Effects of Machining on Fiber-Reinforced Composites . . . *Randy Lee*

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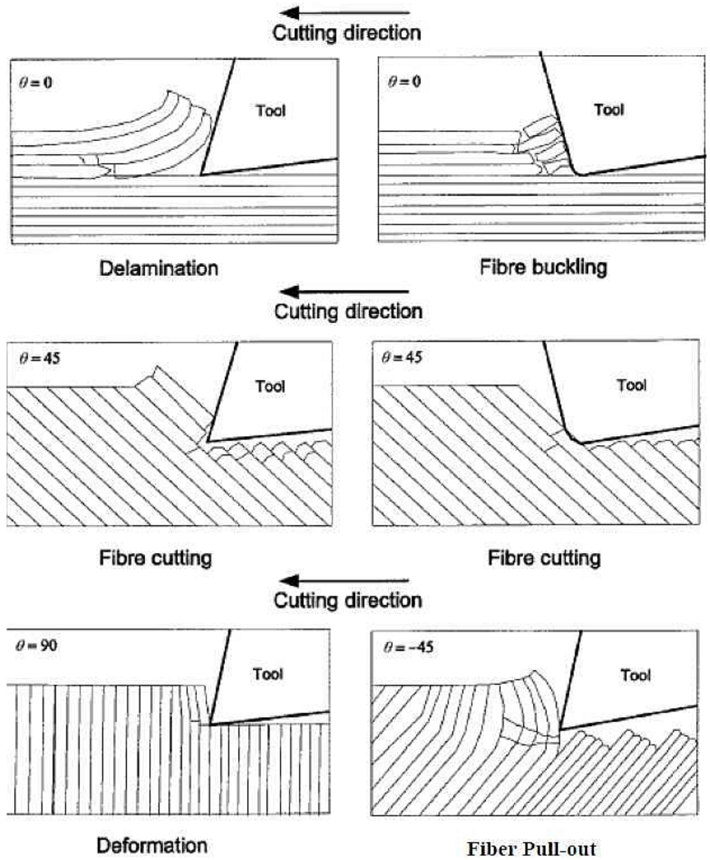
This is a formatted memo which was generated informally in response to concerns raised by colleagues regarding the effects and possible alternatives for machining composite structures, particularly smaller articles fabricated from C-C/SiC material. Many of the following points are common knowledge but are reiterated here to facilitate understanding of the problems and concerns involved during the machining of critical composite articles.

In general, the reinforcement is the structural backbone of a composite system, while the matrix provides minimal structural contributions . . . the primary function of the matrix is to hold the reinforcement fibers together (indeed, it is often referred to as the *binder*). Now, *continuous* reinforcements are the primary factor that differentiates between non-structural materials such as adhesives, putties and molding compounds, and structural reinforced composites, particularly those which are capable of bearing heavy loads and/or thermomechanical shocks (that is, in certain directions – composites are highly anisotropic materials). When the fibers or bundles become discontinuous, the structural reinforcement benefits vanish. In regions where fiber bundles have been damaged due to extraneous factors or mechanisms, the local mechanical quality is substantially degraded while the overall reinforcement integrity of the composite is weakened.

From another important aspect, the longitudinal surfaces of fibers, bundles and tow provide the greatest strengths and the hardest surfaces, while the transverse (diameter) cross-sections of the fiber/bundles are very weak and frail. Obviously, plied bundles, yarns and tow are subject to intra-bundle separations which can lead to fraying and brooming. Additionally, cross-sections of the individual fibers and filaments are vulnerable to microstructural (inter-layer) damage since the graphene basal layers comprising the turbostratic rayon and PAN carbon fibers are generally parallel to the fiber length and are easily separated, laterally. Undoubtedly, cross-sectional deterioration of fiber bundles and reinforcement tow will often lead to undesirable results and can also facilitate catastrophic failures of the composite article at some point during its history.

