only has excellent J_c values, but can be produced in long lengths as well. In 2003, Oxford Superconducting Technology delivered 100 kg of this wire for use in a new dipole magnet at Lawrence Berkeley National Laboratory (LBNL) that recently reached a new dipole field record of nearly 16 T [51].

Development of magnet designs that maintain a relatively low transverse stress on the conductor in the highest field regions, coupled with improvements in Nb_3Sn J_c values, means that Nb_3Sn has the technical performance and cost-effectiveness for use in dipole magnets up to at least 16 T, and perhaps up to 18 or 20 T. Detailed magnet design efforts are underway at several HEP laboratories to determine how far Nb_3Sn can be pushed, and for the selection of the best conductor design for such fields.

After completion of the first phase of ITER, that project is moving ahead with the next phase, which is an international tokamak test facility [46]. The ITER design requires the production of an enormous quantity of Nb_3Sn conductor (approximately 500 tonnes over 3-4 years). The main challenge will be to scale-up the yearly production level from the present 10 tonnes/manufacturer to the required 150 tonnes/manufacturer.

The largest present commercial application of Nb_3Sn is in the area of NMR magnets. This application is growing rapidly and moving steadily to higher fields. Recent improvements in J_e of Nb_3Sn have extended its useful field range to at least 21 T at 1.8K, making persistent mode 900 MHz systems possible. Significant challenges had to be overcome in order to achieve these high fields in a practical NMR magnet. Since this is a very competitive commercial area, magnet construction details are not usually published. Oxford/Varian and Bruker both have working 900 MHz systems with an operating field of 21.1 T and very low field drift rates. The increased performance of recent Nb_3Sn conductors for HEP use at 10-16 T has also benefited very high field applications, because the upper critical field was increased [52]. It is now widely expected that the newer Nb_3Sn conductors will permit 1 GHz NMR magnets to be made of Nb-Ti and Nb_3Sn. This would require persistent operation at a field of about 23.8 T in a material with an upper critical field of about 28-29 T at the operating temperature, a truly challenging but attainable goal.

VI. Niobium-Aluminum

The driver for interest in another A-15 compound Nb_3Al is an even higher magnetic field capability than Nb_3Sn. Two approaches for fabricating Nb_3Al are being developed. The first is analogous to the internal tin process for fabricating Nb_3Sn. A composite of Nb and Al is formed, most often using a jelly roll technique to make the composite sub-element which is extruded. This step is followed by drawing, hexing, and restacking into the final composite with a Cu matrix, which is then drawn to final size and reacted to form Nb_3Al. Sumitomo Electric Industries has used this approach to manufacture several tons of wire for a model coil for the ITER project; so manufacturability has been demonstrated [53]. However, the Nb_3Al layer formed by this diffusion step is off-stoichiometry, and both T_c and H_{c2} are suppressed. This in turn leads to a J_c vs field performance inferior to Nb_3Sn.

An alternate processing route has been developed by the former National Research Institute for Metals (NRIM), now the National Institute for Materials Science (NIMS), in Japan. In this process, a composite of Nb and Al is assembled and drawn to produce a fine-scale composite where the Nb and Al components have 100-200 nanometer dimensions. This composite is heated rapidly to 1900°C and then quenched to produce a bcc solid solution alloy with the stoichiometric Nb_3Al composition. Finally, the composite is heated to about 800°C, where the A15 phase is formed, with high T_c, H_{c2}, and good high-field J_c characteristics. By 1996, the NRIM team scaled up the rapid heat/quench step to produce relatively long lengths using a reel-to-reel system where the wire was resistively heated and then quenched in liquid Ga [54]. Ongoing optimization of this process has steadily raised T_c, J_c and H_{c2} of rapidly quenched Nb_3Al [54, 55]. However, manufacture of the precursor remains challenging, as does the production quenching, addition of stabilization and achieving properties superior to the most recent Nb_3Sn. The NIMS group remains very active in pursuing the possibilities of the raid quench technique and have seeded additional studies on this important material [56].