For these reasons, the longitudinal surfaces of continuous fibers are usually the preferred OML and IML control boundaries of an article by design. Obviously, this is not always possible. Depending on the ply angle involved, the edges (cross-sections) of reinforcement bundles and lamina ply layers can sometimes become vulnerable to interlaminar impingements, particularly damages imparted to article surfaces and edges from physical machining actions and cutting forces. The net effects of these forces depend on several factors, including the radii of the cutting tools (or the media particle edges) and the particular angles of the fiber bundles or plies (relative to the tool). This would be specifically applicable to classical rotary machining-type operations, which are prominent for composite structures. These points are effectively visualized in the following diagrams illustrating the primary modes of fiber breakage under various conditions . . .



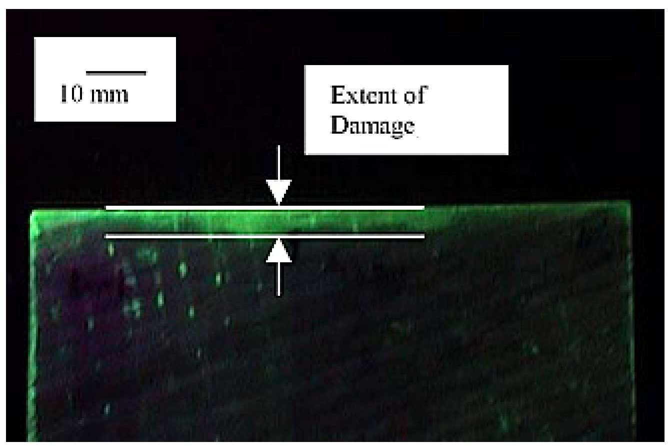
Here, the layers could represent 2-D fabric lamina or the individual filaments comprising bundles within either 2-D or 3-D composite systems. In addition to the basic mode of fiber breakage, more extensive damage can occur during the machining process which penetrates deeper into the material than is shown in these simplified illustrations. While 0° conditions can obviously result in long range delaminations, the drawings do not effectively depict the fact that non-zero degree configurations can also produce edge delaminations (separations or fractures at the fiber-to-matrix interface), or matrix cracks which initiate exclusively within the binder phase. Edge fractures are important in ultra loose weave composite systems, such as articles reinforced with some of the more common 3-D and n-D preforms. Once initiated, it is obvious that such induced fractures can only propagate inward over time. In contrast to isotropic materials, such as metals, the material removal mechanism in fibrous composite structures involves shattering rather than shearing.

In short, the machining of fiber-reinforced composites breaks the fibers to varying extents according to the corresponding failure modes involved. As an extension to these mechanisms, there are penetrating effects which usually result in some level of subsurface damage . . . and these damages are irreversible, regardless of the ply angles or tooling radii involved. In ductile polymer matrix composite (PMC) systems, failure modes due to machining effects could include fiber fracture, fiber pull-out, softening of the matrix, spalling, chipping, delamination, and burning, and may also include unseen secondary damage deeper within the composite body. These defects could manifest themselves during the machining process, during subsequent operations, during the service life of the article, or perhaps during it's maiden flight or initial burn cycle.

Obviously, with most ceramic matrix composite (CMC) systems, the matrix phase does not soften or burn during these processes. For the most part, this also applies to the current C-C/SiC articles under investigation in which fiber fracture, fiber pull-out, spalling and fiber-to-matrix fractures have already occurred on numerous occasions over the period of the program. However, the concern here is whether these defects are influenced by the machining process, or in actuality, if they are physically initiated by that process.

Now these arguments are not intended to denigrate or disregard the requirements for machining of composite materials . . . essentially all composite structures and articles must be machined and/or trimmed accordingly. However, structurally robust designs based on continuous reinforcement concepts strive to limit the machining operations and, by the proper application of mold tooling, mandrel designs and restraint fixturing, are often able to push the primary machining processes out to the periphery of the structure in the forms of trimming, sizing or tag-end removal along regions that are relatively far away from the service areas and control surfaces. With certain configurations, contours and applications, it may be possible to essentially eliminate the requirements for machining by molding and fabricating the articles to net shape and size.