Chevrel phases, with intermediate transition temperatures, were also actively developed for superconducting wire [1], but this effort has now been largely abandoned.

VII. BSCCO-2212

The high temperature superconductor BSCCO-2212, with a T_c of around 90 K, has been eclipsed for most high temperature applications by BSCCO-2223, which has a higher T_c and a higher irreversibility field (see next section). However, BSCCO-2212 has superior J_c at low temperatures (below about 20K) and high fields. For example, using their Pre-Annealing-Intermediate-Rolling (PAIR) process on long length dip-coated tapes, Showa Electric Wire and Cable Co. Ltd., in collaboration with NRIM (NIMS) and the Japan Science and Technology Corporation, have achieved J_c of 7100 A/mm^2 in self field and 3500 A/mm^2 in 10 T parallel to the tape plane [56]. This has enabled application in high field inserts for NMR and other applications [57, 58]. At present, BSCCO-2212 holds the record for producing the highest field with a superconducting magnet. The BSCCO-2212 tape, produced by Oxford Superconducting Technology (OST), and wound into an insert magnet at the US National High Magnetic Field Laboratory (NHMFL), achieved a 5T increase in field in a background field of 20 T from a copper Bitter coil [8].

Also, BSCCO-2212 has been fabricated as a round wire in a silver matrix (Fig. 1c) with excellent critical current-- J_e at 20 T, 4.2 K over 500 A/mm^2 [59]. This wire can be used as a direct substitute for LTS wires in Rutherford-type cables [60] in applications such as accelerator magnets, and the first series of prototype coils have been made [61]. Two issues must be resolved for this application to proceed. First, the cost/performance ratio must be reduced from the roughly \$50/kA-m value at present, to under \$10/kA-m. Second, practical methods for achieving stringent heat treatment control in an industrial environment must be developed. A partial melting step is required to achieve optimum properties, and the temperature must be controlled to within a few degrees at around 870 C. At present, three companies worldwide are manufacturing BSCCO-2212—Showa, Nexans and OST. All companies have demonstrated excellent J_c values. Showa and Nexans [62] have both demonstrated that they can make 1600 m lengths of wire without breakage and can control the heat treatment to achieve uniform properties.

VIII. BSCCO-2223

The high-temperature superconductor $(Bi,Pb)_2Sr_2Ca_2Cu_3O_{10}$, called "BSCCO-2223," with a T_c of about 110 K, is used in the first generation of commercial HTS conductor, a composite of the superconductor with silver [3]. As shown in the cross-section in Figure 1, this first generation HTS wire has a tape-shape, typically 0.2 by 4 mm, consisting of 55 or more tape-shaped filaments, each 10 μm thick and up to 200 μm wide, embedded in a silver alloy matrix. The filaments consist of grains of BSCCO-2223, up to 20 μm long, often arranged in colonies sharing a common c-axis. For some applications, the wire is laminated to stainless steel tapes on either side for enhanced mechanical properties and environmental protection. This kind of HTS wire is manufactured by a number of companies, including American Superconductor Corporation (USA), European Advanced Superconductor GmbH (formerly Vacuumschmelze, Germany), Innova Superconductor Technology (China), Sumitomo Electric Industries Ltd. (Japan), and Trithor GmbH (Germany). Worldwide capacity exceeds 1000 km per year.

While precise details of the industrial production process are not public, the basic powder-in-tube deformation process is schematized in Fig. 7 [3,10]. A powder consisting of a mixture of Pb-containing BSCCO-2212, alkaline earth cuprates, copper oxide and other oxides is initially prepared by an aerosol, spray-dry or simple oxide mixing process; the overall cation stoichiometry is chosen to match that of BSCCO-2223. The powder is packed into a silver tube, which is sealed, evacuated and drawn through a series of dies, elongating it into a hexagonal rod. These rods are next cut, assembled into a multifilament bundle (55-85 filaments are typical) and inserted into a second silver or silver alloy tube. This tube is sealed, evacuated and further drawn into a fine round wire, which is subsequently deformed to a tape shape in a rolling mill, in multiple steps of rolling and heat treatment to texture and react the powder inside the wire to form BSCCO-2223. The superconducting fill factor is typically 30-40%.