In a few shops, *unnecessary* surface finishing of continuous fiber-reinforced composite structures is still practiced by some fabricators (I've seen them first hand), but this practice demonstrates a lack of fully understanding the ramifications of fiber breakage, fiber discontinuities and it seems to ignore the primary purpose of utilizing continuous reinforcements in the first place. Granted, some machining is necessary, but running a grinder or sander across molded surfaces out of habit or based on ill-conceived notions is a bad practice for these types of materials (...for boats and cars, such rough body shop techniques may be OK, but not for aerospace components). One of the primary goals in this industry should be to minimize, and if possible, eliminate all factors that contribute to broken reinforcement fibers. It is not a trivial matter.



Machined surface edge after dye penetrant test on a glass fiber composite article. Taken from, “Secondary Processing of Polymer Matrix Composites”; Inderdeep Singh, Debasis Nayak, Naresh Bhatnagar; Department of Mechanical Engineering; Indian Institute of Technology Delhi, New Delhi, India.

Again, the extent of penetration damage can be substantial and is dependent on a number of factors, particularly the specific substrate material under attack, the reinforcement orientation at the machining interface and the nature of the machining tooling or abrasive. Generally, the affected depth may encompass the outer 100 to 300 mils of the periphery or surface layers. A representative example of this type of damage is shown in the figure given to the right which illustrates the extent of damage along an edge of a common glass fiber-reinforced composite article. Here, the lamina appear to be perpendicular to the machine plane in which maximum subsurface damage is expected as the rotary tooling action applies transverse shear forces to the planes causing fiber bundle fraying, matrix microcracks and interlaminar fractures along the fiber-to-matrix interfaces. This may be typical of polymer matrix composites in general, particularly laminated systems.

In the case of the C-C/SiC pintle articles used in the Orion Launch Abort System motor valve, longitudinal surfaces of the *u-v-w* bundle planes comprise the cross-section of the pintle shaft with the *z* bundles oriented along the shaft length. In this configuration, the *u-v-w* bundles are oriented toward the circumferential machined surfaces of the shaft while their cross-sections directly interact with the machining tools. Across transverse fiber/bundle sections, damage penetration is probably maximized and could even be greater than 60-70 mils. Indeed, this level of penetration has been visually indicated in photos taken of fractured surfaces on the HT-7 pintle shaft (as evaluated in in reports).

In addition, along the shaft region, *u-v-w* fiber breakage (during machining) occurs at varying angles from 0° to 90° to □90°, along an orthogonal plane parallel to the *u-v-w* layers . . . and the mode of breakage varies from one bundle to the next. The prominent damage modes for the FMI articles may well involve fiber pull-out with matrix cracks and likely fiber-to-matrix edge fractures which occur on the tool side at 0° to □90° attack angles. Across the pintle head, *u-v-w-z* breakage expands to 3 dimensions as the tool attacks all four dimensional tows at varying angles. Again, localized fiber-to-matrix edge cracks are likely initiated and penetrate to varying degrees.

It is interesting to note that the ratio of the potential machining damage penetration depth to the shaft diameter might run 6 to 8 times greater than that for the shaft length, and this ratio becomes drastically larger along the conic head section of the pintle proceeding toward the apex. Also, it is obvious that *u-v-w* bundles nearest to the edge all along the shaft section can easily be dislodged or shifted (they are so short). In fact, length and width dimensions of the *u-v-w* fibers closest to the surface (circumference) are directly comparable to the potential depth of machining damage. Thus, it would be no surprise to discover that some of these bundles are completely ejected from the substrate during the machining process. This condition is exemplified along the surface of the head section and again, increases proceeding toward the apex. Undeniably, the apex (or orifice) region is expected to be quite prone to mechanical (and thermal) degradation, and may well be the most sensitive region for damage to occur.

From previous LAS HT work, it has already been established that these articles are loaded with high levels of porosity consisting of both open and closed pores and voids. From work I have personally done in years past via manufacturing trials and porosity testing, it has been confirmed that total pore volume fractions greater than about 8-10% substantially weaken the mechanical properties in these types of materials and often lead to detrimental effects downline, while closed pores and voids are subject to pressurized gas rupture during subsequent heating events.

Most importantly however, localized pore clusters and void agglomerations (whether open or closed) become *extremely* weak regions within the composite body and are very sensitive to mechanical disruptions. In the FMI billets, these clusters are typically localized around certain *u-v-w-z* bundle intersections and appear to be due to inadequate densification methods. All of this has been well expounded on in previous reports (for anyone who read them) and these conclusions should not really be surprising to anyone with at least a semi-technical background on the subject.

In addition, the LAS work has provided clear evidence that these porosity defects are quite prevalent throughout the billets (under the current fabrication and densification schemes), including the peripheral subsurface regions of the as-fabricated articles. It is more than conceivable that spalling or excavation could be initiated when the machining tool impacts regions of the substrate which contain pore clusters just below the surface. Obviously, the presence of such defects in the apex region is a recipe for disaster.

Contrary to certain industry perceptions, there is no practical method that can be applied to replenish fiber continuity, recoup lost mechanical integrity or repair this damage to its previous state. Once fiber breakage occurs, continuity is lost and the local reinforcement strength diminishes. No doubt, application of seal coats and surface adhesive layers during the finishing stages provide multiple benefits to the overall material system, but they cannot repair broken fibers and degraded fiber bundle cross-sections, nor can they realistically rejuvenate matrix-to-fiber delaminations.

Such perceptions may be deceptive in their apparent effects. In essence, these peripheral coating layers only seal the defects by 'fill and fare'. In all likelihood, micro pull-outs, missing bundle segments, buckled regions and/or pot holes probably exist in relative abundance across freshly machined C/C-SiC articles . . . while the coating process 'levels' the surface morphology to smoothness, as it is intended to do. However, if these types of defects are indeed present on the as-machined surfaces, the seal coat and last two PIP layers probably do little more than cover them up.

Another point to consider . . . in most 3-D fibrous networks (including the FMI articles), fiber bundles become fixed in space under *limp* conditions during the preform fabrication and molding processes (rigidization) . . . and there is no real physical interaction established between the bundles throughout the process. On the other hand, in more tightly woven reinforcements, fiber bundles physically interact due to the interlacing weaving patterns which results in interlocks, crimps, layer-to-layer nesting and intermingling among neighboring bundles. This provides a combinatorial effect that is substantially greater than a loosely compiled 3-D array of non-interacting bundles, such as that comprising raw FMI preform billets. Also, for woven fabric reinforcements, the bundles are not so limp . . . crimps and weaving patterns often impart low levels of longitudinal tension along the bundles, and the process of fabric lay-up and composite fabrication generally increase these effects (wrapping, laying, stretching, winding, etc...).

To make matters worse, articles extracted from FMI-produced billets are undesirably low in fiber volume fraction. This is partially reflected in the high open porosity reported by FMI (i.e... ~13% . . . but my experiences teach that the *total* porosity is sometimes closer to twice this level). These points have been well broadcasted and recognized during the last three years. Presumably, you and everyone on your team understand the full ramifications of composite structures comprised of low fiber volumes. However, allow me to reiterate a couple of comments here.

Generally, composites with a minimum of about 50-60% (v/v) continuous fibrous reinforcement are required to meet the requirements for structurally robust articles. In actuality, most structural composites, particularly those bound for aerospace and high performance applications, have fiber volumes in the 65-75% range. It is well substantiated that systems with fiber volumes less than about 40-45% are poor structural candidates. Generally, these types of articles should not be used in demanding performance applications which impose high loads and shocks . . . in this respect, *traditional 3-D articles simply cannot compete on the same level as the higher fiber volume 2-D systems*. Low fiber volumes (and the corresponding low mechanical strengths) are the most pervasive weakness inherent to composite systems which are based on classical pre-woven 3-D architectures. This is their primary shortcoming.

In classical 2-D laminated composites, longitudinal (orthotropic) mechanical properties are quite exceptional (many times greater than 3-D systems), while their interlaminar strengths (ply-to-ply interactions) are undesirably weak. This has always been the weakness of 2-D systems, and it still is today. Alas! . . . 3-D architectures were invented which were supposed to eliminate the concept of 'interlaminar' altogether. However, it was soon discovered that these systems could not replace 2-D lamina because they demonstrated substandard mechanical properties in all directions. In short, these weaknesses are a result of low fiber volume, longitudinal bundle limpness, excessive and poorly distributed porosity. While the latter factor may affect certain 2-D systems, the former two have long been overcome.