Three important insights into this process are as follows. First, the use of silver, a relatively expensive metal, is necessary because silver, like other noble metals, is inert to the superconductor, and particularly because it permits rapid oxygen diffusion at high temperatures without itself oxidizing. This latter property allows control of the oxygen activity in the reaction of the powder to BSCCO-2223. The silver also acts as an electrical and mechanical stabilizer in intimate contact with the superconducting filaments. Secondly, the use of the rolling process to flatten the wire to a tape shape is necessitated by the need to texture the powder, which consists of mica-like grains of BSCCO-2212, along with the less aspected other oxides. High texture is the key to a high current density in the final product. Thirdly, the success of this process with BSCCO rather than other HTS materials appears to be related to easy slip along the weakly bonded double-Bi,Pb-oxide layers; this facilitates texturing during the rolling deformation. The easy slip plane also enables formation of colonies or stacks of many highly aspected grains rotated in apparently random orientations with respect to each other around a common c-axis perpendicular to the CuO planes. This structure is believed to facilitate current flow as discussed in Section III above [27].

J_c of BSCCO-2223-based wire has been an R+D focus worldwide for 15 years, increasing steadily over time, with long-length (>150 m) wires now achieving up to a maximum of 170 A and an average above 150 A in end-to-end I_c at 77 K and self-field in 0.21x4.2 mm tape (Fig. 8) [3,10]. The maximum I_c corresponds to an engineering critical current density of close to 180 A/mm^2. Given a 40% superconductor fill factor in the wire, this corresponds to J_c of 450 A/mm^2. At a temperature of 30 K and 2 T, of interest for example in rotating machinery applications, the current density is enhanced over its 77 K self-field value by almost a factor of 2. These performance levels have been applied successfully in a variety of commercial-level prototypes and appear adequate for most power applications. At this point, the main value of further increasing I_c is reducing the price-performance ratio $/kA-m. First generation HTS wires are currently sold at $150-200/kA-m, and with further manufacturing efficiencies, price-performance of $50/kA-m is expected, even without further increase in I_c.

More progress in I_c seems feasible, based on an improved understanding of the materials science of BSCCO-2223 and its phase conversion, closely correlated to knowledge of where the current flows and how it is limited by microstructural defects [63]. This task has been pursued by the US-DOE-supported Wire Development Group, an industry-national lab-university partnership, during the past dozen years. Recently there has been progress in mitigating two important current-limiting mechanisms. One is residual porosity left over from the dedensification that occurs during the reaction of 2212 phase to 2223 and the cracking that occurs during the intermediate rolling step that is required to remove such porosity. The second is due to unreacted streaks of 2212 phase that occur either as intergrowths within the 2223 phase or as individual grains within a larger colony of ~10 2223 grains which share a common c axis. Porosity and cracking have been remarkably reduced by processing at hydrostatic pressures of up to 150 bar, [63-66] advancing J_c at the appropriate benchmark of 0.1T at 77K to 304 A/mm^2, as shown in Fig. 9. The corresponding zero-field value at 77 K has surpassed 200 A in a 4.2 x 0.21 mm^2 wire. This marks a doubling of J_c in the last 5 years, and the pace is accelerating as fundamental understanding grows. Similarly, improved heat treatments have reduced the amount of residual 2212 phase. Recent magneto-optical current reconstructions (Fig. 10) on a high-J_c monofilament conductor with J_c(77K, sf) = 350 A/mm^2 [63] show that there is still great headroom in the system. Remarkably, J_c locally reaches well over 2500 A/mm^2, 5 times higher than its average value, and regions of J_c exceeding 1000 A/mm^2 are 50 to 100 μm long, several times the ~20 μm grain length. Although there is a tendency for the highest J_c regions to be located at the Ag-superconductor interface, in fact very high J_c regions are located throughout the filament. The J_c distribution is both higher and spatially larger in the over-pressure processed tapes, consistent with strong reduction of porosity and cracking.