Modern approaches for multi-dimensional weaving usually center around more elaborate (and expensive) equipment for generating the preforms (or billets) which is able to improve fiber interactions and net fiber volumes . . . while imparting z-directional (pseudo-interlaminar) reinforcement properties. For larger structures, stitching, needling and fiber placement methods have gained prominence in recent years for producing 3-D reinforced networks utilizing pre-impregnated tow and yarns or dry preform structures which are concurrently impregnated during the fabrication process. Some of these techniques may also decrease tow limpness, permitting greater realization of the fiber's potential tenacity. However, needles and stitches grossly damage the existing fibers during insertion and many of the continuous fibers are completely broken in the process.

By themselves, needling and stitching are not weaving operations . . . they essentially poke holes into an existing 2-D weave, warp-fill preform or fabric lay-up. However, stitching offers greater fiber continuity and network compactability, as well as various stitching styles with interlock and outerlock mechanisms. Needles are just short straight fibers which are injected into the fabric . . . but they may be 'barbed', permitting micro-interlocks with the 2-D fibrous substrate. Since fibers are irreversibly damaged during insertion, both processes degrade the mechanical properties of the original 2-D structure. A balance must be struck between the density of z-directional fibers incorporated into the material and the minimal 2-D mechanical loss acceptable.

For the smaller articles however, the challenges are greater, especially when one or more of the article dimensions is small and directly comparable to the dimensions of the fiber bundles, the pore volume clusters and . . . the depth of potential machining damage. I would suspect that the shaft diameter of FMI pintles is borderline in this regard and the upper apex region is below the threshold. This is what the evidence seems to indicate.

For these articles (and others like them), there are indeed specific techniques and general approaches which can increase the fiber volume fraction, decrease and better distribute the porosity fraction, and reduce or possibly eliminate the machining requirements. Personally, I know of a multitude of improvements and innovative techniques which could be applied in these respects . . . many have already been directly proven to work for these types of materials and some would be developmental (but highly feasible). My strongest resources lie in the areas of molding/rigidization, densification and coating methodologies.

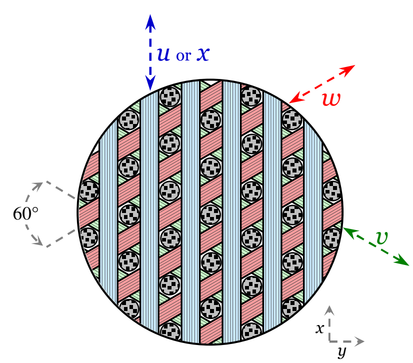
Some of my approaches have already been conveyed in previous reports and will not be rehashed here. To be certain, I spent many years proving out, optimizing and perfecting these techniques and frankly, I should be more cautious in treating such concepts as personal proprietary (fortunately, no one on the LAS team has taken me seriously or placed much importance in those previous writings). By *my* standards however, FMI currently utilizes inferior methods for densification and sealing, and there are specific techniques which could be implemented to make these processes more effective and consistent. If MSFC ever acquires the capability to manufacture these types of articles, I would be glad to demonstrate the superior properties and performance that I could achieve with MSFC-produced pintles, pintle guides, throat inserts and exit cones. At this point however, I would be reluctant to divulge these ideas to any business entity without careful consideration, even if you guys had the opportunity and really *were* serious about improving the current methodologies.

As you should already know, I am not a mechanical engineer by training or by experience. While I have addressed an abundance of mechanical issues over the years and have taken several ME-related graduate courses, when faced with difficult mechanics problems, I relish the opportunity to call upon my ME colleagues for assistance who are much more focused and skilled in those areas than I will ever be. Admittedly, my expertise is in the chemistry and physics of these materials. I have worked with composites since the late 1960s and started experimenting with 3-D preforms and braids produced by FMI, Stackpole and Techniweave in the early 1980s (hands-on manufacturing development and C/C densification). Additionally, I have been directly involved with a number of machining approaches including a variety of rotary methods and some of the more modern water-jet methods, and I have conducted thousands of mechanical tests with heavy emphasis on test method development.