With the strong performance improvement, cost reduction and successful manufacturing scale-up, first generation BSCCO-2223-based HTS wire has proven successful in meeting applications requirements and is being widely used in commercial-level prototypes. No other HTS wire technology is likely to replace it for at least several years [10]. It will be the basis of the first commercial HTS applications, which are expected to be largely in the electric power arena and in magnets where operation at temperatures above 30 K is preferred. The longer-term future for this wire depends on the degree to which its price-performance ratio can be reduced and on the success of alternatives like second generation HTS wire based on YBCO coated conductor.

IX. YBCO coated conductors

The vision for a second generation of HTS wire is based on a quasi-single-crystal layer of $YBa_2Cu_3O_7$, with grain-to-grain misorientations of order 5° or less to enable high current by exploiting the strong grain-to-grain coupling which occurs under these conditions (see Section III)[9,67-74]. Such a continuous YBCO film must be manufactured over long lengths and at low cost with only a nominal Ag fraction [75]. After about a decade of effort, it was shown in 2003 by multiple fabrication routes that continuous processing of lengths of 10-50 m with high critical current properties is possible, opening the path to production scale up. Coated conductors use thin-film technology to apply an epitaxial superconductor layer to a highly biaxially-textured tape-shaped template, whose texture the YBCO replicates. This is a complex, metal-supported, multi-functional, multilayer oxide structure. Cost models for coated conductors, particularly those made by non-vacuum methods, predict values below \$10/kAm (77K, 0T); well below the effective price-performance of copper [75]. Because of the significantly higher irreversibility field of YBCO compared to BSCCO (Fig. 2), second generation HTS wire also offers the opportunity to achieve higher temperature operation in a given magnetic field.

All coated conductors include a flexible substrate, preferably of strong and non-magnetic or weakly-magnetic metal, typically ~50 μm thick, on top of which is a multifunctional oxide barrier or buffer layer, typically less than ~0.5 μm thick, on top of which is the superconducting YBCO layer 1-3 μm thick. A protective Ag layer of a few μm and a thicker Cu protection and stabilization layer complete the conductor [76]. The textured template is created by one of two basic methods, either by texturing the buffer layer by Ion Beam Assisted Deposition (IBAD) [67-71,77], or inclined substrate deposition (ISD) [78,79], or by deformation-texturing the metal substrate with the Rolling Assisted Biaxially Textured Substrate approach [72-74] and applying epitaxial oxide buffer layers (trademarked RABiTS^{TM} by Oak Ridge National Laboratory).

The IBAD method [67] allows for the use of strong, non-magnetic Ni-superalloy or stainless steel substrates on which a textured IBAD layer of aligned yttria-stabilized zirconia (YSZ), gadolinium zirconate (GZO) or magnesium oxide MgO can be deposited [70,77]. This process can achieve good texture (FWHM ~ 10°) with a layer of YSZ that is ~0.5-1 μm thick. However, deposition of IBAD-YSZ is widely considered to be too slow, and thus too expensive, to be commercial. IBAD-MgO, however, produces excellent texture within the first 1-2 nm, and thus deposition may be rapid enough to be commercially cost-effective [70,77]. The x-ray pole-figure full-width-half-maximum (FWHM) of recent IBAD-MgO coated conductors is much better than for IBAD-YSZ, being of order 2-4° and thus genuinely approaching single crystal structure for the YBCO overlayer [70], although performance so far does not fully reflect this unusually high degree of texture, and achieving the required atomic level surface roughness over long lengths has proven challenging. The Inclined Substrate Deposition (ISD) process is more rapid than YSZ-IBAD since there is no re-sputtering during deposition, but in work to date, the texture is not as high [78,79]. ISD may also permit simpler buffer structures. Other ion texturing processes for buffer layer deposition have also been explored [80]. All these methods of producing a textured buffer layer rely on physical vapor deposition processes, whose cost is an obstacle to achieving price-performance competitive with copper.