However, I would hardly refer myself as a qualified consultant in the areas of mechanical shape design or composite machining techniques. Obviously, an astute ME with several years of solid experience in the machining of carbon fiber reinforced CMC structures would be a good resource regarding the most appropriate methods for sectioning and machining these billets (that is, one with direct machine shop experience, not PowerPoint bodhi). This person would know, first hand, how to minimize fiber damage during the various machining approaches and could probably explain the ideas I have attempted to outline here with much greater clarity and correctness.

Since you did ask about machining methodologies (in response to my previous comments), I will end with a couple of thoughts on that topic. It may be feasible to produce near net-shaped dry preform articles using 3-D *braiding* techniques which could either reduce the machining requirements to minor trimming operations or, if woven appropriately, could eliminate the machining process altogether. Admittedly, FMI (or its subsidiary, Intermat) appears to offer some rather archaic approaches in the areas of 3-D braiding and their so-called 'N-D' preform configurations. They are not necessarily the leading company in this field.

It may be fruitful to consider some of the newer, more modern techniques for 3-D construction available from companies like 3TEX. I am not sure whether they can fabricate pintle-like preforms (I have not contacted them in this regard), but these guys appear to have some very precise, state-of-the-art capabilities for producing complex-shaped preforms with fiber volumes in the 65-70% range ! If the opportunity ever arises, this is one source that is definitely worth looking into. It offers the potential to submit net-shaped rigidized preform articles (rather than billets) for densification processing which are comprised completely of continuous fiber-reinforced backbones in all directions with very high fiber volumes and whose machining requirements are essentially nil.



Cross-section of C-C/SiC pintle shaft showing u-v-w planes and bundles (the z-direction is the shaft length running in and out of the paper).

In essence, machining of composites containing *chopped* randomly oriented fibers or mats is trivial . . . but the machining of systems which are reinforced with networks of continuous fibers and fabrics is quite critical.

Articles comprised of C-C/SiC (carbon-carbon/silicon carbide) are modified (bimatrix) CMC systems which start out as 3-D woven fibrous preform billets comprised of four primary fiber bundle orientations, *u*, *v*, *w* and *z*. Each *u*-*v*-*w* bundle set forms a plane where the bundles are oriented at 60° to one another and the planes loosely stack on top of each other along the *z* direction with *z*-directional bundles permeating through *u*-*v*-*w* bundle intersection gaps at a density or spacing of 0.05" between *z* bundles. In essence, the effective fiber volume fraction in the *u*-*v*-*w* plane is different that the effective fiber volume in the *z* direction, but the overall fiber volume of the bulk preform is in the 40-45% range. This forms a very loose 3-D network which is not a true woven structure since none of the bundles actually interact with one another in the form of interlocks, crimps, bends or any other weaving features. All the *u*-*v*-*w*-*z* bundles progress through the preform network as straight reinforcements which only slightly contact one another leaving large voids at all the intersections.

The dry fibrous preform billets are first rigidized, which is probably accomplished via a very gentle impregnation or exposure to a low viscosity thermosetting resin which is then cured/crosslinked and pyrolyzed. This leaves a residue or spotty coating of glassy carbon on the fiber surfaces which comprises the stiff rigidization interphase and becomes the first carbon phase incorporated into the system.

Earlier, C-C/SiC was described as a modified CMC system. Rather than a single homogeneous matrix phase as with most PMC and CMC systems, the C-C/SiC system is 'modified' in the sense that the total matrix actually consist of several phases. In simplistic terms, a C-C phase and SiC phase would comprise a 'bimatrix' system. However, other matrix phases are present in current C-C/SiC under study and should also be accounted for in order to provide a correct description. First, after the carbonization, the polymer used to rigidize the dry preform structure deposits a layer of glassy carbon residue.

To be continued . . .