An advantage of the RABiTS^{TM} approach is texture can be achieved with a low-cost, non-vacuum process. A strong [100] cube texture is introduced into the substrate by conventional rolling and recrystallization. Although the RABiTS^{TM} approach was developed initially with pure nickel [72], substrate materials with more robust mechanical properties and reduced magnetism, notably nickel with 5 atomic % tungsten, have now replaced pure Ni and achieved FWHM below 5° [73].

The buffer layer is also in active development. At present virtually all processes employ a multifunctional buffer in which the functions of seed layer for the metal-oxide interface, a metal and oxygen diffusion barrier, and cap layer lattice-matching to the YBCO interface are separated. Many such architectures are presently vying for selection. Seed layers of Y_2O_3, Gd_2O_3 or NiO have all been employed. YSZ is by far the most common Ni diffusion barrier or buffer, while CeO_2 is generally employed for the cap layer interface to YBCO. IBAD-MgO buffer layers are also in rapid development.

Many different deposition methods for the crucial epitaxial YBCO layer have been developed, all achieving high critical currents in short samples. Thus, the choice among them can be driven by low cost. The leading low-cost alternatives are those based on liquid-phase chemical routes using metal-organic-deposition (MOD) [75,76] or on metal-organic-chemical-vapor-deposition (MOCVD) [81], though advocates of fast electron-beam deposition of either metal constituents or the so-called BaF_2 route remain, as do the advocates of so-called fast pulsed laser deposition. It is also becoming increasingly recognized that the use of higher-T_c versions of the YBCO structure, for example $RBa_2Cu_3O_7$, where R might be Eu or Nd with a T_c of 94-95K could offer significant benefit to 77K properties.

The design of a coated conductor is such that about 50 μm substrate thickness is needed to support 1-5 μm of YBCO, giving a superconductor fraction which is at best 5-10% of the cross-section, as compared to 30-40% in the first generation HTS wire of Figure 1. Present short samples of both IBAD and RABiTS conductors can both exceed 2 MA/cm^2 at 77 K for YBCO layers up to ~2 μm, enabling currents of 400 A/cm-width or 160 A in a 4 mm wide tape, which is comparable to first generation wire. A significant and not well understood obstacle to further increasing I_c is that in many cases J_c of the YBCO layer degrades with increasing thickness. Loss of epitaxy, increasing porosity, a microstructural transition through thickness, and possibly a 2D to 3D transition of the vortex structure once the layer gets thicker than about ~0.5 μm may all play significant roles [82].

Progress on continuous, reel-to-reel processes for making coated conductors has been particularly rapid in the recent year. Continuous 10-100 m lengths of 1 cm wide conductor with end-to-end critical currents as high as 270 A/cm-width A at self field at 77K show that this multi-step fabrication process is being mastered [83]. Uniformity along the length indicates that kilometer-length wires should be possible. Lamination of hardened copper strip to the multilayer of metal substrate, buffer, YBCO and Ag passivation is an important step to achieve a robust, stabilized conductor. In particular, conductor designs such as those shown in Figure 11 include a "neutral axis" conductor in which the YBCO lies at the position of zero tensile or compressive strain during bending, or a "face-to-face" conductor in which two layers of YBCO, coupled through an internal stabilizing copper layer, provide alternative current paths for each other in case of a defect in one [76].

No fundamental technical barriers to the fabrication of long-length coated conductors appear to exist, although much work remains to scale up the process to long lengths. Work will also need to focus on improving flux pinning in YBCO so as to achieve higher J_c in field and a monotonic J_c as a function of field angle. The introduction of these second generation HTS wires into practical applications will be facilitated by designing them with similar size and rating as first generation HTS wires since all present electrical equipment design is being done with first generation wires. Second generation wires are also very attractive for current limiting applications because of the high electric fields that can be reached when the superconductor goes normal in response to a high fault current.

In short, second generation HTS wires show promise for achieving a price-performance ratio below that of copper and for providing new performance features, thus significantly expanding the HTS market. Expected applications are largely in the electric power arena and for magnets where operation at temperatures above 30 K is preferred. Active R&D is underway at Los Alamos, Oak Ridge and Argonne National Laboratories, at the International Superconductivity Technology Center (ISTEC) in Japan, and at a variety of European laboratories and universities. Industrial programs are underway at American Superconductor Corp. and the Intermagnetics General subsidiary SuperPower Inc. in the United States, at Fujikura Ltd., Furukawa Electric Co. Ltd., and Sumitomo Electric Industries Ltd. in Japan, and at Theva GmbH in Europe. However it should be emphasized that it will take several years to achieve a meaningful scale-up to long lengths and adequate production capacity for the commercial market.

X. Magnesium Diboride

In early 2001, the Akimitsu group in Japan discovered that the long-ago-synthesized compound magnesium diboride MgB_2 was a hitherto unappreciated 40K superconductor [84]. Extensive work on MgB_2 has now shown that it is a "conventional" s-wave superconductor but with the unconventional property that it contains two, only very weakly coupled superconducting gaps. This two-gap property is very important to its high-field performance potential [85-87].

A key early discovery was that randomly oriented grain boundaries in MgB_2 are not obstacles to current flow, in spite of the fact that its alternating Mg and B sheet structure makes it electronically anisotropic [88]. The magnitude of this H_{c2} anisotropy is under intense study at present. In single crystals this anisotropy is quite strongly temperature dependent and increases with decreasing temperature to values of 5-7 [89]. In a "clean", very low H_{c2} limit, the zero-temperature perpendicular upper critical field, $H_{c2}^{\perp}(0)$, is only 3-4T, while the equivalent parallel field value, $H_{c2}^{\parallel}(0)$ reaches ~15-17T. Study of dirty-limit films shows that $H_{c2}^{\perp}(0)$ can reach more than 40T, while $H_{c2}^{\parallel}(0)$ exceeds present measurement fields and probably is exceeds 70T [90]. An important conclusion is that the $H_{c2}(T)$ envelope of MgB_2 can exceed that of Nb_3Sn at any temperature, even in the weaker condition of field perpendicular to the B sheets. Because grain boundaries appear to carry current independent of misorientation, round wire conductors without texture are practical, provided the limitation of the anisotropic H_{c2} is accepted. As noted above, this anisotropy actually decreases as H_{c2} is enhanced, primarily because the lower perpendicular H_{c2} increases more rapidly than the parallel value.

MgB_2 can be made by the powder-in-tube (PIT) process and many groups have made prototype wires, using both pre-reacted (*ex situ)* MgB_2 powder and mixtures of Mg and B powders, which must be reacted to MgB_2 *in situ* within the wire [12]. A particularly rapid use of MgB_2 wire has been for low thermal conductivity current leads in the ASTRA-2 satellite [91]. However, it is clear from analyses of the electrical resistivity of many samples, including wires, that bulk MgB_2 is actually very far from being fully sintered or electrically connected [33]. Porosity and wetting by B and perhaps oxygen-rich phases obstruct many grain boundaries [92,93] and porosity is still endemic in wires [12]. Yet even so, J_c exceeds 10^5 A/cm^2 at lower temperatures and fields. In well-connected thin films J_c values as high as 10^7 A/cm^2 have been reported [94]. It thus appears clear that, at least in the temperature range below about 25 K, there is neither a current density, nor an upper critical field barrier to the application of MgB_2. Thus MgB_2 wires could become a credible competitor to LTS-based wires or to BSCCO-based wires used in low-temperature (<25 K) applications. An additional quality is that the H_{c2} transition appears to be rather sharp, much more similar to low-T_c materials [95] than to HTS materials [96]. Thus high n-values and even persistent-mode magnets appear feasible, making MgB_2 a potential NMR or MRI magnet conductor.

Another advantage of MgB_2 is that the raw material costs of both B and Mg are low; reasonable estimates suggest that even for appropriately purified material, they will be several times less than those of Nb-based superconductors [97]. Thus, the principal uncertainties of the technology are the costs and difficulties of developing the powder-in-tube (PIT) composite technology or some alternative. By way of comparison, PIT technology is employed for Nb_3Sn manufacture using powders of $NbSn_2$ inside Nb tubes. Present manufacturing costs of Nb_3Sn conductors using PIT processes are about 5-6 times higher than conventional metal-working approaches to Nb_3Sn composites. Some of this is due the costs of making appropriate starting powders but a large part is also due to the small scale on which PIT-Nb_3Sn is presently made. It is worth noting that the quality and J_c of PIT-Nb_3Sn composites made by ShapeMetal Innovation are very high – conductors containing 200-500 15-25 μm diameter filaments have been made in magnet lengths [52]. Such filaments are completely decoupled, which is not at all true of the filaments in either BSCCO-2212 or –2223 multifilamentary conductors. If MgB_2 is to challenge LTS, the goal must be to replicate this quality, while keeping production costs down.

In summary MgB_2 is a most intriguing new entry into the superconductor wire arena. By making MgB_2 dirty, just as was done earlier when Ti was added to Nb_3Sn, H_{c2} can be greatly enhanced [86,87, 90,98,99]. Indeed the two-gap nature of MgB_2 adds additional versatility to the material, because the low temperature value of H_{c2} is determined by the electronic diffusivity of the dirtier band, while the initial slope of $H_{c2}(T)$ is determined by that of the cleaner band. Carbon-doping of MgB_2 appears to be particularly valuable to enhancing H_{c2} [90,98,99]. One effective way of both enhancing H_{c2} and the flux pinning strength may be by the addition of SiC nanoparticles [100]. Because the fabrication of MgB_2 can occur by conventional metal-working processes, the barrier to experimentation is low. The problems of fabricating MgB_2 conductors commercially are being addressed by Hyper Tech Research Inc. in the US [101], Columbus Superconductors [102] in Europe, Hitachi and NIMS in Japan [103] and many other labs worldwide, as a good recent survey makes clear [12]. Several groups have fabricated more than 100 meter lengths of prototype wires. Assuming costs can be kept low, competition with Nb-base LTS conductors seems likely, while competition with either BSCCO-based or YBCO-based first and second generation HTS conductors will depend on progress in expanding the temperature range where high currents are maintained in several Tesla fields.

XI. Conclusions

The overall picture of superconductor wire is diverse and developing rapidly. The LTS materials Nb-Ti and Nb_3Sn are well established in high-energy physics, commercial MRI and NMR, and many low-temperature magnet applications. Their properties are well developed and their applications well understood. Even though Nb_3Sn will be 50 years old this year, it has seen great progress recently. The very recently discovered MgB_2 has the potential to play a growing role in these applications. BSCCO-2212 wire also has advantages for ultra-high magnetic field use. Great changes are likely in the arena of electric power applications and magnet applications at temperatures above 30 K. Here the HTS materials BSCCO-2223 and YBCO are the strongest candidates, with first generation HTS wire based on BSCCO-2223 already commercially available and applied in a variety of prototypes. Indeed, the first commercial sale of HTS power equipment based on first generation HTS wire has recently occurred for dynamic synchronous condensers supplying reactive compensation in the power grid [104]. This is a harbinger of a coming revolution in the market for superconductor wire.

Tables

Table I. Target specifications for HEP conductor:

J_c (non-copper, 12T): 3000 A/mm^2

J_e (12T): >1000 A/mm^2

Effective filament size: <40 μm

Process unit size: scaleable to >100 kg and average piece lengths >10,000 m in wire diameters of 0.3mm to 1.0 mm

Wire cost: <$1.50/kA-m (12T, 4.2K)

Short heat treatment times: maximum 400 hrs; target 50 hrs for wind and react magnets

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Figure captions

Fig. 1: Representative multifilamentary conductors made from a. Nb47wt.%Ti, b. Nb_3Sn, c. Bi-2212, and d. Bi-2223 and e. MgB_2. The matrices for the conductors are high purity copper for (a), (b) and the outer sheath of (e) and pure silver for (c) and (d). The filaments of MgB_2 are surrounded by 316 stainless steel in (e). Conductors were manufactured by Oxford Instruments - Superconducting Technology (a and c), ShapeMetal Innovation (b), American Superconductor Corporation (d) and Hitachi cable in collaboration with the Natioanal Institute for Materials Science (e).

Fig. 2: Upper critical field (H_{c2}) for Nb47wt.%Ti, Nb_3Sn, and MgB_2, and irreversibility fields (H*) for Bi-2223, and YBCO. Note that the two fields are very close for Nb47wt.%Ti, Nb_3Sn, and MgB_2, (H* is 85-90% of H_{c2}) but widely separated for all cuprate superconductors. For MgB_2, Bi-2223 and YBCO the values plotted are the lower values appropriate to fields perpendicular to the strongest superconducting planes, the B planes for MgB_2 and the CuO_2 planes in the cuprates. Values plotted are the highest credible for each compound.

FIG 3: Critical current density across [001] tilt grain boundaries of YBCO-123 measured at 4.2 K and self field, from multiple studies on bicrystal boundaries. [25] The exponential fall-off of Jc with grain boundary angle creates the underlying requirement for texturing to achieve high currents in HTS wires. (Based on original work of Dimos et al.[24])

Fig 4: Critical current density of conventional and artificial pinning center composites, showing that it is possible to further enhance the current density of Nb-Ti by tuning the pinning center properties. Data is sourced from [41,43,45].

FIG 5: Bronze, powder in tube, and internal Sn conductor processes for fabricating Nb_3Sn. (Figure courtesy of M. Naus, University of Wisconsin).

FIG 6: Cost/performance parameters for filamentary Nb_3Sn conductors used for tokamak plasma containment devices (ITER and KSTAR) and for high field accelerator dipoles (D-20, RD-3, and HD-1). The trends show that cost has come down markedly with time, partly because of production experience and partly because J_c has markedly increased. J_c(12 T, 4.2K) of conductor for HD-1 is about 4 times that for ITER.

FIG 7 : Schematic of the OPIT process for making Bi-2223 superconducting wire such as that shown in Fig. 1d. Starting powder based on Bi-2212 and required Ca-Cu-O balancing ingredients is packed in a silver billet, deformed to a monofilament stack and then rebundled into a multifilament stack. The heat treatment which drives the 2212 to 2223 phase conversion occurs in two or more heat treatments separated by an intermediate rolling step which helps densify the tape.

Fig. 8 : The I_c (77K, self field) distribution from production runs of recent AMSC unlaminated first generation HTS wires based on BSCCO-2223, with 4.2x0.21 mm^2 cross section and >150 m length (courtesy of AMSC). The 170 A result corresponds to an engineering critical current density of over 190 A/mm^2.

FIG 9: Steady progress in understanding the current limiting mechanisms in BSCCO-2223 first generation HTS wires has led to continued progress in raising the current density expressed at the benchmark field of 0.1 T at 77K, at which field the tapes are always fully penetrated by the external field and self-field suppression of the J_c is negligible. The highest value sample plotted above has a strongly self-field limited critical current of 182 A(77K) in zero applied field, while its estimated self-field-free Ic is 235A, corresponding to J_c of 266 A/mm^2. (Courtesy of AMSC and U. Wisconsin-Madison)

Fig. 10: Current distribution reconstructions from magneto-optical images at 77K and 40 mT field parallel to the ab planes of two monofilament AMSC BSCCO-2223 first generation HTS wires. This imaging condition develops a current density that is within 5% of the self-field current density under normal self-field test conditions. The J_c(77K, self-field) values for the 1 atmosphere-processed tape was 39 kA/cm^2 and that for the 148 bar-processed (over-pressure-OP) tape was 48 kA/cm^2. In the left hand pair the upper dark-red-coded areas are 50kA/cm^2 but 200 kA/cm^2 in the left pair, showing that many regions exceed 200 kA/cm^2. (The blue bar in the 1 atm-processed tape is a non-measurable area in the magneto-optical imaging.) Data taken from ref [63].

FIG 11: Illustrative architectures [76] of a neutral axis YBCO coated conductor, showing the copper stabilizer, the solder joining the Cu to the YBCO and the Ni5W textured template.